Weld metal hydrogen-assisted cracking of X70 linepipe steel using E6010 electrodes

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WELD METAL HYDROGEN-ASSISTED CRACKING OF X70 LINEPIPE STEEL USING E6010 ELECTRODES

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ABSTRACT

Manual metal arc welding (MMAW) with cellulosic consumables is still widely used in Australia for field girth welding during the construction of small diameter gas transmission pipelines. Decomposition of the cellulosic electrode coating during welding introduces high levels of hydrogen into the arc and the resulting welds are potentially susceptible to weld metal hydrogen-assisted cold cracking (WMHACC).

Although extensive weld metal hydrogen-assisted cold cracking has not been reported to date, WMHACC is not only possible, but likely under conditions promoting high weld hydrogen concentrations and restraint stresses. The mechanism of WMHACC in cellulosic weld deposits is, however, complex and the factors that influence the initiation and propagation of cracks in the weld metal not well understood to date.

Despite the high likelihood of cracking, successful preheat-free root pass welding of thin-walled API 5L X70 pipelines with E6010 cellulosic electrodes is possible, provided there is strict adherence to the guidelines published in the Australian Standard for pipeline welding, AS 2885.2. This project aims to expand the guidelines given in AS 2885.2 in order to provide the pipeline industry with more confidence that WMHACC can reliably be prevented during the root pass welding of X70 with E6010 electrodes.

In order to determine the effect of welding parameters on WMHACC in root pass welds, a modified version of the Welding Institute of Canada (WIC) test was used. The joint geometry entailed a single V-preparation with an included angle of 60°, and a root gap and root face of maximum 0.8 (± 0.2 mm). The experimental welding parameter matrix was selected to encompass the parameters used in qualified welding procedure specifications in industry at present. The welding current was varied from 120 A to 160 A in increments of 10 A. At each current level, the heat input was varied from 0.3 kJ/mm to 1.0 kJ/mm in increments of 0.1 kJ/mm by adjusting the welding speed from a minimum of 172 mm/min to a maximum of 725 mm/min. This test matrix was repeated three times: without preheat, and at preheat temperatures of 50°C and 80°C.

Hydrogen cracks in the weld metal were observed to initiate exclusively at the wagon tracks at the toes of the weld bead. Cracks propagated downward through the weld metal to stress concentrations located near the root of the weld. The direction of crack growth was, for the main part, through the weld metal with a few exceptions deviating through the heat-affected zones. WMHACC was observed throughout the entire heat input range tested (0.3 to 1.0 kJ/mm), especially in welds deposited without preheat. Preheating reduced the number of cracked samples.

The WMHACC behaviour of the welds was found to be dependent on the heat input, the preheat temperature and the welding current. When WMHACC was viewed simply as a function of the heat input and the preheat temperature, only a small portion of the welding parameter matrix could be regarded as safe. It was concluded that WMHACC can be more accurately explained by isolating the effect of individual welding parameters. The welding current affects the arc force, which in turn influences the weld bead geometry. Shallow, incomplete penetration was observed at low welding currents, introducing an additional stress concentration at the weld root that increased the risk of WMHACC.

Statistical models of the predicted probability of WMHACC as a function of the welding current and welding speed were developed. These models can be used to determine the probability of WMHACC based on the selected welding parameters (welding current, travel speed and preheat temperature).
No correlation was found between welding parameters and solidification (hot) cracking in the current study. It was therefore concluded that the experimental setup used in this investigation was not appropriate to assess the influence of welding parameters on the incidence of solidification cracking during root pass welding of X70 pipeline steel using cellulotic E6010 electrodes. It is recommended that future work focuses on establishing the relationship between solidification cracking and welding parameters using more appropriate tests.

The weld bead geometry, as influenced by the arc force and welding current, in turn influenced the minimum cooling rate, heat input and weld metal hardness required to ensure crack-free weld metal. At lower welding currents, the more severe stress concentration at the weld root due to incomplete penetration decreased the cooling rate required to prevent crack formation. The highest resistance to WMHACC was observed at a welding current of 160 A, the highest current level used in this investigation. This can be attributed to the deeper penetration and thicker weld beads observed at higher current levels.

Heat input and preheat temperature were not found to affect the metallurgical transformation products observed in the microstructures of the weld metal and heat-affected zones to any significant extent. Two forms of primary ferrite were observed in the weld metal, namely allotriomorphic and idiomorphic ferrite. Other forms of ferrite observed in the weld metal included acicular ferrite, Widmanstätten ferrite and upper bainite. The weld metal also contained ferrite-carbide aggregates tentatively identified as degenerate pearlite, granular bainite and lower bainite. Upper bainite dominated the high-temperature heat-affected zone microstructures. Lower bainite, ferrite-carbide aggregate and primary ferrite were also observed. In both the weld metal and heat-affected zone, a gradual coarsening of the prior austenite grain size and the transformation products was observed with an increase in heat input and preheat temperature. More diffusion-controlled transformation products were found in welds that experienced slower cooling rates. Microcracks were observed to propagate through the hard and brittle carbon-enriched microphases and continuous cementite seams, facilitating crack propagation in microstructures formed at faster cooling rates.

Non-metallic inclusions (NMI’s) were mainly found to be spherical, incoherent Mn-Si-Al-Ti-O oxides. The number of these inclusions, the total area covered by these particles and the volume percentage decreased with increasing heat input, while the average inclusion size increased. Non-metallic inclusions reduce the ductility of weld metal, act as irreversible hydrogen trapping sites and can influence the weld metal hardness. Microcracks observed ahead of macrocrack tips showed that crack propagation was associated with the non-metallic inclusions in the microstructure.

Similar trends in residual stress were observed in all welds examined. The residual stress distribution in all three directions (transverse, longitudinal and normal) is strongly affected by the geometry of the weld bead. The residual stress peaked in the centre of the welds, followed by a rapid decrease in the vicinity of the wagon tracks and another increase to a peak in the base metal corresponding to the edge of the single-V joint preparation where the base metal again assumes its full thickness. The decrease in residual stress observed at the wagon tracks can be ascribed to a lower resistance to yielding on cooling due to the smaller cross-sectional area of the weld metal in this area. The heat input and preheat temperature did not affect the residual stress distribution in any significant way.

**KEYWORDS:** WMHACC, X70 pipeline steel, E6010 electrodes, welding parameters, Modified Welding Institute of Canada Test (MWIC), probability of cracking, residual stress of welding, non-metallic inclusions
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<tbody>
<tr>
<td>PRCI</td>
<td>Pipeline research council international</td>
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<tr>
<td>EPCRC</td>
<td>Energy pipelines cooperative research council</td>
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<tr>
<td>UOW</td>
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<td>WMMHACC</td>
<td>Weld metal hydrogen-assisted cold cracking</td>
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<td>HACC</td>
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<td>WM</td>
<td>Weld Metal</td>
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<td>HAZ</td>
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CHAPTER 1

BACKGROUND

1.1. INTRODUCTION

Domestic gas markets in Australia tend to be small and located far from natural gas sources. Gas is therefore transmitted over long distances, and the norm for gas transmission across Australia is small diameter (less than DN500 or outside diameter of 508 mm), thin-walled high-pressure pipelines. This makes the Australian pipeline industry somewhat unique internationally. Reducing pipeline construction costs is critical in justifying gas transmission pipeline projects and Australian pipeline construction practices therefore focus on maximising productivity. Manual metal arc welding (MMAW) with cellulotic consumables is still considered the most economical method of joining small diameter pipelines with wall thicknesses up to 10 mm. Pipeline construction in Australia is therefore dominated by welding of API 5L X70 pipe with cellulotic Exx10 electrodes at high welding speeds and removal of line-up clamps after 50% to 70% completion of the root bead [1,2].

Based on cost and productivity considerations, there is still a strong preference in current Australian pipeline construction projects for the use of manual (shielded) metal arc welding (MMAW or SMAW) with current-controlled power sources [3,4]. Welding of the root pass is the rate-controlling step in pipeline construction and cellulotic consumables are preferred for the root and hot passes due to the deep penetration, forceful spray-type arcs, good arc stability, keyhole capabilities and reduced risk of defects such as slag entrapment and porosity that can be realised [5]. The process is also well suited to accommodating poor pipe fit-up [2,6,7], and allows the use of high welding speeds. This allows the front end of pipeline construction to proceed more rapidly.

Lower strength cellulotic electrodes, classified as E6010 in accordance with AWS A5.1/A5.1M:2012 [5], are routinely used in Australia for root pass welding of API 5L X70 linepipe steel to achieve high toughness and to reduce the likelihood of cracking [8]. Decomposition of the cellulotic electrode coating during welding, however, introduces high levels of hydrogen, up to and even exceeding 40 ml H₂ per 100 g of weld metal [9,10], into the arc and the resulting welds are potentially susceptible to hydrogen-assisted cold cracking (HACC). This cracking phenomenon is associated with loss of ductility at near-ambient temperatures caused by a diffusible solid solution of hydrogen atoms within the crystal structure of the weld metal and surrounding heat-affected zone (HAZ). Preheating can reduce the cooling rate after welding sufficiently to eliminate the risk of cracking by allowing some of the absorbed hydrogen to effuse out into the atmosphere, but the use of preheat is expensive and lowers production rates considerably [11].

The Australian standard for pipeline welding AS 2885.2:2016 considers the risk of weld metal hydrogen-assisted cold cracking (WMHACC) as ‘remote’ when X70 is welded preheat-free in wall thicknesses up to 10 mm, provided the heat input remains above 0.5 kJ/mm, the carbon equivalent is limited to a maximum of 0.40, and both lifting practice and delay times between the start of the root pass and the start of the hot pass are carefully controlled [8]. These welding practices (described in more detail in AS 2885.2-2016 and in Chapter 2) have been used in Australia for many years, and thousands of kilometres of pipeline have been constructed successfully with few issues related to hydrogen-assisted cold cracking.
New participants in the Australian pipeline industry are, however, often more familiar with heavier wall thicknesses, slower welding speeds associated with the use of preheat, extended line-up clamp hold times and welding with low hydrogen consumables, and may not have confidence in existing Australian pipeline construction practices. The Australian approach of welding high-strength pipelines preheat-free with fully cellulosic procedures and high travel speeds is fairly unique internationally. As explained in a review by Roshan et al. (2003) [2], new companies lacking in Australian experience may not follow established procedures when qualifying welding procedure specifications, potentially leading to longer construction times and more expensive low-hydrogen practices.

Since compliance with qualified welding procedures (WPS) is a mandatory quality assurance requirement for fit-for-purpose pipeline girth welds, any WPS imposing unnecessary restrictions on hydrogen control can result in dilution of the economic benefits by imposing over-compensating practices. Roshan et al. showed that welding practices such as preheating, removing the line-up clamp only after completion of 100% of the root pass, reducing welding speeds, or specifying hybrid procedures where the root and hot passes are deposited using cellulosic consumables, while the fill and cap passes are deposited with low hydrogen electrodes, increase the project duration while significantly raising the cost of construction [2].

The above authors showed that for the construction of a 943 km pipeline, imposing preheating practices on top of existing welding practices would delay the project by three days with an associated increase in cost of 13%. Line-up clamp removal at 100% root bead completion increases the project duration by another 22 days with a 9% increase in cost. When welding is carried out using low hydrogen electrodes (E9018), an increase in the project duration of 83 days and 42% higher costs are expected. When all the above constraints are imposed simultaneously, an increase in project duration of 83 days is expected with a total increase in cost of 61%, without any observable change in the incidence of WMHACC [2].

The aims of the current investigation were therefore formulated to lend confidence in current pipeline welding procedures by examining those welding parameters that lead to WMHACC, and to define a safe operating window within the confines of the parameters already used in welding procedure specifications in the field. Although current industry practice is regarded as reasonably safe, the risk of cracking has not been eliminated to date, and the current study aims to define a method for estimating the probability of cracking based on existing and pre-defined risk factors.

When high-strength pipeline steels are welded with cellulosic electrodes, hydrogen-assisted cold cracks often initiate within minutes in the root pass because of hydrogen supersaturation in the weld metal [9]. AS 2885.2 suggests that in modern high-strength pipelines, weld metal hydrogen-assisted cold cracking (WMHACC) may be more likely than heat-affected zone hydrogen-assisted cold cracking (HAZHACC) [8]. The shift from HAZHACC towards WMHACC is attributed to the development of thermo-mechanically controlled-rolled processing (TMCP) of steel specifically tailored to reduce carbon equivalent values, and consequently increase the toughness and weldability of steels [12]. These high-strength low-alloy (HSLA) steels do not attain their strength primarily from alloying, but from a highly refined grain size (resulting from controlled rolling and cooling), transformation strengthening, micro-alloying, precipitation strengthening, transformation strengthening, and tempering after quenching [13-17]. The introduction of API 5L X70 HSLA steel for gas transmission pipelines was prompted by the need for high strength, excellent toughness and good weldability to withstand the high operating pressures in small-diameter, thin-walled pipes [12].
As shown in Figure 1.1.1, four requirements have to be satisfied simultaneously for hydrogen-assisted cold cracking to occur in steel during welding: the presence of a critical amount of diffusible hydrogen, a crack-susceptible microstructure, a critical tensile stress, and a temperature near normal ambient (generally below 200°C). This translates to the diffusible hydrogen content dissolved in the weld metal (absorbed from the arc plasma), both the toughness and fracture toughness of the weld metal (derived from the microstructure) as defined by the resistance to crack initiation and crack propagation, respectively, and the combination of stresses to which the joint is exposed. These stresses include shrinkage stresses, residual stresses and external stresses due to lifting, lowering, and any irregular handling during normal placement and critical tie-ins [18].

![Diagram](image)

**Figure 1.1.1** Factors causing HACC in steels [18].

A crack susceptible microstructure is considered the least critical prerequisite for determining susceptibility to WMHACC, and weld microstructures regarded as having a low susceptibility to hydrogen-assisted cold cracking, such as acicular ferrite, can develop cracks when the hydrogen concentration, local stress intensity and temperature favour crack initiation [19].

Hydrogen is inevitably present during arc welding, and the risk of cracking generally increases if more hydrogen is present in the weld metal. Cellulosic welding electrodes typically contain up to 30% cellulose in the form of purified wood pulp in the coating, which volatilises and combusts in the arc to produce CO₂, CO, H₂, and water vapour [5]. Partial dissociation of molecular hydrogen (H₂) in the arc plasma to form atomic hydrogen (H) raises the hydrogen solubility in the liquid weld metal and leads to rapid hydrogen dissolution in the weld pool [20]. During subsequent cooling and solidification, the hydrogen solubility decreases rapidly and much of the absorbed hydrogen escapes from the solidified weld by diffusion.

The heat input is, however, characteristically low during root pass welding of pipeline steels, mainly due to the high welding speeds required to increase production rates. Lower heat input levels give rise to fast cooling rates, which reduce the time available for hydrogen diffusion and lead to substantial supersaturation of hydrogen in the weld metal on cooling. When welding high-strength pipeline steels with cellulosic electrodes, root cracks may often initiate within minutes, since the root pass is saturated with hydrogen and significant local accumulation of hydrogen is not needed before the onset of cracking. Therefore, an unspecified but wide time interval exists over which HACC cracking can initiate. The greatest risk of HACC is therefore inherent in the welding of the root pass [21].
1.2. **AIMS OF THE INVESTIGATION**

The current study aims to expand the guidelines given in AS 2885.2 in order to provide the pipeline industry with more confidence in the root pass welding of X70 pipeline steel with E6010 electrodes with regard to WMHACC. To more accurately define the conditions required for avoiding WMHACC in root pass welds, an experimental matrix incorporating different arc welding parameters and preheat conditions was developed and the resulting welds tested for WMHACC using the modified Welding Institute of Canada (MWIC) (as discussed in detail in Chapter 3 and 4 of the current study) test for an API 5L X70 grade pipeline steel. The project also aims to improve current understanding of the underlying mechanism of WMHACC and contributory causes to cracking.

This study, designated RP1-04, forms part of the suite of Energy Pipelines Cooperative Research Council (EPCRC) Program 1 research projects, and based on expert industry partnerships defined the following outcomes, aiming to understand the conditions under which WMHACC occurs during the root pass welding of X70 pipelines with E6010 electrodes:

- The main aim of this investigation is to define safe welding parameter envelopes for avoiding WMHACC. The boundaries between safe and unsafe conditions are referred to as crack/no crack boundaries, with the experimental welding parameters selected coinciding with those used in the Australian pipeline industry.
- The project also aims to statistically model the crack/no crack boundaries to assess the probability of cracking within regions that are regarded as “safe”, yet are located close to the boundary. This modelling approach will lend confidence in the positions of the crack/no crack boundaries since the possibility of cracking is always present, especially during extreme lifts or when unintended welding defects are present.
- A further aim is to add to the limited guidelines on the avoidance of WMHACC during the root pass welding of X70 pipelines with E6010 electrodes, described in part 2 (section E9.2) of Australian Standard AS 2885.2 for the construction on gas and liquid petroleum pipelines.

Secondary aims of the investigation were to investigate all contributory factors increasing the risk of WMHACC and to assess how they are influenced by the welding parameters selected for the investigation. These aims include:

- Investigating the variation in weld metal cooling rate with heat input during welding and preheat temperature in terms of the cooling times $\Delta t_{8/5}$ and $\Delta t_{100}$, defined as the time it takes for the weld metal to cool between 800°C and 500°C, and between 800°C and 100°C, respectively. The first influences the microstructure that forms on cooling, the hardness of the weld metal, the presence and characteristics of non-metallic inclusions, and under certain conditions, the residual stress in the welded joint. The latter determines the amount of hydrogen able to effuse from the weld metal on cooling.
- Investigating the weld metal and heat-affected zone hardness as a function of heat input and preheat temperature and defining mathematical models that can be applied to estimate the expected hardness based on heat input. The investigation focused on refining these mathematical models to include the hardness purely as a function of the cooling rate. This accounts for scaling problems between the cooling rates of the MWIC test pieces used in the current investigation and those of actual pipelines during field welding.
- Investigating the transformation products in the weld metal and heat-affected zone of welds as a function of heat input and preheat temperature. This includes identifying the metallurgical phases present in the
room temperature microstructures and assessing their influence on the WMHACC behaviour observed during the formulation of the crack/no crack boundaries.

- Investigating the nature of non-metallic inclusions observed in the weld metal through high-resolution imaging and chemical analysis, and assessing the influence of heat input on the risk these inclusions pose to the incidence of WMHACC.

- Investigating how the heat input, cooling times, weld metal and heat-affected zone hardness, and non-metallic inclusions vary along the statistically modelled crack/no crack boundaries in order to assess the parameters that result not only in uncracked welds, but those parameters combinations that create the highest resistance to WMHACC.

- Investigating the residual stress caused by restraint inherent in the MWIC test pieces during welding and assessing the influence of heat input and preheat temperature on the stress states in the transverse, longitudinal and normal directions across the welds.

- Investigating the fracture surfaces of cracked welds in order to gain a better understanding of the mechanisms involved in crack initiation and propagation, and assessing how weld metal characteristics such as stress concentrations, microstructure and non-metallic inclusions influence WMHACC.

Figure 1.1.2 is a schematic diagram summarising the welding-related factors that affect the incidence of WMHACC in the root pass of X70 pipeline steel welded using E6010 electrodes. It includes factors affecting the WMHACC of MWIC test pieces and actual pipeline girth welds. Investigation of all these factors were considered to be outside the scope of this project. Those factors included as test variables or those that were measured directly during the course of the current investigation are indicated by the chapter numbers in which they are addressed, while those not included in the investigation, either as variables or direct measurements, are indicated in red. Figure 1.1.2 serves as a road map for the current study and as a causal tree of WMHACC, highlighting those factors affecting the cracking behaviour as observed in this study and drawing from the findings of other authors on the topic.

Chapter 2 provides more information on current Australian pipe welding practices, the mechanisms of hydrogen-induced cold cracking and research published to date in open literature.
Figure 1.1.2 Welding-related factors affecting WMHACC susceptibility as a road map to the current study with relevant chapters indicated where individual factors are discussed in terms of the results. All factors highlighted in red were not directly investigated or measured during the course of this project.
1.3. REFERENCES


CHAPTER 2

LITERATURE REVIEW

Although not widely acknowledged, WMHACC during root pass welding of API 5L X70 with cellulosic consumables may be more prevalent than reported in published literature [1]. Indications are that WMHACC is not only possible, but likely under welding conditions favouring high weld hydrogen concentrations and restraint stresses. The mechanism of WMHACC in cellulosic weld deposits is, however, complex and the role of the different factors that influence the initiation and propagation of cracks in the weld metal not well understood at present.

Despite the high likelihood of cracking, successful preheat-free root pass welding of thin-walled pipelines in API 5L X70 steel with E6010 electrodes is possible, provided strict adherence to qualified welding procedures is ensured. As described in Chapter 1, the aim of the current investigation is to minimise the risk of cracking during preheat-free root pass welding by optimising the welding parameters used in field girth welding.

The literature presented in this section provides an overview of the current state of pipeline welding in Australia with regard to WMHACC and presents the background necessary to justify the experimental study presented in the following chapters.

2.1. CURRENT FIELD GIRTH WELDING PRACTICE IN AUSTRALIA

A hydrogen-control approach is generally taken during the welding of high-strength low-alloy steels whenever there is a risk of HACC. This approach entails the use of low-hydrogen consumables, preheating the joint to specified minimum temperatures, maintaining a minimum interpass temperature, and post-weld heating to reduce the amount of hydrogen in the joint and decrease the risk of HACC [2]. This approach is, however, time consuming and expensive. Production time is a major constraint in the context of Australian pipeline welding, and as described in Section 1.1, pipeline welding practices in Australia have been optimised over many years to yield high production and low repair rates.

Based on cost and productivity considerations, there is a strong preference in Australian pipeline welding operations for MMAW with current-controlled power sources [3,4]. Welding of the root pass is the rate controlling step, and cellulosic electrodes are preferred over low-hydrogen consumables for the root and hot pass on account of their good arc stability, excellent penetration, keyhole capabilities and the reduced risk of defects. These procedures are also well suited to accommodate poor pipe fit-up [1,5,6]. For illustration, schematic depictions of the typical front-end pipeline construction procedure used in Australia are shown in Figures 2.1.1 and 2.1.2 [6,8].

Figure 2.1.1 shows the typical setup and steps describing how each individual pipe section is fitted during root pass welding. The entire pipeline is supported on wooden skids above ground to ease access to the joint by the welders depositing the root, hot, and fill passes. The pipe section being fitted for welding is lifted with a crane and positioned over an internal line-up clamp which, in addition to wedges placed within in the root gap, is used to maintain proper fit-up.
Skids are used to support the pipeline above the ground.

An internal line-up clamp is positioned and clamped to the front end of the leading pipe.

A crane inserts the next pipe over the line-up clamp. Wedges are used to correct the root gap and the pipes are restrained with exterior clamps.

The root pass is partially welded with cellulosic consumables.

The front end of the pipeline is lifted by a crane, while a support skid is placed underneath the leading pipe.

The pipe is lowered onto the skid and the front-end team moves to the next weld.

**Figure 2.1.1** Step-wise illustration of the typical mainland pipeline construction procedure followed in Australia [7], reproduced from [6].

The crane holds the new pipe section in place while the welders deposit the root pass with E6010 electrodes. After 50 to 70% completion of the root pass, the crane lifts the pipe section while a support skid is prepared and placed underneath the pipe, which is then lowered onto the skid.

After the welded pipe section is lowered onto the skid, the next pipe section is lifted and fitted using a crane and the line-up clamp is pulled through to the location of the next joint (as shown in Figure 2.1.2). This process is repeated at the front-end of pipeline construction with the welding team depositing the root pass following closely behind. A different team depositing the remainder of the welding passes follows in their wake [8].

**Figure 2.1.2** Schematic illustration of the lifting and lowering operation during the front-end of the mainland pipeline construction process in Australia [8].
Figure 2.1.2 also highlights the slight bending of the pipe during the typical lifting and lowering operations when the support skids are inserted. This induces stresses in the newly deposited root pass that are exacerbated by the fact that only 50 to 70% of the root pass is completed before the lifting and lowering operation. For the duration of the current study these stresses will be referred to as lifting and lowering stresses. Welding is performed manually and progresses in a downhill manner using MMAW and low strength cellulosic consumables. Consumables strength is under-matched for increased toughness in the root pass. Figure 2.1.3 outlines the welding sequence typically followed by the welders. The first welder starts at the 9 o’clock position and welds downhill towards the 6 o’clock position while the second welder welds from the 12 o’clock position towards the 3 o’clock position. The first welder then moves to close the gap between the 12 o’clock and 9 o’clock positions while the second welder closes the gap between 3 and 6 o’clock. Once 50 to 70% of the root pass is complete, the clamp is released and the pipe is positioned by lifting and lowering off onto a support skid [3,9]. Since the time between consecutive lifting and lowering of the joined pipe segments determines the productivity of pipeline fabrication, lifting is done before completion of the root pass. At this stage, the root pass should have sufficient hot ductility and strength to accommodate the lifting operation without cracking [10,11].

![Welding sequence followed by Welder 1 and Welder 2](image)

**Figure 2.1.3** Welding sequence followed by two welders depositing the root pass during pipeline girth welding.

Cracking of the root pass normally appears near the 6 o’clock position, initiating at weld discontinuities. The 6 o’clock position, as indicated in Figure 2.1.3, is considered a critical section of the root pass and often exhibits a more severe stress concentration due to possible surface irregularities and coarsening of the metallurgical microstructure. It contains the last weld metal to be deposited and the position where the welding arc is extinguished, and therefore the final portion of weld metal to solidify, and consequently must accommodate the highest amount of strain due to solidification shrinkage [10]. This part of the weld also experiences the highest stresses during the lifting and lowering operations [3]. The introduction of stress concentrations due to misalignment, careless handling during lifting, and irregularities such as incomplete fusion or lack of penetration may further promote cracking [12].

The use of low heat inputs and fast welding speeds increases productivity, but also introduces a high cooling rate and suppresses hydrogen effusion from the root pass, which favours HACC. With a higher concentration of diffusible hydrogen, the risk of HACC increases and cracks may appear faster [12,13]. To mitigate the detrimental effects of the rapid cooling rates experienced during root pass welding, deposition of the hot pass has to take place within a specified time following completion of the root pass. The hot pass not only reduces
Electrode manufacturers often recommend a maximum delay time between root and hot pass welding to reduce the cooling rate sufficiently to allow enough hydrogen to effuse out of the weld metal. Böhler recommends that deposition of the hot pass occurs within 10 minutes of root pass completion, while Lincoln Electric recommends a maximum delay of 5 minutes. The exact delay time often varies as welding conditions change in the field [11]. The Australian/New Zealand Standard AS2885.2 [14], however, recommends that the delay time does not exceed 8 minutes, which is shortened to 6 minutes where extreme lifting operations are unavoidable.

In an attempt to increase productivity, current field welding practices are increasingly showing overlap with conditions known to promote cracking. Qualified welding procedures are currently in use for preheat-free root pass welding of X70 pipelines in wall thicknesses up to 15.24 mm (but more commonly 12.7 mm) at heat input levels ranging from 0.39 to 1.0 kJ/mm and at welding speeds up to 600 mm/min in [4]. These procedures fall outside the limits (as discussed throughout this study and summarised in section 2.7) of accepted practice for preheat-free welding and may result in excessive cooling rates in the root pass, effectively preventing hydrogen from escaping through diffusion and increasing restraint stresses on cooling.

Recent years have also seen a shift towards larger diameter, thicker wall coal seam pipelines for the export of natural gas. Even though mechanised gas metal arc welding (GMAW) is the preferred process for mainline welding of large diameter pipelines, cellulosic procedures are still widely used for repair and for the welding of tie-ins. Heavier wall thicknesses reduce the safety margin for preheat-free welding and potentially place the pipeline construction industry at risk with regard to weld metal cracking [15].

These observations suggest that the phenomenon of weld metal hydrogen-assisted cold cracking in linepipe steel needs to be revisited, and that clear guidelines are needed to assure the industry that the risk of weld metal cracking during pipeline construction can be controlled. Guidance is needed to ensure that new and existing participants in the Australian pipeline construction industry have confidence in current procedures designed to reduce the risk of WMHACC during field girth welding.

### 2.2. REPORTED CASES OF WMHACC DURING PIPE GIRTH WELDING

As stated earlier, few cases of WMHACC during pipeline field girth welding have been reported to date, although there are suggestions of unreported cases being routinely repaired during pipeline construction. To date, only one example of WMHACC during pipeline girth welding in Australia has been studied in detail.

In the early 1970’s, hydrogen-assisted cold cracks were discovered during the construction of one of the spreads of the Moomba-to-Sydney pipeline, resulting in extensive delays in commissioning. The occurrence of hydrogen-assisted cold cracks in this pipeline shaped much of the local thinking around the influence of hydrogen and as a result the Australian Standard for pipeline welding, AS 2885.2:2016 [14], now places more emphasis on the avoidance of hydrogen-assisted cold cracking than comparable international standards [1]. The pipeline in question utilised relatively thin walled 864 mm (34”) diameter X65 steel with 8.5 mm wall thickness. The pipe was dual sourced from suppliers in Australia and Japan; the Australian pipe being of higher carbon equivalent and rare earth treated. Cracking occurred in pipe from both suppliers, but was more prevalent in the Australian material. Evidence suggests that the majority of the cracks initiated at the fusion boundary of the root/hot pass and at root intrusions in the HAZ close to the fusion boundary of the root pass. Some cracks
were, however, found to have originated in the weld metal. Failure was ultimately attributed to the higher carbon equivalent (CE) of the Australian pipe, with molybdenum segregation in the root runs and root discontinuities due to poor pipe fit-up identified as contributing factors. Cracking in the weld metal was mainly caused by the high CE (0.19 wt% carbon, 0.5 wt% molybdenum) of the cellulosic E7010 consumables employed, which produced hard phases in the as-deposited weld metal originating from undissolved molybdenum-rich precipitates in the electrode coating [3,16]

In recent years, however, several incidents of WMHACC have been reported in girth, pipe-to-fitting and repair welds in North America [17,18], with the Edison Welding Institute investigating two cases of severe WMHACC on behalf of PRCI in 2005 [18]. It is believed that cracking occurred despite close adherence to qualified welding procedures. In both instances, the composition of the weld metal was substantially richer than expected; exhibiting carbon levels higher than 0.20 wt% and carbon equivalents up to 0.66. It was reported that arc length affected the partitioning of alloying elements (particularly carbon, and the deoxidisers manganese and silicon) between the weld metal and slag, but variation in arc length alone could not account for the extremely rich alloy compositions observed in the cracked girth welds. Further investigation revealed that the moisture content of the electrode coating had an even more pronounced effect and that consumables with low moisture levels resulted in the deposition of weld metal with double the expected manganese and triple the expected silicon concentrations.

This is consistent with results reported by Weaver and Ogborn [19] who observed a change in metal transfer towards a more globular mode and a richer weld metal chemistry for cellulosic electrodes with lower moisture contents. The larger droplet size with smaller surface area-to-volume ratio, lower moisture levels and reduced oxygen contents in the weld metal promote increased partitioning of deoxidising elements (such as manganese and silicon) to the weld metal [20]. Consequently, the welder may have no indication that there is a potential problem with the consumables until actual failure occurs. Recommendations based on the outcomes of the aforementioned studies include keeping cellulosic-coated electrodes in cool, shaded areas throughout the workday and discarding any electrodes that have been exposed to hot, dry conditions [17,18].

Australia generally has a hot climate, and since the largest reservoirs of natural gas are located in the outback, the majority of pipelines are constructed under hot and dry conditions. Welding procedures are normally qualified in workshop facilities located in major cities situated on the coast. These procedures are then applied during pipeline construction in much hotter and drier environments with temperatures sometimes exceeding 50°C. This increases the risk of excessive moisture loss from the electrode coating throughout the day, potentially resulting in weld metal with considerably richer chemistry and a higher risk of WMHACC. The Australian pipeline standard, AS 2885.2:2016 [14], does not adequately address issues associated with loss of moisture from cellulosic electrode coatings. Since these issues are unique to cellulosic electrodes, welding supervisors and welders may not be aware of the potential risks. Sealed containers of E6010 electrodes are usually opened on site and used before the end of the shift, but in the odd case where electrodes are left, coupled with a drive towards the implementation of cost reduction practices, the likelihood that safe consumable storage and disposal procedures may not be followed in the field at all increases.

### 2.3. METALLURGICAL FACTORS INFLUENCING THE RISK OF HACC

The risk of weld metal cracking is strongly linked to the chemical composition of both domestically produced and imported pipeline materials and consumables. This is evident from the carbon equivalent (CE) formulas shown in Equations 2.2.1 and 2.2.2. The equation for \( \text{CE}_{\text{HW}} \) was originally developed by the International
Institute of Welding (I IW) for plain-carbon and carbon-manganese steels, whereas the equation for the carbon equivalent was developed by the Japanese Welding Engineering Society for a wider range of low-carbon steels [21]. The carbon equivalent represents the hardenability (or crack susceptibility) of an alloy based on its chemical composition. In these equations, the effect of each element on hardenability is weighted against that of carbon to obtain one representative value. A higher CE value signifies a higher hardenability, which in turn signifies a higher crack susceptibility. An alloy with a high CE value, therefore, tends to exhibit low weldability [21,22]. The relationship between carbon content (wt%), CE, and weldability is shown schematically in Figure 2.2.1.

$$CE_{IW} = C + \frac{Mn}{6} + \frac{Cu+Ni}{15} + \frac{Cr+Mo+V}{5}$$
Eq. 2.2.1

$$P_{cm} = C + \frac{Si}{20} + \frac{Mn+Cu+Cr}{20} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B$$
Eq. 2.2.2

Figure 2.2.1 Relationship between carbon content (wt%), CE, and weldability [21].

As discussed in Section 1.1, the introduction of thermo-mechanically controlled-rolled processing of steel with reduced CE values results in welds displaying higher CE values in the weld metal (WM) than in the heat-affected zone (HAZ). Lowering of the CE in low-alloy pipeline steels raises the austenite (γ) to ferrite (α) transformation temperature in the HAZ to such an extent that the ferrite transformation in the HAZ occurs before that in the weld metal [23,24]. This is significant since the solubility and diffusivity of hydrogen are affected by the transformation from austenite to ferrite.

Figure 2.2.2 shows the Fe-H phase diagram at atmospheric pressure. The diagram shows that hydrogen does not favour any metallurgical phase transformations in pure iron, yet the dotted lines indicate areas of hydrogen solubility, with atomic hydrogen (H) dissolved within the Fe matrix (whether γ or α) to the left of the dotted lines, and recombined molecular hydrogen (H₂) after effusion from the metal, to the right of the dotted lines. As can be seen from the diagram, hydrogen solubility decreases sharply on solidification, continues to decrease with decreasing temperature and reaches extremely low values near ambient temperature.
Figure 2.2.2 Fe-H phase diagram at atmospheric pressure [25].

Figure 2.2.3 Hydrogen solubility in pure iron as a function of temperature and pressure [26].

It is evident from both Figures 2.2.2 and 2.2.3 that hydrogen is much more soluble in austenite than in ferrite. This is due to the face centred cubic (FCC) crystal structure of austenite having a higher volume fraction of interstitial spaces in the iron lattice than the body centred cubic (BCC) crystal structure of ferrite [27], enabling it to accommodate a higher percentage of interstitially dissolved hydrogen. This difference in hydrogen solubility acts as the driving force for the directional diffusion of hydrogen from ferrite to austenite. Compared to ferrite, austenite acts as an irreversible hydrogen trap or sink and retards hydrogen diffusion (as shown in Figure 2.2.4, where at constant temperature, austenite exhibits much lower coefficients of hydrogen diffusion than ferrite).

Figure 2.2.3 is a schematic diagram that emphasises this effect, showing hydrogen solubility in terms of absorbed hydrogen (Ncm³/100 g Fe) as a function of temperature at three different pressure levels (0.1 atm, 1 atm and 10 atm).
Figure 2.2.4 Coefficient of diffusion (cm$^2$/s) of hydrogen in ferrite (F) and austenite (A) in pure iron [26].

With a lower CE in modern linepipe steels, transformation from austenite to ferrite on cooling therefore occurs at a higher temperature in the HAZ, leading to hydrogen being rejected from the HAZ. The hydrogen diffuses towards the austenite ($\gamma$) in the weld metal where a higher solubility exists, thereby increasing the risk of WMHACC on cooling to room temperature [3]. Such a change in transformation behaviour effectively shifts much of the hydrogen-assisted cold cracking risk from the HAZ to the weld metal, especially if root pass defects such as severe undercut or solidification cracks are present in the weld. This phenomenon is illustrated in Figure 2.2.5, where in (a) the $\gamma/\alpha$ transformation temperature is higher in weld metal than in the HAZ and hydrogen diffuses from the ferritic weld metal to the austenitic HAZ, leading to an increased risk of HACC in the HAZ. In (b) the $\gamma/\alpha$ transformation temperature is higher in the HAZ (as is the case in modern linepipe steel) and hydrogen diffuses from the ferritic HAZ to the austenite in the weld metal where there is a consequential increase in the risk of HACC. This does not mean that the base metal and HAZ of X70 steel are completely resistant to HACC [28-31] and the necessary precautions for its avoidance are still required.

The Australian Standard for pipeline welding, AS 2885.2:2016 [14], places more emphasis on the avoidance of hydrogen-assisted cold cracking than comparable international standards. This is because of the occurrence of HACC during construction of the Moomba-to-Sydney pipeline in the 1970’s. The damage was severe and caused extensive delays in commissioning. Despite this, there is continued use of cellulosic consumables during field girth welding of pipelines in Australia [1]. Sufficient guidance is given in the standard on the avoidance of heat-affected zone (HAZ) hydrogen-assisted cold cracking [17,32], but although qualification tests provide some assurance that weld metal hydrogen cracking will not occur during field welding, clear guidelines on reducing the likelihood of weld metal hydrogen-assisted cold cracking (WMHACC) are still lacking. Amongst other objectives, this study therefore aims at defining a safe welding parameter window in which welding can be carried out with minimal risk of root bead WMHACC for X70 pipeline steel, using E6010 cellulosic consumables.
Recent years have seen a number of prominent welding consumable manufacturers introducing changes to the electrode formulations of cellulotic consumables in order to promote the formation of acicular ferrite, improve weld metal strength and increase the overall joint toughness. Modern consumables may contain higher amounts of alloying elements and produce enriched welds with harder microstructures and a higher susceptibility to hydrogen-assisted cold cracking [34]. Due to the wide specification limits for cellulotic consumables in AWS A5.1/A5.1M:2012, E6010 electrodes from different manufacturers, and even different electrode batches from the same manufacturer, may display significant variations in chemistry while still satisfying the classification requirements in AWS A5.1. This concern has been identified and recognised by the local pipeline industry and common practice is to require the consumable manufacturer to provide a certificate with the full chemical analysis of the consumable. More guidance is, however, required on acceptable limits for various alloying additions and impurity elements in E6010 consumables.

2.4. WELD METAL HYDROGEN-ASSISTED COLD CRACKING

2.4.1 Hydrogen Embrittlement

Hydrogen-assisted cold cracking (HACC) is a form of hydrogen embrittlement (HE) and is the designation given to the branch of HE that deals with embrittlement during welding. HACC in the heat-affected zones (HAZ) of carbon steel welds has been the focus of numerous investigations, but very little information is available in published literature on weld metal HACC. It is widely accepted that the factors responsible for weld metal hydrogen-assisted cold cracking (WMHACC) in pipeline steel welds are similar to those responsible for HAZ hydrogen-assisted cold cracking in steels.

As stated in Section 1.1.1, four requirements have to be satisfied simultaneously for hydrogen-assisted cold cracking to occur in steel during welding: the presence of a critical amount of diffusible hydrogen, a crack-susceptible microstructure, a critical tensile stress, and a temperature near normal ambient. Embrittlement occurs during cooling within a critical temperature range, typically around -150 °C to 150°C [2], as hydrogen collects around areas of concentrated stress and lattice defects such as grain boundaries, dislocations, lattice-
precipitate interfaces and vacancies. Embrittlement is characterised by a drastic loss of ductility, i.e. hydrogen causes a reduction in the energy required to fracture the weld metal, evident as a reduction in strain to failure when compared to weld metal in the absence of hydrogen [2,21,35,36].

In addition to the factors listed above, the carbon equivalent of the steel, the initial weld metal hydrogen content, the yield stress of the steel or weld metal, hardness, heat input, preheat temperature, material thickness, joint restraint intensity, notch concentration factor and a number of other factors play a role in determining susceptibility to WMHACC. These factors are schematically illustrated in Figure 2.4.1.

![Figure 2.4.1 Causal factors of HACC [22]](image)

As shown in the second column of Figure 2.4.1, the residual hydrogen content in the weld metal when the critical temperature range for cracking is reached on cooling, H_0, is determined by the initial weld hydrogen concentration, H_0, and the cooling conditions. Approximately 90% of the absorbed hydrogen diffuses out of the weld metal at temperatures above 100°C. Once the weld temperature cools down below 100°C, hydrogen tends to diffuse to and accumulate at weld imperfections where stress and strain are concentrated and crack initiation occurs. Such stress concentrations can take the form of pre-existing cracks, non-metallic inclusions, pores, or geometric stress raisers such as wagon tracks or undercut. The locally accumulated hydrogen content, H_{local}, is influenced by the stress concentration factor: K_t. H_{local} is close to H_R when welding with cellulose consumables. Root bead WMHACC is initiated within minutes after welding because a high concentration of hydrogen is present at 100°C. If the hot- and cap passes of the weld is deposited within the specified inter-pass time (maximum 8 minutes), then the root pass temperature should not fall to within the critical temperature range where WMHACC can occur [37].

2.4.2 Mechanisms of hydrogen embrittlement

Although the fundamentals of the mechanisms of hydrogen embrittlement are still subject to ongoing debate, it is widely accepted that several mechanisms may operate for various types of hydrogen-induced failures. The material, hydrogen source, charging conditions and loading determine the operative mechanism. For a specific mechanism to hold for a certain class of failures, it must address the signature dependencies of hydrogen embrittlement, i.e. it must explain the distinguishing features of HE, even at low hydrogen concentrations, and how it differs from fracture in the absence of hydrogen. It must also explain how hydrogen on an atomistic
scale influences the macro-fracture mechanics and fracture appearance. The proposed mechanism must also be applicable universally within the specific class of failures [38,39].

Many mechanisms have been put forth in the literature to explain the fundamentals of hydrogen embrittlement. Of these, hydrogen enhanced decohesion (HEDE) and hydrogen enhanced localised plasticity (HELP) are the most popular. HEDE advocates a reduction in the cohesive bond between atoms in the presence of hydrogen, which reduces the stress required to initiate brittle fracture [40]. HELP advocates a lowering of the forces necessary to initiate localised dislocation nucleation and growth, reducing the stress needed to cause fracture. There is a third widely accepted model, adsorption induced dislocation emission (AIDE), which advocates the nucleation of dislocations due to a weakening of near-surface interatomic bonds caused by hydrogen adsorbed onto the surface from H₂ atmospheres [40].

According to the HEDE mechanism of hydrogen embrittlement, interstitially dissolved hydrogen diffuses to sites of triaxial stress. These may include surface discontinuities, dislocations and dislocation pile-ups, precipitate interfaces, and crack tips. Once the accumulated hydrogen reaches a minimum concentration, degradation of the cohesive bond between the iron atoms occurs, leading to fracture [21,41-43]. Hydrogen concentrated and trapped at grain boundaries would, in the same manner, cause intergranular decohesion, leading to intergranular fracture [39]. In general, it is theorised that a linear relationship between degradation and the concentration of interstitially dissolved hydrogen exists above the minimum hydrogen concentration necessary for decohesion [44].

The theory of decohesion was formulated based on interpretations of fracture surface appearance, and any evidence of accompanying plasticity was assumed to be a consequence of fracture, rather than an integral part of the mechanism leading to fracture [39,41]. Direct experimental evidence linking the presence of interstitially dissolved hydrogen to the HEDE mechanism has not been published to date due to the difficulty in measuring the atomic bond strength ahead of a growing crack tip. Furthermore, measurements with regard to lattice energies of strained structures in the presence and absence of hydrogen do not indicate a lowering of the lattice energy [45].

On the other hand, published experimental evidence for the HELP mechanism abounds. Figure 2.4.2 was produced by Ferreira et al. [46] and shows a time-lapse image of a dislocation pile-up (white lines and numbers) at a grain boundary, superimposed on an earlier image of the same dislocation pile-up (black lines and numbers). This image was produced by in situ video recording of deformation experiments in a modified transmission electron microscope (TEM) equipped with an environmental cell. The black lines show dislocations produced under vacuum conditions, while the white lines represent the same dislocations (at a constant load) after ingress of H₂ gas into the environmental cell of the TEM. It is evident that the separation distance between the dislocations decreased after introduction of the H₂ gas, suggesting that absorbed and interstitially dissolved hydrogen lowered the elastic interactions between the dislocations, thereby enhancing their mobility. Further experimentation by Robertson [39] and Robertson et al. [47] confirmed this effect for various crystal structures, including FCC, BCC, and HCP (hexagonal close packed) structures, while Birnbaum and Sofronis [45] demonstrated the same effect for edge, screw, and mixed dislocations.

Martin and Fenske et al. [48] showed that, during hydrogen embrittlement of an API 5L X60 steel, river markings typical of quasi-cleavage fracture were caused by plastic processes attributable to the HELP mechanism. Figures 2.4.3(a) and (b) show scanning electron microscope (SEM) images of such river markings, taken at 0° and 55° tilt angles, respectively. Figures 2.4.3(c) and (d) show three dimensional reconstructed
images of the surface topography, produced using fractographs and the Alicona MeX™ software package. From these images, it is clear that the areas between the ridges or river markings are not flat and smooth, but consist of slight undulations across the fracture surface. Samples were then extracted from site specific locations using a combination of focused ion beam (FIB) and conventional milling, normal to the surface and extending from the fracture surface into the material. This allowed the authors to study the microstructure that formed directly beneath the surface.

![Figure 2.4.2](image)

**Figure 2.4.2** Superimposed time-lapse images of moving dislocations. The black numbered lines represent the original dislocations piled up against a grain boundary, while the white numbered lines represent a superimposed image of the same dislocations after ingress of H₂ gas into the environmental cell of the TEM [46].

![Figure 2.4.3](image)

**Figure 2.4.3** Micrographs showing the river markings of a quasi-cleavage fracture surface of an API X60 steel due to hydrogen embrittlement at (a) 0° and (b) 55° tilt angles, with (c) and (d) showing the topographical features on the fracture surfaces, reconstructed in three dimensions using the Alicona MeX™ software package [48].

Figure 2.4.4 shows a bright-field TEM micrograph of a sample extracted directly beneath a ridge of the river markings. It shows densely packed slip planes on multiple slip systems directly beneath the fracture surface. The extensive slip of dislocations observed in the micrograph was aided by the presence of hydrogen. The authors linked this microstructure to the fracture surface topography by deducing that microvoids or
microcracks initiate at corners where slip lines from different slip systems intersect, which then propagate to form the typical ridges seen in Figure 2.4.3. These microvoids eventually coalesce, aided by enhanced dislocation motion. Ridges from conjugate fracture surfaces were also reported to meet ridge-to-ridge instead of ridge-to-valley, which supports the plasticity mechanism, consistent with HELP, proposed by the authors.

![Fracture Surface](image)

**Figure 2.4.4** TEM bright-field micrograph of the microstructure directly beneath and normal to the surface ridges [48].

The HELP mechanism hinges on the segregation of hydrogen to the elastic stress portion of dislocation centres and other obstacles. This modifies the elastic field by increasing it in certain directions and decreasing it in others. This means that the energy barriers to dislocation motion decrease in the same direction as the elastic field, enhancing dislocation mobility in that direction [47]. Birnbaum and Sofronis [45] showed that in BCC metals, dislocation motion was preferentially enhanced in the \{112\} and \{110\} directions.

Martin, Robertson, and Sofronis [49] further showed that quasi-cleavage fracture surfaces of X60 steel at even higher magnifications displayed fine dimples, as shown in Figure 2.4.5. Using the same techniques as in the previous study, these authors produced the TEM bright-field micrographs in Figure 2.4.6, showing dense dislocation forests and loops at depths of (a) 250 nm, and (b) 1500 nm, below the dimpled fracture surface at two different surface features. They, again, attributed the high dislocation densities, extending through several grains beneath the fracture surface, to the hydrogen-enhanced localised plasticity mechanism.

![SEM image](image)

**Figure 2.4.5** (a) SEM image of flat quasi-cleavage fracture surface, and (b) fine dimples at higher magnification [49].

Intergranular fracture is also facilitated by the HELP mechanism. If the grain boundary area contains a higher concentration of hydrogen or if the grain boundaries are weakened in any way, either by precipitation or segregation of various alloying or impurity elements, plasticity-mediated intergranular fracture could occur [39,45]. It has also been suggested that all fracture modes associated with hydrogen embrittlement, i.e. microvoid coalescence, quasi-cleavage and intergranular fracture, undergo fracture related to micro-plasticity.
It follows then that the enhanced plasticity mechanism of hydrogen embrittlement mainly facilitates the deformation and fracture modes that would have occurred in the absence of hydrogen [11,50].

Figure 2.4.6 Microstructures at two different features at depths of (a) 250 and (b) 1500 nm beneath the surface [49].

In short, the HELP mechanism of hydrogen embrittlement lowers the macroscopic ductility of components and results in brittle fracture due to the action of enhanced plastic processes on a microscopic level [45]. Many authors have investigated and attributed fracture of hydrogen-embrittled metals either directly to the HELP mechanism or indirectly by quoting microscopic plastic processes as the reason for failure [51-61] reported that for MMAW weld metal with an acicular ferrite and bainite microstructure, the HELP model of hydrogen embrittlement in a seawater environment appeared to be the more applicable fracture mechanism.

The underlying basis for the adsorption-induced dislocation emission (AIDE) model shares characteristics with both the HEDE and HELP models. AIDE and HEDE both presume a lowering of the interatomic bond strength between atoms in the iron matrix, while AIDE and HELP advocate enhanced localised plasticity due to hydrogen as an essential prerequisite to fracture [62].

Figure 2.4.7 is a schematic representation of the AIDE mechanism. This diagram shows that the adsorption of hydrogen weakens the interatomic bonds of the metal. This weakening is caused by hydrogen adsorbing directly onto the surface, but also by interstitially dissolved hydrogen next to the surface causing weakening a few atomic layers deep. This weakening of atomic forces facilitates the nucleation and injection of dislocations on two slip planes (designated A and B in the sketch). These dislocations are injected into the material ahead of the crack tip in equal amounts, as back-stress from piled-up dislocations from either side would affect dislocation nucleation on the other side of the crack tip. The injected dislocations cause the crack tip to propagate by a distance, Δa, where it then coalesces with plastically growing voids directly in its path. These growing voids ahead of the crack tip may nucleate at vacancy clusters, second-phase particles, dislocation boundaries or at slip band intersections, and give quasi-cleavage fractures a dimpled, or even nano-dimpled appearance at high magnifications [63,64].

Lynch [62] responded to the work of Martin, Robertson, and Sofronis [49], discussed above, and stated that the nano-dimpled structures with the associated dislocation-dense microstructure beneath the surface (Figures
2.4.5 and 2.4.6) would be better explained by the AIDE mechanism, rather than solely by HELP as originally done by the authors.

![Figure 2.4.7](image)

Figure 2.4.7 Schematic representation of the AIDE mechanism of hydrogen embrittlement [63].

In addition to those discussed above, other mechanisms have been proposed to account for the embrittling effect of hydrogen. Dunne and Nolan [11], and Robertson et al. [39] summarised these mechanisms as follows:

- Internal hydrogen gas pressure in microvoids (discussed in greater detail in Section 4.9);
- Hydrogen-induced reduction in surface energy;
- Hydrogen- and deformation-assisted vacancy production;
- Hydrogen-triggered ductile-to-brittle transition;
- Hydride formation and cleavage;
- Hydrogen- and strain-induced phase transformations; and
- Reactants and hydrogen.

Within the context of celluloseic stovepipe welding of X70 pipeline steel with E6010 electrodes, some of these mechanisms may be eliminated or, at least, given secondary status. HEDE, HELP and AIDE could certainly be operative, but hydride formation, for example, would only be a viable mechanism for group V metals and Ti and Zr from group IV (according to the periodic table’s classification) and their alloys. Iron hydride (FeH$_2$) only forms at extreme combinations of temperature and pressure, and dissociates readily at either ambient pressure or temperature or a combination of both [36,47]. Hydrogen-induced reduction of surface energy can be eliminated as a main cause of embrittlement according to Figure 2.4.8.
Figure 2.4.8 Calculated values of surface energies in the \{100\} direction in ferrite and austenite in the absence and presence of hydrogen [65].

Figure 2.4.8 shows that the interstitially dissolved hydrogen in ferrite and austenite lowers the surface energy by about 7.2% and 9.6%, respectively, in the \{100\} direction of both allotropes [65]. This is not enough to cause embrittlement, although it will most likely contribute to the prevailing failure mechanism during HACC, especially in the presence of stress raisers at the surface.

Although hydrogen plays a role in blistering during corrosion, the mechanism of high internal hydrogen pressure within microvoids has been discounted as the leading cause of HACC [21]. On the other hand, hydrogen and deformation-induced vacancy production could potentially be a strong contributor to HACC during stovepipe welding. In most metals, there is an exponential increase in vacancy concentration as the temperature increases [66]. Due to rapid cooling of the root bead during field welding, a large concentration of quenched-in vacancies can be expected at near-ambient temperatures. There is also a high level of restraint inherent in the root bead during cellulosic welding and the formation of deformation-induced vacancies is possible. A model, proposed by Nagumo [67], suggests that the existing vacancy concentration is stabilised and increased in the presence of dissolved and diffusible hydrogen. The agglomeration of dense vacancy clusters would then either contribute or lead to failure. Several authors [39,68] considered microvoid nucleation, growth, and coalescence due to enhanced vacancy formation as a viable mechanism. Such failure is only possible in relation to plasticity, however, and therefore suggests that vacancy clustering either leads to plastic instability, or that a cooperative mechanism exists that effectively combines both [69].

It is not uncommon to find authors attributing the mechanism of hydrogen embrittlement to combinations of proposed mechanisms, most notably HEDE and HELP [70-73].

2.4.3 Weld metal and heat-affected zone microstructures

As shown in the first column of Figure 2.4.1, the microstructure that forms on solidification and cooling mainly determines the hardness of the weld metal. In general, the higher the strength or hardness of the weld, the lower its resistance to HACC. This relationship underpins the use of the carbon equivalent (CE) equations, which quantify the potential hardenability of the base metal (heat-affected zone) and weld metal as a means of predicting susceptibility to HACC [35]. The chemical composition (including the CE) of the diluted weld metal, the cooling rate after welding (as determined by the heat input, plate thickness and preheat temperature), and the availability of nucleation sites for ferrite are the primary factors determining the microstructure of the weld metal.

Cellulosic E6010 weld metal typically consists of a complex mixture of primary ferrite, ferrite-carbide aggregate, Widmanstätten ferrite side-plates, ferrite with aligned second phase, acicular ferrite, bainite and martensite [74]. The hydrogen-assisted cold cracking susceptibility is generally considered to be higher for microstructures containing hard grains of martensite or bainite, coarse prior austenite grain sizes (which increase the hardenability and the amount of segregation at the grain boundaries), slag inclusions or coarse carbide particles along grain boundaries and high dislocation densities. It has been reported that the hydrogen-assisted cold cracking resistance of acicular ferrite is higher than that of upper bainite [75,76].
<table>
<thead>
<tr>
<th>Principal structure classification</th>
<th>Category terminology</th>
<th>Overall</th>
<th>Main</th>
<th>Sub</th>
<th>Component structure description</th>
<th>Comments</th>
</tr>
</thead>
<tbody>
<tr>
<td>Reconstructive transformations (diffusion controlled, with slow rates of reaction)</td>
<td>Ferrite</td>
<td>PF*</td>
<td>PF(GB)</td>
<td>PF(G)*</td>
<td>Grain boundary primary ferrite</td>
<td>Ferrite veins or polygonal grains aligned with prior austenite grain boundaries</td>
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<td>Allo- and martensitic ferrite</td>
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<td>Polygonal ferrite</td>
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<td>Ferrite veins</td>
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<td>PF(N)*</td>
<td>Polygonal primary ferrite non-aligned</td>
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<tr>
<td></td>
<td>Pearlite</td>
<td>P+</td>
<td>P*</td>
<td>FC(P)+</td>
<td>Lamellar pearlite</td>
<td>Nodules of alternate ferrite/cementite lamellae, which are often difficult to resolve under the optical microscope. The structure has a rapid etching response in 2% nital and a generally low hardness. Pearlite may be present as a microconstituent.</td>
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<td>Pearlite lamellae viewed in cross-section. Distorted pearlite lamellae may appear as a dark etching virtually irresolvable ferrite/cementite aggregate known as primary troostite. Difficult to distinguish ferrite/cementite aggregate from bainite.</td>
</tr>
<tr>
<td>Displaceable transformations (shear dominated, with rapid rates of reaction)</td>
<td>Widmanstätten ferrite</td>
<td>WP</td>
<td>WP(GB)</td>
<td>WS(A)*</td>
<td>Widmanstätten ferrite with aligned microphases</td>
<td>Colonies of parallel ferrite laths (or sideplates) with microphases aligned between the laths ranging from pearlite to martensite. Lath boundaries are difficult to resolve. Primary Widmanstätten ferrite grows from the prior austenite grain boundaries, whereas secondary Widmanstätten ferrite grows from allotriomorphic ferrite at the boundary.</td>
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<td>Widmanstätten ferrite sideplates</td>
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<td>FS(N)*</td>
<td>Widmanstätten ferrite with non-aligned microphase</td>
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<td>FS(II)</td>
<td>Intragranular Widmanstätten ferrite sideplates</td>
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<td>FP(II)</td>
<td>Intragranular Widmanstätten ferrite plates</td>
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<td></td>
<td>AF*</td>
<td>Widmanstätten acicular ferrite</td>
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<tr>
<td></td>
<td>Bainite</td>
<td>B</td>
<td>B(GB)</td>
<td>FS(A)*</td>
<td>Bainitic ferrite with aligned carbide</td>
<td>Fine interlocking structure formed by multiple impingements of individual Widmanstätten ferrite plates growing from intragranular inclusions.</td>
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<td></td>
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<td>Bainite sheaves</td>
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<td></td>
<td>FS(N)*</td>
<td>Bainitic ferrite with non-aligned carbide</td>
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<td>FS(II)</td>
<td>Upper Bainite</td>
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<td>FS(LB)*</td>
<td>Lower bainite</td>
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<td></td>
<td>Martensite</td>
<td>M*</td>
<td>M*</td>
<td>M(L)*</td>
<td>Lath martensite</td>
<td>Sheaves of fine bainitic ferrite plates with aligned carbide, which grow from intragranular inclusions. Individual fine plates of bainitic ferrite that grow relatively unimpeded from intragranular inclusions. Very fine interlocking structure formed by multiple impingements of individual bainitic ferrite plates growing from intragranular inclusions. Low carbon martensite with a lath structure and heavily dislocated sub-structure. Lath martensite has a slow etching response in 2% nital and a generally high hardness. Colonies of martensite may form within the prior austenite grains. Smaller colonies may be treated as microphases. Microphases may consist of martensite with retained austenite (MA).</td>
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<td></td>
<td></td>
<td>AF*</td>
<td>Twin martensite</td>
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</table>

*Retained IIW terminology.
Costin et al. [77] studied the influence of hydrogen charging on the fracture behaviour of two microstructural components, acicular ferrite and upper bainite, in E6010 weld metal using micro-fracture tests. He reported that hydrogen charging reduces the yield strength and changes the fracture behaviour from ductile failure with severe notch tip blunting, to brittle fracture characterised by slow crack growth. It is therefore recognised that even a weld with low hardenability or an “ideal” microstructure (such as acicular ferrite) may be susceptible to WMHACC if the hydrogen concentration is high enough [35].

An in-depth discussion of the microstructures observed in the as-deposited weld metal of E6010 electrodes and in X70 heat-affected zones is given in Section 4.8 with comparisons to published data on E6010 and X70 microstructures deposited under similar welding conditions.

In the current study, characterisation of the transformation products or metallurgical phases in the weld metal and heat-affected zone microstructure was carried out using the modified IIW classification scheme reported by Thewlis [78] and is given in Figure 2.4.9. This figure and its contents are discussed in greater depth in Section 4.8.

### 2.4.4 Tensile stresses operating on weld metal

Tensile residual stresses inevitably arise from thermal contraction during cooling and may be supplemented by other stresses developed due to rigidity in the parts being joined. In rigid structures, the restraint imposed on the weld by the surrounding material intensifies the natural contraction stresses. These stresses are concentrated at the toe and root of the weld and at stress concentrations constituted by inclusions, cracks and other defects (represented by $\sigma_{\text{local}}$ in Figure 2.4.1). For thin walled pipes held in place by external or internal clamps, the restraint intensity is likely to be low and similar to levels present in other structural steel welding applications. During pipeline welding, external stresses (such as lifting stresses) may supplement these residual stresses and increase the risk of cracking temporarily [79]. Fortunately, the lifting stress is considered low in normal lifting operations and only becomes severe for extreme lifts. The stress due to clamp removal is considered lower than the stress produced during the lifting operation, and the main stress acting on the root therefore arises from shrinkage or weld residual stress [37].

As stated above, residual stresses resulting from restraint during welding are concentrated at defects and discontinuities, giving rise to the development of a triaxial stress state ahead of the crack tip or root of the defect. Hydrogen is attracted to these high stress intensity sites and diffuses to stress concentrations at lower temperatures, increasing the local diffusible hydrogen concentration, $H_{\text{local}}$. If the local hydrogen concentration reaches a critical level, embrittlement occurs and cracking initiates at strains markedly lower than for hydrogen-free steel [79].

From the preceding discussion it is clear that stress concentrations in the weld microstructure play an important role in facilitating WMHACC by locally concentrating diffusible hydrogen during cooling. Available literature suggests that weld metal hydrogen-assisted cold cracks in linepipe steel welded with cellulosic consumables have three possible origins associated with various stress raisers in the weld, as listed below and described in more detail in subsequent sections.

- Hydrogen-assisted cold cracks may initiate at stress concentrations caused by the joint geometry and fit-up. Wagon tracks, undercut, misalignment and excessive root gaps affect the levels of stress and strain induced at the weld root during joining and lifting, and may affect susceptibility to WMHACC by acting as hydrogen accumulation sites during cooling [37].
- Microcracks may initiate in the weld metal as a result of hydrogen accumulation at stress concentrations formed by various non-metallic inclusions [77]. The chemistry, size, coherence, and volume fraction of these inclusions are likely influenced by small changes in the weld metal alloy composition and impurity content, and may in turn affect the susceptibility of welds to WMHACC. The role of non-metallic inclusions in initiating WMHACC is considered in more detail in Sections 2.4.6 and 4.9.

- Solidification cracks formed in the weld metal at higher temperatures close to the solidus temperature during cooling may extend into the weld metal at lower temperatures through hydrogen-assisted cold cracking [80-83]. These pre-existing solidification cracks act as significant stress raisers due to the high stress concentration factor of the sharp crack tip ($K_t$ in Figure 2.4.1) and serve as preferential hydrogen accumulation sites in the weld metal. Solidification cracks can therefore act as precursors to WMHACC. The influence of solidification cracks on the initiation of WMHACC is considered in more detail in Sections 2.4.5 and 4.5.

The stress inherent in the experimental work performed as part of this study is discussed in more detail in Sections 2.8 and 4.11.

### 2.4.5 Solidification cracking and WMHACC

The relationship between weld metal solidification cracking at higher temperatures and WMHACC at lower temperatures during the weld cooling cycle was reported by Nolan et al. [80-83] as part of a series of short projects performed with support from the Welded Structures CRC. These studies confirmed that root pass welds deposited using cellulosic consumables are not only susceptible to extensive solidification cracking, but solidification cracks were observed to serve as initiation sites for WMHACC. Solidification cracks formed at higher temperatures were shown to extend at lower temperatures through the weld metal of root pass welds through a mechanism of hydrogen-assisted cold cracking (as shown in Figure 2.4.10). More recently, research at the University of Adelaide showed a strong correlation between WMHACC and solidification cracking in cellulosic weld metal during modified WIC testing. Results confirmed that, in susceptible welds, WMHACC predominates under conditions of low preheat and low heat input, with solidification cracking dominating at higher preheat temperatures and heat input levels [84]. Mixed mode cracking (WMHACC and solidification cracking) was observed at intermediate heat input and preheat levels.

![Figure 2.4.10 A centreline solidification crack in the root pass of a cellulosic weld extending into the weld metal through a mechanism of hydrogen-assisted cold cracking [82].](image)
Solidification cracking is a high temperature crack phenomenon that occurs in weld deposits during cooling after welding. Cracking is typically observed to occur at temperatures about 200°C to 300°C below the melting temperature of the weld metal. Solidification cracks form preferentially at the weld centreline or between columnar grains. The partitioning of alloying elements and impurities ahead of the advancing solid-liquid interface and subsequent rejection of these elements to the columnar and dendritic grain boundaries during solidification cause considerable segregation. Highly enriched low melting phases or eutectics produce strongly wetting films at the grain boundaries. These films weaken the structure to the extent that fissures can form at the columnar grain boundaries and weld centreline under the influence of the contraction stresses that develop following solidification [85]. Solidification cracks pose a significant threat to the integrity of pipeline girth welds due to their stress concentrating effect and the attendant risk of extension by hydrogen-assisted cold cracking at lower temperatures. This effect is particularly relevant in the case of cellulosic weld metals where the high levels of diffusible hydrogen introduced during welding increase the risk of WMHACC, thereby posing a threat to the construction and operation of pipelines [82].

Solidification cracking in carbon or carbon-manganese weld metals is promoted by high heat input welding procedures, fast welding speeds, unfavourable joint geometry and by conditions of high restraint. The likelihood of solidification cracking in low-alloy ferritic weld metals deposited on low-alloy steel by low dilution processes is generally considered to be low, but research has shown that the root pass of pipeline girth welds deposited with cellulosic consumables is sensitive to the development of solidification cracks. It is likely that the high force associated with cellulosic consumables, the fast welding speeds used in root pass pipeline girth welding and the geometry of the root weld configuration produce favourable conditions for the development of solidification cracks [82].

Although MMAW with cellulosic electrodes has dominated the girth welding of the small diameter, thin-walled pipelines historically used in Australia for the transmission of natural gas for decades, reported instances of solidification cracking during pipeline construction are rare and evidence of failure mostly anecdotal. It is possible that solidification cracking occurs during field welding, but remains undetected, or that solidification cracks are mostly removed by dressing and subsequent deposition of the hot pass. Reliance on grinding and hot pass welding to remove solidification cracks from the root pass poses a small, but significant risk to the integrity of the weld. If the crack is deep enough, it may not be removed completely, leaving a residual defect that can act as a stress raiser and initiation site for subsequent hydrogen-assisted cold cracks [80-83].

The susceptibility of steel welds to solidification cracking during pipeline girth welding is dependent on a number of factors, briefly considered below.

- **Welding speed**: Welding speed is an important factor influencing susceptibility of welds to solidification cracking. Welding at higher speeds requires current levels near or at the top of the range recommended for a given consumable, and this is thought to be a major factor in increasing the risk of solidification cracking during pipeline construction. Solidification patterns in welds are strongly affected by the welding speed. Low welding speeds tend to promote the formation of elliptical weld pools, allowing the solidifying columnar grains to curve in behind the moving heat source [85], as shown in Figure 2.4.11.
New grains nucleate continuously to maintain growth along preferential crystallographic growth directions, resulting in grain refinement and a reduction in centreline segregation. High speed welds, on the other hand, tend to produce tear drop shaped weld pools in which the columnar crystals grow in parallel, straight rows towards the weld centreline. This results in sudden and abrupt changes in the growth direction of the solid grains at the weld centreline and promotes extensive segregation. A long, straight-sided columnar grain structure tends to be weaker under stress than the more equiaxed, finer grain structure formed at lower travel speeds. Susceptibility to solidification cracking is therefore increased at faster welding speeds due to the coarser grain size and high levels of segregation at the weld centreline [85]. It is worth noting that whilst it is common practice around the world to limit the welding speed to a maximum of 350 mm/min during root pass welding of pipeline girth welds, higher welding speeds of 450 to 600 mm/min are common in Australia. Oshita et al. [86] reported that crack susceptibility increases as the welding speed approaches 500 mm/min, and that a critical travel speed (approximately 330 mm/min) below which cracking does not occur) exists for pipeline girth welds joined using MMAW with cellulosic electrodes. Figures 2.4.11(a) and (b) show a summary of Oshita’s results.

![Diagram of Low and High welding speed](image)

**Figure 2.4.11** The effect of welding speed on the shape of the weld pool and the direction of crystal growth during solidification [85].

![Graphs of Travel Speed vs Carbon Content](image)

**Figure 2.4.12** Critical travel speed and carbon content causing solidification cracking in MMAW welded root beads deposited with cellulosic consumables on (a) X70 plate welded in the vertical position and (b) girth welds in X70 pipe sections welded in the downhand position (δ in each image indicates joint shrinkage) [86].
Oshita’s results are based on (a) plate thicknesses in the range of 15.2 to 25 mm, and (b) pipe with wall thicknesses in the range of 15.9 to 19.1 mm. These results therefore merely serve as an indication of the critical conditions necessary for solidification cracking in the current study as the plate thickness used in this investigation (discussed in detail in Chapter 3) was 10 mm.

Contrary to Oshita’s results, Nolan et al. [82] observed cracking in X70 root pass welds deposited using E6010 consumables at welding speeds as low as 300 mm/min, suggesting that pipeline girth welds may be at risk of solidification cracking under existing welding conditions used in the field.

- **Heat input:** As a result of epitaxial solidification, the grain size of the weld metal is inherited from the grain growth zone of the HAZ [85]. High heat input welds therefore demonstrate the highest amount of grain growth during the weld thermal cycle. A coarse weld metal grain size is not only detrimental to the mechanical properties of the joint, but significantly increases the level of grain boundary segregation. The use of higher welding currents to compensate for the fast welding speeds during pipeline girth welding may therefore increase susceptibility to solidification cracking by promoting grain growth and segregation in the weld metal.

- **Joint geometry:** Root passes in pipeline girth welds deposited using MMAW and cellulosic consumables are full penetration welds with a tendency for the solidification fronts to extend from either side of the joint preparation and meet at the weld centreline. This geometry renders the root pass welds prone to solidification cracking by increasing the level of segregation in the last liquid to solidify at the weld centreline. The risk of solidification cracking during pipeline construction is exacerbated when pipe is lifted onto skids to accommodate the next pipe length resulting in additional tensile strains across the weld bead. Current Australian practice for land-based pipeline construction involves lifting and lowering the partially welded pipe onto a supporting skid once 50 to 70% of the root pass has been completed. This lifting action results in high tensile stresses and strains at the underside of the root weld. Early clamp release and lifting after partial completion of the root pass substantially increases productivity, but the weld may still be at a high temperature during the lifting procedure. Due to the decrease in ductility observed in welds caused by the presence of low melting liquid films at the grain boundaries, the lifting strain experienced by the root pass of a pipeline girth weld has been shown to approach the elongation to failure for E6010 weld metal at a temperature of 1375°C [82].

- **Chemistry and segregation:** The occurrence of solidification cracking in the weld metal is closely related to segregation of alloying elements and impurities ahead of the solid-liquid interface at columnar grain boundaries and at the weld centreline. Elements that segregate strongly between the liquid and solid during solidification tend to concentrate in the last remaining liquid and promote the formation of low melting phases and eutectics.

The extent of segregation during solidification for various elements can be approximated by examining the partitioning coefficient, \( k \), of the element in iron. Different alloy combinations produce different \( k \) values, the highest level of segregation occurring for the smallest values of \( k \). In the case of steel, for example, any alloying constituent in the weld deposit that exhibits a wide solidification range with iron is likely to have a low \( k \) value. Some approximate values of \( k \) for various elements in iron, as determined from their binary equilibrium solubilities, are given in Table 2.4.1 [87].
Table 2.4.1 Estimations of partitioning coefficients of elements in δ-iron [87].

<table>
<thead>
<tr>
<th>Element</th>
<th>Al</th>
<th>B</th>
<th>C</th>
<th>Cr</th>
<th>Co</th>
<th>Cu</th>
<th>H</th>
<th>Mn</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>k</td>
<td>0.92</td>
<td>0.05</td>
<td>0.13</td>
<td>0.95</td>
<td>0.90</td>
<td>0.56</td>
<td>0.32</td>
<td>0.84</td>
<td>0.80</td>
</tr>
<tr>
<td>Ni</td>
<td>0.80</td>
<td>N</td>
<td>O</td>
<td>P</td>
<td>Si</td>
<td>S</td>
<td>Ti</td>
<td>W</td>
<td>V</td>
</tr>
</tbody>
</table>

According to this table, the elements most likely to segregate in steel are S, O, B, P, C, Ti, N and H, in that order. Of these elements, sulphur is often considered to be the most detrimental to solidification cracking resistance because it readily combines with Fe and Mn to form (MnFe)S, a compound with a low melting point that readily spreads along grain boundaries. At lower temperatures, the higher alloy content in the intercellular or interdendritic regions may increase the local hardenability and give rise to the formation of microstructural regions with higher crack susceptibility. Strong partitioning of hydrogen during solidification increases the local hydrogen concentration at the solidification front, further increasing the risk of hydrogen accumulation at the stress concentration caused by the presence of a solidification crack.

Another important factor affecting the extent of segregation in carbon steels is the amount of carbon present in the alloy [88]. Figure 2.4.13 shows the upper left-hand corner of the Fe-Fe3C phase diagram. When the carbon content is below 0.1 wt%, the metal solidifies as primary δ-ferrite. At higher carbon contents, the primary crystals are δ-ferrite, but just below 1500°C, a peritectic reaction takes place and the remainder of the weld solidifies as austenite. The maximum solubility of sulphur in δ-ferrite is relatively high (0.18%), but in austenite it is significantly lower (0.05%). Consequently, there is a possibility at carbon contents higher than 0.1% that sulphur will be rejected to the grain boundaries of the primary austenite grains, promoting solidification cracking.

![Figure 2.4.13 A section of the Fe-Fe3C equilibrium phase diagram showing the peritectic reaction (Redrawn from [85]).](https://example.com/figure2413)

E6010 cellulosic consumables for pipeline girth welding typically contain around 0.1% carbon. Small variations in consumable chemistry can therefore affect the primary solidification behaviour of the weld and influence the solidification crack susceptibility. Phosphorus has much the same effect on the brittle temperature range as sulphur, and hence on solidification cracking. The formation of a phosphide eutectic is unlikely, but phosphorus segregates to the grain boundaries and act by either lowering the melting point of the interdendritic regions or by reducing grain boundary cohesion [88].

This is contrary to the findings reported by Oshita et al. [86], shown in Figure 2.4.12, indicating that solidification cracking is more prevalent in steels with carbon contents below 0.1%. Oshita attributed this
effect to the shrinkage-induced stress resulting from the phase transformation from δ-ferrite to γ-austenite on cooling. As steels with lower carbon contents solidify as primary δ-ferrite, the difference in volume resulting from the transformation to γ-austenite results in a lateral shrinkage stress that promotes solidification cracking.

Any study on weld metal solidification cracking and its influence on the initiation of WMHACC should therefore consider the effect of impurity elements, such as sulphur and phosphorus, as well as deliberate alloying additions, such as carbon and nickel, on the initiation of solidification cracks during cooling. The influence of electrode composition and flux formulation on WMHACC is considered in more detail in section 2.5.

The preceding discussion suggests that conditions during the field girth welding of root pass welds may be conducive to the formation of solidification cracks in cellulosic joints. These solidification cracks, if not removed completely by root pass grinding and hot pass welding, may promote WMHACC on cooling.

2.4.6 Non-metallic inclusions and their effect on WMHACC

High densities of non-metallic inclusions and precipitates in high-strength steels not only serve as preferential nucleation sites for acicular ferrite formation, but also act as hydrogen accumulation sites or traps in the microstructure at lower temperatures during the weld cooling cycle. Diffusible hydrogen is attracted to these sites due to the triaxial stress state that develops in the vicinity of the inclusions, and hydrogen accumulation occurs at lower temperatures. These traps can remove diffusible hydrogen from the weld, reducing susceptibility to WMHACC, but if the diffusible hydrogen content in the weld metal is sufficiently high and the local hydrogen concentration increases above the critical level required for cracking, the hydrogen traps can act as crack initiators [35].

Depending on the binding energy of the trap, hydrogen can be released slowly into the surrounding microstructure or become completely immobile. Hydrogen traps are generally divided into two groups based on their binding energies [89]:

- Weak (reversible) traps are characterised by binding energies lower than 30 kJ/mol, and typically include interstitial lattice sites, elastic stress fields, dislocations, stress fields around substitutional atoms, grain boundaries, vacancies and tempered martensite.
- Strong (irreversible) traps are characterised by binding energy levels higher than 60 kJ/mol and include microvoids and non-metallic inclusions such as Fe₂O₃, Fe₃O₄, MnS, Al₂O₃, SiO₂, MnO and Ce₂O₃.

In multipass welding the presence of irreversible hydrogen traps is considered beneficial as the heat introduced by subsequent welding passes releases only a minor part of the hydrogen contained in the traps. On cooling the released hydrogen is readily recaptured in strong traps before it can diffuse to potential crack initiation sites. Weak traps, however, easily and repeatedly release a considerable amount of hydrogen, even at lower temperatures. The diffusible hydrogen can accumulate at stress concentrations and increases the risk of WMHACC [35].

Weld metal deposited using cellulosic consumables can contain high densities of non-metallic inclusions, as well as carbide and nitride precipitates that act as effective hydrogen traps and modify the diffusivity of hydrogen at temperatures near or below ambient. The inclusions and precipitates that form in the molten weld pool are often trapped in the interdendritic or intercellular regions during solidification. In addition, there may

32
be carbides and nitrides that precipitate during cooling as a result of solid state transformations in the weld metal [90]. The interfaces of these non-metallic inclusions with the steel matrix can be considered irreversible trapping sites for hydrogen, and the location and morphology of these particles have a significant effect on HACC susceptibility [91].

Oxide inclusions, in particular, have high binding energies for hydrogen and are capable of trapping hydrogen at temperatures well above 100°C. Theoretically, a favourable distribution of non-metallic inclusions, particularly oxides, can effectively trap hydrogen and reduce its diffusible content sufficiently to prevent HACC when the temperature reaches the critical range on cooling [35,92]. If hydrogen is, however, present in sufficient quantities and its local concentration increases above the critical level, these hydrogen-rich inclusion interfaces can act as crack initiation sites. An increase in weld metal oxygen content has been found to promote intergranular fracture in weld metals due to an increase in the number of oxide inclusions at the prior austenite grain boundaries [93]. These oxide inclusions act as hydrogen traps and local stress concentrations, allowing intergranular cracking by decohesion at the oxide-matrix interfaces [94,95]. The weld metal oxide content is a strong function of the consumable formulation and the amount of deoxidisers added to the electrode flux. This can vary significantly between consumables and is considered in more detail in section 2.5.

Hydrogen accumulation at the stress fields surrounding non-metallic inclusions have been shown to promote decohesion and microplasticity by effectively lowering the local lattice strength until it is reduced to the magnitude of the stress concentration at the crack tip [94]. Non-metallic inclusions are often observed at the fracture surfaces of welds that failed by HACC, and are associated with the initiation and propagation of micropores, quasi-cleavage cracks and intergranular cracks. These cracks can propagate by further localised interaction with diffusible hydrogen in the stress field ahead of a crack tip [35]. Previous work at the University of Adelaide confirmed that non-metallic inclusions are often associated with the formation of hydrogen-induced microcracks in cellulosic weld metal, as shown in Figure 2.4.4. The complex inclusions in this figure were identified as iron-manganese oxides containing silicon, aluminium and titanium, and are most likely the products of deoxidation reactions during cooling [96].

![Figure 2.4.4 Non-metallic inclusions observed at the fracture surface and crack tip in weld metal deposited using an E6010 cellulosic consumable [96].](image-url)
Incoherent particles are often associated with HACC failure. The boundary conditions determining whether incoherent particles have a favourable or detrimental effect on WMHACC are, however, not clear. The consumable and flux formulations used during electrode production, as well as the welding conditions, affect the chemistry of the inclusions and precipitates and the diffusion of hydrogen. Quintana et al. [97] characterised the inclusions formed in an open arc weld deposit and showed that the formation of inclusions in these welds is a complex process dependent on the concentration of the different deoxidisers and nitrogen scavengers in the system. Different solidification conditions due to variations in the process parameters also affect the inclusion properties. A fine distribution of spheroidal inclusions decreases the likelihood of exceeding the critical hydrogen content required for crack initiation and hence reduces HACC susceptibility [92]. However, due to the high oxygen potential and low desulphurisation capacity of slags formed by cellulosic electrodes, cellulosic deposits typically exhibit lower inclusion populations with larger inclusion sizes than welds produced with basic electrodes [89].

The manganese/silicon ratio in the weld is also important in controlling the inclusion content. These elements are strong deoxidisers, and their partitioning between the weld metal and slag during welding is influenced by the oxygen content of the weld pool. Short arc lengths can reduce oxygen pick-up during welding, increasing the resulting alloy content of the weld metal. The detrimental effect of excess deoxidisers (manganese and silicon) in solution in the weld metal due to the reduced moisture contents of cellulosic consumables in hot, arid environments was discussed earlier, and can be attributed to the partitioning of these elements between the weld metal and the slag during welding. The actual levels of manganese and silicon in the weld metal can affect the inclusion volume fraction because of their influence on the melting range of the inclusions and the ease with which inclusions are removed from the liquid metal into the slag [89].

2.5. **CONSUMABLE FORMULATION**

As described above, the electrode and flux formulations of cellulosic consumables influence the partitioning of alloying elements between the weld metal and the slag, as well as the non-metallic inclusion content and the level of microsegregation in the weld. E6010 consumables have been available for many decades and both the core wire and cellulosic coating compositions have been developed to deliver a combination of good running characteristics and a weld deposit with the desired chemistry and mechanical properties. The risk of WMHACC under normal conditions is therefore considered to be low. Furthermore, to ensure that the risk of HACC is designed out at the procedure development stage, AS 2885.2:2016 [14] requires the welding procedure to be qualified using the worst possible HACC scenario, utilising preheat temperatures (if applicable) and heat input or burn-off rates at or near the lower end of the range to be qualified. Additionally, to simulate the worst-case scenario common practice during procedure qualification is to deposit the root and hot passes, followed by a 12 to 24 hour delay before non-destructive testing and deposition of the remaining passes.

AS 2885.2:2016 [14] states that the performance of pipeline girth welds made with cellulosic electrodes depends heavily on the design formulation and on the manufacturing quality assurance of the electrodes. Unfortunately, the specifications used in the classification of cellulosic electrodes do not always meet the needs of the pipeline industry [74].

The requirements for classification of carbon steel electrodes for manual metal arc welding are described in AWS A5.1/A5.1M: 2012 “Specification for carbon steel electrodes for shielded metal arc welding” [98]. This specification lists the minimum weld metal mechanical property requirements and chemical composition limits
that must be satisfied before the relevant AWS electrode classification can be used. In addition to the mechanical property requirements listed below, the chemical composition of weld metal deposited using E6010 electrodes has to fall within the limits shown in Table 4.5.1.

Minimum tensile strength: 60 ksi (430 MPa)
Minimum yield strength: 48 ksi (330 MPa)
Minimum % elongation: 22%

No impact requirements are listed unless specifically requested by the user.

Table 4.5.1 Chemical composition requirements for E6010 weld metal in accordance with AWS A5.1/A5.1M: 2012 [98] (weight percentage; single values represent maximum limits).

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
<th>Combined Mn+Ni+Cr+Mo+V</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.2</td>
<td>1.2</td>
<td>1.0</td>
<td>Not specified</td>
<td>Not specified</td>
<td>0.3</td>
<td>0.2</td>
<td>0.3</td>
<td>0.08</td>
<td>Not specified</td>
</tr>
</tbody>
</table>

It is evident from Table 4.5.1 that the acceptable chemical composition range for classification of a consumable as E6010 is very wide and covers a range of IIW carbon equivalents up to 0.5 and \( P_{ce} \) values up to 0.34 (not including the effect of boron). Although AS 2885.2 recommends a maximum carbon concentration of 0.17% for all cellulosic consumables, a maximum carbon equivalent is not specified in the standard for the lower strength E6010 electrodes. Of further concern is the absence of specified maximum limits for the impurities, sulphur and phosphorus. These elements are known to segregate strongly and promote cracking in the weld metal.

Due to the wide specification limits for cellulosic consumables, E6010 electrodes from different manufacturers, and even different electrode batches from the same manufacturer, may display significant variations in chemistry while still satisfying the requirements in AWS 5.1 for classification as E6010. It has been reported that weld metal deposited using commercial cellulosic E8010 consumables from different manufacturers and production batches showed significant variations in chemistry, resulting in a range of carbon equivalent values, microstructures and HACC susceptibility [34]. In recent years, a number of prominent welding consumable manufacturers have introduced changes to the electrode formulations of cellulosic consumables, with more recent consumable formulations containing higher amounts of alloying elements, producing enriched welds with harder microstructures and a higher susceptibility to hydrogen-assisted cold cracking. While investigating two batches of the same brand of commercial E8010 electrodes from different years (2003 and 2007), it was observed that not only did the manufacturer introduce boron into the coating, but the amounts of carbon, manganese and copper were increased, resulting in welds with significantly higher carbon equivalent values [34]. If there are significant differences in chemical composition between batches, the welding procedure is typically requalified to reduce the risk of hydrogen-assisted cold cracking. This is, however, not clearly specified in AS 2885.2, which lists changes in electrode type, size, and manufacturer or trade name as essential variables for procedure qualification, but not batch changes.

In addition to what is described above, small changes in consumable chemistry within the required specification limits can affect the properties of the weld metal in a number of ways:

- Alloying elements can affect the transformation behaviour of the weld metal during cooling. Alloying additions increase the hardenability of the weld deposit, as characterised by the carbon equivalent or \( P_{ce} \).
values. An increase in hardenability with an associated increase in deposit hardness can increase susceptibility to WMHACC, particularly in coarse-grained welds with high levels of segregation. The presence of higher alloying contents in the weld metal can also affect the transformation from austenite to ferrite on cooling and may influence the diffusion of hydrogen.

- Alloying additions and impurities influence the volume fraction, size and chemistry of non-metallic inclusions that form in the weld metal on cooling.
- Alloying elements and impurities with high partitioning coefficients segregate strongly to grain boundaries and to the weld centreline on solidification and can affect the grain boundary strength and cohesion. This renders the grain boundaries more prone to solidification cracking and fissuring on cooling.
- Changes in flux formulation, specifically in relation to the presence of metallic deoxidisers such as ferromanganese and ferrosilicon, can affect the deoxidation of the weld pool and the partitioning of oxygen between the weld metal and slag. Research has shown that lower concentrations of manganese in the weld pool increase the density of coarse oxide inclusions. The presence of a higher volume fraction of coarse oxide particles reduces the fracture toughness of the deposit and promotes hydrogen-induced decohesion and increased susceptibility to hydrogen-assisted cold cracking.
- Although more common in higher strength electrodes, trace amounts of boron may be present in weld metal deposited using more recent cellulosic welding consumables. Boron is a strong hardenability agent, even at very low concentrations, and is known to decrease grain boundary cohesion, thereby effectively weakening the grain boundary regions [99].

Small changes in consumable chemistry can therefore have a significant influence on the susceptibility of X70 steel welds to WMHACC, and it is important that uncertainty about the consistency of batches of electrodes from the same manufacturer, as well as equivalent E6010 consumables from various manufacturers with different formulations and levels of quality assurance during production, be considered.

2.6. **FIELD CONDITIONS**

In addition to welding parameters, consumable chemistry, microstructure and joint geometry (described in the preceding sections), it is important to recognise that variation in procedure variables encountered in the field may also play a role in determining susceptibility to WMHACC during the field welding of high-strength pipe [100]. The factors include:

- Operator skill, especially if the joint geometry is not ideal (excessive misalignment or uneven root gap), requiring adjustment of the welding technique to ensure sound full penetration welds.
- Weather conditions (high winds, temperature extremes and rain can affect the consumable moisture level).
- Welding in difficult terrain.
- Removal of the line-up clamp earlier than specified by AS 2885.2.
- The type of power source used (welds made using inverter power sources may be more sensitive to HACC than those produced with generator-driven power sources) [17].
- Moving the pipe before deposition of the required number of welding passes.
- Highly restrained conditions experienced when welding critical tie-ins.
2.7. SUMMARY OF THE FACTORS INFLUENCING WMHACC

The preceding discussion indicates that WMHACC is a more complex phenomenon than heat-affected zone HACC, and therefore more difficult to reliably predict and control. As shown earlier, the occurrence of WMHACC during field girth welding of X70 linepipe steel is influenced by a number of factors, briefly summarised below.

- **Preheat:** Preheating decreases the cooling rate after welding and allows more time for hydrogen to diffuse from the weld before the critical temperature range for cracking is reached on cooling. In more highly alloyed steels preheating also promotes the formation of higher temperature transformation products and softer microstructures. The preheat temperature (or the temperature of the base material prior to welding in the case of preheat-free welding) plays an important role in determining WMHACC susceptibility. Although preheating is not popular during field girth welding in Australia due to reduced production rates, it has been shown that crack-free cellulosic E6010 welds can be produced under a range of conditions by preheating to temperatures above 30°C [79] or 40°C [37].

- **Wall thickness:** According to AS 2885.2-2016 [14], with existing procedures, small diameter X70 gas transmission pipelines can be welded successfully without preheat in wall thicknesses of 10 mm or below, provided lifting practice and the delay time between the start of the root pass and the start of the hot pass are controlled, the heat input does not fall below 0.5 kJ/mm and the carbon equivalent is limited to a maximum of 0.40. An increase in pipe wall thickness will increase the cooling rate of the weld (if not compensated for by a higher heat input and/or preheat temperature), and reduce the time available for hydrogen to escape through diffusion. A heavier wall thickness is also likely to increase the restraint stresses in the joint, increasing susceptibility to WMHACC and solidification cracking.

- **Weld metal hydrogen content:** As described in Section 2.4.1, the residual hydrogen content in the weld metal when the critical temperature range for cracking is reached on cooling, H\textsubscript{R}, is determined by the initial weld metal hydrogen content, H\textsubscript{0}, and the cooling conditions. Although most of the absorbed hydrogen escapes from the weld during cooling, a higher initial hydrogen content increases the likelihood that the critical H\textsubscript{R} to initiate cracking will be reached at lower temperatures. Cellulosic consumables introduce enough hydrogen into the arc to supersaturate the weld metal, with accumulation at defects and discontinuities further increasing the local hydrogen content, H\textsubscript{local}, on cooling. Since the weld is supersaturated in hydrogen, the initial weld metal hydrogen content is of less importance than the cooling conditions in determining susceptibility to WMHACC.

- **Heat input and welding speed:** The fast welding speeds used during root pass welding to increase productivity tend to result in low heat inputs that are likely to increase the cooling rate of the weld and prevent effective hydrogen degassing. Research has, however, shown that in contrast to HAZHACC, increasing the heat input during welding is not necessarily beneficial in the case of WMHACC [101]. Excessive heat inputs and welding speeds increase the risk of forming solidification cracks that act as precursors for WMHACC, while fast welding speeds affect the bead geometry by promoting undercut, wagon tracks and window formation.

- **Composition and microstructure:** A higher alloying element content in the weld metal raises the hardenability (as quantified by higher CE or P\textsubscript{cm} values) and promotes the formation of harder, more crack susceptible microstructures. A crack susceptible microstructure is, however, considered the least important factor in determining susceptibility to WMHACC, and even weld microstructures regarded as having a
low susceptibility to HACC (such as acicular ferrite) can develop cracks when the hydrogen concentration, local stress intensity and temperature favour crack initiation. The weld metal non-metallic inclusion content and the deoxidation reactions that take place in the weld are more likely to play an important role.

- **Hot pass delay time:** The hot pass is deposited within a specified time after completion of the root pass to mitigate the detrimental effects of the rapid cooling rates experienced during root pass welding. The hot pass not only reduces the overall cooling rate and increases the time available for hydrogen diffusion, but also reduces the stress in the joint by increasing the weld throat thickness. If the delay time between the deposition of the root pass and the hot pass is longer than specified, the weld may be allowed to cool down below the critical temperature for crack initiation, promoting the formation of weld metal hydrogen-assisted cold cracks.

- **Lifting practice:** Lifting stresses caused by extreme lifting conditions (more than two pipe lengths lifted off the skids to shift the line-up clamp to the next joint) supplement the residual restraint stresses that develop during welding and are believed to increase the risk of WMHACC. The lifting action causes high tensile stresses and strains at the underside of the root weld and may promote solidification cracking if the weld is still at a high temperature. Lifting stresses are, however, believed to be low during normal lifting operations (no more than two pipe lengths off the skids) [102].

- **Level of restraint:** Tensile stresses inevitably arise from thermal contraction during cooling and may be supplemented by other stresses developed as a result of rigidity in the parts being joined. For thin walled pipes held in place by external or internal clamps, the restraint intensity is likely to be low, but an increase in wall thickness will increase the rigidity of the joint and the resulting restraint stresses. Higher levels of joint restraint are therefore expected to increase susceptibility to WMHACC.

- **Root bead geometry:** Weld restraint stresses generally concentrate at the weld toe and root, resulting in high local stresses, $\sigma_{\text{local}}$, that may promote hydrogen accumulation. Excessive misalignment, root undercut, slag inclusions and wagon tracks act as additional stress raisers and these root bead discontinuities have been shown to serve as preferential WMHACC initiation sites. The weld throat thickness will affect the nominal stress level in the weld, with an increase in weld throat thickness resulting in a decrease in the stress level in the joint.

Many of the factors described above influence WMHACC susceptibility by controlling the cooling rate after welding. The cooling time between 800°C and 500°C, $\Delta t_{8/5}$, affects the weld metal microstructure that forms on cooling (with harder microstructures increasing the risk of cracking), whereas the cooling time to 100°C, $\Delta t_{100}$, quantifies the time available for hydrogen diffusion from the weld before the critical temperature range for cracking is reached. Subsequently in Chapter 4 of the current study, $\Delta t_{100}$ is defined slightly different, due to experimental difficulties, as the time to cool from 800°C to 100°C. The cooling time to 100°C will also determine the required delay time before deposition of the hot pass, with very fast cooling rates reducing the available window for depositing the hot pass. High cooling rates brought about by low heat inputs, fast welding speeds, heavy wall thicknesses and welding with little or no preheat, reduce the time available for hydrogen effusion and increase the risk of forming crack-sensitive microstructures. Very slow cooling rates should, however, also be avoided to reduce the risk of grain growth, segregation and solidification cracking.
2.8. MODIFIED WELDING INSTITUTE OF CANADA TEST (MWIC)

Several tests exist for assessing the hydrogen-assisted cold cracking susceptibility of steel welds, and depending on how they are executed, can give reliable information of either a qualitative (such as crack/no crack regions based on process variables) or quantitative (such as parameter curves for limiting the occurrence of cold cracking) nature. These can be divided into two categories, depending on the method by which restraint stresses are imposed on the welded joint, namely self-restrained or externally loaded tests. The inherent stiffness or constraint on free shrinkage forms part of the design of the self-restrained cold cracking tests and can be varied depending on the structural design of the test. An intentional mechanical stress is applied to the externally loaded cold cracking tests, which should be varied according the relative stress intensity of the test sample’s real-life counterpart. The stress intensities generated during these two types of tests are designed to reproduce the residual stress experienced by actual welds in industry [103].

Included in the self-restrained category of cold cracking tests are those that are grouped under ‘slotted tests’, including the Tekken-, U-Weld, Leigh and Slot Weld tests, the tests designated as ‘circular slotted tests’, including the Circular Patch and Circular Groove tests, and those that utilise single-run overlay welds, including the Gapped Bead-on-Plate (GBOP) and Batelle Under bead Cracking tests. There are self-restrained tests designed specifically for testing cold cracking susceptibility in fillet welds and these include the Controlled Thermal Severity (CTS), Cruciform, Bending Restraint Weld Cracking (BRC) and Corner Joint Weld Cracking (CJC) tests. Instrumented self-restrained tests include the Rigid Restraint (RRC) and Instrumented Restraint Cracking (IRC) tests. Apart from these there are tests designed for specific applications and conditions such as the Bead Bend, Welding Institute of Canada (WIC), Schnadt-Fisco and Root Pass Butt Weld tests. The externally loaded tests include the Implant, Tensile Restraint Cracking (TRC), Longitudinal Tensile Restraint (LTP) and Augmented Strain Cracking (ASC) tests [103].

The Welding Institute of Canada (WIC) test was developed for testing the cold crack susceptibility of joints that undergo orbital welding, specifically stave pipe welding of pipeline steels. It is the method of choice for testing the hydrogen-assisted cold cracking susceptibility of the weld metal, rather than the HAZ, of root welds in pipelines. The WIC test is economical, reproducible and is suitable for butt welded joints. It can be performed for root/single run or multipass welds and lends itself to simple microscopic investigation. It is also widely used for acceptance and weldability tests [103].

This test method was selected for the current investigation since it is suitable for testing the cold cracking susceptibility of high-strength low-alloy steel weld metal and a wide range of variables can be isolated for investigation such as the plate thickness, welding process, welding parameters (welding speed and welding current), heat input, preheat temperature, cooling rate and critical hydrogen concentration.

Although hot (solidification) cracking has also been observed during the WIC test [84], it is mainly suited for testing cold cracking susceptibility. There are other tests (both self-restrained and externally restrained) better suited to testing hot cracking susceptibility [104] and it is recommended that the scope of the current study be expanded to incorporate appropriate hot crack susceptibility tests since the influence of welding parameters on hot cracking could not be determined accurately with the MWIC test. This will be discussed in greater detail in Section 4.5.

The standard WIC test setup is illustrated in Figure 2.8.1 with the dimensions of the test pieces indicated on the schematic. The test weld is deposited manually in the V-groove (indicated by the number 4 in Figure 2.8.1)
between the parent plates (API 5L X70 was used in the current investigation). The X70 plates are welded to a stiffener T-piece (number 3 in Figure 2.8.1) before the test weld is performed. These welds are termed anchor welds and are indicated by the number 5. The length of the X70 plates not anchored to the T-piece (in the vicinity of the test weld) by means of welding is called the restraint length and is indicated in Figure 2.8.1 as the 25 mm long unwelded sections on either side of the V-groove.

The level of restraint imposed on the test weld is the result of a combination of the residual stresses due to weld shrinkage and transformational stress. It can be adjusted by changing the restraint length of the test piece from 25 mm to 150 mm, with a shorter restraint length resulting in a higher restraint intensity, and hence higher residual stress [103]. It should, however, be noted that an increase in the restraint length reduces the cooling rate in addition to lowering the restraint intensity. A reduction in cooling rate promotes hydrogen effusion from the weld metal, and varying the restraint length alone is therefore not sufficient for isolating the effect of varied restraint intensity on WMHACC [79, 105].

According to Alam et al. [79], the restraint stress based on the restraint intensity in the WIC test can be estimated by means of Equation 2.8.1, where \( \sigma_w \) is the restraint stress (also called the working stress) and \( R_F \) is the restraint intensity, given by Equation 2.8.2, where \( L \) is the restraint length, \( E \) is Young’s modulus, \( h \) is plate thickness, \( B \) is the length of the weld, and \( A \) is the restraining plate cross section. Equation 2.8.1 is altered, however, in the presence of any stress concentration, including surface discontinuities in the form of mismatch, undercut (wagon tracks) or shallow penetration, and should be rewritten in the form of Equation 2.8.3, where \( \sigma_L \) is the working stress (now designated as the local stress associated with the stress concentration), and \( k \) is the stress concentration factor.

\[
\sigma_w = 0.04R_F \quad \text{Eq. 2.8.1}
\]
\[ R_F = \frac{Eh}{3.1L(1 + \frac{hB}{d})} \quad \text{Eq. 2.8.2} \]

\[ \sigma_L = k(\sigma_w) \quad \text{Eq. 2.8.3} \]

After a delay of 24 hours following welding, the test weld is removed from the test piece and sectioned into smaller pieces for optical microscopy.

From Equation 2.8.2, a decrease in the restraint length, \( L \), leads to an increase in the restraint intensity, whereas an increase in plate thickness results in an increase in the stress intensity [79]. These equations are, however, of an empirical nature and do not take into account the yield strength of the material (since the material will yield when the stress level exceeds the yield strength, especially at higher temperatures), nor does it take into account welding parameters such as the heat input.

The WIC test has been shown to accurately represent the restraint conditions experienced during field girth welding of pipelines, but according to Alam et al. [79], a restraint length of 100 mm is more representative of the local stress conditions experienced by the root pass on an actual pipeline girth weld. Equivalently, a 25 mm WIC restraint length with a preheat temperature of 80°C adequately represents the stresses generated during actual field welding. Unfortunately, the standard WIC test does not lend itself to automation and due to the inherent variations associated with manual welding, test repeatability cannot always be assured. A major drawback of the design of the WIC test sample is the short length of the test weld (50 mm). It does not compensate for difficulty in arc starting and welding defects such as end craters.

A 3 mm wafer made from medium-carbon steel is used as a spacer between the X70 plates and the stiffener T-piece and the anchor weld is deposited over it, as indicated in Figure 2.8.1. This leaves a 3 mm gap between the X70 plate and the stiffener T-piece in the vicinity of the restraint length. This small gap constitutes another drawback in the design of the WIC test since it does not allow easy access to the root of the weld for instrumentation of the test weld. Turbulent gas flow underneath the moving welding arc has also been observed to compromise weld quality [106].

To resolve these problems, Kurji et al. [106] suggested several modifications to the design of the standard WIC test piece, as shown in Figure 2.8.2. To resolve the problems associated with a short test welds, run-on/run-off tabs, fabricated using medium-carbon steel and prepared with the same V-groove as the test plate, are tacked onto the test plate (indicated by the number 2 in Figure 2.8.2). With smooth transitions between the run-on and run-off tabs and the test weld itself, the full length of the test weld can be deposited with sufficient quality to be analysed.

Medium carbon steel was selected for the run-on and run-off tabs to preserve the amount of X70 steel available for this study. These tabs are removed after the welding is completed and discarded. It is readily available and cost-effective and their chemistry do not interfere with that of the critical portion of the weld to an appreciable degree.

Another critical modification made to the design of the WIC test sample was to machine a groove with a radius of 10 mm in the stiffener T-piece directly below the test weld (number 4 in the insert shown in Figure 2.8.2). This groove allows for the egress of gasses from the tip of the arc without sacrificing weld quality. It also allows for access to the root for instrumentation of the weld. To compensate for any loss in stiffening from the
T-piece, the thickness of the backing plate of the T-piece (number 4 in Figure 2.8.2) was increased from 20 mm to 40 mm.

![Diagram of the T-piece test setup](image)

**Figure 2.8.2** Modified Institute of Canada (MWIC) test with (1) test/parent plate (API 5L X70), (2) run-on/run/off tabs (mild steel of 10 mm thickness), (3) spacer (3 mm ± 0.5), (4) backing plate on T-piece (of 40 mm thickness), and (5) stiffener of T-piece. The included angle is 60-70° [106].

Experimental test studies, supplemented by finite element analyses, were conducted to compare the standard WIC and modified WIC (MWIC) test pieces. The modified design was confirmed not to alter the restraint or thermal conditions of the test pieces in any significant way, but was shown to improve the repeatability and quality of the test welds [106].

As discussed in Chapter 3, the modified WIC test design was used throughout the current study and, as discussed in Section 4.3, the results obtained were comparable to those of other studies that incorporated the MWIC design [107].

### 2.9. WELDING TRIALS USING THE MWIC DESIGN

Kurji et al. [107] used the modified Welding Institute of Canada test at the University of Adelaide in a series of experiments aimed at delineating a safe welding parameter window for avoiding WMHACC of the root pass when welding API 5L X70 linepipe steel with E6010 electrodes. These authors welded X70 plate of 10 mm and 20 mm thickness with 4 mm diameter E6010 electrodes. A restraint length of 25 mm was used for all test welds.

Due to experimental difficulties experienced during testing, the University of Adelaide data set mostly covers the lower limits of the welding current, heat input and travel speed ranges used in industry, and only four of the University of Adelaide data points overlap with conditions used during field welding. The MWIC test data set therefore needs to be expanded significantly to cover the full range of welding parameters used during root pass welding in the field and to allow for meaningful conclusions to be drawn.
A total of 14 MWIC test pieces were welded at the University of Adelaide for 10 mm thick X70 plate at heat input levels ranging from 0.48 to 0.82 kJ/mm, and at preheat temperatures of 25°C, 40°C, and 50°C, whereas 25 MWIC test pieces were welded for 20 mm thick X70 plate at heat inputs ranging from 0.37 to 1.11 kJ/mm and preheat temperatures of 25°C, 50°C, and 100°C.

The test welds were sectioned in such a way that six surfaces could be prepared for observation with an optical microscope at a magnification of 400X. If a crack was observed and was longer than 5% of the height of the weld bead, the weld was considered to be cracked.

In both the 10 mm and 20 mm thick test batches, a threshold heat input level of 0.3 kJ/mm was observed below which the welds were of insufficient quality due to lack of penetration and incomplete sidewall fusion. The results from the 10 mm welded MWIC test pieces are shown in Figure 2.9.1(a) and those for 20 mm thick samples in Figure 2.9.1(b). Higher preheat temperatures and higher heat input levels during welding were shown to reduce susceptibility to WMHACC.

![Figure 2.9.1](image_url)

**Figure 2.9.1** Delineation of safe welding parameter windows (crack/no crack boundaries) as a function of heat input and preheat temperature determined using MWIC testing for (a) 10 mm thick X70, and (b) 20 mm thick X70 plates. A restraint length of 25 mm was used for both plate thicknesses [107].

Cracks observed in this study included both WMHACC (cold) and solidification (hot) cracks. The dotted lines in Figures 2.9.1(a) and (b) represent a boundary, below and to the left of which cracking was observed to occur and, above and to the right of which no cracking was observed. Kurji et al. [107] went further and expressed the boundaries shown in Figure 2.9.1 mathematically, given here as Equations 2.9.1 (for 10 mm thick plate) and 2.9.2 (for 20 mm thick plate).

\[
PH = 100 - 100HI \quad \text{Eq. 2.9.1}
\]

\[
PH = 70 - 50HI \quad \text{Eq. 2.9.2}
\]

Equations 2.9.1 and 2.9.2 give the minimum preheat temperature (PH) in °C to which the 10 mm and 20 mm thick plate should be heated, respectively, as a function of the heat input (HI) in kJ/mm to avoid cracking.

The investigation described in this thesis uses a similar approach to that of Kurji et al. [107] to define a safe welding parameter envelope for the avoidance of WMHACC, although limited to 10 mm thick X70 plate at a restraint length of 25 mm. As shown in Section 4.3, the results shown in Figure 2.9.1(a) were incorporated in the results of the current study and expanded upon.
2.10. REFERENCES


[15] Fletcher, L. RP01-02 - Identification of the controlling factors that affect the boundaries of whether a weld will be susceptible to WMHACC - Research proposal. Energy Pipelines CRC. 2010.


CHAPTER 3

EXPERIMENTAL PROCEDURE

The experimental procedure described in this chapter was designed to address the primary and secondary aims of the project, as explained in more detail in Chapter 1. The MWIC test, described in Chapter 2, was used at the dimensions specified in Figure 2.8.2, to determine the position of a series of crack/no crack boundaries as a function of welding parameters for the root pass of 10 mm thick API 5L X70 steel welded with E6010 consumables. The experimental welding matrix was designed on the basis of the welding parameter ranges (including welding current, heat input and travel speed) currently used during the field girth welding of X70 pipelines in Australia, as well as the guidelines given in AS 2885.2, the Australian Standard for pipeline welding [1]. Within these boundaries, a well-defined transition was observed from welding parameters that promote WMHACC in the root beads of pipeline welds, to those that did not result in cracking.

In addition to establishing crack/no crack boundaries, the experimental procedure was widened to investigate some of the factors that contribute to the occurrence of cracking during root pass welding. The relationship between these contributory factors was schematically illustrated by means of the WMHACC causal tree presented in Figure 1.1.2 in Chapter 1. Experiments therefore focused on measuring the weld metal cooling rates for cracked and uncracked samples, from which the hydrogen diffusion behaviour could be inferred. The microstructures of the weld metal and heat-affected zone, the hardness of the weld deposit, and the characteristics of inclusions as a function of changing welding parameters were also investigated. Since cracking can only occur in the presence of a tensile stress, residual stress levels were measured transversely across selected MWIC test welds using a neutron diffraction technique. The role of solidification cracking and its link to WMHACC were also investigated and are discussed in more detail in Chapter 4.

More information on the experimental procedure followed during the course of this investigation is given in the remainder of this chapter.

3.1. WELDING TRIALS

The experimental welding parameter matrix was designed on the basis of guidelines published in the Australian standard for pipeline welding [1] for the avoidance of WMHACC, as well as information gleaned from welding fabricators involved in pipeline welding in Australia [2]. The investigation into welding parameters focused on establishing a boundary that separates combinations of parameters (and their consequent heat input levels) into two different groups. The first group consists of those welding parameters that resulted in WMHACC during testing, and the second group those welds that were crack-free. This boundary is henceforth referred to as a crack/no crack boundary. It is hoped that the establishment of a safe window of welding parameters for the prevention of WMHACC will assist welders and welding coordination personnel involved in the field welding of X70 pipelines, especially during the qualification of welding procedure specifications, and lend confidence in qualified welding procedure specifications.

The investigation also incorporated preheating as a calibration between the cooling rates observed during the MWIC test and those of preheat-free girth welds on actual pipelines. As discussed in section 2.8, MWIC testing using a restraint length of 25 mm and a preheat temperature of 80°C has been shown to be more representative of the conditions during root bead welding in the field and therefore a preheat to 80°C was used in this
investigation. An intermediate preheat temperature of 50°C was also used in the current investigation in order establish the change in WMHACC behaviour with regards to temperature.

Testing consisted of mechanised manual metal arc welding (MMAW) of modified Welding Institute of Canada (MWIC) test pieces, as described in Section 2.8 of this study. The results presented throughout this study pertain to API 5L X70 base metal from a single batch to ensure a uniform chemical composition. The X70 material was supplied in the form of 20 mm thick plate and was machined down from both sides to a final thickness of 10 mm to conserve any centreline segregation that may have resulted from the continuous casting and rolling processes. A restraint length of 25 mm was used in all MWIC testing.

The chemical composition of the low-carbon, low-alloy X70 plate, as given by the supplier, is shown in Table 3.1.1. The carbon equivalent for this steel, calculated in accordance with equation 2.2.1, is 0.38. The root bead of each MWIC sample was deposited using 4 mm diameter E6010 electrodes from a single batch and supplier (trade name Lincoln Electric Pipeliner 6p(TM)) with a chemical composition range, specified by the supplier, as shown in Table 3.1.2. Table 3.1.3 gives the chemical composition of the weld metal, measured using Optical Emission Spectrometry, for a layered weld representative of weld metal with negligible dilution. The chemical analysis given in Table 3.1.3 yields a carbon equivalent of 0.26 and a Pcm value of 0.176.

| Table 3.1.1 Chemical composition (wt%) of X70 parent/base material (as given by the supplier), balance Fe. |
|---|---|---|---|---|---|---|---|---|---|---|
| C | Mn | Si | S | P | Nb | Ti | Cu | Ni | Mo | Cr | Al | V |
| 0.052 | 1.55 | 0.21 | 0.011 | 0.0097 | 0.041 | 0.012 | 0.15 | 0.19 | 0.18 | 0.026 | 0.039 | 0.029 |

| Table 3.1.2 Composition range (wt%) of the E6010 welding electrode used (as given by the supplier), balance Fe. |
|---|---|---|---|---|
| C | Mn | Si | S | P |
| 0.085-0.15 | 0.35-0.55 | 0.15-0.25 | 0.01-0.02 | 0.005-0.01 |

| Table 3.1.3 Chemical composition (wt%) of the weld metal deposited using E6010 electrodes, balance Fe. |
|---|---|---|---|---|---|---|---|---|---|---|
| C | Mn | Si | S | P | Ni | Cr | Mo | Cu | V | Ti | Nb | Al | B |
| 0.14 | 0.45 | 0.11 | 0.01 | 0.01 | 0.04 | 0.03 | 0.02 | 0.06 | 0.01 | 0.01 | <0.01 | <0.006 | <0.0005 |

The welding parameters used in this study incorporate parameter ranges from a series of qualified welding procedure specifications used in field welding of X70 pipelines by Australian pipeline construction companies (shown in Table 3.1.4), as well as restrictions and guidelines given in AS 2885.2 (Clause C9). The only reference AS 2885.2 makes to the heat input range required to avoid WMHACC is recommendation of a minimum electrode burn-off rate equivalent to a heat input level of 0.5 kJ/mm.

| Table 3.1.4 Welding parameter ranges according to qualified welding procedure specifications [2]. |
|---|---|---|---|---|---|
| Process | Electrode type | Electrical characteristics | Travel speed | Heat input |
| Electrode diameter | AWS classification | Polarity | Welding current | Arc voltage | 400-475 mm/min | 0.41-1.00 kJ/mm |
| MMAW | 4 mm | AWS A5.1 E6010 | DC+ | 130-170 A | 25-30 V | |

The Australian standard (S 2885.2) specifies the heat input of welding to be calculated by an arc efficiency of 1. Table 3.1.4 above follows the same convention.
Table 3.1.4 indicates that welding currents in the range of 130 to 170 A, travel speeds between 400 and 475 mm/min, and heat inputs ranging from 0.41 to 1.0 kJ/mm are currently detailed in welding procedure specifications used for field girth welding of X70 by Australian pipeline construction companies. Based on these parameters, a welding parameter matrix (a summary of which is shown in Table 3.1.5) was developed for the MWIC tests performed during the course of this project. The full welding parameter matrix used is given in Appendix A. As shown in Table 3.1.5, the welding current was varied from 120 A to 160 A in increments of 10 A (the maximum welding current of 170 A specified in Table 3.1.4 could not be achieved experimentally in MWIC samples without sacrificing root bead quality). At each current level, the heat input was varied from 0.3 kJ/mm to 1.0 kJ/mm in increments of 0.1 kJ/mm by adjusting the welding speed from a minimum of 172 mm/min to a maximum of 725 mm/min - the maximum speed achievable in the equipment used during the welding trials. This test matrix was repeated three times: at room temperature and at preheat temperatures of 50°C and 80°C.

<table>
<thead>
<tr>
<th>Current (A)</th>
<th>Voltage (V)</th>
<th>Travel speed (mm/min)</th>
<th>Preheat (°C)</th>
<th>HI (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>120</td>
<td>25</td>
<td>173-576</td>
<td>RT, 50°C, 80°C</td>
<td>0.3, 0.4, 0.5 ... 1.0</td>
</tr>
<tr>
<td>130</td>
<td>25</td>
<td>207-650</td>
<td>RT, 50°C, 80°C</td>
<td>0.3, 0.4, 0.5 ... 1.0</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>214-714</td>
<td>RT, 50°C, 80°C</td>
<td>0.3, 0.4, 0.5 ... 1.0</td>
</tr>
<tr>
<td>150</td>
<td>27</td>
<td>234-724</td>
<td>RT, 50°C, 80°C</td>
<td>0.3, 0.4, 0.5 ... 1.0</td>
</tr>
<tr>
<td>160</td>
<td>29.5</td>
<td>283-725</td>
<td>RT, 50°C, 80°C</td>
<td>0.38, 0.4, 0.5 ... 1.0</td>
</tr>
</tbody>
</table>

As shown in Table 3.1.5, a minimum heat input of 0.3 kJ/mm was not experimentally achievable at a welding current of 160 A due to limitations on travel speed, and the minimum achievable heat input at this current level was 0.38 kJ/mm.

The heat input (or welding energy) equation prescribed in AS 2885.2 [1] (and reprinted here as equation 3.1.1) was used throughout this study. It was rearranged in the form of equation 3.1.2 to determine the welding speed necessary to achieve the desired heat input at each current level, where $Q$ is the heat input (or arc energy) (kJ/mm), $E$ is the arc voltage (V), $I$ is the welding current (A), and $V$ is the welding speed (mm/min). The arc voltage was not controlled during this study, but was measured at each welding current level so that the required welding speed could be calculated using equation 3.1.2.

$$Q = \frac{EI}{V} \times \frac{60}{1000}$$  \hspace{1cm} \text{Eq. 3.1.1}

$$V = \frac{60}{Q} \times \frac{60}{1000}$$  \hspace{1cm} \text{Eq. 3.1.2}

For clarification, Table 3.1.6 shows an excerpt from the full welding parameter matrix that was used during the welding trials (extracted from Appendix A) for the full heat input range tested at a constant current level of 140 A, as well as the welding speeds used to achieve the desired heat input levels.

Where possible, extra MWIC samples were welded at half increments in the heat input range to more accurately position the measured crack/no crack boundaries.

Figure 3.1.1 shows the experimental test rig used in this study. Manual metal arc welding (MMAW) was performed using a voltage-controlled welding machine with direct current and electrode positive polarity (DC+). The MWIC test piece was clamped in a bench vice attached to a moveable lathe, which in turn was
connected to a variable speed drive motor. The MWIC test pieces were levelled with a clinometer to ensure consistent flat, downhand (1G) welding. During welding the MWIC test pieces were moved at the desired speed relative to the electrode, which was kept stationary and gravity fed into the V-groove of the MWIC test piece at an angle of 15° to the vertical (dragging).

Table 3.1.6 Excerpt from the full weld parameter matrix tested, at a set current level of 140 A and a preheat temperature of 50°C.

<table>
<thead>
<tr>
<th>Current (A)</th>
<th>Voltage (V)</th>
<th>Speed (mm/min)</th>
<th>Preheat (°C)</th>
<th>HI (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>140</td>
<td>25.5</td>
<td>714</td>
<td>50</td>
<td>0.3</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>536</td>
<td>50</td>
<td>0.4</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>429</td>
<td>50</td>
<td>0.5</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>371</td>
<td>50</td>
<td>0.6</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>306</td>
<td>50</td>
<td>0.7</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>268</td>
<td>50</td>
<td>0.8</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>238</td>
<td>50</td>
<td>0.9</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>214</td>
<td>50</td>
<td>1.0</td>
</tr>
</tbody>
</table>

Figure 3.1.1 Welding test rig for mechanised manual metal arc welding (MMAW) of MWIC test pieces.

The welding speed was controlled by means of a dial controller on the variable speed motor drive. Initially the speed at which the lathe moved (i.e. the welding speed) at each dial setting of the variable speed drive was measured. At each dial setting, the distance the lathe moved during a 20 second time interval was measured and recorded. Each measurement was repeated three times and the average distance was used to calculate the speed (mm/s) at which the lathe moved. This speed was then converted to mm/min and plotted against the dial setting, as shown in Figure 3.1.2. The equation of the trend line shown on the graph, where \( y \) is the lathe (welding) speed (mm/min) and \( x \) is the dial setting, could then be rewritten in the form of Equation 3.1.3 to determine the dial setting needed to produce the required welding speed.

\[
x = \frac{y + 4.1333}{71.444} \quad \text{Eq. 3.13}
\]

Testing was performed at room temperature (RT), and repeated at preheat temperatures of 50°C and 80°C in order to track the shift in crack/no crack boundaries with temperature. Preheating was performed using a
resistance heated ceramic pad connected to a thermal controller, placed over the MWIC test piece, as shown in Figure 3.1.3. The temperature of the MWIC test piece was raised above the required preheat temperature to allow enough time for the sample to be placed in the bench vice and levelled prior to welding.

The preheat temperature was measured using an infrared scanner with a measurement temperature range of -38°C to 520°C, an accuracy of ±2°C, and a distance-to-spot ratio of 8:1. It was used at close range in all cases. To compensate for a reduction in accuracy from the effects of surface reflectivity, the MWIC test piece surface was painted in the areas where temperature measurements were taken. Both black and white paints were used as these colours have different thermal absorption tendencies. Figure 3.1.4 shows where the paint was applied to the surface of the test pieces. White paint was used on one side of the V-groove and black on the other. Throughout the course of the experiments, an average difference of 3°C was measured between the black and white regions. Welding was performed when the average temperature of the black and white painted regions reached the preheat temperatures of either 50°C or 80°C.

![Figure 3.1.2 Moveable lathe speed (welding speed) (mm/min) vs. the motor’s variable speed-drive dial setting.](image)

According to AS 2885.2, Subclause 12.9 [1], the preheat temperature must be maintained and measured at a minimum distance of 75 mm away from the weld centreline prior to welding. The preheat temperatures were
therefore measured at the centreline of the MWIC test pieces and at a distance of 75 mm from the centre of the V-groove, as shown in Figure 3.1.4.

The cooling rates of the weld metal were measured for each MWIC test piece by lancing an R-type thermocouple into the molten weld pool. A Pt/Pt-13% Rh thermocouple with a wire diameter of 0.5 mm was encased in a twin bore ceramic insulating tube with the hot junction protruding about 5 mm from the tip and plunged into the molten weld pool behind the moving welding arc. The thermocouple was connected to a pico-data logger that recorded the temperature of the cooling weld metal in one second intervals. Figure 3.1.5(a) shows the thermocouple before welding, and (b) after solidification of the weld pool. The cooling rates of the heat-affected zones were not measured in the current study.

![Figure 3.1.4 Locations of preheating temperature measurements.](image)

![Figure 3.1.5 R-type thermocouple (a) before, and (b) after plunging into the molten weld pool behind the welding arc.](image)

### 3.2. SECTIONING OF THE MWIC TEST PIECES AFTER WELDING

After welding, the run-on/run-off tabs were removed and the critical section of the MWIC test piece, which consisted of the root bead, the heat-affected zone and a significant portion of X70 base/parent material, was removed using water jet cutting, as shown in Figure 3.2.1. Figure 3.2.2(a) shows the critical test sample after water jet cutting, while Figure 3.2.2 (b) shows the test sample after further sectioning using an abrasive cut-off wheel. Three sections were made through the test sample to reveal six root bead cross-sections. These
surfaces were subsequently prepared for metallographic investigation to facilitate detection of WMHACC, microstructural investigation and Vickers hardness testing.

The test pieces were not sectioned where the thermocouple had been plunged into the weld bead to avoid the effects of contamination from the thermocouple on the incidence of cracking, microstructure, and hardness measurements.

![Critical section of root bead](image1)

![Water jet cutting lines](image2)

**Figure 3.2.1** Positions where MWIC test piece was sectioned using water jet cutting.

![Section of root bead](image3)

(a) (b)

**Figure 3.2.2** Section of root bead (a) after water jet cutting and (b) after cutting with an abrasive cut-off wheel.

### 3.3. METALLOGRAPHIC PREPARATION

Cutting with the abrasive wheel divided the critical sections of the MWIC test pieces into four samples, as shown in Figure 3.2.2(b). The two samples removed from the central portion of the critical section each had two surface areas available for investigation, while the two samples from the edges had one surface each available for examination. These samples were ground and diamond polished to a 1 μm finish, followed by etching using 2% Nital (2% nitric acid and 98% ethanol) to reveal the microstructure. All metallographic samples were examined visually using an optical microscope.
3.4. **SCANNING ELECTRON MICROSCOPY**

Samples from a selection of MWIC test pieces were examined using a high-resolution JSM-7001F thermal field emission scanning electron microscope (SEM). Table 3.4.1 shows the welding parameters of the samples selected for the SEM investigation. Samples were selected from the batch of MWIC test pieces welded at room temperature (RT) and from the batch welded at a preheat temperature of 80°C. These samples represent the extremes in the preheat range used in this study with a representative spread on the welding parameters in order for any trends in the occurrence of non-metallic inclusions (NMI’s), with changes in both preheat temperatures and welding parameters, can be clearly discerned.

For all SEM investigations, a working distance of 10 mm and a probe current of 7 nA were used. The SEM investigation focused mainly on the non-metallic inclusions present in the weld metal. The investigation included imaging of the non-metallic inclusions in both secondary electron (SE) and back-scattered electron (BSE) imaging modes, energy dispersive spectroscopic (EDS) analysis of inclusions, EDS mapping of both inclusions and various microstructural features (with a minimum scanning time of at least one hour), and imaging of fracture surfaces (SE). Inclusions on polished and etched surfaces were found to be more visible and identifiable in BSE imaging mode, whereas inclusions on fracture surfaces were clearer in SE imaging mode.

<table>
<thead>
<tr>
<th>Current (A)</th>
<th>Voltage (V)</th>
<th>Speed (mm/min)</th>
<th>Preheat (°C)</th>
<th>HI (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>120</td>
<td>24</td>
<td>432</td>
<td>RT</td>
<td>0.4</td>
</tr>
<tr>
<td>120</td>
<td>24</td>
<td>257</td>
<td>RT</td>
<td>0.7</td>
</tr>
<tr>
<td>120</td>
<td>24</td>
<td>212</td>
<td>RT</td>
<td>0.85</td>
</tr>
<tr>
<td>120</td>
<td>24</td>
<td>180</td>
<td>RT</td>
<td>1.0</td>
</tr>
<tr>
<td>130</td>
<td>25</td>
<td>259</td>
<td>RT</td>
<td>0.8</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>546</td>
<td>RT</td>
<td>0.4</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>437</td>
<td>RT</td>
<td>0.5</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>318</td>
<td>RT</td>
<td>0.7</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>218</td>
<td>RT</td>
<td>1.0</td>
</tr>
<tr>
<td>160</td>
<td>29.5</td>
<td>648</td>
<td>RT</td>
<td>0.4</td>
</tr>
<tr>
<td>160</td>
<td>29.5</td>
<td>390</td>
<td>RT</td>
<td>0.7</td>
</tr>
<tr>
<td>160</td>
<td>29.5</td>
<td>278</td>
<td>RT</td>
<td>1.0</td>
</tr>
<tr>
<td>120</td>
<td>24</td>
<td>432</td>
<td>80</td>
<td>0.4</td>
</tr>
<tr>
<td>120</td>
<td>24</td>
<td>257</td>
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<td>0.7</td>
</tr>
<tr>
<td>120</td>
<td>24</td>
<td>173</td>
<td>80</td>
<td>1.0</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>536</td>
<td>80</td>
<td>0.4</td>
</tr>
<tr>
<td>140</td>
<td>25.5</td>
<td>306</td>
<td>80</td>
<td>0.7</td>
</tr>
<tr>
<td>160</td>
<td>29.5</td>
<td>708</td>
<td>80</td>
<td>0.40</td>
</tr>
<tr>
<td>160</td>
<td>29.5</td>
<td>472</td>
<td>80</td>
<td>0.60</td>
</tr>
</tbody>
</table>

The batch of samples welded at room temperature was the main focus of the investigation to establish trends between the observed non-metallic inclusions and changes in heat input. Selected samples from the batch welded at 80°C were added to investigate the change brought about by a higher preheat temperature.
For the sake of consistency, all images taken in BSE imaging mode were taken at a magnification of 1300X. This magnification was chosen because it included a large enough area of the microstructure to allow for the observation of a representative distribution of non-metallic inclusions. These images of non-metallic inclusions were further analysed using the open source online image processing software package ImageJ™.

With ImageJ™, specific information on non-metallic inclusions could be extracted from the SEM BSE images. This includes the non-metallic inclusion count (number of particles observed in each image), area percentage of particles in the image, and the average size of the particles in the image. A general equation for all particle morphologies, in random orientations, and viewed optically in two dimensions is that the area percentage can be equated to the volume percentage of the particles present [3], which was used as a means of quantifying the inclusion content in the remainder of this study. The following set of images represents a series of screenshots that shows the steps taken to extract the relevant information on non-metallic inclusions using the image analysis software.

![Screenshot](image)

**Figure 3.4.1** Screenshots from ImageJ software of step-wise analysis of inclusion particles. (Sample welded at a welding current of 140 A, a welding speed of 306 mm/min, a heat input of 0.7 kJ/mm, and a preheat temperature of 80°C).

Figure 3.4.1(a) shows a scanning electron micrograph taken using BSE imaging mode. The black dots in the image are the non-metallic inclusions under investigation. The first step in using the ImageJ™ software is to calibrate its measuring tool to the scale bar shown in the BSE image. The image is then cropped to contain only microstructural features, as shown in Figure 3.4.1(b). The threshold of the image is adjusted to produce a
binary image that displays the non-metallic inclusion particles in black on a white background, as shown in Figure 3.4.1(c). Traces of microstructural features are often left behind in the binary image as a very finely pixelated structure. To remove the noise introduced by this effect into the data and to focus on the most critical size distribution of non-metallic inclusions, the software is then set to ignore all particles with a cross sectional area smaller than 0.01 μm². During the final step in the analysis process, the software outlines each of the identified particles in red, as shown in Figure 3.4.1(d), and uses only these particles in the subsequent analyses.

3.5. HARDNESS MEASUREMENTS

Hardness values of the weld metal and heat-affected zones were measured using a calibrated Vickers hardness indenter with an applied load of 1 kg. Where average weld metal and heat-affected zone hardness values are presented in Chapter 4, the values were averaged from 15 to 20 individual measurements made at random. The heat-affected zone hardness values were measured at distances of 20 to 40 μm from the fusion line (the fusion line was often hard to accurately identify). Individual measurements were made at least 60 μm apart to avoid the effects of work hardening. Where hardness profiles are presented in this study, individual hardness measurements were taken at distances of 60 μm apart with a load of 0.5 kg.

3.6. RESIDUAL STRESS MEASUREMENTS

To determine the effects of heat input and preheat temperature on the post-weld residual stress levels in the root beads deposited in the MWIC test pieces, quantitative residual stress measurements were performed on a representative set of MWIC test pieces. The experiments were conducted on the KOWARI strain scanner in the OPAL (Open-Pool Australian Lightwater Reactor) research facility at the Australian Nuclear Science and Technology Organisation (ANSTO). Table 3.6.1 shows the welding parameters of the selected samples.

<table>
<thead>
<tr>
<th>Effect of preheat temperature at constant heat input</th>
<th>Current (A)</th>
<th>Voltage (V)</th>
<th>Speed (mm/min)</th>
<th>Preheat (°C)</th>
<th>HI (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>150</td>
<td>26</td>
<td>334</td>
<td>RT</td>
<td>0.7</td>
<td></td>
</tr>
<tr>
<td>150</td>
<td>26</td>
<td>334</td>
<td>50</td>
<td>0.7</td>
<td></td>
</tr>
<tr>
<td>150</td>
<td>26</td>
<td>334</td>
<td>80</td>
<td>0.7</td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Effect of heat input at constant preheat temperature</th>
<th>Current (A)</th>
<th>Voltage (V)</th>
<th>Speed (mm/min)</th>
<th>Preheat (°C)</th>
<th>HI (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>150</td>
<td>26</td>
<td>486</td>
<td>50</td>
<td>0.5</td>
<td></td>
</tr>
<tr>
<td>150</td>
<td>26</td>
<td>334</td>
<td>50</td>
<td>0.7</td>
<td></td>
</tr>
<tr>
<td>150</td>
<td>26</td>
<td>234</td>
<td>50</td>
<td>1.0</td>
<td></td>
</tr>
</tbody>
</table>

Table 3.6.1 shows two sets of samples. The first set of samples include MWIC test pieces welded at different preheat temperatures (RT, 50°C, and 80°C) at a constant heat input of 0.7 kJ/mm. These samples were selected to determine whether the preheat temperature influences the residual stress levels in the MWIC test pieces. The second set of samples contained MWIC test pieces welded at a constant preheat temperature (50°C) using different heat input levels (0.5, 0.7, and 1.0 kJ/mm). These samples were selected to determine the effect of changing heat input on the residual stress levels in the welded MWIC test pieces.

The samples selected for residual stress measurements were required to be free from any form of cracking, including both WMHACC and solidification cracking, as cracking relieves residual stresses [4]. The crack/no crack boundaries established in the first part of this study (discussed in detail in Section 4.3) were used as
guidelines for the selection of welding parameters for the preparation of samples for residual stress measurements. No cracking was observed in the heat input range tested at a preheat temperature of 50°C and a current level of 150 A. Therefore, this preheat temperature and current level were selected and kept constant while the heat input was varied (by varying the welding speed) during the welding of the second set of MWIC test pieces in Table 3.6.1. To complete the first set of samples, it was observed that at a heat input level of 0.7 kJ/mm all samples fell on the crack free side of the crack/no crack boundaries for both room temperature and a preheat temperature of 80°C. Therefore, welding two additional MWIC test pieces at these two preheat conditions, both at 0.7 kJ/mm heat input, completed the set.

Figure 3.6.1 shows a schematic cross-section of the weld metal and indicates the positions where residual stress measurements were taken. As shown, two sets of measurements were taken for each sample at two different offset levels from the underside of the joint. The first set was at an offset of 1.6 mm from the underside of the joint and the second at an offset of 3 mm from the underside of the joint. Measurements were taken transversely across the root bead, starting at a distance of 3 mm to the left of the centreline of the root bead and made at 1 mm increments. Between 18 and 23 measurements were made per sample. The aperture size on the beam collimator was set so that the gauge volume for each measurement taken was 2x2x2 mm. At each incremental point, the residual stress was measured in three orthogonal directions, i.e. in the normal (x), longitudinal (y), and transverse (z) directions, as shown schematically in Figure 3.6.2.

![Figure 3.6.1](image1)

**Figure 3.6.1** Positions of residual stress measurements at offset levels of 1.6 mm and 3 mm from the underside and 3 mm to the left of the centreline.

![Figure 3.6.2](image2)

**Figure 3.6.2** Orthogonal directions in which the stress was measured at each incremental point.

It should be noted that residual stresses are not measured directly. Rather, the diffraction angle between the incident and reflected neutron beams is measured and related to the lattice spacing of the crystal structure of the material. The angle of diffraction is related to the lattice spacing at each point of measurement using the Bragg Equation (equation 3.6.1), where \( \lambda \) is the wavelength of the incident beam, \( d \) is the lattice spacing of the strained crystal lattice in the direction of measurement, and \( \theta \) is the angle of diffraction [4,5].
At each point shown in Figure 3.6.1, the angle of diffraction was measured in all three directions \((x, y, z)\) and with the value of the wavelength \((\lambda)\) known, the values of the strained lattice spacing could be determined in each direction \(d_{xx}, d_{yy},\) and \(d_{zz}\) by means of Equation 3.6.1.

The unstrained lattice spacing, \(d_0\), was determined separately by preparing cubes of unstrained X70 pipeline steel and E6010 weld metal, each with a volume of 10x10x10 mm. The cube of X70 steel was sectioned from the plate material used to produce the MWIC test pieces and the cube of E6010 weld metal was prepared by building up layers of weld metal from which the cube could be removed. Both cubes were solution annealed by austenitising at 950°C for 1 hour, and furnace cooled for stress relief. It was assumed that in the unstrained lattice, the \(d_0\) value is independent of the direction of scanning and therefore an average of the \(d_0\) values for all three directions was taken as the unstrained \(d_0\) value for both the X70 and E6010 metal cubes, respectively. To ensure the \(d_0\) values are accurate, similar phases need to be present in both the annealed and as-welded microstructures of the samples tested. It was not within the capabilities of the heat treatment furnaces to ensure similar microstructures in the annealed microstructure as within the as-welded structure. Mostly ferrite and some pearlite would be present within the annealed microstructure and various morphologies of ferrite was observed in the as-welded microstructures (as discussed in greater detail in section 4.8). Although this inconsistency exists, it is assumed that since both microstructures are essentially ferritic, the discrepancy between the actual \(d_0\) value of the weld metal would not differ significantly from that measured in the annealed sample.

The strain at each point of measurement could then be determined in the three directions using equations 3.6.2, 3.6.3, and 3.6.4.

\[
\varepsilon_{xx} = \frac{(d_{xx} - d_0)}{d_0} \quad \text{Eq. 3.6.2}
\]

\[
\varepsilon_{yy} = \frac{(d_{yy} - d_0)}{d_0} \quad \text{Eq. 3.6.3}
\]

\[
\varepsilon_{zz} = \frac{(d_{zz} - d_0)}{d_0} \quad \text{Eq. 3.6.4}
\]

During the calculations, points of measurement were compared to the etched microstructures of sectioned and polished samples. This ensured that, where points of measurement were located in the weld metal, the \(d_0\) value for the weld metal was used in the calculation, while the \(d_0\) value for the X70 pipeline steel was used where points of measurement were located in the heat-affected zone or the parent/base material.

Residual stress calculations from diffraction measurements are based on continuum mechanics, therefore, once the lattice strains in the gauge volume are known, they can be related to macro-stresses in the material using Hooke’s Law [5]. Equations 3.6.5, 3.6.6, and 3.6.7 were used to calculate the macroscopic residual stress in all three orthogonal directions, where \(E\) is the Young’s modulus, and \(\nu\) is the Poisson’s ratio of the material.

\[
\sigma_{xx} = \frac{E}{(1 + \nu)(1 - 2\nu)} \left[ (1 - \nu)\varepsilon_{xx} + \nu(\varepsilon_{yy} + \varepsilon_{zz}) \right] \quad \text{Eq. 3.6.5}
\]

\[
\sigma_{yy} = \frac{E}{(1 + \nu)(1 - 2\nu)} \left[ (1 - \nu)\varepsilon_{yy} + \nu(\varepsilon_{zz} + \varepsilon_{xx}) \right] \quad \text{Eq. 3.6.6}
\]

\[
\sigma_{zz} = \frac{E}{(1 + \nu)(1 - 2\nu)} \left[ (1 - \nu)\varepsilon_{zz} + \nu(\varepsilon_{xx} + \varepsilon_{yy}) \right] \quad \text{Eq. 3.6.7}
\]
Uncertainties in the measurement of the lattice spacing values \(d_{xx}, d_{yy}, d_{zz}\), and \(d_0\) are always present during neutron diffraction scanning due to thermal, grain size and texture effects. Measurements are also dependent on the scanning/neutron counting time of the measurement as longer times lead to higher neutron counts at the peak positions [5] and are equivalent to the calculated standard deviations in the measurements. Uncertainties in the measurement of lattice spacing lead to uncertainties in the calculated strains and consequently, the calculated residual stresses. The uncertainties in the strains and stresses were calculated using an error propagation method [5] and are given by Equations 3.6.8 to 3.6.10, where \(\Delta d_{xx}, \Delta d_{yy}, \Delta d_{zz}\), and \(\Delta d_0\), are the standard deviations in the strained and unstrained lattice spacing measurements.

\[
\Delta \varepsilon_{xx} = \frac{d_{xx}}{d_0} \sqrt{\left(\frac{\Delta d_{xx}}{d_{xx}}\right)^2 + \left(\frac{\Delta d_0}{d_0}\right)^2}
\]

Eq. 3.6.8

\[
\Delta \varepsilon_{yy} = \frac{d_{yy}}{d_0} \sqrt{\left(\frac{\Delta d_{yy}}{d_{yy}}\right)^2 + \left(\frac{\Delta d_0}{d_0}\right)^2}
\]

Eq. 3.6.9

\[
\Delta \varepsilon_{zz} = \frac{d_{zz}}{d_0} \sqrt{\left(\frac{\Delta d_{zz}}{d_{zz}}\right)^2 + \left(\frac{\Delta d_0}{d_0}\right)^2}
\]

Eq. 3.6.10

Uncertainties in the residual stress calculations are shown as error bars on the measured residual stress curves given in Section 4.9 of this study, and were calculated using equations 3.6.11-3.6.13.

\[
\Delta \sigma_{xx} = \frac{E}{(1+v)(1-2v)} \sqrt{(1 - v)^2 \Delta^2 \varepsilon_{xx} + v^2 \Delta^2 \varepsilon_{yy} + v^2 \Delta^2 \varepsilon_{zz}}
\]

Eq. 3.6.11

\[
\Delta \sigma_{yy} = \frac{E}{(1+v)(1-2v)} \sqrt{(1 - v)^2 \Delta^2 \varepsilon_{yy} + v^2 \Delta^2 \varepsilon_{xx} + v^2 \Delta^2 \varepsilon_{zz}}
\]

Eq. 3.6.12

\[
\Delta \sigma_{zz} = \frac{E}{(1+v)(1-2v)} \sqrt{(1 - v)^2 \Delta^2 \varepsilon_{zz} + v^2 \Delta^2 \varepsilon_{xx} + v^2 \Delta^2 \varepsilon_{yy}}
\]

Eq. 3.6.13

The Von Mises stress, given by Equation 3.6.14 as \(\sigma_M\), was also determined from the calculated residual stress values. The Von Mises criterion states that when a component experiences a triaxial stress state and the Von Mises stress exceeds the yield strength of the material, as tested under simple tension, the material will yield (plastically deform) [6]. Therefore, the yield strength of both the E6010 weld metal (as given by the supplier) and that of the X70 plate material (given by the supplier in the as-rolled condition) were compared to the calculated Von Mises stress.

\[
\sigma_M = \frac{1}{\sqrt{2}} \sqrt{(\sigma_{xx} - \sigma_{yy})^2 + (\sigma_{xx} - \sigma_{zz})^2 + (\sigma_{yy} - \sigma_{zz})^2}
\]

Eq. 3.6.14

The results of the experiments described in this chapter are presented in Chapter 4.

3.7. REFERENCES


CHAPTER 4

RESULTS AND DISCUSSION

This chapter describes the results of the experimental procedure presented in Chapter 3 and explores the benefits of the project outcomes to the local pipeline industry. It highlights the importance of establishing reliable crack/no crack boundaries as a function of readily measurable welding parameters, and recommends guidelines for avoiding WMHACC during root pass welding of 10 mm thick X70 pipeline steel using E6010 cellulosic electrodes. Several factors contributing to the occurrence of WMHACC are also discussed in this section.

Throughout this study, 10 mm thick API 5L X70 plate material was used as substrate, and therefore all results pertain to pipelines with a maximum wall thickness of up to 10 mm. Single batches of X70 plate and E6010 electrodes were used throughout the course of this investigation, and a restraint length of 25 mm was utilised for all MWIC test pieces. As described in Chapter 2, the investigation focused on the effects of preheat temperature and welding parameters, in particular heat input, travel speed and welding current, on the incidence of WMHACC during root pass welding.

As discussed in section 1.1, the high diffusible weld metal hydrogen content of welds deposited using cellulosic consumables is the main precipitating factor for WMHACC in the presence of restraint stresses. Hydrogen contents were not measured directly during the course of this study, and any discussion relating to the influence of hydrogen diffusion rates in the welds is therefore inferred from the effect of the preheat temperature, heat input and welding parameters on the weld cooling rates.

This chapter also discusses additional factors that can give rise to WMHACC. It highlights those factors that increase the risk of cracking in terms of the prerequisites for HACC, i.e. rapid cooling rates (with direct bearing on the weld metal hydrogen content), high hardness levels in the weld metal and heat-affected zones, the weld metal microstructure, other microstructural features such as non-metallic inclusions, and the presence of residual stresses. The influence of heat input and preheat temperature on these factors is also investigated.

This chapter therefore aims to combine the fundamental factors that contribute to cracking into a single interconnected causal tree for WMHACC.

4.1. WELDING OF MWIC TEST PIECES

The welding parameter matrix summarised in Table 3.1.5 was used to generate a series of welded MWIC test pieces for further investigation. As described in Chapter 3, the minimum and maximum welding parameter limits shown in Table 3.1.5 were determined from a series of qualified welding procedure specifications used by pipeline construction companies across Australia. During the course of this investigation, it was observed that weld quality degraded severely at heat input levels below 0.3 kJ/mm and above 1.0 kJ/mm. Figure 4.1.1 shows images of MWIC test pieces welded at heat input levels below 0.3 kJ/mm where, as a consequence of the low heat input, the electrode burn-off rate was too low and a continuous weld bead could not form. The discontinuities in the weld beads are termed “windows” and typically result when enough heat is not available to melt the consumable electrode (E6010) at the rate required to fill the root gap. Root beads
deposited at heat input levels below 0.3 kJ/mm also tended to display lack of sidewall fusion and poor root penetration.

Figure 4.1.1 MWIC root welds deposited at a heat input below 0.3 kJ/mm displaying defects/discontinuities termed “windows.”

Figure 4.1.2 shows root beads deposited at heat input levels above 1.0 kJ/mm (using welding currents higher than 160 A). At these high heat input levels, defective root beads formed displaying discontinuities similar in appearance to the windows observed in Figure 4.1.1. The discontinuities in the root beads shown in Figures 4.1.2(a) and (b) were caused by melt-through, i.e. excessive amounts of base metal melted at the root of the V-groove due to overheating during welding. The higher welding heat input, coupled with an increase in arc force associated with higher welding currents (discussed in more detail in Section 4.3), resulted in the large discontinuities evident in these images.

Figure 4.1.2 MWIC test pieces welded at heat input levels above 1.0 kJ/mm (welding currents exceeding 160 A) displaying discontinuities because of melt-through.

Figure 4.1.3 shows the typical quality of root beads achievable in the heat input range of 0.3 to 1.0 kJ/mm. Root beads, such as those shown in Figure 4.1.3(a) and (b), were deemed of sufficient quality if the bead was continuous, exhibited a smooth surface profile, was centred in the middle of the V-groove, and displayed
adequate sidewall fusion (although sidewall fusion could only be verified once the welds were sectioned during metallographic preparation).

![Image](image1.jpg)

![Image](image2.jpg)

**Figure 4.1.3** Root beads of sufficient quality, deposited within the heat input range of 0.3 to 1.0 kJ/mm, specifically at (a) 0.6 kJ/mm, and (b) 0.4 kJ/mm.

The depth of penetration and weld bead profiles varied significantly with heat input and welding current within the range of the welding parameter values used. These observations are discussed in more detail in Section 4.2.

### 4.2. OBSERVATION OF CRACKS

Figure 4.2.1 shows two macrographs of etched root bead welds sectioned from MWIC test pieces. The weld in (a) was deposited at a heat input of 1.0 kJ/mm (at a welding current of 150 A and a travel speed of 234 mm/min), whereas the weld in (b) was deposited at a heat input of 0.5 kJ/mm (at a current of 150 A and a welding speed of 486 mm/min). Both welds were preheated to 50°C.

![Image](image3.jpg)

![Image](image4.jpg)

**Figure 4.2.1** Typical root bead morphologies of welds sectioned from the MWIC test pieces of (a) a high heat input (1.0 kJ/mm) and (b) an intermediate heat input (0.5 kJ/mm) weld. Scale is in mm.

These macrographs show the typical features of root beads deposited on MWIC test pieces with E6010 electrodes, such as the deep undercut at the toes of the weld (also referred to as wagon tracks). Incomplete or partial root penetration, evident in Figure 4.2.1(b), was observed in welds deposited at low heat input levels, and also in welds performed at high heat input levels, but low welding currents. Partial penetration in the
root creates a severe stress concentration. As discussed in Section 2.4.1, hydrogen has a propensity to diffuse towards stress raisers such as surface irregularities where a naturally high stress concentration already exists. If the hydrogen concentration reaches a critical level, these irregularities act as preferential crack initiation sites in the presence of a sufficiently high tensile stress. At higher heat input levels, root beads with larger weld metal cross sectional areas are deposited with more pronounced, and slightly more rounded wagon tracks. The sharper radius of the wagon tracks at lower heat input levels increases the inherent stress concentration at the toes of the weld.

Both WMHACC and solidification cracking are likely occurrences during the root pass welding of X70 pipelines with E6010 electrodes [1]. Both types of cracking were observed during this investigation, although WMHACC was more prevalent. Figure 4.2.2 shows examples of both WMHACC and solidification cracking as observed in root beads deposited without preheat. The welds were mounted and polished, but not etched, to highlight the cracks. The images in Figure 4.2.2 show the typical cracking behaviour displayed in most of the welds investigated throughout this study. The image pairs (c) and (d), (e) and (f), and (g) and (h) represent different surfaces of the same welds for heat inputs of 0.6 kJ/mm, 0.3 kJ/mm and 0.4 kJ/mm, respectively.

In the images shown it is evident that cracks mostly originate at the stress concentration presented by the wagon tracks and then propagate through the weld metal, as in Figures 4.2.2(a), (c), (d), (e) and (h). In some instances, cracks originate at the sharp stress concentrations presented by partial penetration at the weld root, as in (e), and in severe cases cracks extend fully from the wagon tracks to these sharp corners at the underside of the root bead as in (b), (f) and (g). Cracks are also observed to change their growth direction drastically, as shown in (a) and (g).

Even though the images in Figures 4.2.2(c) and (d) represent different cross-sections of the same weld (heat input 0.6 kJ/mm, welding current 130 A and travel speed 325 mm/min) they display cracks of distinctly different lengths. According to the outline of the Bluescope Steel Standard Procedure for the sectioning of MWIC test pieces, [2], any crack extending through less than 5% of the throat thickness of the root bead can be disregarded and will likely be repaired by the hot pass. These large differences in observed crack lengths stress the importance of preparing more than one cross-sectional surface area per weld.

An attempt was made to classify the crack lengths of welds that displayed WMHACC based on “total percentage of throat thickness cracked.” No observable trends with regard to the total fraction of the throat thickness cracked as a function of welding parameters were observed, and such observations were disregarded in the remainder of this study due to excessively large statistical variances and standard deviations. Any weld that contained a crack extending more than 5% through the thickness of the throat on any of the six surfaces investigated per weld was therefore regarded as cracked.

The images shown in Figures 4.2.2(e) and (f) are micrographs of two cross-sections of the same weld (heat input 0.3 kJ/mm, welding current 140 A and welding speed 714 mm/min). Both micrographs show WMHACC originating from wagon tracks, but the micrograph in (f) also shows evidence of solidification cracking concurrent with WMHACC. Solidification cracking does not seem to appear in isolation, but is often associated with WMHACC (discussed in more detail in Section 4.5). Solidification cracking is a high-temperature cracking phenomenon caused by segregation of elements such as sulphur and phosphorus to grain boundaries and the weld centreline, leading to the formation of weak, low melting temperature eutectic phases that readily crack under the influence of contraction and restraint stresses. On cooling, the centreline
cracks in (f) therefore originated at a higher temperature than the weld metal hydrogen-assisted cold cracks that initiated at the wagon tracks at temperatures closer to room temperature [1]. The presence of both types of cracks in a single weld suggests that solidification cracking did not relieve the restraint stress enough to prevent WMHACC at lower temperatures on cooling. Solidification cracks can also continue propagating through the weld metal in the form of weld metal hydrogen-assisted cold cracks since the crack tip of a centreline crack creates another severe stress concentration in the weld metal.

![Images of cracks](image1.png)  
(a) Heat input 0.6 kJ/mm, welding current 120 A and welding speed 300 mm/min, no preheat.  
(b) Heat input 0.5 kJ/mm, welding current 130 A and welding speed 390 mm/min, no preheat.  
(c) Heat input 0.6 kJ/mm, welding current 130 A and welding speed 325 mm/min, no preheat.  
(d) Heat input 0.6 kJ/mm, welding current 130 A and welding speed 325 mm/min, no preheat.  
(e) Heat input 0.3 kJ/mm, welding current 140 A, welding speed 714 mm/min, no preheat.  
(f) Heat input 0.3 kJ/mm, welding current 140 A, welding speed 714 mm/min, no preheat.  
(g) Heat Input 0.4 kJ/mm, welding current 140 A, welding speed 550 mm/min, no preheat.  
(h) Heat Input 0.4 kJ/mm, welding current 140 A, welding speed 550 mm/min, no preheat.

**Figure 4.2.2** Examples of WMHACC and solidification cracking observed on polished and unetched welds. Most cracks observed are WMHACC except for the two centreline cracks observed in f, which are solidification cracks.
Figure 4.2.3 shows a polished and etched microstructure in which the characteristic microstructural features of the root bead are highlighted. This micrograph shows a weld metal hydrogen-assisted cold crack that originated at the stress concentration caused by the presence of a wagon track at the toe of the root bead. It is evident that the wagon tracks have irregular profiles, which further increase the local stress concentrations.

![Figure 4.2.3](image)

**Figure 4.2.3** Typical weld metal hydrogen-assisted cold crack originating from a wagon track (heat input 1.0 kJ/mm, welding current 120 A, and welding speed 180 mm/min, preheat free).

The weld metal shows typical columnar grains extending towards the weld centreline of the root bead. A finer metallurgical microstructure exists within the columnar grains. The weld metal and heat-affected zone microstructures will be discussed in more detail in Section 4.8 of this study.

Figures 4.2.4(a) to (d) shows additional images of cracks belonging to the same weld, including two images, (c) and (d), taken on a high-resolution JSM-7001F thermal field emission scanning electron microscope (FEG-SEM). The weld in Figure 4.2.3 was deposited at a heat input of 1.0 kJ/mm, welding current of 120 A and a welding speed of 180 mm/min.

Figure 4.2.4(a) shows a crack that initiated at the wagon track shown in Figure 4.2.3 and propagated through the weld metal before deviating to the heat-affected zone where, in this case, crack propagation was arrested. The image also shows a second crack propagating upwards through the weld metal from the underside of the root bead where it originated due to partial penetration (b). Evident in (c) and (d) is the fibrous nature of the propagating crack, which suggests that the type of fracture encountered in these images is not fully cleavage (brittle), but probably a mixed ductile/brittle mode of fracture.

Cracks often propagate in a stepwise manner and Figure 4.2.4(c) shows a sharp change in the direction of crack propagation. Steps and sharp turns in the crack path may be due to the propagating crack changing its direction towards more brittle/crack sensitive parts of the microstructure. Sharp changes in direction could also be attributed to two cracks originating in separate parts of the microstructure, such as at a wagon track and a site of partial penetration (as was seen in (a)) that linked during fracture.

Figure 4.2.5 shows another example of a weld with two types of cracking: WMHACC in (a) to (c) and solidification cracking in (d). The weld was prepared at a heat input of 0.8 kJ/mm, a welding current of 130 A and a welding speed of 259 mm/min. Images (b) and (c) show the same weld metal hydrogen-assisted cold crack as in (a), but on different sectioned surfaces. The crack shown in (a) propagated partly through the weld metal, then continued for a short distance as a hairline crack before being arrested within the weld.
metal. The crack shown in (b) and (c) is a crack that extended through the full length of the weld throat and propagated from the wagon track at the toe of the weld to the stress concentration caused by partial penetration at the root of the weld. It propagated through the heat-affected zone before returning to the weld metal. This crack could have originated at the wagon track and propagated towards the root of the weld bead, or it could have originated at both the wagon track and the root and propagated to link as previously discussed.

![Images](a),(b),(c),(d)

**Figure 4.2.4** WMHACC in a weld deposited at a heat input of 1.0 kJ/mm, welding current of 120 A, and welding speed of 180 mm/min, preheat free.

The image in Figure 4.2.5(d) shows a hairline solidification crack in the centre of the weld. This short solidification crack was not observed on any of the other five cross-sectional surfaces prepared from this weld. As solidification cracks are generally not found near the surface of the weld bead and are typically located in the centre of the weld, it is assumed that solidification cracks are not readily repaired by the deposition of the hot pass and are therefore regarded as significant flaws, regardless of the percentage of the throat through which it propagated.

Figure 4.2.6 shows a weld metal hydrogen-assisted cold crack in a weld prepared at a heat input of 0.5 kJ/mm, a welding current of 140 A, and a welding speed of 437 mm/min. All cracks observed in this weld originated in the weld metal at the wagon tracks and propagated from the weld metal, through the heat-affected zone, before returning to the region of partial penetration at the root of the weld metal.
Figure 4.2.5 Weld metal hydrogen-assisted cold crack and a solidification crack (welded at a heat input of 0.8 kJ/mm, welding current of 130 A, and welding speed of 259 mm/min, preheat free).

Figure 4.2.6 Weld metal hydrogen-assisted cold crack in a weld prepared at a heat input of 0.5 kJ/mm, welding current of 140 A, and welding speed of 428 mm/min, preheat free.
Figure 4.2.7 shows a weld metal hydrogen-assisted cold crack in a weld prepared at a heat input of 0.4 kJ/mm, where the crack propagation path was through the weld metal without deviating into the heat-affected zone. The images in (a) and (c) show the crack extending from the wagon track at the toe of the weld, and the images in (b) and (d) show the crack extending from the root of the weld where there is a stress concentration due to partial penetration. In (c) and (d) it can be seen that there is a sharp kink in the crack propagation path. As before, the crack either changed direction as it propagated from the wagon track, or it originated at both the wagon track and the root of the weld and linked during propagation. The crack propagated in a stepwise manner.

![Figure 4.2.7 Weld metal hydrogen-assisted cold crack in a weld prepared at a heat input of 0.4 kJ/mm, welding current of 150 A and a welding speed of 608 mm/min.](image)

Figure 4.2.8 shows a few more examples of solidification cracks observed in welds deposited at different heat input levels. The images in (e) and (f) are high-resolution SEM micrographs, with image (f), which was taken at a slightly higher magnification than (e), showing a small weld metal hydrogen-assisted cold crack initiating at the tip of the solidification crack.

REFERENCES


4.3. **WMHACC – CRACK/NO CRACK BOUNDARIES**

As stated earlier, WMHACC was observed to be more prevalent throughout this study than solidification cracking, and is the main focus of the investigation. Solidification cracking will be discussed in more detail in Section 4.5. The following section focuses on developing guidance with regard to the selection of welding parameters that will result in WMHACC-free welds. The aim of this section is, therefore, to separate the welding parameters that result in root bead weld metal cracking from those that result in crack-free root beads.

The first attempt at establishing a crack/no crack boundary as a function of welding parameters for WMHACC is illustrated in Figure 4.3.1. Following the format proposed by the University of Adelaide [1,2], this figure shows the crack/no crack boundary as a function of heat input and preheat temperature. For each preheat temperature, the boundary was positioned at a heat input value calculated as the average between the maximum heat input level that resulted in WMHACC and the minimum heat input that resulted in a crack-free weld. The boundary was placed in such a way that all incidences of WMHACC are below and to the left of the boundary. For preheat-free welds, cracking was observed over the entire heat input range tested (0.3 to 1.0 kJ/mm).

Above and to the right of the boundary line (at high heat input levels and high preheat temperatures), crack-free welds were observed. The region below and to the left of the boundary line (at low heat input levels and
low preheat temperatures) contains all the welds that exhibited WMHACC, although there is considerable overlap between cracked and uncracked welds. No WMHACC was observed at high heat input levels and preheat temperatures because an increase in both these parameters reduces the cooling rate of the root bead. This allows more time for hydrogen to diffuse out of the weld metal, lowering the risk of WMHACC.

The crack/no crack boundary shown in Figure 4.3.1 represents the minimum heat input level for different preheat temperatures (room temperature, 50°C and 80°C) above which the likelihood of WMHACC occurring is considered to be low. This includes any combination of welding parameters that exceeds the minimum heat input level at the specified preheat temperature. Mathematically, the boundary is represented by equation 4.3.1, where \( y \) is the heat input (kJ/mm) and \( x \) is the preheat temperature (°C). Equation 4.3.1 shows no relation to previously published equations since the data is more populated over a wider range of parameters. Although the data from previous studies, as shown in Figure 4.3.2, correlates well with the current findings.

![Figure 4.3.1 Crack/no crack boundary (for WMHACC only) as a function of heat input and preheat temperature.](image)

\[
y = -0.0167x + 1.7833 \quad \text{Eq. 4.3.1}
\]

When a particular heat input level and preheat temperature are specified on a welding procedure specification (WPS), equation 4.3.1 can be used prior to qualification to ascertain whether there is a risk of WMHACC.

Figure 4.3.2 combines the crack/no crack data presented in Figure 4.3.1 (designated “UOW” in this figure) with the results of MWIC tests performed at the University of Adelaide for 10 mm thick plate and a restraint length of 25 mm (designated “UOA”) [1,2]. The data from UOA are mostly compatible with the data from the current study as it does not impact the position of the crack/no crack boundary established in Figure 4.3.1 in any significant way.

Below and to the left of the boundary line there is extensive overlap between cracked and uncracked welds, as shown in Figures 4.3.1 and 4.3.2. The overlap in this region of the graph indicates that the cracking behaviour is more complex than can be explained solely by a change in weld cooling rate. A different approach is therefore presented in the following figures, where the crack/no crack boundary is presented as a
function of the welding current and welding speed at a constant preheat temperature, rather than heat input and preheat temperature.

![Graph](image1)

**Figure 4.3.2** Crack/no crack boundary (for WMHACC only) as a function of heat input and preheat temperature, incorporating the data from the University of Adelaide.

Figure 4.3.3 shows the crack/no crack boundary for welds deposited without preheat. In this format there is considerably less overlap between cracked and uncracked welds. Cracking is most prevalent at high welding speeds and low welding currents.

![Graph](image2)

**Figure 4.3.3** Crack/no crack boundary as a function of welding speed and welding current for welds deposited without preheat.

The crack/no crack boundary shown in Figure 4.3.3 was determined in a similar way as the crack/no crack boundary shown in Figure 4.3.1. The boundary was positioned midway between the maximum welding speed that resulted in crack-free weld metal and the minimum travel speed that resulted in WMHACC, and
connected across the welding current range. The boundary formed in this way proved to be non-linear, but was indicated on the figure as a linear trend line. Because the boundary was established as a linear trend line across a non-linear data set, cracked welds were found below and to the right of the trend line. The trend line was taken as the crack/no crack boundary and shifted manually without changing the slope until all cracked welds were located above the line.

As the cooling rate increases with an increase in welding speed and a decrease in welding current, more hydrogen is trapped within the weld metal, embrittlement becomes more severe and the incidence of WMHACC increases (towards the upper left-hand corner in Figure 4.3.3). Conversely, as the heat input (welding energy) increases at slower welding speeds and higher welding currents, more hydrogen diffuses out of the weld metal and the incidence of WMHACC decreases (towards the lower right-hand corner of Figure 4.3.3). Although relying on the cooling rate alone does not adequately describe the observed trends as was evidenced in Figure 4.3.1.

Relevant data points generated by the University of Adelaide for preheat-free welds in 10 mm thick X70 and a restraint length of 25 mm are included in Figure 4.3.4. The data from the University of Adelaide agree well with the data from the current study, with all welds showing WMHACC situated to the left of the crack/no crack boundary established in Figure 4.3.3.

![Graph showing crack/no crack boundary as a function of welding speed and welding current](image)

**Figure 4.3.4** Crack/no crack boundary as a function of welding speed and welding current for welds deposited without preheat (University of Adelaide data points included).

Figures 4.3.5 and 4.3.6 show the crack/no crack boundaries for welds deposited at preheat temperatures of 50°C and 80°C, respectively. No University of Adelaide data points for these preheat temperatures are available.

As considered in more detail in Section 4.6, preheating prior to welding lowers the weld cooling rate allowing more hydrogen to diffuse out of the weld metal before the temperature is reached where cracking occurs. The higher the preheat temperature, the slower the cooling rate. This shifts the crack/no crack boundaries to lower heat input levels (i.e. higher welding speeds and lower welding currents).
The crack/no crack boundaries, presented as a function of welding speed and welding current, provide a more practical guide for avoiding WMHACC than the crack/no crack boundaries shown as a function of heat input and preheat temperature. Welding current and travel speed are readily measured and controlled in the field. Since the arc voltage is not expected to change significantly provided the arc length is maintained within acceptable limits, this suggests that the heat input, and therefore the weld cooling rate, can be readily controlled to avoid cracking.

![Graph](image)

**Figure 4.3.5** Crack/no crack boundary as a function of welding speed and welding current for welds deposited at a preheat temperature of 50°C.

![Graph](image)

**Figure 4.3.6** Crack/no crack boundary as a function of welding speed and welding current for welds deposited at a preheat temperature of 80°C.

For the three preheating conditions used in this study, i.e. no preheat (room temperature taken as an average of 25°C), 50°C, and 80°C, the crack/no crack boundaries shown in Figures 4.3.3, 4.3.5, and 4.3.6 indicate the maximum welding speed that can safely be maintained at different current levels without risking WMHACC. Since the crack/no crack boundaries are shown as straight lines in Figures 4.3.3, 4.3.5 and 4.3.6, equations...
4.3.3, 4.3.4 and 4.3.5 can be used to describe the boundaries, where \( y \) is the welding speed (mm/min) and \( x \) is the welding current (A).

\[
\begin{align*}
y_{RT} &= 17.34x - 2065.3 \\
y_{50^\circ C} &= 15.546x - 1684.7 \\
y_{80^\circ C} &= 16.199x - 1639
\end{align*}
\]  
**Eq. 4.3.3** **Eq. 4.3.4** **Eq. 4.3.5**

Prior to the qualification of any WPS, equations 4.3.3 to 4.3.5 can be used to calculate the maximum welding speed that will result in crack-free root bead welds for a specific preheat temperature and welding current. If, for example, a welding current of 130 A is specified without any preheating, equation 4.3.3 predicts that crack-free welds can safely be produced at welding speeds below 189 mm/min, whereas for a welding current of 150 A and no preheat, welding speeds below 536 mm/min can be used to produce crack-free root beads. Equation 4.3.5 predicts that for a welding current of 130 A at a preheat temperature of 80\(^\circ\)C, welding speeds below 467 mm/min produce crack-free welds.

Figure 4.3.7 shows the three crack/no crack boundaries established for room temperature and preheat temperatures of 50\(^\circ\)C and 80\(^\circ\)C on a single graph. This figure highlights the shift in the crack/no crack boundaries towards higher welding speeds as the preheat temperature increases. As already stated, the boundaries are based on linear trend lines across the interface between cracked and uncracked welds, with the slopes remaining constant with an increase in preheat temperature. Increasing the preheat temperature from room temperature to 50\(^\circ\)C results in an average increase in allowable welding speed of about 155 mm/min at each current level. With an increase in preheat temperature from room temperature to 80\(^\circ\)C, the corresponding increase in allowable welding speed is around 273 mm/min for each welding current.

![Figure 4.3.7 Shift in crack/no crack boundaries with an increase in preheat temperature.](image)

According to Alam et al. [3], the cooling rates of the root beads deposited in WIC test pieces at room temperature are more conservative (faster) than those experienced during the welding of actual pipelines. This ensures that the crack/no crack boundaries shown in Figure 4.3.7 are conservative and provides a higher level of confidence that crack-free welds will be produced in the field if these guidelines are followed.
It is important to note that the behaviour of welds performed at similar heat input and preheat levels are not always consistent, resulting in the overlap observed between cracked and uncracked welds deposited at the same heat input levels (as shown in Figures 4.3.1 and 4.3.2). This suggests that the heat input, and hence the cooling rate, is not the only factor determining whether WMHACC occurs, and that the WMHACC sensitivity can be managed with appropriate adjustment in the welding current. This behaviour can be explained by examining the geometry of the weld cross-sections.

Tables 4.3.1 and 4.3.2 show macrostructures of polished and etched welds deposited without preheat and at a preheat temperature of 50°C. These tables map the change in weld geometry as a function of the welding current and welding speed in the same order as they appear in Figure 4.3.3. Some welds (such as those with \( H_I = 0.9 \text{ kJ/mm} \) and \( H_I = 1.0 \text{ kJ/mm} \) at a current level of 130 A) showed clean breaks and in such cases only the side containing the weld metal was metallographically prepared.

Tables 4.3.1 and 4.3.2 indicate that, at constant heat input levels, subtle changes occur in the geometry of the root beads with increasing welding current. At constant heat input, the depth of penetration increases as the welding current increases. As the welding current approaches 160 A (the maximum welding current examined in this study), the root bead progressively bows out at the underside of the joint, eliminating the stress concentration due to partial penetration at the root. These changes in root bead geometry are in agreement with the findings of Rokhlin et al. [4] and Yang et al. [5] who found that increasing arc force (due to higher welding currents) causes deeper depression of the weld pool surface during welding, and consequently increased depth of penetration. This is explained in more detail below.

The current-density distribution in the arc strongly influences the arc force exerted on the weld pool surface [6]. The magnetic field generated by the current flow exerts an electromagnetic force on the charge carriers (electrons) in the arc. These forces are known as Lorentz forces and in the case of a cylindrical arc, exhibit only radial components. The stronger the welding current, the stronger the magnetic field and hence, the Lorentz forces. These forces increase the pressure in the arc, and higher pressure is therefore generated within the welding arc at higher welding currents [5].

The radial component of the Lorentz force decreases as the radius of the arc increases and in the case of a typical welding arc, the radius increases towards the weld pool (in the direction of the workpiece). This variance in the arc radius causes a gradient in the pressure that develops, with lower pressure found closer to the workpiece. The arc plasma is accelerated through this pressure system from high pressure to low pressure. This generates an arc force in the direction of the workpiece (and weld pool) and thus, a higher welding current results in a stronger arc force directed at the weld pool [4-6].

Rokhlin et al. [4] found that the arc force has a quadratic dependence on the welding current. These authors demonstrated that impingement of the plasma jet causes depression in the weld pool surface, and that a linear relationship exists for weld pool depression and arc force. They also found that pool depression directly affects depth of penetration. As the arc force increases, the depth of the weld pool increases and more molten metal is pushed out towards the sides. This results in the plasma jet (heat source) being closer to the solid/liquid interface at the bottom of the pool, causing more efficient melting of this region and hence, increasing the depth of penetration [4,5]. This justifies the deeper weld pools observed for higher welding currents at constant heat input levels observed in Tables 4.3.1 and 4.3.2.

Yang et al. [5] demonstrated that the arc length varies with the depth of pool depression. With an increase in the arc length, the arc voltage is expected to increase slightly. Increasing the welding current will therefore
also result in a higher arc voltage, as shown in the experimental matrix in Table 3.1.5. During the course of this project, the slight increases in measured arc voltage were compensated for by means of alterations in the welding speed (as described in Section 3.1).

**Table 4.3.1** Weld geometry as a function of welding current (A) and welding speed (mm/min) for welds deposited without preheat.

<table>
<thead>
<tr>
<th>Welding current, A</th>
<th>120 A</th>
<th>130 A</th>
<th>140 A</th>
<th>150 A</th>
<th>160 A</th>
</tr>
</thead>
<tbody>
<tr>
<td>V = 725 mm/min</td>
<td><img src="image1.png" alt="Image" /></td>
<td><img src="image2.png" alt="Image" /></td>
<td><img src="image3.png" alt="Image" /></td>
<td><img src="image4.png" alt="Image" /></td>
<td><img src="image5.png" alt="Image" /></td>
</tr>
<tr>
<td>HI = 0.3 kJ/mm</td>
<td><img src="image6.png" alt="Image" /></td>
<td><img src="image7.png" alt="Image" /></td>
<td><img src="image8.png" alt="Image" /></td>
<td><img src="image9.png" alt="Image" /></td>
<td><img src="image10.png" alt="Image" /></td>
</tr>
<tr>
<td>V = 607 mm/min</td>
<td><img src="image11.png" alt="Image" /></td>
<td><img src="image12.png" alt="Image" /></td>
<td><img src="image13.png" alt="Image" /></td>
<td><img src="image14.png" alt="Image" /></td>
<td><img src="image15.png" alt="Image" /></td>
</tr>
<tr>
<td>HI = 0.4 kJ/mm</td>
<td><img src="image16.png" alt="Image" /></td>
<td><img src="image17.png" alt="Image" /></td>
<td><img src="image18.png" alt="Image" /></td>
<td><img src="image19.png" alt="Image" /></td>
<td><img src="image20.png" alt="Image" /></td>
</tr>
<tr>
<td>V = 428 mm/min</td>
<td><img src="image21.png" alt="Image" /></td>
<td><img src="image22.png" alt="Image" /></td>
<td><img src="image23.png" alt="Image" /></td>
<td><img src="image24.png" alt="Image" /></td>
<td><img src="image25.png" alt="Image" /></td>
</tr>
<tr>
<td>HI = 0.5 kJ/mm</td>
<td><img src="image26.png" alt="Image" /></td>
<td><img src="image27.png" alt="Image" /></td>
<td><img src="image28.png" alt="Image" /></td>
<td><img src="image29.png" alt="Image" /></td>
<td><img src="image30.png" alt="Image" /></td>
</tr>
<tr>
<td>V = 371 mm/min</td>
<td><img src="image31.png" alt="Image" /></td>
<td><img src="image32.png" alt="Image" /></td>
<td><img src="image33.png" alt="Image" /></td>
<td><img src="image34.png" alt="Image" /></td>
<td><img src="image35.png" alt="Image" /></td>
</tr>
<tr>
<td>HI = 0.6 kJ/mm</td>
<td><img src="image36.png" alt="Image" /></td>
<td><img src="image37.png" alt="Image" /></td>
<td><img src="image38.png" alt="Image" /></td>
<td><img src="image39.png" alt="Image" /></td>
<td><img src="image40.png" alt="Image" /></td>
</tr>
<tr>
<td>V = 257 mm/min</td>
<td><img src="image41.png" alt="Image" /></td>
<td><img src="image42.png" alt="Image" /></td>
<td><img src="image43.png" alt="Image" /></td>
<td><img src="image44.png" alt="Image" /></td>
<td><img src="image45.png" alt="Image" /></td>
</tr>
<tr>
<td>HI = 0.7 kJ/mm</td>
<td><img src="image46.png" alt="Image" /></td>
<td><img src="image47.png" alt="Image" /></td>
<td><img src="image48.png" alt="Image" /></td>
<td><img src="image49.png" alt="Image" /></td>
<td><img src="image50.png" alt="Image" /></td>
</tr>
<tr>
<td>V = 225 mm/min</td>
<td><img src="image51.png" alt="Image" /></td>
<td><img src="image52.png" alt="Image" /></td>
<td><img src="image53.png" alt="Image" /></td>
<td><img src="image54.png" alt="Image" /></td>
<td><img src="image55.png" alt="Image" /></td>
</tr>
<tr>
<td>HI = 0.8 kJ/mm</td>
<td><img src="image56.png" alt="Image" /></td>
<td><img src="image57.png" alt="Image" /></td>
<td><img src="image58.png" alt="Image" /></td>
<td><img src="image59.png" alt="Image" /></td>
<td><img src="image60.png" alt="Image" /></td>
</tr>
<tr>
<td>V = 200 mm/min</td>
<td><img src="image61.png" alt="Image" /></td>
<td><img src="image62.png" alt="Image" /></td>
<td><img src="image63.png" alt="Image" /></td>
<td><img src="image64.png" alt="Image" /></td>
<td><img src="image65.png" alt="Image" /></td>
</tr>
<tr>
<td>HI = 0.9 kJ/mm</td>
<td><img src="image66.png" alt="Image" /></td>
<td><img src="image67.png" alt="Image" /></td>
<td><img src="image68.png" alt="Image" /></td>
<td><img src="image69.png" alt="Image" /></td>
<td><img src="image70.png" alt="Image" /></td>
</tr>
</tbody>
</table>
Table 4.3.2 Weld geometry as a function of welding current (A) and welding speed (mm/min) for welds deposited at a preheat temperature of 50°C.

<table>
<thead>
<tr>
<th>Welding current, A</th>
<th>120 A</th>
<th>130A</th>
<th>140 A</th>
<th>150 A</th>
<th>160 A</th>
</tr>
</thead>
<tbody>
<tr>
<td>V = 576 mm/min H1 = 0.3 kJ/mm</td>
<td>V = 650 mm/min H1 = 0.3 kJ/mm</td>
<td>V = 714 mm/min H1 = 0.3 kJ/mm</td>
<td>V = 725 mm/min H1 = 0.3 kJ/mm</td>
<td>V = 725 mm/min H1 = 0.38 kJ/mm</td>
<td></td>
</tr>
<tr>
<td>V = 432 mm/min H1 = 0.4 kJ/mm</td>
<td>V = 487 mm/min H1 = 0.4 kJ/mm</td>
<td>V = 550 mm/min H1 = 0.4 kJ/mm</td>
<td>V = 607 mm/min H1 = 0.4 kJ/mm</td>
<td>V = 708 mm/min H1 = 0.4 kJ/mm</td>
<td></td>
</tr>
<tr>
<td>V = 346 mm/min H1 = 0.5 kJ/mm</td>
<td>V = 390 mm/min H1 = 0.5 kJ/mm</td>
<td>428 mm/min H1 = 0.5 kJ/mm</td>
<td>V = 486 mm/min H1 = 0.5 kJ/mm</td>
<td>V = 556 mm/min H1 = 0.5 kJ/mm</td>
<td></td>
</tr>
<tr>
<td>V = 300 mm/min H1 = 0.6 kJ/mm</td>
<td>V = 325 mm/min H1 = 0.6 kJ/mm</td>
<td>V = 371 mm/min H1 = 0.6 kJ/mm</td>
<td>V = 397 mm/min H1 = 0.6 kJ/mm</td>
<td>V = 472 mm/min H1 = 0.6 kJ/mm</td>
<td></td>
</tr>
<tr>
<td>V = 257 mm/min H1 = 0.7 kJ/mm</td>
<td>V = 295 mm/min H1 = 0.7 kJ/mm</td>
<td>V = 318 mm/min H1 = 0.7 kJ/mm</td>
<td>V = 334 mm/min H1 = 0.7 kJ/mm</td>
<td>V = 405 mm/min H1 = 0.7 kJ/mm</td>
<td></td>
</tr>
<tr>
<td>V = 225 mm/min H1 = 0.8 kJ/mm</td>
<td>V = 243 mm/min H1 = 0.8 kJ/mm</td>
<td>V = 267 mm/min H1 = 0.8 kJ/mm</td>
<td>V = 304 mm/min H1 = 0.8 kJ/mm</td>
<td>V = 354 mm/min H1 = 0.8 kJ/mm</td>
<td></td>
</tr>
<tr>
<td>V = 200 mm/min</td>
<td>216 mm/min</td>
<td>238 mm/min</td>
<td>V = 270 mm/min</td>
<td>V = 315 mm/min</td>
<td></td>
</tr>
</tbody>
</table>
The effect of increased depth of penetration due to weld pool depression resulted in root bead geometries with less severe stress concentrations at the root of the weld as the welding currents increased. Even though the root bead creates its own inherent stress concentration, it is less severe than the sharp stress concentrations resulting from partial penetration and the incidence of cracking therefore decreases with increasing welding current (at constant heat input levels).

It is also evident from Tables 4.3.2 and 4.3.1 that the application of preheat increases the overall depth of penetration even further, as expected from an increase in the heat generated during welding. The preheat temperature therefore influences the WMHACC behaviour of the welds by slightly altering the weld bead geometry, in addition to reducing the cooling rate.

Figure 4.3.8 shows a direct comparison of two welds, both deposited at a heat input of 0.7 kJ/mm (at a welding current of 150 A and a welding speed of 334 mm/min). Figure 4.3.8(a) was deposited without preheat, whereas (b) was preheated to a temperature of 80°C. The increased depth of penetration is evident in the preheated weld.

![Figure 4.3.8 Comparison of weld bead geometries and depth of penetration for welds deposited at a constant welding heat input of 0.7 kJ/mm (welding current 150 A and welding speed of 334 mm/min) at (a) room temperature and (b) a preheat temperature of 80°C.](image)

Variations in the root gaps of butt joints may also play a role in determining the sensitivity of root beads to WMHACC. Kumar et al. [7] found that the geometry of the root bead is strongly affected by the root gap. These authors reported that the depth of penetration increased and under-filling became more pronounced as the root gap increased, but that the widths of weld beads and heat-affected zones remained unchanged. With a change in root bead geometry, especially the depth of penetration (as seen in Tables 4.3.1 and 4.3.2), the stress concentration in the root of the weld is reduced, with wider root gaps leading to deeper penetration.

Section 5 of the Australian Standard AS2885.2-2016 [8] outlines the limitations on joint design in pipeline welding. Section 5.2 stipulates the maximum allowable range on root gaps for joining pipes of equal thickness in the form of Figure 4.3.9. In order to be in compliance with the standard, a root gap in the range of 1.4±0.6 mm (translating to an allowable gap of 0.8 to 2.0 mm) is prescribed. The current study used
MWIC test pieces with a constant root gap of 0.8 mm as per the modified WIC test piece design proposed by Kurji et al. [9]. Weld fabricators in industry may qualify their welds using wider root gaps, which will result in deeper penetration and potential mitigation of a small portion of the risk of WMHACC. The results of the current study therefore only pertain to narrow root gaps, which is in itself the worst case scenario for the creation of a root bead stress concentration due to lack of penetration. Wider root gaps often resulted in burn-through and discontinuous welds in the initial experimental trials.

![60° to 70°](image)

**Figure 4.3.9** Allowable root gap for butt welds according to AS2885.2 [8].

REFERENCES


4.4 MODELING OF CRACK/NO CRACK BOUNDARIES USING LOGISTIC REGRESSION

4.4.1 Odds and probability of WMHACC

The crack/no crack boundaries shown in Figures 4.3.3, 4.3.5 and 4.3.6 and described mathematically by equations 4.3.3 to 4.3.5 separate uncrazed samples (to the right of the boundary) from cracked samples (to the left of the boundary). Each boundary was established by drawing a linear trend line midway between the maximum welding speed that resulted in a crack-free weld and the minimum welding speed that resulted in cracking at each current level. The slope of the trend line was then kept constant while the intercepts were changed until all the cracked samples were located to the left of the boundary line. Some experimental scatter, however, remains in the data near and to the left of the boundary line.

The scatter in the data is statistically significant because it indicates that the odds of finding evidence of WMHACC change as the boundary is approached. The odds of WMHACC occurring diminishes as the boundary is approached from the left and the odds of finding WMHACC-free samples increases as one moves away from the boundary to the right. With the degree of scatter observed, it is therefore reasonable to assume that the odds of observing WMHACC in samples situated on the original crack/no crack trend line (before changing the intercepts) approach 50/50. As one moves further away from the boundary (down and to the right), the odds of finding cracked samples decrease.

The aim of this investigation is to provide guidance to industry on the avoidance of WMHACC during field welding. Taking the experimental scatter in the results into account, it would be unreasonable to claim that no WMHACC would occur to the right of the crack/no crack boundary, even if no cracking had been observed experimentally. In light of the changing odds of cracking, it is more practical to model the probability of WMHACC as a function of welding parameters, rather than to claim a boundary that separates cracked and crack-free samples as a well-defined line.

The model used to predict the probability of cracking is based on the experimental results discussed in Section 4.3, which, in essence, form a series of experiments with a binary outcome (samples are either cracked or uncracked). Because the binary outcome of the experiment (cracked/uncracked) is dependent on the welding parameters, it is referred to as a dependent variable; and due to its nature, also called a binary variable or dichotomous variable, or simply the outcome variable. The most popular method of modelling binary dependent variables is statistical logistic regression. The aim of the model derived in this section is to predict the binary outcome of the experiment, but in order to accomplish this, it first predicts the odds of cracking as a function of the welding parameters. The model then uses the odds of cracking to predict the probability of WMHACC with a change in welding parameters [1].

A predicted probability of cracking of 0.5 (or 50%) is then used by the model as a cut-off probability, also known as a binary classifier, to classify a sample as either cracked or uncracked. In the model, if a sample has a predicted probability of cracking greater than the cut-off, the sample is assumed to be cracked. If it has a predicted probability of cracking less than the cut-off probability, it is assumed to be uncracked [1]. In this method, the statistically calculated crack/no crack boundary will be represented by a line of predicted probability of cracking of 50%.
In logistic regression, the cracked/uncracked outcome is called the binary or dichotomous dependent variable of the model, and the predicted probability of cracking is referred to as the continuous dependent variable (because it takes on continuous/real values). These variables are dependent on the model input variables, which in this case, are the welding parameters (welding speed, welding current and preheat temperature). These input variables are referred to as independent variables, and because they affect the outcome of the model directly, they are also called predictors or covariates [1].

In the model, the binary dependent variable is assigned one of two outcome values, “1” for cracked samples and “0” for uncracked samples. The odds of outcome condition “1” are determined throughout the model and related to the probability of outcome condition “1”, which is the predicted probability of cracking.

In general statistics, the odds of an event happening (cracking in this case) can be calculated by means of the ratio shown in Equation 4.4.1 [1].

\[
\text{Odds}_{\text{cracking}} = \frac{\text{Number of cracked samples}}{\text{Number of uncracked samples}}
\]

Eq. 4.4.1

If more samples are uncracked than cracked, the odds are smaller than 1. If more samples are cracked than uncracked, the odds are greater than 1. If an equal number of samples are cracked and uncracked, the odds of cracking are equal to 1. Therefore, odds smaller than 1 indicate that the likelihood of an event (cracking) happening is low or diminishing as a function of the predictors/independent variables (welding parameters), whereas odds greater than 1 indicate that the likelihood of that event (cracking) occurring is high or is increasing as a function of the predictors/independent variables. To calculate the probability, based on the odds, Equation 4.4.2 is generally used [1].

\[
\text{Probability}_{\text{cracking}} = \frac{\text{Odds}_{\text{cracking}}}{1 + \text{Odds}_{\text{cracking}}}
\]

Eq. 4.4.2

For an equal number of cracked and uncracked samples (and therefore odds of cracking of 1), Equation 4.4.2 yields a probability of cracking of \(1/(1+1) = \frac{1}{2}\), which is 0.5. This can be converted to a percentage by multiplying the probability by 100 - in this example a probability of cracking of 50%.

These equations are impractical to use in this investigation because Equations 4.4.1 and 4.4.2 require that each entry in the experimental matrix shown in Table 3.1.5 be repeated a number of times, and the odds and probability for each entry calculated based on the ratio of cracked to uncracked samples. The odds of cracking can, however, still be predicted by fitting the data to a standard statistical model using logistic regression. Therefore, whenever reference is made to the probability of cracking in this section, it refers to the predicted probability of cracking.

During logistic regression, the binary outcome of the MWIC tests, i.e. samples are either cracked or uncracked (as shown in Figures 4.3.3, 4.3.5 and 4.3.6), is fitted to a continuous linear model given by Equation 4.4.3, where the natural logarithm (ln) of the odds is given as a linear function of the predictors/independent variables. \(B_0\) is the intercept of the linear function on the ordinate axis and is equal to the natural log of the odds when all predictor variables are equal to 0, and \(X_0\) designates the individual predictor variables where \(X_1\) is the welding current and \(X_2\) is the welding speed. \(B_n\) represents constant values equal to the amount by which each predictor influences the value of the natural log of the odds, also referred to as constants of variation [1].

\[
\ln(\text{odds}) = B_0 + B_1X_1 + ... + B_nX_n
\]

Eq. 4.4.3
Equation 4.4.3 can then be rewritten in the form of Equation 4.4.4 to calculate the predicted odds of cracking as a function of the predictors, where $e$ is Euler’s number (base of the natural logarithm) with a value of approximately 2.71828. Equation 4.4.5 can be used to calculate the predicted probability based on the predicted odds [1].

$$\text{Odds}_{\text{cracking}} = e^{B_0 + B_1 X_1 + ... + B_n X_n} \quad \text{Eq. 4.4.4}$$

$$\text{Probability}_{\text{cracking}} = \frac{e^{B_0 + B_1 X_1 + ... + B_n X_n}}{1 + e^{B_0 + B_1 X_1 + ... + B_n X_n}} \quad \text{Eq. 4.4.5}$$

Even though the ln(odds) function is linear, the probability output from Equation 4.4.5 is a sigmoidal S-shaped function. Equation 4.4.3 is used because it is more practical to fit the data to a linear model, while Equation 4.4.5 serves as the function that converts the odds to a probability, and is called the logistic function. Because Equation 4.4.3 serves as a link between the fitted data and the non-linear sigmoidal probability function, Equation 4.4.3 is called a logit link function, and will be referred to as such in the remainder of this section.

The regression analyses, i.e. the values of the unknown constants $B_0$ and $B_1 ... B_n$ in the equations above as well as statistical data on how well the model fits the experimental data, were obtained through the use of the statistical software package IBM SPSS Statistics™, which is widely used for the purposes of statistical analysis and regression. The results described in this section are discussed in terms of the IBM SPSS Statistics™ output. The output is given, where possible, in the same format as given by the software.

For the sake of simplicity, WMHACC was modelled using logistic regression for each of the three preheat conditions, i.e. no preheat, and preheat temperatures of 50°C and 80°C with welding current and welding speed as the two predictor/independent variables. A comprehensive model was also developed that models WMHACC with all three predictor variables (welding speed, welding current and preheat temperature) combined into a single model.

### 4.4.2 Bivariate correlations

The IBM SPSS Statistics™ software package allows one to run a preliminary step, called a bivariate correlation, before the logistic regression is carried out. This is a test of the statistical significance of the effect of selected predictor variables on the dichotomous outcome of the model. The bivariate correlation analysis determines the strengths of the relationships between the predictors and the outcome and allows for the preliminary inclusion or exclusion of any predictor variables. The strength of the relationships (also called linkages) between each predictor variable and the outcome variable is given a value between -1 and +1. This value correlates the strength of the relationship between the variables and is called the correlation coefficient. It is also referred to as the Pearson’s product-moment correlation coefficient or simply the Pearson’s $r$-value or Pearson’s correlation coefficient.

If the Pearson correlation coefficient has a positive value, it indicates a positive relationship, i.e. a unit increase in the predictor variable would contribute positively to the outcome of the model, and vice versa. In the current investigation, the welding parameters serve as predictor variables and a positive correlation coefficient for any parameter suggests that a unit increase in that parameter favours/contributes to the occurrence of WMHACC - the bigger the correlation coefficient, the stronger the contribution. A negative correlation for a predictor variable indicates a negative relationship (or linkage), suggesting that a unit
increase in that parameter would reduce the likelihood of WMHACC. The more negative the correlation coefficient, the stronger the relationship. Since the occurrence of WMHACC is not linked only to the heat input, as demonstrated in the previous section, but rather to the individual welding parameters, a bivariate correlation analysis is extremely useful in isolating the welding parameters with the strongest effect on the occurrence of cracking. In the current study, a welding parameter is considered to have a strong relationship with the outcome of the model if it exhibits a Pearson’s correlation coefficient with an absolute value equal to or greater than 0.1.

If the Pearson’s correlation coefficient is equal to zero, it means that no relationship exists between the variables, i.e. the predictor variable has no influence on the outcome variable. The Pearson’s correlation coefficient is, however, based on a linear correlation between the variables. If a non-linear relationship exists between the variables, the Pearson’s correlation coefficient would also be calculated as zero.

Tables 4.4.1 to 4.4.3 show the results of bivariate correlation analysis for the three data sets corresponding to no preheat, a preheat of 50°C and a preheat of 80°C. Because the heat input (HI) is a function of both the travel speed and the welding current according to Equation 3.1.1, it was not included in the logistic regression analyses. It was, however, added to the bivariate correlation analyses for the sake of interest to compare its correlation with WMHACC to that of the welding current and travel speed.

Table 4.4.1 Correlations between predictor variables (welding parameters) and the outcome variable for samples welded without preheat.

<table>
<thead>
<tr>
<th></th>
<th>Current</th>
<th>Speed</th>
<th>HI</th>
<th>Cracking</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cracking Pearson Correlation</td>
<td>-0.532**</td>
<td>0.293</td>
<td>-0.455**</td>
<td>1</td>
</tr>
<tr>
<td>Sig. (2-tailed)</td>
<td>0.000</td>
<td>0.057</td>
<td>0.002</td>
<td></td>
</tr>
<tr>
<td>N</td>
<td>43</td>
<td>43</td>
<td>43</td>
<td>43</td>
</tr>
</tbody>
</table>

** Correlation is significant at the 0.01 level (2-tailed).

Table 4.4.2 Correlations between predictor variables (welding parameters) and the outcome variable for samples welded at 50°C preheat.

<table>
<thead>
<tr>
<th></th>
<th>Current</th>
<th>Speed</th>
<th>HI</th>
<th>Cracking</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cracking Pearson Correlation</td>
<td>-0.572**</td>
<td>0.216</td>
<td>-0.439**</td>
<td>1</td>
</tr>
<tr>
<td>Sig. (2-tailed)</td>
<td>0.000</td>
<td>0.181</td>
<td>0.005</td>
<td></td>
</tr>
<tr>
<td>N</td>
<td>40</td>
<td>40</td>
<td>40</td>
<td>40</td>
</tr>
</tbody>
</table>

** Correlation is significant at the 0.01 level (2-tailed).

In Tables 4.4.1 to 4.4.3, the 2-tailed significance is the result of statistical hypothesis testing, where the results from the current correlations are compared to a null hypothesis (in which no correlation exists between the predictors and the outcome). It is a means of testing the statistical significance of the Pearson’s correlation. The smaller (closer to zero) the sig. (2-tailed) value, the higher the statistical significance of the correlation. In other words, the smaller the sig. value, the more reliable the correlation coefficient. The value of N is the number of entries in the test matrix (or the number of samples in the batch tested).
Table 4.4.3 Correlations between predictor variables (welding parameters) and the outcome variable (WMHACC) for samples welded at 80°C preheat.

<table>
<thead>
<tr>
<th></th>
<th>Current</th>
<th>Speed</th>
<th>HI</th>
<th>Cracking</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cracking</td>
<td>-0.236</td>
<td>0.450**</td>
<td>-0.481**</td>
<td>1</td>
</tr>
<tr>
<td>Sig. (2-tailed)</td>
<td>0.143</td>
<td>0.004</td>
<td>0.002</td>
<td></td>
</tr>
<tr>
<td>N</td>
<td>40</td>
<td>40</td>
<td>40</td>
<td>40</td>
</tr>
</tbody>
</table>

** Correlation is significant at the 0.01 level (2-tailed).

In Tables 4.4.1 to 4.4.3, the value of 1 assigned to the column with the heading “Cracking” is simply the IBM SPSS Statistics™ software recognising that WMHACC has a perfect linear relationship with itself; i.e. for every weld showing evidence of WMHACC, the incidence of WMHACC increases by 1.

As expected, the welding current and heat input both have negative Pearson correlation coefficients, meaning that both parameters have a negative relationship with WMHACC. For every unit increase in their values, the likelihood of WMHACC diminishes. The negative correlation of welding current and heat input with WMHACC is particularly strong in Tables 4.4.1 and 4.4.2, with welding current showing the stronger negative correlation in each case. Welding speed has a positive correlation with WMHACC at all three preheat levels, but the effect of travel speed is particularly strong in Table 4.4.3.

Taking the absolute values of the ratios between the Pearson’s correlation coefficients tabulated above (|$r_{\text{current}/\text{speed}}$| or |$r_{\text{speed}/\text{current}}$|), reveals that increasing the welding current has a 1.82 times stronger effect on the occurrence of WMHACC than lowering the welding speed (|-0.532/0.293|) when welding without preheat. At a preheat of 50°C, increasing the welding current has a 2.65 times stronger effect than lowering the welding speed (|-0.572/0.216|), whereas at a preheat temperature of 80°C, an increasing welding current has an effect 0.52 times that of lowering the welding speed (|-0.236/0.45|). Or, stated differently, at a preheat temperature of 80°C, lowering the welding speed is about 1.9 times more effective at preventing WMHACC than increasing the welding current, (|0.45/-0.236|).

These correlation ratios show that it is safer to use welding current values at the higher end of the allowable range, and confirms the results shown in Tables 4.3.1 and 4.3.2, where an increase in welding current resulted in notable changes in the root bead geometry, which in turn affected the WMHACC behaviour. At a preheat of 80°C, which facilitated the removal of most of the hydrogen from the weld metal due to the slower cooling rate and ensured deeper overall penetration, the situation changes somewhat and the welding speed has a stronger correlation.

These results correlate well with the trends shown in Figures 4.3.1, 4.3.3, 4.3.5 and 4.3.6. Higher heat inputs and preheat temperatures appear to suppress WMHACC, confirming the negative correlation between WMHACC and heat input. Higher welding currents reduce the risk of WMHACC at all preheat temperatures (negative correlation), whereas higher travel speeds increase the risk of cracking (positive correlation).

As show in Tables 4.3.1 and 4.3.2, welding current shows a stronger correlation with WMHACC than travel speed in welds performed without preheat or at a preheat of 50°C. An increase in welding current influences WMHACC in two ways. An increase in the welding current causes an increase in the heat input, resulting in higher electrode burn-off rates and slower cooling rates. This allows more hydrogen to diffuse out of the weld metal on cooling. With an increase in welding current there is also an associated increase in the arc force, favouring weld pool depression and deeper penetration (even at constant heat input values). This alters
the root bead geometry by decreasing the severity of the stress concentration caused by incomplete penetration at the root of the welds.

Welding speed only appears to affect the heat input, with faster welding speeds resulting in lower heat input levels which, in turn, result in lower electrode burn-off rates and faster cooling rates. A faster cooling rate causes more hydrogen to be trapped within the weld metal, promoting crack initiation at wagon tracks and other stress concentrations.

4.4.3 Logistic regression of MWIC test data for samples welded without preheat

After establishing the strength of each predictor by means of bivariate correlation, logistic regression was performed. During this regression, the logit link function (Equation 4.4.3) was fitted to the three MWIC data sets for cracked/uncracked samples. Equation 4.4.3 can be rewritten in the form of Equation 4.4.6 to represent the predictor variables relevant to this investigation.

\[
\ln(\text{odds of cracking}) = B_0 + B_1(\text{Current}) + B_2(\text{Speed})
\]

Eq. 4.4.6

This equation was fitted separately for each of the three preheat conditions in this study (no preheat, and preheat temperatures of 50°C and 80°C). IBM SPSS Statistics™ renders, as outputs to the data fitting, values for the intercept constant \(B_0\), the constants of variation \(B_1\) and \(B_2\), and statistical information pertaining to the goodness-of-fit of the model compared to the original data.

The first step is to set up a null hypothesis model in which the constants of variation \(B_0\) are equal to zero so that no indicator variables are included in the model. This is done in order to establish a level of statistical significance for the fitted model.

Table 4.4.4 shows the classification of samples into cracked (1) and uncracked (0) according to the null hypothesis model. Samples with a predicted probability of cracking less than 0.5 (50%) are classified as uncracked, and those with probabilities greater than 0.5 are classified as cracked.

<table>
<thead>
<tr>
<th>Observed</th>
<th>Predicted Cracking</th>
<th>Percentage Correct</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>0.00</td>
<td>1.00</td>
</tr>
<tr>
<td>No Cracking</td>
<td>0</td>
<td>21</td>
</tr>
<tr>
<td>Cracking</td>
<td>1.00</td>
<td>0</td>
</tr>
<tr>
<td>Overall Percentage</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Table 4.4.4 Classification of the null hypothesis model for samples welded without preheat.

After the MWIC tests, 21 of the samples welded without preheat showed no evidence of cracking (0). The null hypothesis model predicts that all 21 samples will crack (1). This gives a percentage correct predictions of 0.0%. On the other hand, 22 samples cracked (1) during MWIC testing, and the null model predicts that all 22 samples will crack (1). This gives a percentage correct prediction of 100%. Therefore, with the null hypothesis model predicting that all 43 samples should crack, it has a total accuracy (overall percentage correct) of 51.2%. Thus, the null model will be correct 51.2% of the time if used for predicting WMHACC.

The second step in the logistic regression analysis is to fit the logit link function to the experimental data. The output of this step is presented in Table 4.4.5. It shows the values of the intercept constant \(B_0\) and the
constants of variation \((B_1 \text{ and } B_2)\) of the fitted logit link function for samples welded without preheat along with data relating to their statistical significance.

**Table 4.4.5** Variables in the logit link function for samples welded without preheat, where S.E. represents the standard error and Sig. the statistical significance.

<table>
<thead>
<tr>
<th></th>
<th>(B)</th>
<th>S.E.</th>
<th>Sig.</th>
<th>(\exp (B))</th>
<th>95% confidence interval for (\exp (B))</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>Lower</td>
</tr>
<tr>
<td>Step 1(^a)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Current ((B_1))</td>
<td>-0.242</td>
<td>0.076</td>
<td>0.001</td>
<td>0.785</td>
<td>0.676</td>
</tr>
<tr>
<td>Speed ((B_2))</td>
<td>0.017</td>
<td>0.006</td>
<td>0.003</td>
<td>1.017</td>
<td>1.006</td>
</tr>
<tr>
<td>Constant ((B_0))</td>
<td>27.174</td>
<td>8.710</td>
<td>0.002</td>
<td>6.33 x 10(^{12})</td>
<td></td>
</tr>
</tbody>
</table>

\(^a\) Variable(s) entered on step 1: Current, speed.

Using Table 4.4.5, the logit link function for the natural log of the odds of cracking at room temperature can be written in the form of Equation 4.4.7, with current in A and speed in mm/min.

\[
\ln(\text{odds}_{\text{cracking}}) = 27.174 - 0.242(\text{Current}) + 0.017(\text{Speed})
\]  

Eq. 4.4.7

Based on Equation 4.4.4, Equation 4.4.7 can now be rewritten in the form of Equation 4.4.8 to give the odds of WMHACC as a function of the welding parameters. From Equation 4.4.5, the odds of WMHACC can be written as Equation 4.4.9 to give the logistic regression model for the predicted probability of WMHACC as a function of the welding parameters. The percentage probability of WMHACC can be determined by multiplying the output of Equation 4.4.9 by 100.

\[
\text{Odds}_{\text{cracking}} = e^{27.174 - 0.242(\text{Current}) + 0.017(\text{Speed})}
\]  

Eq. 4.4.8

\[
\text{Probability}_{\text{cracking}} = \frac{e^{27.174 - 0.242(\text{Current}) + 0.017(\text{Speed})}}{1 + e^{27.174 - 0.242(\text{Current}) + 0.017(\text{Speed})}}
\]  

Eq. 4.4.9

The probability of WMHACC occurring in the root pass of a pipeline girth weld performed without preheat with E6010 consumables can now be calculated using Equation 4.4.9 for API 5L X70 pipelines with wall thicknesses up to 10 mm.

The S.E. value in Table 4.4.5 is the standard error of the estimates and is an average of the fitting error, or rather a function of the average of the distances on the regression graph between the experimental data and the fitted regression line represented by Equation 4.4.7. As described earlier, the Sig. values represent the statistical significance, with values less than 0.05 indicating a high level of significance for each indicator variable. The significance values can be used to justify rejection of the null hypothesis model, and are dependent on the number of samples used in the investigation (a total of 43 tests in the current investigation). The more samples, the more statistically significant the result of the regression model. It is also dependent on the percentage accuracy of both the null hypothesis model and the current model. The greater the difference in the percentage accuracy between the current model and the null hypothesis model, the more statistically significant the values of the \(B\) constants (the smaller the values of Sig.) and the more strongly the null hypothesis model is rejected. The \(\exp (B)\) value is also called the odds ratio and represents the magnitude of change in the odds brought about by a unit increase in the predictor variable.

Simply stated, the \(B\) values in Table 4.4.5 track changes in the \(\ln(\text{odds})\) of WMHACC, and the \(\exp (B)\) values track changes in the odds of cracking. A negative \(B\) value will result in an \(\exp (B)\) value less than 1,
which is indicative of diminishing odds of WMHACC as the predictor variable increases. A positive \( B \) value will result in an exp \((B)\) value greater than 1, which is indicative of increasing odds of WMHACC with an increase in the predictor variable. The more the value of exp \((B)\) diverges from 1, the greater the effect on the odds of WMHACC. Since the divergence of the exp \((B)\) from a value of 1 is greater for current than for speed, the results in Table 4.4.5 agree well with the bivariate correlations of Table 4.4.1, which suggested that welding current has a stronger effect on the occurrence of WMHACC than welding speed. The last two columns in Table 4.4.5 give the 95% confidence intervals for the calculated exp \((B)\) values, indicating that there is a 95% certainty that the predicted value for exp \((B)\) in each case falls within the range defined by the lower and upper bounds in the table.

### 4.4.4 Tests for the goodness-of-fit between predicted and observed data for samples welded without preheat

Table 4.4.6 compares the number of predicted incidences of cracking to the number of observed incidences of cracking for the logistic regression model described by Equation 4.4.9. In Table 4.4.6, the binary classifier (cut-off probability) is a predicted probability of 0.5 (50%); i.e. samples with a probability of cracking greater than 0.5 is classified as cracked (1) and samples with a probability less than 0.5 is classified as uncracked (0).

<table>
<thead>
<tr>
<th>Observed</th>
<th>Predicted Cracking</th>
<th>Percentage Correct</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>0.00</td>
<td>1.00</td>
</tr>
<tr>
<td>Step 1</td>
<td>No Cracking 21</td>
<td>17</td>
</tr>
<tr>
<td></td>
<td>Cracking 22</td>
<td>2</td>
</tr>
<tr>
<td>Overall Percentage</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

*a The cut value is 0.50.

For the 21 crack-free samples observed experimentally, the logistic regression model predicts 17 uncracked samples, giving a percentage correct predictions of 81%. For the 22 samples observed to have cracked during testing, the model predicts 20 cracked samples, resulting in a percentage correct predictions of 90.9%. This yields a total overall accuracy of 86%.

Figures 4.4.1 and 4.4.2 display a comparison between the observed and predicted incidences of WMHACC. Figure 4.4.1 shows the observed results of MWIC tests performed without preheat and is similar to Figure 4.3.3. Figure 4.4.2 shows the predicted occurrence of WMHACC determined by means of Equation 4.4.9 with a probability of 0.5 as the binary classifier (or cut-off probability).

Since Figure 4.4.2 is based on the probability of WMHACC, the fundamental implication is that if the welding trials were repeated an infinite number of times, a graph representing the average of the incidences of WMHACC would resemble Figure 4.4.2, rather than Figure 4.4.1.
There is a good visual correlation between the observed and predicted WMHACC behaviour shown in Figures 4.4.1 and 4.4.2, although further testing of the statistical significance of the fit between the predicted and experimentally observed data is required (shown subsequently). It is evident from Figures 4.4.1 and 4.4.2 that the moderate accuracy of the model (86%) stems largely from the general amount of noise in the data. The predictive model cancels out the noise, separating the cracked and uncracked parts of the graph along a boundary line representing a 50% probability of cracking (since a probability of 0.5 serves as the binary classifier cut-off value).

As shown in Tables 4.4.7 to 4.4.9, it is more practical to generate single values from various statistical tests that capture the goodness-of-fit of logistic regression models. This also simplifies the estimation of the predictive power of the model. It is good practice to quote the goodness-of-fit data with every model created using logistic regression. IBM SPSS Statistics™ uses two separate approaches to test for the goodness-of-fit and therefore gives two sets of statistical data for interpretation. The first is the pseudo R-square values given in Table 4.4.7.

<table>
<thead>
<tr>
<th>Table 4.4.7 Pseudo R² values for samples welded without preheat.</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Cox &amp; Snell R Square</strong></td>
</tr>
<tr>
<td>0.539</td>
</tr>
</tbody>
</table>
The R^2 values given in Table 4.4.7 are called pseudo R^2 values as they differ from the R^2 values obtained from least square regression (where an R^2 value is obtained from the goodness-of-fit of a trend line through plotted data). The R^2 values in Table 4.4.7 show the goodness-of-fit between predicted dichotomous outcomes and observed dichotomous outcomes. In other words, it only compares predicted cracking (1) to observed cracking and samples predicted not to have cracked (0) to observed crack-free samples. The Cox & Snell R^2 has a maximum upper limit of 0.75 and because it does not equate to an upper bound value of 1, is hard to interpret for non-statisticians. The Nagelkerke R^2, however, has an upper limit of 1 for a perfect model [1], and the proposed predictive model can therefore be said to have an R^2 value of 0.719.

The second data set given by IBM SPSS Statistics™ are results pertaining to the Hosmer and Lemeshow test for goodness-of-fit. It is a test used extensively for logistic regression, especially in risk-prediction models. The Hosmer and Lemeshow test can therefore be used to calculate how well the predicted incidences of WMHACC correlate to the observed cases of cracking. In the Hosmer and Lemeshow test, the dichotomous outcomes of the prediction model are grouped into ten separate groups, based on similar predicted probabilities. Therefore, all samples falling into the same group have roughly the same probability of cracking. The number of samples expected (predicted by the model) to have cracked (1) and not cracked (0) in each group is then compared to the number of samples observed to have cracked (1) and not cracked (0) for that same group, as shown in the contingency table for the Hosmer and Lemeshow test in Table 4.4.8.

<table>
<thead>
<tr>
<th>Groups</th>
<th>Cracking = 0.00</th>
<th>Cracking = 1.00</th>
<th>Total</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Observed</td>
<td>Expected</td>
<td>Observed</td>
</tr>
<tr>
<td>1</td>
<td>1</td>
<td>4</td>
<td>3.988</td>
</tr>
<tr>
<td>2</td>
<td>2</td>
<td>4</td>
<td>3.942</td>
</tr>
<tr>
<td>3</td>
<td>3</td>
<td>4</td>
<td>3.798</td>
</tr>
<tr>
<td>4</td>
<td>4</td>
<td>4</td>
<td>3.388</td>
</tr>
<tr>
<td>5</td>
<td>5</td>
<td>1</td>
<td>2.496</td>
</tr>
<tr>
<td>6</td>
<td>6</td>
<td>2</td>
<td>1.560</td>
</tr>
<tr>
<td>7</td>
<td>7</td>
<td>1</td>
<td>0.978</td>
</tr>
<tr>
<td>8</td>
<td>8</td>
<td>0</td>
<td>0.505</td>
</tr>
<tr>
<td>9</td>
<td>9</td>
<td>1</td>
<td>0.274</td>
</tr>
<tr>
<td>10</td>
<td>10</td>
<td>0</td>
<td>0.071</td>
</tr>
</tbody>
</table>

From the data shown in Table 4.4.8, a chi-square (χ^2) value is then calculated, which correlates the difference between the expected and observed data into a single value. The general statistical expression for the χ^2 value is given by Equation 4.4.10 (Hosmer et al 2013).

\[
\chi^2 = \sum_{i=1}^{n} \frac{(\text{observed}_i - \text{expected}_i)^2}{\text{expected}_i} \quad \text{Eq. 4.4.10}
\]

As can be seen from Equation 4.4.10, the smaller the χ^2 value, the better the fit between the model’s predicted outcomes and the experimentally observed outcomes. As a stand-alone value, the χ^2 number is, however, only relevant in comparison with other models. Hosmer and Lemeshow altered the equation for χ^2 slightly to that shown in Equation 4.4.11, where n_i is the total number of observations (cracked and uncracked) in the i^{th} group [1].

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\[ x^2 = \sum_{i=1}^{10} \frac{(\text{observed}_i - \text{expected}_i)^2}{\text{expected}_i \times (1 - \text{expected}_i/n_i)} \]  
Eq. 4.4.11

Equation 4.4.11 is then used to calculate the \( x^2 \) value given in Table 4.4.9. On its own it does not give any deeper insights into the goodness-of-fit, but is calculated here as a means of determining the \( p \)-value, or the statistical relevance of the model. The \( p \)-value is estimated through hypothesis testing and the \( x^2 \) value is an input into the calculation. The df (degrees of freedom) value listed in the table is equal to the number of groups in the Hosmer and Lemeshow test (10) minus the number of predictors (independent variables) from the logistic regression (welding current and travel speed = 2). Therefore, a Hosmer and Lemeshow \( x^2 \) value of 6.307 with eight degrees of freedom gives a \( p \)-value of 0.613.

**Table 4.4.9** Results from the Hosmer and Lemeshow test for goodness-of-fit of the predictive model for samples welded without preheat.

<table>
<thead>
<tr>
<th>Chi-square</th>
<th>df</th>
<th>Sig. (( p )-value)</th>
</tr>
</thead>
<tbody>
<tr>
<td>6.307</td>
<td>8</td>
<td>0.613</td>
</tr>
</tbody>
</table>

The \( p \)-value shows the statistical significance of the model, and in the Hosmer and Lemeshow test, a \( p \)-value larger than 0.05 indicates a statistically significant model with good predictive capability. A \( p \)-value greater than 0.05 is also indicative of good repeatability of the results. The \( p \)-value of the current predictive model is 0.613, well above the minimum of 0.05 and, therefore it can be assumed that the results are repeatable and that the model shows an acceptable degree of fit with the observed experimental data.

### 4.4.5 Guidance on parameter selection based on acceptable probabilities of cracking for samples welded without preheat

As shown in Figure 4.4.2, the logistic regression model reduces the noise in the experimental test results as the probability of cracking approaches 50%. A well-defined boundary, representing a predicted probability of cracking of 50%, therefore demarcates the areas of the graph where the probability of cracking is in favour of WMHACC on one side and against WMHACC on the other. The probability of cracking then increases as one moves away from the boundary on the side where WMHACC is favoured, and *vice versa*. This boundary can be determined using the following method.

Figure 4.4.3 shows a typical sigmoidal S-shaped graph of crack probability (described in Section 4.4.1), in this case for a welding current of 150 A and no preheat. This sigmoidal graph of predicted probability as a function of the welding speed and was determined by applying Equation 4.4.9, in which the welding current was kept constant while the welding speed was varied. This was repeated for each welding current used in this study. The welding speeds that resulted in predicted probabilities of cracking of 90%, 50% and 10% were then determined for each welding current used. These data points are shown in Figure 4.4.4 as a function of welding current and travel speed.

As shown in Figure 4.4.4, trend lines representing probabilities of 90%, 50% and 10% for WMHACC formed straight lines. These lines of probability are represented mathematically by Equations 4.4.12 to 4.4.14 where \( y \) is the welding speed (mm/min) and \( x \) is the welding current (A). These equations are represented graphically in Figure 4.4.5.
Figure 4.4.3 Sigmoidal S-shaped graph of the predicted probability of cracking as a function of welding speed at a constant welding current (150 A).

Figure 4.4.4 Predicted probabilities of WMHACC of 90%, 50% and 10% as a function of welding parameters for MWIC samples welded without preheat.

\[ y_{90\%} = 13.889x - 1418 \]  
Eq. 4.4.12

\[ y_{50\%} = 14.432x - 1615.3 \]  
Eq. 4.4.13

\[ y_{10\%} = 14.475x - 1754 \]  
Eq. 4.4.14

Using Equations 4.4.12 to 4.4.14, the maximum welding speed \( y \) yielding probabilities of WMHACC of 90%, 50% or 10% can be determined at any given welding current. These equations can therefore be used by pipeline manufacturers during the development and qualification of welding procedure specifications to determine the upper limits on welding speeds once a welding current range has been selected.

Equation 4.4.9 can also be used by pipeline manufacturers to determine whether certain pre-selected welding parameters will result in an acceptable probability of cracking. The acceptable risk (probability) of cracking will have to be determined prior to parameter selection. Alternatively, Figure 4.4.5 can be used as a graphical aid to estimate the probability of failure associated with specific combinations of welding parameters. In the interest of thoroughness, Figure 4.4.5 can eventually be populated with probability lines between 10%, 50% and 90%.
Figure 4.4.5 Lines of predicted probability (90%, 50% and 10%) of cracking as a function of welding parameters for MWIC samples welded without preheat.

Figure 4.4.6 shows the lines of predicted probability superimposed on the original crack/no crack data. This figure stresses the importance of working with acceptable probabilities of cracking well below than 50%, or as far from the crack/no crack boundary as is reasonably possible. At 130 A two cracked samples are situated to the right and below the 50% probability line, and one cracked sample is situated on the line. Although all three samples displayed predicted probabilities of cracking of 50% and less (therefore a lower risk of cracking), all three still cracked. It is therefore important that each pipeline manufacturer decides on a level of acceptable risk during field welding. Figure 4.4.5 and Equation 4.4.9 can then be used to determine welding parameter combinations that yield acceptable probabilities of WMHACC during preheat-free root pass welding.

Figure 4.4.6 Lines of predicted probability (90%, 50% and 10%) of cracking superimposed on the original MWIC data set for samples welded without preheat.
It is also evident from Figure 4.4.6 that the scatter in the crack/no crack data is largely contained within the 10% and 90% probability lines. For this reason, the 10% and 90% probability lines are used to demarcate safe and unsafe regions. The risk of WMHACC is considered remote, although not zero, to the right of and below the 10% probability line, and high to the left and above the 90% probability line. Such lines of probability can be regarded as more realistic and practical crack/no crack boundaries than a single line separating cracked and uncracked samples.

4.4.6 Guidance on parameter selection based on acceptable probabilities of cracking for samples welded at a preheat of 50°C

Figures 4.4.7 and 4.4.8 provide a graphical comparison between the observed and predicted occurrence of cracking for the data set consisting of MWIC samples welded at a preheat of 50°C. For the sake of brevity, the goodness-of-fit, statistical significance and accuracy of the model are described in Appendix B. The overall accuracy of the model’s predictions compared to observed data was found to be 92.5%, which is higher than the accuracy of the model developed for preheat-free welding. The improved accuracy can, to an extent, be ascribed to the lower level of scatter in the original data for samples welded at a preheat temperature of 50°C.

![Figure 4.4.7](image1.png)

**Figure 4.4.7** Experimentally observed incidence of WMHACC in samples welded at a preheat temperature of 50°C.

![Figure 4.4.8](image2.png)

**Figure 4.4.8** Predicted incidence of WMHACC for samples welded at a preheat temperature of 50°C.
Using the method described in Section 4.4.5, the predicted welding speeds that result in 10%, 50% and 90% probabilities of WMHACC were determined for each welding current, as shown graphically in Figure 4.4.9. The equations for the trend lines shown in Figure 4.4.9 are given by Equations 4.4.15, 4.4.16, and 4.4.17 for 10%, 50% and 90% probability of cracking, respectively, where $y$ is the welding speed (mm/min) and $x$ is the welding current (A).

![Graph of predicted probabilities](image)

**Figure 4.4.9** Predicted probabilities of WMHACC of 90%, 50% and 10% as a function of the welding parameters for MWIC samples welded at a preheat temperature of 50°C.

\[
y_{90\%} = 17.117x - 1741.9 \quad \text{Eq. 4.4.15}
\]

\[
y_{50\%} = 17.5x - 1874.1 \quad \text{Eq. 4.4.16}
\]

\[
y_{10\%} = 17.371x - 1932.1 \quad \text{Eq. 4.4.17}
\]

Equations 4.4.15 to 4.4.17 are shown graphically in Figure 4.4.10, and relative to the measured MWIC test data in Figure 4.4.11. The lines of probability appear to be closer together than those observed for the preheat-free model in Figure 4.4.5. This can be attributed to the smaller area of uncertainty between the 10% and 90% probability lines where most of the experimental scatter is found. All scatter was again contained within the 10% and 90% probability lines, demarcating the area of high uncertainty. To avoid WMHACC, welding parameters should preferably be selected to fall to the left and below the 10% probability line in the area that is considered safe. The closer to the bottom right hand corner of the graph the welding parameters are located, the lower the probability of cracking during root bead welding. However, this is merely a recommendation and it is left to the discretion of the pipeline manufacturers to decide on an acceptable level of risk for each project.

**4.4.7 Guidance on parameter selection based on acceptable probabilities of cracking for samples welded at a preheat of 80°C**

A similar method was used to determine the probability of failure as a function of welding parameters for MWIC test samples welded at a preheat temperature of 80°C. As shown in Appendix B, the prediction accuracy for the current model is 95%. The high level of predictive accuracy is evident in Figures 4.4.12 and
4.4.13, which compare the experimentally observed occurrences of WMHACC at a preheat temperature of 80°C to the predicted incidences of cracking.

![Image of graph](image1)

**Figure 4.4.10** Lines of predicted probability (90%, 50% and 10%) as a function of welding parameters for MWIC samples welded at a preheat temperature of 50°C.

![Image of graph](image2)

**Figure 4.4.11** Lines of predicted probability (90%, 50% and 10%) superimposed on the original data set of MWIC samples welded at a preheat temperature of 50°C.

The 10%, 50% and 90% probability trend lines for cracking are shown in Figures 4.4.14 and 4.4.15, and represented mathematically by Equations 4.4.18, 4.4.19, and 4.4.20. Again, the lines appear to be more closely spaced than the lines for the preheat-free model since less scatter was observed between the 10% and 90% probability lines in this data set.

\[
y_{90\%} = 11.316x - 851.43 \quad \text{Eq. 4.4.18}
\]

\[
y_{50\%} = 11.456x - 938.19 \quad \text{Eq. 4.4.19}
\]

\[
y_{10\%} = 11.375x - 1007.1 \quad \text{Eq. 4.4.20}
\]
Figure 4.4.12 Experimentally observed incidence of WMHACC in samples welded at a preheat temperature of 80°C.

Figure 4.4.13 Predicted incidences of WMHACC for samples welded at a preheat temperature of 80°C.

Figure 4.4.14 Predicted probabilities of WMHACC of 90%, 50% and 10% as a function of the welding parameters for MWIC samples welded at a preheat temperature of 80°C.
Figure 4.4.15 Lines of predicted probability (90%, 50% and 10%) as a function of welding parameters for MWIC samples welded at a preheat temperature of 80°C.

As discussed in sections 2.8 and 3.1, MWIC testing using a restraint length of 25 mm and a preheat temperature of 80°C has been shown to be more representative of the conditions during root bead welding in the field and therefore a preheat to 80°C was used in this investigation. An intermediate preheat temperature of 50°C was also used in the current investigation in order establish the change in WMHACC behaviour with regards to temperature.

4.4.8 Probability distributions for samples welded preheat-free and with preheats of 50°C and 80°C

Figures 4.4.16 and 4.4.17 are two-dimensional and three-dimensional plots comparing the predicted probability distributions for WMHACC for welds deposited without preheat and at preheat temperatures of 50°C and 80°C, where the shift from safe welding parameters combinations to unsafe combinations is highlighted by a change from dark blue to dark red. Higher welding currents and lower welding speeds favour crack-free welding at all preheat levels.

4.4.9 A comprehensive predictive model

To simplify the prediction of probabilities of cracking, it is possible to combine all of the main welding parameters pertaining to the welding of root beads on pipelines into a single and comprehensive model. Instead of having three models developed at different preheat temperatures, the preheat temperature can be included in the logistic regression analysis to yield a single equation with welding current, travel speed and preheat temperature as independent predictor variables. Combining all three predictor variables into one model may reduce the model accuracy or predictive power compared to the individual models, but at the same time the higher number of observations on which the model is based results in improved statistical significance.

Table 4.4.10 shows the results of a bivariate correlation analysis of all three data sets combined (a total of 123 samples (N) including 43 samples in the preheat-free data set and 40 samples each in the 50°C and 80°C preheat data sets. Table 4.4.10 again highlights the importance of welding current, with an increase in welding current having the most pronounced effect on preventing WMHACC. The heat input is included
here only for the sake of interest since it is a combination of current and speed. As expected, the preheat temperature also has a relatively strong effect.

Figure 4.4.16 Two-dimensional plot of predicted probability of WMHACC as a function of the weld speed and welding current for welds deposited (a) without preheat, and at preheat temperatures of (b) 50°C, and (c) 80°C.
Figure 4.4.17 Three-dimensional plot of predicted probability of WMHACC as a function of the weld speed and welding current for welds deposited (a) without preheat, and at preheat temperatures of (b) 50°C; and (c) 80°C.

Table 4.4.10 Correlations between predictor variables (welding parameters) and the outcome variable (WMHACC) for the comprehensive model.

<table>
<thead>
<tr>
<th></th>
<th>Current</th>
<th>Speed</th>
<th>HI</th>
<th>Cracking</th>
<th>Preheat</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cracking Pearson Correlation</td>
<td>-0.437**</td>
<td>0.281**</td>
<td>-0.421**</td>
<td>1</td>
<td>-0.369**</td>
</tr>
<tr>
<td>Sig. (2-tailed)</td>
<td>0.000</td>
<td>0.002</td>
<td>0.000</td>
<td>0.000</td>
<td>0.000</td>
</tr>
<tr>
<td>N</td>
<td>123</td>
<td>123</td>
<td>123</td>
<td>123</td>
<td>123</td>
</tr>
</tbody>
</table>

** Correlation is significant at the 0.01 level (2-tailed).

Taking the ratios of the absolute values of the effects of the independent variables listed in Table 4.4.10, it is evident that increasing welding current has a 1.6 times (|-0.437/0.281|) stronger influence on preventing WMHACC than a reduction in welding speed. This is due to the dual effect of increasing heat input with increasing welding current, causing a slower cooling rate and more hydrogen effusion from the weld metal, as well as an increase in arc force, which leads to a reduction in the severity of the stress concentration due to incomplete penetration at the root of the weld.

As discussed in Section 4.6, preheating prior to welding has a much stronger effect on reducing the cooling rates after welding than does increasing the welding current. Nonetheless, the welding current still has a 1.2 times stronger effect in the bivariate correlation (|-0.437/-0.369|) than preheating, due to the dual effect discussed above. This could be of interest to pipeline construction companies who wish to prevent WMHACC but is constrained by the economics of pipeline construction not to apply preheating. The risk of WMHACC could be mitigated to an extent by increasing the welding current, with an associated reduction in welding speed to maintain the desired heat input, if the heat input is affected significantly by such an adjustment.

More detailed information on the statistical significance, accuracy and goodness-of-fit of the comprehensive model is given in Appendix B. The model has a prediction accuracy of 91.1% and much smaller standard errors associated with each variable than observed on average for the individual models, leading to narrower bounds on the 95% confidence intervals for the predicted exp (B) values.
The logit link function for the comprehensive model is represented by Equation 4.4.21, whereas the predicted odds and probability of WMHACC are given by Equations 4.4.22 and 4.4.23, respectively, with welding current in A and travel speed in mm/min. Preheat temperature (in °C) is included in the equations as a third variable.

\[
\ln(\text{odds}_{\text{cracking}}) = 37.781 - 0.311(\text{Current}) + 0.021(\text{Speed}) - 0.111(\text{Preheat}) \quad \text{Eq. 4.4.21}
\]

\[
\text{Odds}_{\text{cracking}} = e^{37.781 - 0.311(\text{Current}) + 0.021(\text{Speed}) - 0.111(\text{Preheat})} \quad \text{Eq. 4.4.22}
\]

\[
\text{Probability}_{\text{cracking}} = \frac{e^{37.781 - 0.311(\text{Current}) + 0.021(\text{Speed}) - 0.111(\text{Preheat})}}{1 + e^{37.781 - 0.311(\text{Current}) + 0.021(\text{Speed}) - 0.111(\text{Preheat})}} \quad \text{Eq. 4.4.23}
\]

This comprehensive model for predicting the probability of cracking is a more practical tool for welding coordination personnel and fabricators qualifying welding procedure specifications. If one considers that the preheat temperature is simply equivalent to the temperature of the pipe at the start of welding, the term “Preheat” in Equation 4.4.23 can be used interchangeably with “Temperature”, as shown in Equation 4.4.24. The term “Temperature” then indicates the ambient temperature or the temperature to which the pipe is raised, in the absence or presence of any deliberate preheating. Welding in the outback in summer, where the ambient daytime temperature may be considerable higher than that experienced at the coast, may widen the safe operating window and allow the use of faster welding speeds without increasing the probability of cracking.

\[
\text{Probability}_{\text{cracking}} = \frac{e^{37.781 - 0.311(\text{Current}) + 0.021(\text{Speed}) - 0.111(\text{Temperature})}}{1 + e^{37.781 - 0.311(\text{Current}) + 0.021(\text{Speed}) - 0.111(\text{Temperature})}} \quad \text{Eq. 4.4.24}
\]

If, for example, a welding procedure specification specifies a welding current of 150 A and a travel speed of 400 mm/min, the heat input is approximately 0.6 kJ/mm (assuming the arc voltage at this current is around 27 V). If, at the time of welding, the ambient temperature is 42°C (and assuming that the surface temperature of the pipe was measured to be the same), Equation 4.4.24 predicts a probability of WMHACC of 0.006 (0.6%). If however, the welding parameters remain exactly the same but the ambient temperature drops to 10°C, Equation 4.4.24 predicts a significantly higher WMHACC probability of 17.1%. If this probability exceeds the maximum allowable probability of cracking determined by the pipeline construction company or end user, adjustments will have to be made to the welding parameters to bring the probability of WMHACC down to an acceptable limit. Since welding current has the most pronounced effect on minimising the probability of WMHACC, an increase in welding current to 160 A would increase the heat input marginally to 0.65 kJ/mm and reduce the probability of WMHACC to 0.009 (0.9%) at 42°C. If the heat input is, however, required to remain constant at 0.6 kJ/mm, an increase in welding current to 160 A would require a reduction in travel speed to 432 mm/min, resulting in a probability of cracking of 0.018 (1.8%).

REFERENCES

4.5 SOLIDIFICATION CRACKING

Solidification cracking in the root pass of pipeline girth welds may increase the risk of weld metal hydrogen-assisted cracking by introducing a sharp stress concentration within the weld metal. Solidification cracks have been shown to initiate at elevated temperatures and propagate as hydrogen-assisted cold cracks at lower temperatures. An example of this was shown in Section 4.2 of this study.

Solidification cracking is caused by segregation of impurity elements such as sulphur and phosphorus to the liquid phase ahead of the advancing solidification interface on cooling. These elements combine with iron during the final stages of solidification to form eutectic phases with low melting temperatures. Thin films of liquid persist between crystals and at the weld centreline where the advancing columnar crystals meet, and when the weld metal contracts during cooling, the films rupture unless backfilled by liquid. These fissures may develop into cracks that readily propagate under the influence of solidification stresses [1,2].

The carbon content of the steel strongly affects the segregation of impurity elements. The δ-ferrite phase has a higher solubility for elements such as sulphur and phosphorus than γ-austenite [1,2] and segregation to the liquid is therefore less pronounced when solidification is ferritic. From the Fe-C phase diagram, solidification is predicted to be predominantly ferritic at carbon contents less than around 0.08%.

Nolan et al. [2] reported that in the root bead of API 5L X70 welded with E6010 electrodes, significant changes in cracking behaviour were not observed within the composition range of 0.013 to 0.085% C, 0.004 to 0.028% S, and 0.016 to 0.033% P. Since the carbon content of the weld metal evaluated in the current investigation was 0.14%, and therefore outside the range considered low risk by Nolan et al. [2], solidification is expected to be at least partially austenitic. A sulphur content of 0.01% promotes segregation and the formation of low melting eutectic phases, causing the centreline cracking behaviour summarised in Figure 4.5.1 for preheat-free welds.

From Figure 4.5.1, it is evident that solidification cracking in this study was not exclusively governed by either the travel speed or the welding current used. Solidification cracking was observed in various samples throughout the entire range of travel speeds (200 to 724 mm/min) and welding currents (120 A to 160 A) evaluated. It was not possible to establish a clear boundary between areas of the graph that contained cracked samples and areas that were crack-free. Therefore, based on Figure 4.5.1, no clear guidance can be given on the avoidance of solidification cracking.

In Figure 4.5.2, the crack/no crack boundary representing a 50% probability of WMHACC for pre-heat free welds (Figure 4.4.4) is superimposed on Figure 4.5.1. It is evident that the area where solidification cracking occurs at low and intermediate travel speeds overlaps with the area on the graph where WMHACC was not observed. There appears to be no direct visual correlation between the incidence of WMHACC and that of solidification cracking, although in Section 4.2 of this investigation and in other studies [3], solidification cracks were observed to act as initiation sites for WMHACC.

As discussed in Section 2.4.5, Ohshita et al. [4] reported that the risk of solidification cracking increases as the welding speed approaches 500 mm/min. These authors used X60 and X70 steels in plate thicknesses between 15.2 and 25.0 mm, and pipe wall thicknesses between 15.2 and 19.1 mm. Solidification cracking was not observed below a critical welding speed of 330 mm/min in plate material and 300 mm/min in pipe girth welds, suggesting that a strong correlation exists between solidification cracking and welding speed.
Easterling [5] confirmed that there is a higher risk of solidification cracking with an increase in travel speed. Tear drop-shaped weld pools with columnar grains growing in parallel, straight rows toward the weld centreline tend to form at high welding speeds. This promotes higher levels of segregation at the weld centreline and a linear crack propagation path, which favours centreline solidification cracking. On the other hand, Nolan et al. [1,2] observed solidification cracking at travel speeds as low as 300 mm/min (the slowest welding speed tested in their study), suggesting that solidification cracking is possible over a wider range of welding speeds than previously believed.

![Graph showing welding speed vs welding current.](image1.png)

**Figure 4.5.1** Incidence of solidification (centreline) cracking in MWIC test pieces welded without preheat.

![Graph showing welding speed vs welding current.](image2.png)

**Figure 4.5.2** Boundary representing a 50% probability of WMHACC superimposed on data points showing the incidence of solidification cracking.

High heat input levels have also been reported to increase the risk of solidification cracking [1,2]. A high heat input promotes the formation of a coarse weld metal grain structure, resulting in higher levels of
segregation to the centreline grain boundaries. From the above discussion, high welding speeds and high heat input levels are expected to promote solidification cracking during welding.

Figure 4.5.3 shows the incidence of solidification cracking in preheat-free welds as a function of the heat input. There is significant overlap between cracked and uncracked samples over the entire range of heat inputs tested. Contrary to expectations, solidification cracking does not appear to have a strong correlation with heat input.

![Graph](image)

**Figure 4.5.3** Solidification cracking as a function of heat input and preheat temperature.

Table 4.5.1 shows the heat input levels of welds showing centreline cracking with the associated welding speeds. This confirms that solidification cracking was observed over the entire range of heat input levels tested. A small majority of the observed incidences of solidification cracking in preheat-free welds (7 out of the observed 12 cracked welds) occurred at heat input levels above 0.5 kJ/mm. Another seven welds showed evidence of centreline cracking at welding speeds above 400 mm/min (not the same seven welds). Most of the samples that cracked at high welding speeds were associated with intermediate to low heat input levels, while the samples that cracked at high heat input levels were associated with intermediate to low welding speeds (as seen from Figure 4.5.4). From these results, it is evident that a weak relationship exists between solidification cracking and the welding parameters (heat input and welding speed) used in this investigation.

The results of this investigation are therefore not in complete agreement with published literature. Ohshita *et al.* [4] did, however, report that an increase in restraint intensity increases the risk of solidification cracking. This is due to free shrinkage of weld metal being opposed by external restraint. The higher the external restraint intensity, the higher is the likelihood of solidification cracking. The level of restraint inherent to the MWIC test (at a restraint length of 25 mm) has been reported to be an order of magnitude higher than that obtained during actual pipeline girth welding [6]. It has been noted [7] that solidification cracking occurs whenever the shrinkage and restraint stresses exceed the hot fracture strength of the semi-solid metal during solidification. The high restraint inherent to the MWIC test is therefore expected to dominant other factors to some extent, hence weakening the expected correlation between solidification cracking and welding parameters.
Table 4.5.1 Heat inputs of samples showing solidification cracking with the corresponding welding speeds.

<table>
<thead>
<tr>
<th>Heat input (kJ/mm)</th>
<th>Welding speed (mm/min)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.3</td>
<td>650</td>
</tr>
<tr>
<td>0.3</td>
<td>714</td>
</tr>
<tr>
<td>0.4</td>
<td>432</td>
</tr>
<tr>
<td>0.4</td>
<td>708</td>
</tr>
<tr>
<td>0.5</td>
<td>486</td>
</tr>
<tr>
<td>0.6</td>
<td>472</td>
</tr>
<tr>
<td>0.7</td>
<td>405</td>
</tr>
<tr>
<td>0.8</td>
<td>258</td>
</tr>
<tr>
<td>0.85</td>
<td>212</td>
</tr>
<tr>
<td>0.9</td>
<td>200</td>
</tr>
<tr>
<td>0.9</td>
<td>270</td>
</tr>
<tr>
<td>1.0</td>
<td>283</td>
</tr>
</tbody>
</table>

Figure 4.5.4 Solidification cracking as a function of heat input and welding speed.

Oshita *et al*. [4] also reported that even though root beads with very low carbon contents (typically in the range of 0.05 to 0.10%) solidify as δ-ferrite, they still have a high susceptibility to solidification cracking. Subsequent cooling induces the phase transformation from δ-ferrite to γ-austenite, which introduces an additional shrinkage stress due to the volume difference between the two phases. This contributes to the already high shrinkage stress caused by external restraint, further dominating any correlations between solidification cracking and welding parameters.

No trends or meaningful correlations could therefore be found between the welding parameters and solidification cracking in the current investigation. It can be concluded that the experimental setup used in the current study was not suitable for assessing the influence of welding parameters on the incidence of solidification cracking during root pass welding of X70 linepipe steel using cellulosic E6010 electrodes.
It is therefore recommended that future work focuses on establishing trends between solidification cracking and welding parameters in a study designed primarily for detecting this type of weld metal defect, preferably using one of the existing tests specifically designed for the testing of hot/solidification cracking of welds [8]. If MWIC test pieces are to be used in a future study, the restraint length should be varied along with the welding parameters to determine the thresholds where constraint, or stress intensity, becomes the dominating factor in determining correlations between solidification cracking and welding parameters.

Due to the randomised nature of the occurrence solidification cracking in this study, it is further recommended that the test matrix be repeated several times to establish not only the occurrence of cracking, but the extent of variance in the results.

REFERENCES


4.6 MEASURED COOLING RATES OF WELDING

As described in Section 3.1, the cooling rates after welding were measured by lancing the molten weld pool directly behind the moving arc with a Pt-Pt-13% Rh (R-type) thermocouple. The cooling rate is influenced by the heat input during welding, rather than by individual welding parameters. Since the focus of the current study is the hydrogen-assisted cold cracking of the weld metal, the cooling rates of the heat-affected zones were not measured, but are assumed to follow the same general trends as those measured in the weld metal.

Two critical cooling regimes play a role in determining the WMHACC behaviour of root welds, namely the cooling rate through the temperature range from 800°C to 500°C, and the cooling rate from solidification down to 100°C [1].
The cooling rate between 800°C and 500°C affects the hardness of the weld metal and heat-affected zone by influencing the metallurgical transformation products that form on cooling. During fast cooling, harder phases like bainite or martensite are favoured. During slower cooling, various morphologies of ferrite form, and at intermediate cooling rates, combinations of bainite and ferrite may result. This will be discussed in greater detail in Section 4.8. The cooling rate between 800°C and 500°C also affects the relative coarseness of phases, with faster cooling rates resulting in a finer microstructure, which is inherently harder than a coarser microstructure containing the same phases. The time it takes to cool between 800°C and 500°C is designated as the $\Delta t_{8/5}$ time [2-4].

The second way the cooling rate influences the WMHACC behaviour is by controlling the amount of hydrogen that effuses out of the weld metal before the critical temperature range where WMHACC occurs (around 100°C to 150°C) is reached [5]. This is a longer time interval than the $\Delta t_{8/5}$ time. This time interval is designated as the $\Delta t_{100}$ time and is defined as the cooling time from solidification of the weld pool to a temperature of 100°C.

Accurate solidification temperatures could not be determined during the course of this investigation, and the thermocouples started registering temperatures on the cooling curves at temperatures between 1400°C and 900°C, depending on the sensitivity of the thin thermocouple wires to the high temperatures. The time available for hydrogen effusion down to 100°C was therefore defined differently in the current study as the cooling time between 800°C and 100°C, designated as $\Delta t_{8/1}$.

Figure 4.6.1 shows the influence of heat input on the cooling curves for heat inputs of 0.4, 0.6 and 1.0 kJ/mm. The time it takes for the weld metal to cool through the critical temperature ranges ($\Delta t_{8/5}$ and $\Delta t_{8/1}$) can be read off from these graphs. Higher heat inputs increase the cooling time of the weld metal, both in terms of the $\Delta t_{8/5}$ and the $\Delta t_{8/1}$ times. An increase in heat input is therefore expected to decrease the hardness of the weld metal (by favouring the formation of softer higher temperature transformation products) and to decrease the weld metal hydrogen content (by allowing more time for hydrogen to diffuse out of the welds).

![Figure 4.6.1](image)

**Figure 4.6.1** The effect of varying heat input levels on the weld cooling curves. The welds were deposited without preheat at a constant welding current of 130 A and at heat input levels of 0.4 kJ/mm (welding speed 487 mm/min), 0.6 kJ/mm (welding speed 325 mm/min), and 1.0 kJ/mm (welding speed 207 mm/min).
Figure 4.6.2 shows the effect of preheat on the cooling rate of the weld metal. The welds were deposited at a heat input of 0.4 kJ/mm using no preheat, and preheat temperatures of 50°C and 80°C, respectively. Both the $\Delta t_{6/5}$ and the $\Delta t_{6/1}$ cooling times are affected by preheating. The $\Delta t_{6/1}$ appears to be more strongly affected by the preheat temperature as there is a drastic increase in the time required for the weld metal to cool to 100°C when preheated. More diffusible hydrogen is therefore expected to effuse out of the weld metal as the preheat temperature increases, lowering the risk of WMHACC. The $\Delta t_{6/5}$ time also increases with an increase in preheat temperature, but the effect is less pronounced.

In the current study, various combinations of welding parameters were used to deposit welds at the same heat input levels. Multiple cooling curves were therefore measured at each heat input. This allowed for average trends in the $\Delta t_{6/5}$ and the $\Delta t_{6/1}$ cooling times to be established as a function of increasing heat input for all three preheat conditions. Figures 4.6.3 to 4.6.5 show the $\Delta t_{6/5}$ and the $\Delta t_{6/1}$ cooling times as a function of heat input for the three different preheat conditions.

Accurate temperature measurements for every weld deposited were often problematic due to the thermocouples being lanced by hand behind the moving arc. In certain cases, such as at the lower heat input levels where the welding speed was at a maximum, plunging the hot junction of the thermocouple successfully into the moving weld pool was challenging, especially with poor visibility through the welding helmet. At the higher end of the heat input range, plunging the thermocouple into the molten weld pool was often unsuccessful due to melt-off of the thin thermocouple wires. Several data points are therefore missing at the low and high ends of the heat input ranges, yet enough points are available to establish a trend in each case.

![Temperature vs. Time Graph](image)

**Figure 4.6.2** Effect of preheat temperature on the cooling curves of weld metal deposited at a constant heat input of 0.4 kJ/mm (welding current 150 A, welding speed 607 mm/min).

The trend lines shown in Figures 4.6.3 to 4.6.5 for the $\Delta t_{6/5}$ and $\Delta t_{6/1}$ cooling times are shown graphically in Figures 4.6.6(a) and (b). It is evident that both the heat input and the preheat temperature strongly affect the cooling times after welding, with the cooling times increasing linearly with heat input at all three preheat levels. As observed earlier, preheat appears to affect the $\Delta t_{6/1}$ cooling time more strongly than $\Delta t_{6/5}$.
Figure 4.6.3 Trends in $\Delta f_{8/5}$ (a) and $\Delta f_{8/1}$ (b) as a function of the heat input for welds deposited without preheat.

Figure 4.6.4 Trends in $\Delta f_{8/5}$ (a) and $\Delta f_{8/1}$ (b) as a function of the heat input for welds deposited at a preheat temperature of 50°C.

Figure 4.6.5 Trends in $\Delta f_{8/5}$ (a) and $\Delta f_{8/1}$ (b) as a function of the heat input for welds deposited at a preheat temperature of 80°C.
**Figure 4.6.6** Influence of heat input and preheat on the $\Delta t_{8/5}$ and $\Delta t_{8/1}$ cooling times of the weld metal.

Figure 4.6.7(a) and (b) shows the effect of the preheat temperature on the $\Delta t_{8/5}$ and $\Delta t_{8/1}$ cooling times, respectively (note that the time scales are different). For each cooling interval, the effect of increasing preheat temperature is shown for two heat input levels, namely 0.5 and 1.0 kJ/mm. The critical cooling times increase strongly with an increase in the preheat temperature. It is again evident that preheating affects the $\Delta t_{8/1}$ cooling time more strongly than $\Delta t_{8/5}$. This trend is consistent with the findings of Alam et al. [1] who reported that the time required to cool down to 100°C was more strongly affected by the preheat temperature than the $\Delta t_{8/5}$ cooling time, resulting in more hydrogen effusing out of the weld metal at higher preheating temperatures.

**Figure 4.6.7** Influence of the preheat temperature on the $\Delta t_{8/5}$ (a) and $\Delta t_{8/1}$ (b) cooling times for welds deposited at two different heat input levels (0.5 and 1.0 kJ/mm).

Relating these results back to section 4.4, according to the bivariate correlation analysis conducted on the welding parameters (which included the welding current, travel speed, heat input and preheat temperature),
as described in Section 4.4.12 with correlation parameters given in Table 4.4.10 in section 4.4.9, heat input has a more pronounced effect on the incidence of WMHACC than preheat. Preheat temperature, heat input and welding current had Pearson’s correlation coefficients of -0.369, -0.421, and -0.437, respectively. The more negative correlation coefficient of heat input compared to preheat temperature, suggests that increasing heat input is more effective in reducing the odds of WMHACC occurring.

Increasing the heat input and preheat temperature increases the cooling rate of the weld metal and affects WMHACC by reducing both the hardness (discussed in Section 4.7) and the diffusible hydrogen content of the weld metal. From section 1.1, these are two major contributors to WMHACC.

Increasing the welding current serves a dual purpose in that it increases the arc force and hence the weld pool depression, effectively lowering the intensity of the stress concentration at the root of the weld due to partial penetration. If the welding speed is not adjusted accordingly, increasing the welding current also increases the heat input, which reduces the cooling rate of the weld metal. This dual effect accounts for the higher Pearson’s correlation coefficient for heat input compared to preheating.

Figure 4.6.8 shows the boundary line representing a 50% probability of WMHACC for preheat-free welding. The corresponding theoretical heat input for data points along this line, included in Figure 4.6.8, can be determined using Equation 3.1.1. It is evident from Figure 4.6.8 that at welding currents below 130 A, the theoretical heat input required to ensure a 50% probability of cracking is higher than the maximum heat input used in this investigation. This can be attributed to lack of penetration at low welding currents. The more severe the lack of penetration, the higher the heat input required to achieve a cooling rate that results in weld metal hardness and hydrogen content low enough to resist WMHACC.

![Figure 4.6.8](image)

*Figure 4.6.8* Calculated heat input along a boundary representing a 50% probability of WMHACC for welds deposited without preheat.

As the welding current approaches 130 A, the required theoretical heat input decreases rapidly. Between welding currents of 130 and 140 A, a more gradual decrease in the heat input that results in a 50% probability of WMHACC is evident, and as the welding current continues to increase, the heat input starts levelling off. These results suggest that there are two separate regimes in Figure 4.6.8, with two different phenomena affecting the WMHACC behaviour.
In the first regime (at low welding currents), the WMHACC behaviour is dependent on weld bead geometry. Lack of penetration causes an additional stress concentration at the root of the weld, exacerbating the WMHACC tendency of the weld metal. A higher heat input, and hence cooling rate, is required to compensate for this effect, resulting in softer weld metal with a lower concentration of diffusible hydrogen more able to resist cracking.

In the second regime (at high welding currents), the arc force is sufficient to achieve a level of weld pool depression that results in full penetration welds, alleviating the stress concentration in the root of the weld. The risk of WMHACC is therefore no longer dependent on the root bead geometry, but is mainly determined by the cooling rate, and therefore the hardness and diffusible hydrogen content of the weld metal.

From the heat input values along the 50% probability line, the $\Delta t_{0.5}$ and $\Delta t_{0.1}$ cooling intervals along the boundary could in turn be determined. These cooling times are represented graphically in Figures 4.6.9(a) and (b). Because both the $\Delta t_{0.5}$ and $\Delta t_{0.1}$ cooling intervals increase linearly with an increase in heat input, as shown in Figure 4.6.6, similar trends are evident along the 50% probability line to those observed for heat input. Both the $\Delta t_{0.5}$ and $\Delta t_{0.1}$ cooling intervals rapidly decrease as the welding current increases to around 130 A. At higher welding currents the cooling times are less dependent on the current level. This confirms that at lower current levels, where the incidence of WMHACC shows a stronger dependency on weld bead geometry, a slower cooling rate (longer cooling time) is necessary to compensate for partial root penetration, resulting in lower hardness and lower diffusible hydrogen contents.

![Graphs showing $\Delta t_{0.5}$ and $\Delta t_{0.1}$ cooling intervals vs welding current](image)

**Figure 4.6.9** Calculated $\Delta t_{0.5}$ (a) and $\Delta t_{0.1}$ (b) cooling intervals along the boundary representing a 50% probability of WMHACC for welds deposited without preheat.

Figures 4.6.10 and 4.6.11 show the required heat input along the boundary line resulting in a 50% probability of WMHACC at preheat temperatures of 50°C and 80°C. The same trend is evident as in Figure 4.6.8, suggesting that at lower welding currents, WMHACC is strongly affected by the root bead profile, whereas at higher welding currents cooling rate effects dominate.

At higher preheat temperatures, lower heat input levels are required to achieve a 50% probability of cracking, and the decrease in required heat input with an increase in welding current is more gradual than
that observed in Figure 4.6.8. The higher preheat temperatures increase the overall depth of penetration of the weld metal, as was shown in Table 4.3.2 of Section 4.3, lowering the dependence of WMHACC on root bead geometry to some extent. An increase in preheat temperature also increases the cooling rate, allowing more hydrogen to diffuse out of the weld metal on cooling.

Figures 4.6.12 and 4.6.13 show the $\Delta_t/5$ and $\Delta_t/8/1$ cooling times as they vary along the boundary resulting in a 50% probability of WMHACC at preheats of 50°C and 80°C, respectively. Both the $\Delta_t/8/5$ and $\Delta_t/8/1$ cooling times show similar trends to those observed in Figure 4.6.9. Because higher preheat temperatures result in deeper overall weld penetration, the WMHACC behaviour is less dependent on the geometry of the weld metal, resulting in lower required heat inputs, and hence cooling rates, for a 50% probability of WMHACC.

Preheating increases both the time required for the weld metal to cool down from the liquid state to 100°C (i.e. $\Delta_t/8/1$), which defines the time available for hydrogen effusion from the weld metal, as well as the cooling interval/time (defined by $\Delta_t/8/5$). The final austenite transformation products in E6010 welds typically consist of either ferrite (in its various morphologies), or mixtures of ferrite and bainite (as discussed in more detail in Section 4.8). These transformation products all have BCC crystal structures [2-4] giving them a much lower hydrogen solubility limit than FCC austenite. The lower solubility limit for hydrogen in the BCC crystal structure is a strong driving force for hydrogen effusion out of the weld. With increasing preheat temperature, more time is spent in the $\Delta_t/8/5$ cooling interval and more hydrogen effuses out of the weld metal. This results in a weaker dependency of WMHACC on the $\Delta_t/8/4$ cooling interval.

![Figure 4.6.10](image-url)  
**Figure 4.6.10** Calculated heat input along the boundary representing a 50% probability of WMHACC for welds deposited at a preheat temperature of 50°C.

The heat input during welding also affects the burn-off rate of the welding electrode, effectively determining the amount of metal deposited in the weld bead [5]. Welding at higher heat input levels result in more metal being deposited and a thicker weld throat. The throat thickness of the weld determines how much stress the weld can withstand before crack initiation, with thicker throats being more resistant to cracking.
Figure 4.6.11 Calculated heat input along the boundary representing a 50% probability of WMHACC for welds deposited at a preheat temperature of 80°C.

Figure 4.6.12 Calculated $\Delta t_{8/5}$ (a) and $\Delta t_{8/1}$ (b) cooling intervals along the boundary representing a 50% probability of WMHACC for welds deposited at a preheat temperature of 50°C.

Figure 4.6.14 shows the throat thickness of welds deposited without preheat, measured in that part of the weld where crack initiation was observed, i.e. vertically down from the bottom of the wagon track for all welds deposited.

As can be seen from Figure 4.6.14, there is a roughly linear increase in the throat thickness as the heat input increases. The throat thickness as a function of the heat input can be represented by the trend line shown in Figure 4.6.14, given here as Equation 4.6.1, with $y$ the throat thickness in mm and $x$ the heat input in kJ/mm.

$$y = 1.0887x + 1.1332$$  \hspace{1cm} \text{Eq. 4.6.1}
Figure 4.6.13 Calculated $\Delta t_{8/5}$ (a) and $\Delta t_{8/1}$ (b) cooling intervals along the boundary representing a 50% probability of WMHACC for welds deposited at a preheat temperature of 80°C.

Figure 4.6.14 Throat thickness of weld beads deposited without preheat, measured vertically down from the bottom of the wagon track.

Using Equation 4.6.1, the throat thickness can be calculated as a function of the heat input as it varies along the boundary line representing a 50% probability of WMHACC (shown in Figure 4.6.15). It is evident that at low welding currents where WMHACC is strongly dependent on the weld geometry, a thicker throat is required to ensure a 50% probability of cracking. At higher welding currents, cooling rate effects dominate.

The results presented in this section explains the high negative Pearson’s correlation coefficient for increasing welding current. For welds deposited at room temperature, the required heat input for a 50% probability of WMHACC more than doubled as the welding current was reduced from its maximum value of 160 A to the minimum of 120 A. The required $\Delta t_{8/5}$ cooling interval increased by more than six times to maintain a 50% probability of WMHACC with a decrease in welding current, while the required $\Delta t_{8/1}$ time increased by more than three times and the required throat thickness almost doubled.
Figure 4.6.15 Calculated weld throat thickness required for welds deposited without preheat to ensure a 50% probability of WMHACC.

REFERENCES

4.7 HARDNESS MEASUREMENTS

As discussed in Section 2.4 of the literature review, there are three major requirements for the occurrence of hydrogen-assisted cold cracking: the presence of dissolved hydrogen, a high enough tensile stress, and a microstructure susceptible to cracking. The hardness of the weld metal therefore plays an essential role in determining the susceptibility of the microstructure to WMHACC.

In Section 4.2 it was shown that the WMHACC crack propagation paths are not confined to the weld metal, but often deviate through the heat-affected zones. Figures 4.2.4(a), 4.2.5(a) and (b), and 4.2.6 showed that cracks in different sections of the same sample all initiate in the weld metal, but veer off to propagate through either the weld metal or the heat-affected zone.

Crack initiation invariably occurs within the weld metal at pre-existing stress concentrations, most notably the wagon tracks at the toes of the weld metal, but crack propagation paths are determined by a more complex mechanism. Cracks tend to propagate preferentially through harder sections of the microstructure as the lower toughness associated with an increased hardness favours brittle, rapid, low energy crack growth [1]. The hardesses of the weld metal and heat-affected zones are therefore of the utmost importance in understanding WMHACC behaviour.

Figures 4.7.1(a) and (b) show hardness profiles across two weld beads, deposited at heat input levels of (a) 0.3 kJ/mm and (b) 1.0 kJ/mm. Measurements were made with a 0.5 kg load about a third of the weld height from the top of the weld, with individual indentations about 60 μm apart. The width of the weld is indicated by the vertical lines on the graphs.

![Hardness profiles across the weld metal (WM) and heat-affected zones (HAZ) of welds deposited preheat-free at heat input values of (a) 0.3 kJ/mm (welding current 130 A, welding speed 650 mm/min) and (b) 1.0 kJ/mm (welding current 160 A, welding speed 283 mm/min) (HV: hardness on the Vickers scale).](image)

The average base metal hardness of the X70 parent material was 210 HV. The hardness increased through the heat-affected zones approaching the fusion line, and reached a maximum at the fusion boundary. The maximum hardness values recorded were around 330 HV, but generally averaged at about 300 HV. Similar
values for X70 heat-affected zones were reported by Hanzaei et al. [2] More scatter was observed in the weld metal hardness measurements than in the heat-affected zone as the microstructure showed greater variance (the weld metal contained different morphologies of ferrite and mixtures of ferrite and bainite, whereas the heat-affected zones consisted mainly of bainitic transformation products) (as discussed in Section 4.8).

The hardness of the weld metal and heat-affected zones of the deposited welds were found to be a function of the heat input (since the chemical composition of the weld metal was not a variable in the current study), and hence the cooling rate. Figure 4.7.1 indicates that, as the heat input increased from 0.3 kJ/mm to 1.0 kJ/mm, the hardness decreased accordingly, with much lower hardnesses measured in both the weld metal and heat-affected zones of welds deposited at higher heat inputs. This is due to the slower cooling rates at high heat inputs, resulting in coarser transformation products with different ratios of ferrite and bainite.

Hardness levels were measured for all welds made at each heat input level (0.3 to 1.0 kJ/mm) over the entire current range tested. An example of the average measured hardness values as a function of the heat input is given in Figure 4.7.2 for a welding current of 150 A. The 95% confidence intervals were determined for each averaged measurement and are indicated as error bars in the figure.

![Figure 4.7.2 Weld metal and HAZ hardness for welds deposited preheat-free at a welding current of 150 A.](image)

Because the hardness of the weld metal and heat-affected zones were observed to be a function of only the heat input, the hardness measurements for each welding current could then be plotted on the same graph, as shown in Figure 4.7.3. No error bars were included on Figure 4.7.3 to avoid excessive clutter. Although there is a degree of scatter in the data, the general trends show that both the hardness of the weld metal and heat-affected zones decreasing linearly with an increase in heat input. The trend lines for hardness ($HV$) as a function of the heat input ($x$) (kJ/mm) can be mathematically expressed by means of Equation 4.7.1 for the weld metal and Equation 4.7.2 for the heat-affected zone.

\[
HV_{WM} = -82.158x + 296.4 \quad \text{Eq. 4.7.1}
\]

\[
HV_{HAZ} = -80.315x + 325.35 \quad \text{Eq. 4.7.2}
\]
Figure 4.7.3 Hardness measurements for the weld metal (WM) and heat-affected zones (HAZ) of welds deposited without preheat.

From Figure 4.7.3 it is evident that higher hardness values for both the weld metal and heat-affected zones are found at the lower end of the heat input range. As discussed in the literature review in Section 2.4.3, a harder microstructure is more susceptible to WMHACC, especially if the microstructure was formed at high cooling rates, not allowing hydrogen to fully effuse out of the weld metal. A higher incidence of WMHACC is therefore expected at the lower end of the heat input range, as observed in Figures 4.3.1 and 4.3.3 where WMHACC dominated at low travel speeds and welding currents where low heat input levels prevailed. The strong Pearson correlation coefficient of -0.421 observed for the predictor variable of heat input in Table 4.4.10 in section 4.4.9 can be attributed to the effect of heat input on the hardness of the welds, the cooling rate, and the increase in depth of penetration with an increase in the welding current.

The hardness of the heat-affected zone is consistently higher than that of the weld metal by at least about 30 HV, which could possibly explain the occasional observation of crack paths veering towards and propagating partially through the heat-affected zones. If a crack initiates at a wagon track and its initial direction of propagation is vertically down towards the underside of the joint, it could grow through a section of the harder heat-affected zone if the shape of the weld metal is either vertically concave, such as shown in Figure 4.2.6 of Section 4.2, or vertically convex, such as in Figure 4.2.5. A crack could also initiate at a wagon track, propagate vertically down through the heat-affected zone and change direction back towards the weld metal if a favourable stress concentration exists in or near the root of the weld, as seen in Figures 4.2.5(b) and (c).

As discussed in Section 4.2, cracks observed to have changed direction could have initiated from both stress concentrations (at the wagon track and in or near the root of the weld) and changed direction to join. Such cracks could also propagate without meeting, as shown in Figure 4.7.4 for a weld in which the crack growth direction appeared to be indiscriminate of the hardness of the weld metal and heat-affected zone.

Figure 4.2.5(a) in Section 4.2 shows a crack from the same sample as shown Figures 4.2.5(b) and (c), but propagating through the weld metal only. The crack propagation path could simply be following the shortest route from the stress concentration at the wagon track to that in or near the root of the weld. It was observed, however, that in certain samples, such as those shown in Figures 4.2.5(b) and (c), the crack veers from the
weld metal towards the harder heat-affected zone, before turning back to the stress concentration at the root of the weld.

![Image](image_url)

**Figure 4.7.4** WMHACC initiated at two stress concentrations that propagated vertically down and up in a preheat-free weld deposited at a heat input of 0.3 kJ/mm (current of 150 A and travel speed of 725 mm/min).

Based on experimental observations, it cannot be stated with confidence that a propagating crack will consistently veer towards the harder heat-affected zone in order to take advantage of the lower toughness of the HAZ. A crack could also be said to change its propagation path towards the harder heat-affected zone if there is evidence of a higher hydrogen concentration in the heat-affected zone than in the weld metal, but since hydrogen contents were not measured in this study, this could not be confirmed. The available evidence suggests that a higher hardness in the heat-affected zone would merely facilitate crack propagation near or towards it. The hardness levels of the weld metal and heat-affected zone are therefore not considered a major factor in determining WMHACC crack propagation paths.

Barbaro [3] remarked on similar observations of the crack propagation paths in root pass welds made in X80 steel with cellulose consumables. Crack paths were reported to initiate at stress concentrations within the weld metal and propagate through both the weld metal and heat-affected zones. It was noted that crack propagation paths were influenced by microstructural characteristics such as the presence of allotriomorphic ferrite (grain boundary ferrite ‘veins’) and the prior austenite grain size. Further remarks on microstructure will be made in Section 4.8.

Figures 4.7.5 and 4.7.6 show the hardness of the weld metal and heat-affected zones of welds deposited at preheat temperatures of 50°C and 80°C. The difference in hardness between the weld metal and heat-affected zones ranged on average between 30 and 40 HV, with the hardness displaying the same linear decrease in hardness with an increase in heat input, and hence cooling rate.

The hardness trend lines for all three preheat conditions are shown in Figure 4.7.7. The same trends are evident. The difference in hardness between weld metal deposited at the minimum heat input of 0.3 kJ/mm and the maximum heat input of 1.0 kJ/mm was consistently about 55 HV for all three preheat conditions. An average difference of about 55 HV was also found between the hardnesses of the heat-affected zones of welds deposited at the minimum heat input level of 0.3 kJ/mm and the maximum of 1.0 kJ/mm for all three preheat conditions. For both the weld metal and heat-affected zones at all three preheat conditions there is a decrease in hardness of about 7 HV for an incremental increase in heat input of 0.1 kJ/mm.
Preheat has a similar effect on the hardness levels. There is an average difference in hardness between weld metal deposited without preheat and at a preheat of 50°C of about 16 HV, and between welds deposited at preheat temperatures of 50°C and 80°C of about 10 HV.

![Graph](image)

**Figure 4.7.5** Hardness measurements of the weld metal (WM) and heat-affected zones of welds deposited at a preheat temperature of 50°C.

![Graph](image)

**Figure 4.7.6** Hardness measurements of the weld metal (WM) and heat-affected zones of welds deposited at a preheat temperature of 80°C.

The changes in hardness are fairly consistent, suggesting that changes in cooling rate brought about by variations in heat input did not result in major changes in microstructure. The small and consistent decreases in hardness with incremental increases in the heat input may be associated with an increase in the coarseness of the phases present in the weld metal and heat-affected zones. This will be discussed in more detail in Section 4.8.

The small differences in hardness between welds deposited at the three preheat temperatures cannot alone account for the large differences in WMHACC behaviour observed in Figures 4.3.3, 4.3.5 and 4.3.6. It is therefore reasonable to assume that increasing preheat temperature predominantly affects the WMHACC behaviour by increasing the cooling times, shown in Figure 4.6.7. A higher preheat temperature resulted in
exponentially longer cooling times, allowing more hydrogen to effuse out of the weld metal and effectively lowering the risk of WMHACC. The effusion of hydrogen from the weld metal resulting from preheating therefore appears to have a stronger effect on the WMHACC behaviour than the small decrease in hardness with increasing preheat temperature.

![Graph showing hardness trends vs. heat input](image)

**Figure 4.7.7** Calculated hardness trends as a function of heat input for welds deposited without preheat and at preheat temperatures of 50 and 80°C.

Using Equations 4.7.1 and 4.7.2, the hardness along the boundary representing a 50% probability of WMHACC could be determined, as shown in Figure 4.7.8. The hardness levels required to maintain a 50% probability of WMHACC increase rapidly as the welding current approaches 130 A. As explained earlier, this can be attributed to the presence of the stress concentration at the root of the weld due to lack of penetration at lower welding currents. With WMHACC dominated by the weld bead geometry due to this stress concentration, lower hardness levels are required to maintain a 50% probability of WMHACC in this current range. As the welding current increases and the stress concentration at the root of the weld becomes less severe, the hardness levels required to ensure a 50% probability of failure start to level out at around 263 HV for the weld metal and 293 HV for the heat-affected zone.

Figures 4.7.9 and 4.7.10 show the hardness of the weld metal and heat-affected zone of welds deposited at preheat temperatures of 50°C and 80°C, respectively, along the boundary representing a 50% probability of WMHACC. With an increase in preheat temperature, deeper penetration was observed at the weld root, resulting in a less severe stress concentration at the root of the welds at low welding currents. The increase in hardness as the welding current approaches 130 A is therefore less severe than for samples deposited without preheat. In welds preheated to 50°C the hardnesses start to level off to values of just under 260 HV for the weld metal, and just over 290 HV for the heat-affected zone. At a preheat temperature of 80°C, the hardness values level off much sooner, with final hardness values for the weld metal just below 258 HV and for the heat-affected zone just below 294 HV at a welding current of 160 A.
Figure 4.7.8 Calculated WM and HAZ hardness values along the boundary representing a 50% probability of WMHACC for welds deposited without preheat.

![Figure 4.7.8](image)

Figure 4.7.9 Calculated WM and HAZ hardness values along the boundary representing a 50% probability of WMHACC for welds deposited at a preheat temperature of 50°C.

![Figure 4.7.9](image)

Since the hardness of the weld metal and heat-affected zones is a function of the heat input, it is also a direct function of the critical $\Delta t_{85}$ cooling time. Figure 4.7.11 shows the relationship between the hardness of the weld metal and the heat-affected zone and the critical cooling time $\Delta t_{85}$. The hardness can be expressed mathematically as a function of the cooling time $\Delta t_{85}$ by means of Equations 4.7.3 and 4.7.4, with $x$ being the $\Delta t_{85}$ cooling time in seconds.

\[
HV_{WM} = -8.1516x + 273.94 \quad \text{Eq. 4.7.3}
\]

\[
HV_{HAZ} = -7.8669x + 306.47 \quad \text{Eq. 4.7.4}
\]
**Figure 4.7.10** Calculated WM and HAZ hardness values along the boundary representing a 50% probability of WMHACC for welds deposited at a preheat temperature of 80°C.

**Figure 4.7.11** Calculated WM and HAZ hardness values as a function of the calculated cooling time $\Delta t_{85}$ for welds deposited without preheat and at preheat temperatures of 50 and 80°C.

The heat input and the associated critical cooling time $\Delta t_{85}$ can be calculated for any acceptable probability of failure. Once the cooling times required for the desired probability of WMHACC are known, the resulting hardness values can be calculated according to Equations 4.7.3 and Equations 4.7.4. These hardness values can then be used to ensure that the hardness values in pipeline welds fall within a range that will guarantee a reasonable level of safety.

The hardness levels measured in the current study may also be used to specify limits for the avoidance of WMHACC since no mention is made to such limitations within the Australian Standard (AS 2885.2).
REFERENCE


4.8 WELD METAL AND HAZ MICROSTRUCTURES

The metallurgical phases present in the weld metal and heat-affected zone are determined by the chemical composition and the cooling rates after welding, as discussed in Section 4.6. In this section, the microstructures of these two components will be analysed in depth in terms of the phases present and their comparative coarseness. The hardness values, presented in Section 4.7, did not show any major fluctuations as a function of heat input and preheat temperature, rather linear and consistent changes in hardness were observed. Because the hardness of steel is directly related to chemical composition and the metallurgical phases present, the weld metal and heat-affected zones in this study are expected to have similar metallurgical microstructures with only gradual changes in the phases present and their relative coarseness as the welding parameters change.

Throughout this section, attempts will be made to correlate the observed phases with the modified International Institute of Welding (IIW) classification scheme described by Thewlis [1] and presented in Figure 2.4.9 of the literature review for low-carbon steel welds. The interpretation of certain structures will also be compared to the findings of other published authors.

Comparisons are made in this section between the microstructures of welds deposited at low and high heat input levels with no preheat or at a preheat temperature of 80°C. Microstructures at the extremes of the heat input and preheat ranges were selected for microstructural classification to highlight any differences between these welds. A few weld metal and heat-affected zone microstructures of welds deposited at intermediate heat input levels are also included to highlight gradual changes with increasing heat input or preheat levels.

Optical and high-resolution scanning electron microscopy techniques were used throughout this study to classify the transformation products present in the weld metal and heat-affected zone microstructures. Accurate classification of certain phases is often complicated by the limitations of these techniques and further investigation by EBSD and transmission electron microscopy could clarify the ambiguities in many cases shown here. The application of these techniques was beyond the scope of the current study.

As shown in Table 3.1.3, the carbon and manganese contents of the weld metal were 0.14% and 0.45% by weight, respectively. These values indicate that the weld metal is a low-carbon steel with a carbon equivalent, CE_{IIW}, of 0.24 (according to Equation 1.1.1) and a P_{cm} value of 0.18 (according to Equation 1.1.2). The low carbon equivalent values suggest that the weld metal has low hardenability. The CE_{IIW} and P_{cm} values of the X70 base metal are 0.38% and 0.16%, respectively. Slight differences in the levels of dilution between welds deposited at the minimum and maximum heat input levels in the test matrix shown in Table 3.1.5 are expected, with an increase in the level of dilution as the heat input increases.

4.8.1 Microstructures of welds deposited without preheat

Figures 4.8.1 and 4.8.2 show optical photomicrographs of the microstructure of a weld deposited without preheat at a heat input of 0.4 kJ/mm. Since all micrographs were taken at high magnification, two micrographs were selected for the discussion in order to accurately represent the phases present and their relative morphologies.
Both micrographs primarily consist of different forms and morphologies of ferrite (α). Box 1 in both figures indicates allotriomorphic ferrite in the form of grain boundary primary ferrite, designated as PF(GB). It is the first phase to form below the \(A_e3\) temperature during continuous cooling from austenite (γ) in alloys with low hardenability. Allotriomorphic ferrite forms at low undercoolings along the prior austenite grain boundaries during the \(\Delta T/5\) cooling interval and involves a fully reconstructive phase transformation from the FCC austenite crystal structure to the BCC ferrite crystal structure. It is a diffusion-controlled transformation where substitutional alloying element partitioning occurs in bulk across the transformation interface, followed by subsequent long-range carbon redistribution. Carbon is an interstitial alloying element and since the BCC ferrite crystal lattice has a lower solubility limit for carbon, the FCC austenite phase ahead of the transformation interface becomes enriched in carbon during the transformation [1]. Because allotriomorphic ferrite is the first phase to form from austenite at higher temperatures on cooling, it can be identified with greater ease than some of the other transformation products.

The allotriomorphic ferrite in Figures 4.8.1 and 4.8.2 is present in the form of semi-continuous ferrite veins and polygonal ferrite. Allotriomorphic ferrite has a lower hardness than other ferrite constituents observed in low-carbon steel weld metal, but also exhibits lower toughness [2]. It has been noted [3,4] that for this reason, WMHACC cracks often propagate preferentially along allotriomorphic ferrite veins on prior austenite grain boundaries. This phenomenon was not observed during the current study, with cracks appearing to propagate intragranularly despite the presence of allotriomorphic veins.

A distinction is made in literature between two types of primary ferrite: allotriomorphic ferrite that forms along prior austenite grain boundaries (intergranular primary ferrite) and idiomorphic ferrite (intragranular primary ferrite) [1]. The latter, designated as PF(I), nucleates and grows from intragranular non-metallic inclusions (NMI) after the austenite grain boundaries are fully occupied and is therefore expected to be much
finer than allotriomorphic ferrite. If it is of a size approximately three times larger than the surrounding intragranular ferritic constituents, such as acicular ferrite laths, it can be regarded as allotriomorphic ferrite originating from a prior austenite grain boundary underneath or above the plane of observation.

![Image of microstructure](image)

**Figure 4.8.2** Optical micrograph of the microstructure of a weld deposited without preheat at a heat input of 0.4 kJ/mm (welding current 120 A, welding speed 432 mm/min).

Only tentative identification of idiomorphic primary ferrite is made in the optical micrographs, as shown in box 7 of Figure 4.8.1, since the resolution of these images is not sufficient to enable firm identification. Clearer identification was subsequently made on images obtained from high resolution scanning electron microscopy.

Box 4 in both Figures 4.8.1 and 4.8.2 indicates Widmanstätten ferrite side-plates that nucleated at the prior austenite grain boundaries, designated as WF(GB). This type of ferrite requires a slightly higher driving force and forms at a higher undercooling than allotriomorphic ferrite. This transformation product generally forms as the next step after allotriomorphic ferrite during continuous cooling. Since there is an increase in the carbon content ahead of the γ/α transformation interface during faster cooling, the growth of allotriomorphic ferrite is effectively hindered and only cooperative growth of ferrite needles/side-plates can penetrate through the carbon-enriched barrier in the austenite grain [1,2].

Side-plates have typical length-to-width aspect ratios in the order of 10:1. This type of growth is rapid and resembles a shear transformation. A para-equilibrium exists at the transformation interface where little to no diffusion of substitutinal elements occurs, with carbon redistribution occurring to the sides of the growing needles. The redistribution of carbon leads to an aligned high-carbon microphase between the ferrite needles, which can be cementite, ferrite/carbide aggregate (pearlite), residual austenite, bainite or martensite in higher carbon alloys. Furthermore, Widmanstätten ferrite plates are separated by low angle grain boundaries that can be difficult to resolve using light microscopy [1,2].
In all of the micrographs presented in this section, Widmanstätten ferrite plates are identified as WF(GB) whether they nucleated on prior austenite grain boundaries or from grain boundary allotriomorphic ferrite. A distinction is made by Thewlis [1] based on the nucleation site, with the former designated as primary grain boundary Widmanstätten ferrite and the latter as secondary grain boundary Widmanstätten ferrite. A similar distinction is made for intragranular Widmanstätten ferrite nucleated on non-metallic inclusions. In this section, all intragranularly nucleated Widmanstätten ferrite is designated as WF(I), although plates nucleated directly on NMI’s are referred to as primary intragranular Widmanstätten ferrite and plates nucleated on idiomorphic ferrite are called secondary intragranular Widmanstätten ferrite.

Widmanstätten ferrite forms within the temperature range of shear transformation products during continuous cooling, yet with a para-equilibrium across the transformation interface. It exhibits a higher hardness than primary allotriomorphic and idiomorphic ferrite, but lower than bainite. The structure can, however, be confused with upper bainite, which requires a higher driving force and forms at a slightly higher undercooling than Widmanstätten ferrite.

Bainite forms as a partial shear/displacive transformation product accompanied by an invariant plane strain mechanism in the temperature range where reconstructive/diffusion-controlled transformations are slow. No diffusion of substitutional elements occurs across the transformation interface with rapid carbon redistribution into the remaining austenite between plates. Due to the rapid carbon diffusion, the transformation is not purely displacive and it has been postulated that a para-equilibrium also exists across the advancing γ/α interface. Bainite sheaves resemble parallel arrays with cementite as an aligned second phase, and exhibit a higher dislocation density than Widmanstätten ferrite [1].

Bainite, in addition, is expected to have a finer structure than Widmanstätten ferrite, because it forms at a higher undercooling (lower transformation temperature) during continuous cooling, which could ease its identification. The aligned microphase can also facilitate the identification of the two phases, since it is expected that cementite is almost exclusively aligned with upper bainite and the aligned phase between Widmanstätten ferrite plates may range from pearlite to martensite and retained austenite. The relative coarseness of these phases is often convoluted by the plane of observation and microphases are not readily resolved using an optical microscope.

In the weld metal micrographs in the current study, Widmanstätten ferrite is often clearly identifiable as a phase originating from the prior austenite grain boundaries or from allotriomorphic ferrite, and in some cases from intragranular NMI’s, yet its similar appearance to upper bainite often renders the two phases indistinguishable.

Ghoshachi et al. [5] identified both Widmanstätten ferrite and upper bainite in the weld metal microstructures of X70 welded with E6010 cellulosic electrodes under similar welding conditions. Both phases are therefore expected to be present in the microstructures of welds in the current investigation, but in order to make a definitive distinction, EBSD and TEM techniques should be employed. In the modified IIW classification scheme by Thewlis [1], both phases are identified as ferrite with an aligned second phase FS(A). It is common in literature for different phases with similar appearance to be given the same designation. Different designations are often assigned to a single phase [6]. Whenever side-plates cannot be clearly identified as Widmanstätten ferrite or bainite in the weld metal, they are generally given the designation FS(A), such as in box 3 in Figure 4.8.1 and boxes 2 and 6 in Figure 4.8.2.
The plane of observation often complicates the identification of ferrite side-plates with aligned second phase. If the observation is made longitudinally along the ferrite plates, the second phase appears as fine needles with an elongated and aligned second phase, as shown in box 4 of Figure 4.8.1. When the observation is made transversely or diagonally across the plates, however, the phases are seen as irregularly shaped ferrite grains, with carbides scattered throughout. The ferrite plates are separated by low angle boundaries that cannot be clearly resolved using an optical microscope and the shape of the carbides may appear roughly spherical when viewed transversely, or as short, elongated needles from a diagonal viewpoint [1]. Such observations are designated as ferrite side-plates with a non-aligned second phase FS(NA), as indicated in box 2 in Figure 4.8.1 and box 5 in Figure 4.8.2. The only difference between FS(A) and FS(NA) is the plane of observation.

Acicular ferrite (AF) is shown in boxes 3 and 5 in Figure 4.8.1, box 3 in Figure 4.8.2 and box 1 in Figure 4.8.3, which is a high-resolution SEM micrograph of weld metal deposited without preheat at a low heat input (0.4 kJ/mm). The term ‘acicular’ suggests a needle-like structure, as shown in Figures 4.8.1 to 4.8.3, yet this is not always the case. The structure often resembles short plates [1,2]. Acicular ferrite consists of hard impingements of intragranularly nucleated transformation products, with fine distributions of NMI’s (typically ≤ 5μm in diameter) acting as the main nucleation sites. Mutual impingement occurs during the growth of these intragranular products, resulting in a fine interlocking structure of plates or needles with an average aspect ratio of about 4:1 [1].

![Figure 4.8.3](image_url)

*Figure 4.8.3* High-resolution FEG-SEM micrograph of a weld metal microstructure deposited preheat-free at a heat input of 0.4 kJ/mm (welding current 140 A, welding speed 536 mm/min).

The interlocking nature of the intragranular structure increases the toughness of welds as it promotes the deflection of propagating cracks and contributes effectively to crack arrest. This gives commercial importance to acicular ferrite and it is considered more beneficial to the toughness of weld metals than other transformation products [1,7].
Acicular ferrite can consist of different transformation products nucleating intragranularly on NMI’s when the primary nucleation sites, such as grain boundaries and grain surfaces, are fully occupied by polygonal or allotriomorphic ferrite. Acicular ferrite may therefore consist of intragranularly nucleated Widmanstätten ferrite or acicular bainite [1,8] although Abson [8] maintains that acicular ferrite grows by a reconstructive mechanism (diffusion-controlled) that reaches completion before the temperature reaches the displacive transformation regime and intragranular bainite forms. The modified IIW classification refers to the interlocked structures simply as AF. Therefore, even though the possibility of both structures is discussed in the current study, they are referred to throughout as AF.

If the statement by Abson [8] holds true, then intragranular acicular ferrite and intergranular bainite would differ by the dislocation densities inherent in the two separate phase transformation regimes, and intragranularly precipitated bainite would increase the hardness of weld metals. Ghomaschi et al. [5] showed that bainite laths are separated by low angle boundaries and acicular ferrite by high angle boundaries. These authors used techniques such as EBSD-IQ mapping superimposed on inverse pole figures (IPF) to illustrate the drastic orientation difference between acicular ferrite plates when compared to the low orientation difference between bainite laths in order to accurately distinguish between the two phases. They confirmed that both phases were present in the weld metal microstructures of X70 welded with E6010 electrodes under conditions similar to the current study.

Babu and Bhadeshia [7] demonstrated the interchangeable nature of acicular ferrite and bainite in the same steel during identical isothermal transformation treatments. This implies that acicular ferrite and bainite form in the same temperature range during continuous cooling. It has been shown by the same authors that, as during the transformation from austenite to bainite, the transformation to acicular ferrite is purely displacive with no diffusion of substitutional elements across the γ/α transformation interface. Carbon redistribution into the remaining austenite occurs after the reaction is complete, when the carbon-enriched austenite can transform into a variety of microphases. It has also been argued by several authors that acicular ferrite is nothing more than intragranularly nucleated bainite [9,10].

Bainite formation is favoured in steels with small austenite grains and grain boundaries free from allotriomorphic ferrite. Acicular ferrite forms in large austenite grains when the grain boundaries are occupied by primary ferrite and a fine distribution of intragranular NMI’s is present to act as nucleation sites [7].

Also shown in Figure 4.8.3 are primary Widmanstätten ferrite side-plates, nucleated on both a previous austenite grain boundary (PAGB) (indicated by box 5), and individual needles nucleated intragranularly on NMI’s, designated as WF(I), with the average aspect ratio of 10:1 (indicated by box 2).

Figure 4.8.4 shows a higher magnification SEM micrograph of weld metal deposited without preheat at a heat input of 0.4 kJ/mm. It shows similar phases to those seen in the previous micrographs but, in addition, it facilitates positive identification of idiomorphic ferrite (as indicated by box 7), where the idiomorphic ferrite grains are within the size range of less than about three times the thickness of surrounding transformation products. The ferrite grain indicated by the upper arrow of box 7 has a NMI embedded near the edge. This could possibly be the NMI that acted as nucleation site for the grain, with the ferrite grain growing until its growth was impeded by other transformation products. This is only a tentative observation since the ferrite grain could have nucleated on a larger NMI either below or above the plane of observation.
Figure 4.8.4 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited preheat-free at a heat input of 0.4 kJ/mm heat input (welding current 160 A, welding speed 708 mm/min).

Acicular ferrite is also evident in this figure, indicated by boxes 1 and 8. Both arrows from box 1 point towards NMI’s that could possibly have acted as nucleation sites for the acicular ferrite, although these are not always visible within the two-dimensional views presented by light and scanning electron microscopy.

Boxes 4 and 5 indicate the presence of microphases between the Widmanstätten and acicular ferrite needles, resulting from carbon saturation of the remaining austenite during the transformation to ferrite. As stated before, these microphases can consist of a range of products from both reconstructive and displacive transformations, such as residual austenite, pearlite, bainite or martensite. Higher magnifications of microphases are shown in subsequent micrographs.

Boxes 2 and 6 in Figure 4.8.4 indicate intragranular Widmanstätten ferrite although differentiation between primary (nucleated on NMI’s) and secondary (nucleated on idiomorphic ferrite) WF is unclear. In some cases, it appears as though the transformation proceeded by sympathetic nucleation and growth from other plates, as indicated by the top two arrows of box 6, although the nucleation site (whether idiomorphic ferrite or NMI) could also lie outside the field of observation.

Box 1 in Figure 4.8.5 and box 6 in Figure 4.8.6 show a transformation product consisting of ferrite-carbide aggregate (FC). These SEM micrographs were taken of weld metal deposited preheat-free at heat input levels of 0.5 kJ/mm and 0.7 kJ/mm, respectively. This transformation product could be classified as either periodic pearlite or granular bainite as it consists of blocky and irregular ferrite grains with a fine carbide precipitate distributed throughout the grains. The carbides indicated by the bottom arrow of box 6 appear to be in regular arrays.
In low-carbon steels the carbon content is not sufficient for the growth of lamellar pearlite, yet during cooling, cooperative growth of ferrite and cementite occurs below the $\text{Ar}_1$ temperature, resulting in periodic pearlite that resembles ferrite grains with fine cementite particles scattered throughout. The modified IIW classification by Thewlis [1] refers to this type of ferrite-carbide aggregate as degenerate pearlite. Degenerate pearlite of similar appearance was identified by Ghomashchi et al. [5] in the weld metal of X70 welded with E6010 electrodes. It is therefore expected that this transformation product would also be present in the microstructures observed during the current study.

Zajac et al. [11] found a similar transformation product, which formed at temperatures just below the $\text{Ar}_1$, and is comparable to the FC product indicated by box 6 in Figure 4.8.6. Through FEG-SEM and EBSD techniques, they identified this product as degenerate pearlite. As the transformation temperature decreased, the structure became more chaotic, with what they termed ‘debris of cementite’ scattered throughout the ferrite grains, similar in appearance to the cementite morphologies indicated by box 5 in Figure 4.8.6.

The same authors classified the various bainitic morphologies in low-carbon linepipe alloys and found the structure of granular bainite to be similar to the phase observed in these figures. The major difference between granular bainite and degenerate pearlite is the nature of the second phase and the size of the cementite particles. The second phase was composed of cementite debris, martensite/residual austenite (M/A), and mixtures of various incomplete transformation products. They also found evidence of a dislocation substructure within the grains [11]. Microstructures containing increasing amounts of granular bainite are therefore expected to show an associated increase in hardness.
Figure 4.8.6 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited preheat-free at a heat input of 0.7 kJ/mm (welding current 120 A, welding speed 257 mm/min).

In the current study, and as shown subsequently, this phase, simply designated FC in the micrographs, increases in amount as both the heat input and preheat temperature increase. The hardness values presented in Section 4.7 showed steady and linear decreases in hardness as the heat input and preheat temperatures increased. Therefore, in the absence of TEM and EBSD analyses, it is inferred that although phases such as lower and granular bainite are a likely occurrence in the weld metal during this investigation, the FC transformation product observed in the microstructures is rather a form of degenerate pearlite.

Box 8 in Figure 4.8.5 shows the presence of a microphase that is similar in appearance to the degenerate pearlite discussed above. An increase in the amount of these microphases with an increase in heat input and preheat temperature will be shown subsequently. This transformation product, in addition to the FC product shown in box 5 in Figure 4.8.6, show that the transformation products resulting from carbon redistribution can be of various forms and morphologies, yet are assumed to be a form of cementite precipitation.

Box 3 in Figure 4.8.7 show extremely fine carbide distribution within grains of acicular and Widmanstätten ferrite. Such a fine carbide distribution could be the result of supersaturation of carbon within the grain in the temperature range where diffusion is sluggish. The carbide distribution resembles the structure of lower bainite, which forms at a transformation temperature below that of upper bainite and therefore requires a larger driving force and higher undercooling [1].

Both acicular and Widmanstätten ferrite forms in temperature ranges where displacive transformation products start to dominate and carbon redistribution occurs to various extents, although carbon redistribution within Widmanstätten ferrite is expected to be more efficient since its transformation temperature is higher than that of acicular ferrite.
Figure 4.8.7 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited preheat-free at a heat input of 0.7 kJ/mm (welding current 120 A, welding speed 257 mm/min).

As stated before, Babu and Bhadeshia [7] showed that acicular ferrite and bainite may form in the same temperature range, and the phases shown in Figure 4.8.7 could therefore be intragranular lower bainite nucleated heterogeneously on inclusions. Abson [8] noted that bainite colonies nucleated on intragranular inclusions may resemble acicular ferrite and Widmanstätten ferrite with a fine precipitate structure within the laths, since the size of intragranular bainitic colonies is restricted by the proximity and impingement of similar colonies and other intragranular transformation products. It would be more accurately termed lower bainite, although in box 3 it is given the designation FC since lower bainite is only tentatively identified in this section.

Figure 4.8.8 shows similar structures resembling intragranular lower bainite, as indicated by box 2. The phase indicated by box 1 could also be lower bainite nucleated as a result of carbon redistribution between acicular and Widmanstätten ferrite side-plates. It could also be a fine degenerate pearlite. According to Thewlis [1] lower bainite has a darker etch response under an optical microscope than other transformation products consisting of ferrite due to the fine distribution of intragranular carbides. The darker regions between the acicular ferrite needles shown in Figure 4.8.2 could therefore consist of a mixture of transformation products including lower bainite, as was found by Ghomashchi et al. [5].

Furthermore, Figure 4.8.8 shows primary idiomorphic ferrite, indicated by boxes 4 and 5, intragranular Widmanstätten ferrite, although it is unclear whether it is primary or secondary Widmanstätten ferrite (box 7), acicular ferrite (box 8) and ferrite side-plates with an aligned second phase (box 9).
Figures 4.8.1 to 4.8.8 show a very fine microphase between the transformation products. In most cases, the exact nature of the microphases is unclear and it could be any of the various possibilities discussed previously resulting from carbon saturation of the remaining austenite.

Although the cooling rates of the heat-affected zones were not measured during the course of this investigation, the heat-affected zones are expected to cool more rapidly than the weld metal due to their close proximity to the heat sink posed by the unaffected base metal. More displacive transformation products are therefore expected in the heat-affected zone microstructures. The hardness measurements in Section 4.7 and the microstructures presented in this section only present data pertaining to the high-temperature heat-affected zones adjacent to the weld fusion lines, since crack propagation was observed mostly within the weld metal and occasionally through the high-temperature heat-affected zones.

Figure 4.8.9 shows the heat-affected zone (HAZ) of a weld deposited preheat-free at a heat input of 0.4 kJ/mm. The microstructure consists predominantly of upper bainite, designated as FS(UB), and a smaller scattering of what is assumed to be lower bainite, designated as FS(LB). The intragranular carbide structure within lower bainite is extremely fine and difficult to resolve optically, but this constituent does have a darker etch response than the surrounding ferrite products [1].
**Figure 4.8.9** Optical micrograph of the heat-affected zone microstructure of a weld deposited preheat-free at a heat input of 0.4 kJ/mm heat input (welding current 130 A, welding speed 487 mm/min).

Hanzaei et al. [12] used optical and scanning electron microscopy together with ImageJ™ image analysis software to determine the amounts of the different transformation products observed in the heat-affected zones of welded X70 linepipe steel. These authors mainly found bainite, although they did not distinguish between upper and lower bainite, along with 10 to 20% martensite. The presence of these phases was not confirmed by any other techniques. They reported hardness values for the high temperature heat-affected zones of around 300 HV, similar to the findings of the current study.

Martensite could not be unambiguously identified in the heat-affected zones of the current study, although the possibility of finding extremely small martensitic islands remains. The CEHW of 0.38% of the base metal suggests moderate hardenability in the heat-affected zone and therefore small amounts of martensite is a possibility. At hardness values of around 300 HV, the likelihood of finding martensite in the high-temperature heat-affected zone is, however, small.

Box 1 in Figure 4.8.9 shows primary ferrite (PF). At a carbon level of 0.052% and a Pcm value of 0.16 in the X70 base metal, some ferrite is expected to be present within the heat-affected zones. Boxes 3 and 6 point to ferrite side-plates with non-aligned second phases, most likely upper bainite that nucleated and grew in a direction transverse to the plane of observation.

Figure 4.8.10 is a high-resolution SEM micrograph of the heat-affected zone of a low heat input weld (0.5 kJ/mm) deposited without preheat. Upper bainite is highlighted by boxes 2 and 4, where side-plates with an aligned second phase are clearly visible. The constituents highlighted by boxes 1 and 3 are tentatively identified as lower bainite since a fine carbide distribution appears to be present within the grains. The etch response is also slightly different from that of the surrounding structure. This constituent could, however, also be fine intragranular upper bainite formed at a slightly lower transformation temperature than the upper
bainite nucleated on the prior austenite grain boundaries (PAGB), indicated by box 2 and the bottom arrow of box 4, during continuous cooling.

![Image](image.png)

**Figure 4.8.10** High-resolution FEG-SEM micrograph of the heat-affected zone microstructure of a weld deposited preheat-free at a heat input of 0.5 kJ/mm (welding current 140 A, welding speed 428 mm/min).

The prior austenite grain boundaries are more clearly visible within the heat-affected zones than in the weld metal. This is indicative of the displacive nature of the grain boundary transformation products in the high-temperature heat-affected zone. Allotriomorphic ferrite is the first phase to nucleate and grow on prior austenite grain boundaries (PAGB) within the weld metal via a purely diffusion-controlled transformation. Allotriomorphic ferrite tends to grow into the austenite grains where secondary transformation products subsequently nucleate, either sympathetically on the ferrite/austenite interface or heterogeneously within the grains. Displacive transformations have a strong shear component, involving the coordinated movement of atoms that cannot cross grain boundaries in the same way the atoms in a diffusion-controlled transformation can, and hence the PAGB is much more visible within the HAZ than the weld metal [2].

Contributing to this effect is the fact that the weld metal is fully austenitic before the transformation products form on continuous cooling. The HAZ, on the other hand, is re-austenitised before the transformation products form on cooling. The extent to which the HAZ is re-austenitised and the level of atomic redistribution vary with the maximum temperature reached and the time spent within the austenite temperature range on heating.

Figure 4.8.11 shows the HAZ microstructure of a low heat input (0.5 kJ/mm) weld. Upper and lower bainite are the only two apparent phases present. Boxes 1 and 3 indicate lower bainite containing both fine intragranular and interlath carbides.
Figures 4.8.11 High-resolution FEG-SEM micrograph of the heat-affected zone microstructure of a weld deposited preheat-free at a heat input of 0.5 kJ/mm (welding current 140 A, welding speed 428 mm/min).

Figures 4.8.12 and 4.8.13 show weld metal microstructures of a weld deposited preheat-free at a heat input of 1.0 kJ/mm. These micrographs are highly similar in appearance, but were selected for the discussion to show the uniformity of the weld microstructures.

Allotriomorphic ferrite is indicated by boxes 3 and 6 in Figure 4.8.12, and box 4 of Figure 4.8.13. Both figures show the presence of Widmanstätten ferrite, with primary grain boundary Widmanstätten ferrite indicated by the bottom two arrows of box 6 in Figure 4.8.12 and the bottom arrow of box 1 in Figure 4.8.13. Secondary grain boundary Widmanstätten ferrite, nucleated on allotriomorphic ferrite veins, is indicated by box 3, the top arrow of box 6 in Figure 4.8.12, and the middle arrow of box 1 in Figure 4.8.13. Acicular ferrite is present in both micrographs, indicated by box 5 in both figures, yet appears much coarser than the acicular ferrite observed in the low heat input weld metal microstructures shown in Figures 4.8.1 and 4.8.2. Ferrite side-plates, with both aligned and non-aligned second phases, (which could be either Widmanstätten ferrite or upper bainite intersected transversely) are also present throughout the microstructures.
Figure 4.8.12 Optical micrograph of a weld metal microstructure deposited preheat-free at a heat input of 1.0 kJ/mm (welding current 150 A, welding speed 234 mm/min).

Figure 4.8.13 Optical micrograph of a weld metal microstructure deposited preheat-free at a heat input of 1.0 kJ/mm (welding current 150 A, welding speed 234 mm/min).

Figures 4.8.14 to 4.8.17 are high-resolution SEM micrographs of high heat input weld metal deposited without preheat. Since, in the weld metal, the approximate location of the prior austenite grain boundaries could only be inferred by the relative size and shape of the allotriomorphic ferrite grains and, in some cases,
the primary grain boundary Widmanstätten ferrite packets, the PAGB’s are not always clear at the magnifications used in the SEM images. This complicates the identification of grain boundary and intragranular Widmanstätten ferrite. The focus of the current study, however, does not deem this distinction as important since it does not change the hardness or the resistance to WMHACC of the weld metal in any significant way.

![Image](image.jpg)

**Figure 4.8.14** High-resolution FEG-SEM micrograph of a weld metal microstructure deposited preheat-free at a heat input of 0.8 kJ/mm (welding current 130 A, welding speed 243 mm/min).

Box 1 in Figure 4.8.14 shows coarse grains of ferrite, larger than approximately three times the thickness of the surrounding transformation products and therefore assumed to be allotriomorphic ferrite nucleated on the PAGB. Box 2 shows a ferrite grain designated as idiomorphic ferrite. It is possible that this grain is a part of the allotriomorphic ferrite grain indicated by box 1, yet it appears to have nucleated on an intragranular inclusion visible at the top right of the grain. If this is the case, it would more accurately be designated as idiomorphic rather than allotriomorphic ferrite.

Widmanstätten ferrite is also present in Figures 4.8.14 to 4.8.16, and box 5 in Figure 4.8.15 shows primary Widmanstätten ferrite nucleated on an intragranular non-metallic inclusion. The ferrite-carbide aggregate previously discussed as being either degenerate pearlite or granular bainite in the low heat input welds can also be seen in the high heat input weld metal shown in Figures 4.8.14 and 4.8.16. Microphases associated with this type of transformation product resembles pearlite as can be seen in box 3 of Figure 4.8.14 and box 7 in Figure 4.8.16. These microphases are observed in the intragranular regions between ferrite side-plates, or as an aligned second phase due to carbon enrichment in these areas. The pearlitic nature of this second phase is not typical of lower bainite, which would be more reminiscent of thin cementite laths between sheaths or grains [1].
Figure 4.8.15 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited preheat-free at a heat input of 1.0 kJ/mm (welding current 120 A, welding speed 173 mm/min).

As shown by box 3 in Figure 4.8.16, the microstructure contains plates resembling acicular ferrite, but with both a fine intragranular distribution of carbides and an aligned carbide phase on the boundaries. It is similar in appearance to the plates indicated by box 3 in Figure 4.8.7 and could be a form of lower bainite. Carbide-free acicular ferrite is also present in Figure 4.8.16, as indicated by box 2. Figure 4.8.17 shows a higher magnification of the fine microphase resembling pearlite observed in weld metal deposited preheat-free at a heat input of 1.0 kJ/mm.
Figure 4.8.16 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited preheat-free at a heat input of 1.0 kJ/mm heat input (welding current 120 A, welding speed 173 mm/min).

Figure 4.8.17 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited preheat-free at a heat input of 1.0 kJ/mm heat input (welding current 120 A, welding speed 173 mm/min).
Figures 4.8.18 and 4.8.19 are optical photomicrographs of the high-temperature heat-affected zones of a weld deposited without preheat at a heat input of 1.0 kJ/mm. Clear prior austenite grain boundaries (PAGB) are distinguishable. Boxes 1 and 5 in Figure 4.8.19 show what appears to be ferrite with a fine carbide distribution, which could possibly be a form of granular lower bainite. Boxes 2 and 6 in Figure 4.8.18 show ferrite side-plates with an aligned second phase resembling upper bainite, and boxes 3 and 4 show side-plates with fine islands of carbides also assumed to be upper bainite sectioned transversely across the plates, designated as ferrite side-plates with non-aligned second phase, FS(NA).

![Figure 4.8.18 Optical micrograph of the heat-affected zone microstructure of a weld deposited preheat-free at a heat input of 1.0 kJ/mm heat input (welding current 120 A, welding speed 173 mm/min).](image)

Figure 4.8.19 shows the approximate location of the fusion line of the weld as a dotted line with the weld metal to the right of the line and the high-temperature heat-affected zone to the left. Also evident in this figure is a weld metal hydrogen-assisted cold crack that propagated through the weld metal. The direction of crack propagation does not appear to be determined by the metallurgical phases present, with the crack propagating through acicular ferrite and ferrite side-plates.

Ferrite grains are evident in the heat-affected zone, designated as primary ferrite (PF) in box 1, with bainite colonies that transformed on cooling from the remaining austenite. Upper bainite is evident, designated as FS(NA) and FS(UB) in boxes 2 and 3, respectively. Box 4 highlights ferrite with a fine carbide distribution, similar to box 1 in Figure 4.8.18, which could be a form of granular lower bainite. This is in agreement with the findings of Beidokhti et al. [13], who reported the presence of bainite and primary ferrite within the high-temperature heat-affected zones of welded X70.
Figure 4.8.19 Optical micrograph of the heat-affected zone microstructure of a weld deposited preheat-free at a heat input of 1.0 kJ/mm (welding current 120 A, welding speed 173 mm/min).

Figures 4.20 and 4.21 are high-resolution SEM micrographs of the high-temperature heat-affected zones of welds deposited without preheat at a heat input of 1.0 kJ/mm. These micrographs show upper bainite of various morphologies (due to alignment of the ferrite side plates relative to the plane of observation) and side-plates of different thicknesses.

The second arrow in box 1 of Figure 4.8.20 shows ferrite side-plates with aligned second phase with a much finer structure than that indicated by the first and fourth arrow originating from the same box. The bainite highlighted by the top arrow most likely nucleated first on the austenite grain boundary and grew into the top austenite grain, and hence reduced the energy available for further nucleation from the same boundary. The finer bainite structure indicated by the second arrow appears to have nucleated later at a lower transformation temperature and is, as a result, much finer. It is also possible for these finer bainite plates to have nucleated sympathetically on the ferrite side-plate beneath it, since the grain boundary energy from the boundary above it has been lowered by the earlier nucleation event.

Figure 4.8.21 shows upper bainite as well as possible lower bainite indicated by box 3. There appears to be a fine carbide distribution within the ferrite sheaths coupled with aligned cementite on the boundaries. Towards the centre of the micrograph there appears to be large ferrite grains with a second phase on the boundaries resulting from carbon redistribution, designated as primary ferrite (PF) by box 4.
**Figure 4.8.20** High-resolution FEG-SEM micrograph of the heat-affected zone microstructure of a weld deposited preheat-free at a heat input of 1.0 kJ/mm.

**Figure 4.8.21** High-resolution FEG-SEM micrograph of the heat-affected zone microstructure of a weld deposited preheat-free at a heat input of 1.0 kJ/mm (welding current 140 A, welding speed 214 mm/min).

Figures 4.8.22 (a) to (f) compare the relative coarseness of the microstructures of welds deposited without preheat at low and high heat inputs. As evident from Figures 4.8.1 to 4.8.21, the phases present in the weld
metal and heat-affected zones of welds deposited at high heat input do not differ significantly from those in low heat input welds. In all the microstructures, primary allotriomorphic ferrite was present as the first phase to form from the austenite in the weld metal on cooling. Other transformation products such as idiomorphic ferrite, ferrite side-plates with an aligned second phase (Widmanstätten ferrite or upper bainite), acicular ferrite, possible lower bainite and ferrite-carbide aggregates resembling either granular lower bainite or degenerate pearlite, were also observed throughout the entire heat input range for welds deposited without preheat. The high-temperature heat-affected zone microstructures did not show much variation either, with upper bainite dominating and to a lesser extent, possible lower bainite, ferrite-carbide aggregate and primary ferrite.

The micrographs in Figures 4.8.22(a) and (b) show a comparison between the relative coarseness of the prior austenite grain boundaries (indicated by dotted lines) in the weld metal (as far as their location could be determined at the magnification used) for a low heat input weld and a high heat input weld. As expected, the weld deposited at a heat input of 1.0 kJ/mm shows a coarser columnar prior austenite grain structure than the weld deposited at 0.4 kJ/mm.

According to Bhadeshia et al. [2], larger prior austenite grains promote intragranular nucleation on non-metallic inclusions (and therefore acicular ferrite), whereas smaller prior austenite grains favour grain boundary nucleation of transformation products with subsequent growth into grain interiors. The amount of allotriomorphic ferrite also shifts the nucleation behaviour of transformation products, with grain boundary nucleation of primary ferrite consuming the available grain boundary energy necessary for second phase nucleation.

It is therefore expected that with an increase in prior austenite grain size (with increasing heat input), as shown in Figure 4.8.22, the relative amounts of the transformation products present will change, even though no change in the types of phases were observed. At higher heat input levels, nucleation with subsequent growth and thickening of allotriomorphic ferrite veins along prior austenite grain boundaries is expected to alter the amount of acicular ferrite observed within the grains. A significant increase in the amount of acicular ferrite was not obvious from the micrographs since the acicular ferrite plate thickness is also affected by the heat input, as can be seen in Figures 4.8.22(c) and (d).

There is a noticeable increase in the coarseness of not only the prior austenite grain boundaries in this study, but also of the transformation products in the weld metal. This is also evident in the micrographs of the high-temperature heat-affected zones shown in Figures 4.8.23(e) and (f).
Figure 4.8.22 Optical micrographs comparing the relative coarseness of phases and the prior austenite grain size in the weld metal and heat-affected zones of welds deposited without preheat at heat inputs of 0.4 kJ/mm (a), (c), (d), and 1.0 kJ/mm (b), (d), and (f).
4.8.2 Microstructures of welds deposited at a preheat temperature of 80°C

Figure 4.8.23 is an optical micrograph of the weld metal of a weld deposited at a heat input of 0.4 kJ/mm and a preheat temperature of 80°C. There is no apparent change in the types of observable transformation product present when compared to micrographs for low heat input welds deposited without any preheat, as shown in Figures 4.8.1 and 4.8.2.

![Optical micrograph of weld metal microstructure](image)

**Figure 4.8.23** Optical micrograph of a weld metal microstructure deposited at a heat input of 0.4 kJ/mm (welding current 150 A, welding speed 607 mm/min) and a preheat temperature of 80°C.

Boxes 2 and 8 indicate the presence of allotriomorphic ferrite nucleated on prior austenite grain boundaries with ferrite side-plates nucleated on the thin allotriomorphic veins indicated by box 7. These side-plates are assumed to contain a non-aligned second phase due to the fragmented appearance of the second phase constituent. This transformation product is most likely secondary grain boundary Widmanstätten ferrite intersected at an odd angle. Box 3 indicates small packets of ferrite side-plates with an aligned second phase, while box 5 shows small packets of side-plates with non-aligned second phase. These side-plates apparently nucleated both on grain boundaries and intragranularly. The likelihood exists that at least some of these side-plate colonies are upper bainite, although using the lower hardnesses reported in Section 4.7 for preheated weld metal as a comparative guide, less bainite is expected to be present in Figure 4.8.23 than in Figures 4.8.1 and 4.8.2. Acicular ferrite is also present in the intragranular regions as indicated by boxes 1 and 9, as are needles of intragranular Widmanstätten ferrite (box 4).

Figures 4.8.24 to 4.8.28 are high-resolution SEM micrographs of the same weld metal (i.e. deposited at a heat input of 0.4 kJ/mm and preheat temperature of 80°C). Shown in Figure 4.8.24 is a low magnification micrograph taken in the vicinity of a prior austenite grain boundary (PAGB) (indicated by box 7). It shows primary grain boundary Widmanstätten ferrite, indicated by box 5, that nucleated at the austenite grain boundary, and ferrite side-plates with both aligned (boxes 3 and 8) and non-aligned (box 1) second phases,
which may be small colonies of intragranular Widmanstätten ferrite or upper bainite. The remainder of the grains are occupied by intragranularly nucleated transformation products such as acicular ferrite (box 6), Widmanstätten ferrite needles (box 2) and idiomorphic ferrite (box 4). Microphases are present between plates of intragranular transformation products, but the magnification is too low to identify these phases.

Figure 4.8.24 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 0.4 kJ/mm (welding current 140 A, welding speed 536 mm/min) and a preheat temperature of 80°C.

The micrograph in Figure 4.8.25 shows a portion of the fusion line (FL), with its approximate location highlighted, at low magnification. The high-temperature heat-affected zone is evident to the right of the dotted line and the weld metal to the left.

The weld metal microstructure consists mainly of acicular ferrite, with a much finer structure closer to the fusion line where the cooling rate in the weld metal was most rapid. Coarser acicular ferrite is evident further away from the fusion line where the cooling rate was slower. Also evident is an allotriomorphic ferrite vein along a prior austenite grain boundary, indicated by box 5. Secondary grain boundary Widmanstätten ferrite needles extend from the allotriomorphic ferrite vein, indicated by box 6. Also evident in the weld metal microstructure are intragranular Widmanstätten ferrite needles, indicated by box 2, with aspect ratios in the order of approximately 10:1.

In the heat-affected zone to the right of the dotted line, coarse ferrite side-plates with an aligned second phase resembling upper bainite are evident closer to the fusion line (indicated by box 7). Box 8 shows ferrite side-plates with a more shaded etch effect, which could be due to the presence of intragranular carbides in addition to interlath carbides. This may be lower bainite, but the presence of this constituent could not be conclusively confirmed in the absence of other imaging techniques.
Figure 4.8.25 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 0.4 kJ/mm (welding current 140 A, welding speed 536 mm/min) and a preheat temperature of 80°C.

Figure 4.8.26 is a high magnification photomicrograph of the weld metal, showing the typical transformation products discussed thus far. Ferrite side-plates predominate in the microstructure, which could be either Widmanstätten ferrite needles or upper bainite. The side-plates indicated by boxes 3 and 8 are most likely Widmanstätten ferrite due to their high aspect ratios and the phases observed between the needles. Carbide, acicular ferrite (indicated by the bottom two arrows of box 6) and idiomorphic ferrite (indicated by box 2) are present between packets of needles, whereas the side-plates indicated by boxes 4 and 5 have a distinctly aligned second phase, most likely cementite, and could possibly be upper bainite.

Acicular ferrite, nucleated on NMI’s is also evident, indicated by box 1 and the top arrow of box 6. Idiomorphic ferrite is indicated by the top two arrows of box 2, although the ferrite indicated by the top arrow could possibly be allotriomorphic ferrite since it is only partially within the view of the micrograph and appears to be larger than 3 times the size of the other intragranular transformation products.

Figure 4.8.27 shows a higher magnification of the area across a prior austenite grain boundary. Box 7 shows a thin allotriomorphic vein that nucleated on the austenite grain boundary, with secondary grain boundary Widmanstätten ferrite indicated by box 8. Acicular ferrite is indicated by box 1 and 9, while the ferrite side-plates with aligned second phase, indicated by box 3, are probably also Widmanstätten ferrite.
Figure 4.8.26 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 0.4 kJ/mm (welding current 140 A, welding speed 536 mm/min) and a preheat temperature of 80°C.

Figure 4.8.27 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 0.4 kJ/mm (welding current 140 A, welding speed 536 mm/min) and a preheat temperature of 80°C.

The top arrow of box 4 in Figure 4.8.27 shows a ferrite-carbide aggregate, similar in appearance to the ferrite-carbide aggregate observed in the SEM micrographs of welds deposited at room temperature. As
stated before, this phase could be either granular lower bainite or a form of degenerate pearlite. There appears to be an increase in the amount of this transformation product in the weld metal as the Δt/5 cooling time increases (in other words, as the cooling rate decreases). There is also a corresponding decrease in the weld metal hardness as the cooling rate increases, as shown in Figure 4.7.11, which could be due to an increase in the amount of softer transformation products and general coarsening of the microstructure.

All of the arguments stated above indicate that the transformation product in question is more likely to be a form of degenerate pearlite than lower bainite. A transformation product with a similar appearance is observed throughout the microstructure. Degenerate pearlite is indicated by the top arrow of box 4 of Figure 4.8.27. The grains appear to be semi-polygonal and are also seen in Figure 4.8.28, indicated by box 8 and the bottom two arrows of box 2. The general coarseness of these structures suggests that they formed at reasonably high transformation temperatures, again suggesting that it may be a form of pearlite, rather than bainite.

Another ferrite-carbide aggregate, similar in appearance to acicular ferrite, is indicated by the bottom arrow of box 4 in Figure 4.8.27 and in Figure 4.8.28 by box 7 and the top arrow of box 2. In this case, it is assumed to be a form of lower bainite, since acicular ferrite shares a common transformation temperature range with bainite, as discussed before.

![Figure 4.8.28](image_url)

**Figure 4.8.28** High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 0.4 kJ/mm (welding current 140 A, welding speed 536 mm/min) and a preheat temperature of 80°C.

Figure 4.8.29 shows an optical micrograph of the high-temperature heat-affected zone of a low heat input (0.4 kJ/mm) weld deposited at a preheat temperature of 80°C. The black dotted line indicates the approximate location of the fusion line between the weld metal (to the right of the line) and the heat-affected zone (to the left of the line). There is some primary ferrite visible in the high-temperature heat-affected zone,
as indicated by boxes 3 and 4, although the primary ferrite indicated by box 4 appears to have grown epitaxially from allotriomorphic ferrite veins in the weld metal.

![Figure 4.8.29 Optical micrograph of a heat-affected zone microstructure deposited at a heat input of 0.4 kJ/mm (welding current 150 A, welding speed 607 mm/min) and a preheat temperature of 80°C.](image)

The remainder of the high-temperature heat-affected zone appears to consist mainly of upper and lower bainite. Upper bainite is indicated by boxes 1 and 2 as ferrite side-plates with an aligned second phase FS(UB), and as ferrite side-plates with a non-aligned second phase FS(NA) by boxes 4 and 6. Lower bainite is assumed to be present and indicated by boxes 2 and 8 since these constituents show a darker etch response.

Figure 4.8.30 shows a high-resolution SEM micrograph of the high-temperature heat-affected zone, with similar phases present. Boxes 1 and 4 indicate upper bainite as ferrite side-plates with an aligned second phase. Lower bainite is identified by its finer structure and different etch response and is indicated by boxes 2 and 5. Box 3 highlights islands of primary ferrite.

Figure 4.8.31 is an optical micrograph of the weld metal of a weld deposited at a heat input of 0.8 kJ/mm and a preheat temperature of 80°C. It shows similar transformation products as seen in earlier microstructures. Boxes 4 and 6 indicate allotriomorphic ferrite veins nucleated on prior austenite grain boundaries and box 3 highlights secondary Widmanstätten ferrite that nucleated from the allotriomorphic ferrite indicated by the top arrow of box 4. Large islands of idiomorphic ferrite are also visible and indicated by boxes 2 and 5. These intragranular ferrite islands are much coarser than the idiomorphic ferrite observed in the weld metal microstructures of preheat-free welds due the slower cooling rate resulting from preheating. Acicular ferrite is also present, indicated by boxes 1 and 7, although much coarser and thicker than the acicular ferrite observed previously in preheat-free welds.
Figure 4.8.30 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 0.4 kJ/mm (welding current 140 A, welding speed 536 mm/min) and a preheat temperature of 80°C.

Figure 4.8.31 Optical micrograph of a weld metal microstructure deposited at a heat input of 0.8 kJ/mm (welding current 130 A, welding speed 243 mm/min) and a preheat temperature of 80°C.

Figures 4.8.32 and 4.8.33 are high-resolution SEM micrographs of weld metal deposited at a heat input of 0.7 kJ/mm and preheated to 80°C. These images highlight the prolific nature of ferrite-carbide aggregates in
welds deposited at higher heat inputs and preheat temperatures. Boxes 2 and 6 in Figure 4.8.32 and box 5 in Figure 4.8.33 highlight these FC transformation products, but it is clear that these constituents are widespread in the microstructures. In Figure 4.8.33, the middle arrow of box 7 indicates ferrite side-plates with an aligned second phase. The interiors of the ferrite side-plates show evidence of a fine carbide distribution. A similar constituent is highlighted by both arrows of box 1, appearing to be side-plates with a fine carbide distribution. The top two arrows of box 5 in the same figure point towards what appears to be allotriomorphic ferrite, but this constituent shows the same carbide distribution within the grains. Allotriomorphic, idiomorphic and acicular ferrite are also present in the two micrographs.

**Figure 4.8.32** High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 0.7 kJ/mm (welding current 120 A, welding speed 257 mm/min) and a preheat temperature of 80°C.

Figures 4.8.34 and 4.8.35 are high-resolution SEM micrographs coupled with EDS carbon maps of selected areas of the microstructures of weld metal deposited at a preheat temperature of 80°C and heat inputs of 0.7 kJ/mm and 1.0 kJ/mm, respectively. In Figure 4.8.34, carbon enrichment is evident within the granular structure in the centre of the image, as well as along the boundaries separating different transformation products. The granular transformation product containing a fine carbide distribution below a vein of allotriomorphic ferrite and is relatively coarse compared to the surrounding transformation products. This suggests that it formed at a comparatively higher transformation temperature than expected for any form of lower bainite, and with the carbides arranged in regular arrays, is assumed to be degenerate pearlite.

In Figure 4.8.35, carbon enrichment is again evident in the upper right-hand corner of the image where a fine carbide distribution is present within the microstructure. This figure shows secondary Widmanstätten ferrite that nucleated on the allotriomorphic ferrite veins covering the prior austenite grain boundaries. The microphases in both figures also show carbon enrichment, as expected, since these phases formed as a result of carbon redistribution into the remaining austenite between transformation products on cooling.
Figure 4.8.33 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 0.7 kJ/mm (welding current 140 A, welding speed 306 mm/min) and a preheat temperature of 80°C.

Figure 4.8.34 (a) High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 0.7 kJ/mm (welding current 140 A, welding speed 306 mm/min) and a preheat temperature of 80°C, and (b) EDS map of the carbon distribution in the same microstructure.
Figure 4.8.35 (a) High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 1.0 kJ/mm (welding current 120 A, welding speed 173 mm/min) and a preheat temperature of 80°C, and (b) EDS map of the carbon distribution in the same microstructure.

Figure 4.8.36 is a high-resolution SEM micrograph of weld metal preheated to 80°C and deposited at a heat input of 1.0 kJ/mm. It shows the typical transformation products seen throughout the weld metal microstructures observed thus far, i.e. allotriomorphic ferrite (box 6), idiomorphic ferrite (box 5), both grain boundary (box 4) and intragranular (box 9) Widmanstätten ferrite, acicular ferrite (box 11) and a scattering of microphases (box 1).

Figure 4.8.36 High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 1.0 kJ/mm (welding current 120 A, welding speed 173 mm/min) and a preheat temperature of 80°C.
Also evident in Figure 4.8.36 is the presence of finely distributed carbides scattered through some of the transformation products. The bottom arrow of box 8 indicates a granular structure with what appears to be a fine distribution of carbides in regular arrays, whereas the rest of the microstructure seems to have a fine irregular scattering of carbides. Box 7 indicates the presence of ferrite side-plates with an aligned second phase, also showing fine carbides in the interior of the side-plates. The top two arrows of box 2 appear to indicate allotriomorphic ferrite veins and box 12 indicates intragranular Widmanstätten ferrite needles also containing such carbides. This can be attributed to incomplete carbon redistribution during the formation of these transformation products, and subsequent carbide precipitation.

Figure 4.8.37 shows a high magnification micrograph of a pearlitic microphase observed between idiomorphic ferrite islands, ferrite side-plates and acicular ferrite. A distinct lamellar structure is indicated by box 4, while the bottom arrow of box 1 indicates a structure more reminiscent of degenerate pearlite.

![Micrograph](image)

**Figure 4.8.37** High-resolution FEG-SEM micrograph of a weld metal microstructure deposited at a heat input of 1.0 kJ/mm (welding current 120 A, welding speed 173 mm/min) and a preheat temperature of 80°C.

Figures 4.8.38 and 4.8.39 are optical photomicrographs showing the high-temperature heat-affected zones of welds deposited at heat input levels of 0.6 kJ/mm and 0.8 kJ/mm, respectively, and preheated to a temperature of 80°C. The transformation products in the high-temperature heat-affected zones do not appear to change significantly with an increase in preheat temperature, both at low and high heat input levels.

In the images below, ferrite side-plates with aligned second phase predominate. This constituent is most likely upper bainite, interspersed with small colonies of lower bainite. The lower bainite indicated in Figures 4.8.38 and 4.8.39 consists of ferrite plates or sheaths with a slightly darker etch response.

Figure 4.8.39 also shows small islands of primary ferrite, indicated by box 7. Throughout this study, an increase in the primary ferrite content of the high-temperature heat-affected zone was observed with a
decrease in cooling rate. This is in agreement with the gradual decrease in hardness observed in the high-temperature heat-affected zone with an increase in heat input (shown in Figure 4.7.7).

![Image](image-url)

**Figure 4.8.38** Optical micrograph of the heat-affected zone microstructure of a weld deposited at a heat input of 0.6 kJ/mm (welding current 120 A, welding speed 300 mm/min) and a preheat temperature of 80°C.

High-resolution scanning electron micrographs of the high-temperature heat-affected zones of welds preheated to 80°C and deposited at heat inputs of 0.7 kJ/mm and 1.0 kJ/mm, respectively, are shown in Figures 4.8.40 and 4.8.41. Similar phases to those observed in Figures 4.8.38 and 4.8.39 are evident. Lower bainite, indicated by box 2 in Figure 4.8.40, contains both a fine carbide distribution within the ferrite plates and thicker carbides between the plates. In addition to lower bainite, the heat-affected zones also contain upper bainite and primary ferrite.

The lower magnification photomicrograph in Figure 4.8.41 shows the high-temperature heat-affected zone after welding at a high heat input (1.0 kJ/mm) and a preheat of 80°C. Upper bainite predominates, with islands of primary ferrite scattered through the grains. Lower bainite is not readily apparent. It is therefore tentatively identified by box 7, although the magnification is too low to unambiguously verify its presence.
Figure 4.8.39 Optical micrograph of the heat-affected zone microstructure of a weld deposited at a heat input of 0.8 kJ/mm (welding current 120 A, welding speed 225 mm/min) and a preheat temperature of 80°C.

Figure 4.8.40 High-resolution FEG-SEM micrograph of the heat-affected zone microstructure of a weld deposited at a heat input of 0.7 kJ/mm (welding current 140 A, welding speed 306 mm/min) and a preheat temperature of 80°C.
Figure 4.8.41 High-resolution FEG-SEM micrograph of the heat-affected zone microstructure of a weld deposited at a heat input of 1.0 kJ/mm (welding current 120 A, welding speed 173 mm/min) and a preheat temperature of 80°C.

Figure 4.8.42 displays a higher magnification of the central part of the image presented in Figure 4.8.41, confirming the presence of upper bainite as distinct ferrite side-plates with an aligned second phase (boxes 2 and 5). Box 1 identifies upper bainite as ferrite side-plates with a non-aligned second phase. Primary ferrite is also present, as indicated by box 3.

Figure 4.8.43 contains a collection of optical photomicrographs taken at low magnification illustrating changes in the relative coarseness of the weld metal and high-temperature heat-affected zone microstructures for low heat input welds (micrographs on the left of Figure 4.8.43) and high heat input welds (micrographs on the right of Figure 4.8.43). In the weld metal micrographs, shown in Figures 4.8.43(a) to (d), the approximate positions of the prior austenite grain boundaries are indicated by white dotted lines.

Figures 4.8.43(a) and (b) show the microstructures of weld metal deposited without preheat at heat inputs of 0.4 kJ/mm and 1.0 kJ/mm, respectively. Those in (c) and (d) show weld metal microstructures deposited at the same heat input levels, but preheated to 80°C. The effect of heat input on the prior austenite grain size can be seen when the micrographs are viewed from left to right, and the effect of preheating becomes evident when the micrographs are viewed from top to bottom.

The micrographs in Figures 4.8.43(e) and (f) show the microstructures of the high-temperature heat-affected zones of welds deposited without preheat at heat inputs of 0.4 kJ/mm and 1.0 kJ/mm, respectively, while (g) and (h) show the same heat-affected zone microstructures deposited at a preheat temperature of 80°C.
Figure 4.8.42 High-resolution FEG-SEM micrograph of the heat-affected zone microstructure of a weld deposited at a heat input of 1.0 kJ/mm (welding current 120 A, welding speed 173 mm/min) and a preheat temperature of 80°C.

(a) WM (preheat-free) – 0.4 kJ/mm
(b) WM (preheat-free) – 1.0 kJ/mm
(c) WM (80°C) – 0.4 kJ/mm
(d) WM (80°C) – 1.0 kJ/mm
Figure 4.8.43 Optical photomicrographs comparing the relative coarseness of phases and prior austenite grains in the weld metal and heat-affected zones of welds deposited at different heat inputs and preheats. Microstructures deposited without preheat are (a) weld metal at a heat input of 0.4 kJ/mm, (b) weld metal at a heat input of 1.0 kJ/mm, (e) heat-affected zone at 0.4 kJ/mm, and (f) heat-affected zone at 1.0 kJ/mm. Microstructures deposited at a preheat temperature of 80°C are (c) weld metal at a heat input of 0.4 kJ/mm, (d) weld metal at 1.0 kJ/mm, (g) heat-affected zone at 0.4 kJ/mm, and (h) heat-affected zone at 1.0 kJ/mm.

As expected, coarsening of the prior austenite grain size becomes more pronounced with the application of preheat. Even though these micrographs were taken at low magnifications and the transformation products within the prior austenite grains are not fully resolved, there is an indication that higher heat inputs and preheat temperatures also coarsen the transformation products.

The results shown in this section suggest that the same transformation products are present in microstructures deposited at low and high heat input levels, and that the application of a low temperature preheat of up to 80°C does not alter the type of transformation products formed. As the weld cooling rate increases with increasing heat input and preheat temperature, the hardnesses of the weld metal were shown to decrease steadily (Figure 4.7.7). This can be ascribed to a general coarsening of the microstructure, as seen in Figure 4.8.43, and possibly a gradual shift in the balance of the phases observed from more displacive transformation products governed by shear transformation mechanisms to reconstructive transformation products governed by diffusional processes.
The type of transformation product formed plays an important role in determining resistance to hydrogen-assisted cold cracking. Transformation products need to be considered on the basis of hardness, toughness and efficacy of hydrogen trapping. It has been demonstrated [12,14] that a low hardness is not a sufficient guarantee for resistance to hydrogen-induced cracking (HIC) since microstructures, previously regarded as HIC resistant, can become susceptible to cracking in the presence of excessive amounts of diffusible hydrogen.

Park et al. [15] studied the hydrogen trapping efficiencies of different microstructural constituents in low-carbon pipeline steel with emphasis on diffusivity, permeability, hydrogen solubility and the relative sensitivities of these constituents to hydrogen-induced cracking. They found the lowest trapping efficiency in microstructures consisting of degenerate pearlite/ferrite, an increased efficiency in bainite/ferrite structures, and maximum trapping efficiency in acicular ferrite. Li et al. [16] confirmed acicular ferrite to be the most effective microstructural constituent for hydrogen trapping in low-carbon steels.

These authors [15,16] demonstrated, however, that even though acicular ferrite displays the slowest diffusion rates and permeability for hydrogen (i.e. the highest trapping efficiency), bainite/ferrite microstructures still showed the least resistance to HIC due to the comparatively high toughness of acicular ferrite, even in the presence of diffusible hydrogen. On the other hand, Costin et al. [17] showed that the apparent toughness of acicular ferrite is reduced significantly in the presence of diffusible hydrogen in the weld metal of X70 pipeline steel welded with E6010 electrodes.

In the current investigation, cracking was observed over the entire heat input range (and hence hardness range) evaluated, with welds deposited at low welding currents being most susceptible to WMHACC (as shown in Figures 4.3.3 to 4.3.6). Crack initiation was found to be a strong function of the weld bead geometry and the cooling rate. Crack propagation was observed regardless of the transformation products formed. Acicular ferrite played a possible role in mitigating crack propagation and may have facilitated crack arrest in some of the welds. The propagation paths of large macrocracks were, however, not observed to be sensitive to the transformation products present.

Microcracks were observed near the crack tips of arrested macrocracks, and propagated through the carbon-enriched microphases and cementite aligned with and surrounding the ferritic transformation products. Such microcrack propagation can be seen in Figure 4.8.44, which is a collection of high-resolution SEM micrographs. The images on the left are micrographs taken in secondary electron (SE) imaging mode and those on the right in back scattered electron (BSE) imaging mode to reveal the contrast between cracks and the underlying microstructure more clearly.

From these images it can be seen that hard, brittle carbon-enriched phases and continuous cementite seams facilitate crack propagation more readily than ferritic transformation products. Mohtadi-Bonab et al. [18] demonstrated that hydrogen-induced cracks propagate along similarly hard and brittle phases resulting from carbon redistribution in X70 pipeline steel due to the lower fracture toughness of these regions, and confirmed that such cracks can easily propagate between non-metallic inclusions, as discussed in the next section of this study.

It is not only the lower toughness of continuous cementite seams, such as those found in the aligned phases in the current study, that favours WMHACC crack propagation. Du et al. [19] showed that elongated and banded cementite structures showed sharp variations in hydrogen diffusivity depending on the direction of measurement. They showed that the interfaces of such elongated or banded cementite structures act as
effective hydrogen sinks so that diffusion in a direction longitudinal to the band showed a higher hydrogen diffusion rate than directions perpendicular to the banded cementite. Such could be the case in the current study, where cementite oriented horizontally with respect to the surface of the weld where hydrogen effusion from the weld metal is expected, or aligned second phases with a large volume fraction of cementite surrounding a ferritic transformation product, effectively trap hydrogen and enhance crack propagation.

**Figure 4.8.44** High-resolution FEG-SEM (SEI and BSE) images of microcrack propagation along hard carbon-enriched phases and cementite aligned with and surrounding ferritic transformation products in the weld metal of a weld deposited preheat-free at a heat input of 1.0 kJ/mm (welding current 120 A, welding speed 173 mm/min).
It is widely recognised that finer microstructures exhibit higher toughness than coarse microstructures. As shown in Figure 4.8.43, there is definite coarsening of the microstructures with a reduction in the cooling rate after welding. The decrease in toughness due to coarsening of the microstructure could to some extent be offset by a shift in the transformation products to more phases resulting from reconstructive transformations at slower cooling rates as opposed to phases resulting from displacive transformations at high cooling rates. This effect is evident in the hardness results presented in Section 4.7 of the current study. Displacive transformation products are also expected to exhibit a higher dislocation density, with dislocations acting as reversible hydrogen traps [20], slowing down the diffusion of hydrogen and increasing the risk of WMHACC.

Finer microstructures also have higher grain boundary areas, which accelerate hydrogen diffusion, but present more trapping sites for hydrogen at boundary nodes and junctures [21]. Mohtadi-Bonab et al. [22] used the hydrogen microprint technique and reported that the hydrogen concentration was higher in the grain boundaries than inside the grains in pipeline steels. This dual effect complicates any explanation of the effect of microstructure on WMHACC susceptibility and yet, WMHACC proliferated in weld metal with finer microstructures. If the larger grain boundary area resulted in a mean increase in the hydrogen mobility, it was also offset by the shorter time available for hydrogen to effuse out of the weld metal, hence leading to more severe WMHACC in welds exhibiting faster cooling rates.

REFERENCES


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4.9 NON-METALLIC INCLUSIONS

4.9.1 Physical attributes of non-metallic inclusions

During manual metal arc welding of steels, turbulence in the welding arc promotes oxygen contamination in the weld pool. Elements with a high affinity for oxygen, such as manganese and silicon in the case of E6010 electrodes, are added to the welding electrodes to act as deoxidisers. These elements react with the dissolved oxygen in the weld metal to form an oxide slag that covers the solidifying weld pool and shields it from further atmospheric contamination. Large inclusions drift to the surface while the weld metal is still liquid and are incorporated in the slag, especially at slower cooling rates. Smaller particles, however, may be trapped during solidification and remain within the weld metal microstructure as small particles referred to as non-metallic inclusions (NMI’s) [1,2].

NMI’s play a vital role in determining the fracture toughness of welds. Depending on their size, distribution and chemical composition, inclusions can have a dual effect on the hydrogen-induced cracking (HIC) behaviour of welds [3]. As discussed previously, these inclusions act as nucleation sites for intragranular acicular ferrite, which enhances the toughness of welds without sacrificing strength, but NMI’s may also act as both reversible and irreversible hydrogen traps. These particles can cause microvoid nucleation and coalescence during ductile fracture and cleavage-type failure during brittle fracture. The balance between these factors is extremely delicate, with the size distribution and coherency of such particles being the main determining factors. A mean particle size of about 0.4 μm is crucial for influencing the transformation products found in weld metal, while larger particles, especially in non-uniform distributions like clusters, tend to be detrimental to the toughness of welds [4].

In this section, the weld metal non-metallic inclusions are examined with reference to their influence on weld-metal hydrogen-assisted cold cracking (WMHACC) behaviour. The average inclusion size, number counts, volume percentage and their chemical compositions were analysed and are presented in the context of their relevance to the crack/no crack boundaries determined in Section 4.4 of the current study.

Figure 4.9.1 contains a collection of high-resolution SEM micrographs showing examples of the typical NMI’s observed within the weld metal microstructures. As can be seen, there are large variations in the sizes of observed NMI’s in both welds deposited without preheat, (a) to (d), and welds deposited at a preheat temperature of 80°C, (e) to (h), although most of the inclusions in each image appear to be smaller than 1 μm in diameter.

NMI’s in weld metal microstructures deposited at a preheat temperature of 50°C were not studied, as the purpose of this section is simply to compare the variations in the characteristics of NMI’s for welds deposited at the lowest and highest extremes of the preheat temperature ranges, and their possible effects on the WMHACC behaviour of these welds.

The SEM micrographs presented in Figure 4.9.1 were obtained using a high resolution FEG-SEM in secondary electron imaging mode (SEI). This imaging mode reveals surface details of the specimens, since secondary electrons are excited from a thin surface layer of the subject material through inelastic interactions between beam electrons and the surface atoms of the material. Electrons from the incident beam itself are also scattered from deeper within the material as a consequence of elastic interactions between the incident beam and the subsurface atoms of the material. These electrons are of a higher energetic state and their
scattering is a function of the atomic number, $Z$, of the atoms they interact with. When captured, back-scattered electrons can produce a higher contrast, but lower resolution, image of the microstructure. To facilitate analysis of the NMI's, images were therefore captured in both secondary (SE) and back scattered (BSE) electron imaging modes.
Figure 4.9.1 High-resolution FEG-SEM photomicrographs of non-metallic inclusions in the weld metal microstructures of welds deposited preheat-free at heat input levels of (a) 0.4 kJ/mm, (b) 0.5 kJ/mm, (c) 0.8 kJ/mm, and (d) 1.0 kJ/mm, and at a preheat temperature of 80°C for heat input levels of (e) 0.4 kJ/mm, (f) 0.4 kJ/mm, (g) 0.7 kJ/mm, and (h) 1.0 kJ/mm.

Figure 4.9.2 displays a selection of images taken in both imaging modes to highlight the differences between the two methods. The images on the left are micrographs taken in secondary electron (SE) imaging mode and those on the right were taken in back scattered (BSE) electron imaging mode. The micrographs on the right reveal the presence of NMI’s not readily visible in the micrographs on the left and, as discussed in Section 3.4 of this study, facilitated the analysis of the morphology of NMI’s using image analysis software.

Inclusions observed in this study are generally spherical in shape and Figure 4.9.2 shows that the NMI’s range widely in size and are dispersed randomly throughout the microstructure. Depending on the amount and shape of the inclusions, they can have a negative impact on the ductility and toughness of steel and can increase the risk of hydrogen-induced cracking (HIC). Inclusions of irregular shape such as elongated or needle-like particles, in addition to those with high incoherency mismatches between the inclusion surfaces and the surrounding metal matrix, are especially detrimental to toughness and by extension, the WMHACC response of the steel [2,5].

Spherical non-metallic inclusions are preferred over non-spherical inclusions, such as sharp, needle-like manganese sulphide (MnS). A spherical shape exhibits a low stress concentration factor and many authors and researchers [3,6,7] observed that the likelihood of HIC cracks initiating at spherical inclusions is low when compared to non-spherical NMI’s, on condition that the interfaces between the NMI’s and the metal matrix are coherent. Although WMHACC was not observed to initiate at NMI’s in the weld metal, these inclusions did play a role in fracture, as discussed later.

The average inclusion count per microscopic view, the inclusion size, the area covered by inclusions per microscopic view, and the volume percentage of inclusions within the weld metal were determined as explained in Section 3.4.
Figure 4.9.2 High-resolution FEG-SEM micrographs of non-metallic inclusions in the weld metal microstructures taken in both SE (left column) and BSE (right column) imaging modes for welds deposited without preheat at heat input levels of (a) and (b) 0.4 kJ/mm, (c) and (d) 0.7 kJ/mm, (e) and (f) 0.8 kJ/mm, and welds deposited at a preheat temperature of 80°C for heat input levels of (g) and (h) 0.7 kJ/mm, and (i) and (j) 1.0 kJ/mm.

Figure 4.9.3 shows the average number of inclusions per microscopic view at a magnification of 1300X. The equations for the trend lines are included in the figure, where \( y \) is the average number of inclusions observed and \( x \) is the heat input in kJ/mm. The number of inclusions observed decrease with increasing heat input for welds deposited without preheat and at a preheat of 80°C.

During solidification, non-metallic inclusions are trapped within the weld metal. The number of inclusions observed in the welds are therefore influenced by the cooling rate of the liquid weld metal before solidification. In the first instance, at the smaller undercoolings associated with slower cooling rates, the driving force for nucleation is lower and fewer nucleation sites are available for non-metallic inclusions within the liquid weld metal. Elements added for the purpose of deoxidising the liquid weld pool, such as Mn and Si, can move through the liquid over long distances towards nucleation sites when solidification of the weld metal is delayed. As the temperature decreases, the driving force for the growth of existing nuclei
becomes larger than for the nucleation of new oxide particles and fewer, but coarser, de-oxidation products (non-metallic inclusions) are observed within the weld metal with an increase in heat input [8].

![Average Inclusion Count per View](image)

**Figure 4.9.3** Average number (count) of inclusions per microscopic view at a magnification of 1300X as a function of heat input for inclusions observed in the weld metal of welds deposited without preheat and at a preheat temperature of 80°C.

Secondly, when the liquid weld metal is heated to higher temperatures with an increase in heat input, more inclusions are able to float out of the weld metal to combine with the slag layer on the surface of the weld.

In Figure 4.9.3 the position of the trend lines representing the number of observed weld metal inclusions suggests that welds deposited at a higher preheat temperatures have on average more inclusions than welds deposited preheat-free. Since preheated weld metal cools slower than non-preheated weld metal, this observation does not seem to be in agreement with the earlier discussion. This could be due to the amount of scatter in the experimental results or an artefact of the technique employed in analysing the inclusions. As shown in Figure 3.4.1(c), the threshold of the SEM micrographs were adjusted in order to increase the contrast between the matrix and the NMI’s. Depending on the etch response of the microstructures, some microstructural features, especially the more darkly etched microphases, were also outlined in black. To compensate for this, the software was set up to ignore all microstructural features (including spherical particles) with a cross-sectional area smaller than 0.01 μm². This value may have been too low, resulting in more noise being introduced for welds etched for slightly longer times.

Figure 4.9.4 shows the average size of non-metallic inclusions as a function of the heat input. There appears to be an increase in the size of the inclusions as the heat input, and hence the cooling rate, increases. As discussed before, this trend is expected as the driving force for the coarsening of existing nuclei becomes greater than for the driving force for the nucleation of new ones and existing particles tend to grow as the cooling rate decreases. The trend lines shown in Figure 4.9.4 suggest that the particle size increases with preheat, but the difference is very slight and there is convergence of the two trend lines at higher heat input levels. The amount of scatter in the measurements makes it difficult to draw meaningful conclusions on the role of preheating on the average particle size.
**Figure 4.9.4** Average inclusion size as a function of heat input for inclusions observed in the weld metal of welds deposited without preheat and at a preheat temperature of 80°C.

Figures 4.9.5 and 4.9.6 show the average total area covered by the inclusions in SEM micrographs at a magnification of 1300X and the volume percentage of inclusions as a function of heat input, respectively. Both the total area covered by inclusions and the volume percentage decrease with an increase in heat input, with preheated welds showing higher coverage and volume fractions.

**Figure 4.9.5** Total area of inclusions per microscopic view at a magnification of 1300X as a function of heat input for inclusions observed in the weld metal of welds deposited without preheat and at a preheat of 80°C.

As noted earlier, there is a large amount of scatter in the measurements shown in Figures 4.9.3 to 4.9.6. The error bars in the figures show the 95% confidence intervals for the measurements and have reasonably small variation ranges. This is due to the large number of images analysed in order to determine the average value for each measurement, yet the scatter between measurements remain large. This could be due to measurement effects or the effect of the arc force on the non-metallic inclusion content of weld metal.

The results obtained in this section are presented as a function of the heat input since heat input (and hence cooling rate) is expected to have the most significant influence on NMI formation during cooling after welding. As shown in Section 4, changes in the arc force resulting from a change in welding current affects
the flow properties and hence the geometry of the weld beads. The flow of liquid metal in the weld pool could therefore also affect the physical attributes of the non-metallic inclusions, causing additional scatter in the results. Despite the observed scatter, the general trends observed in the NMI measurements as a function of heat input remain valid. A recommendation for further work is to analyse a larger subset of samples representing the full current range at the heat inputs selected since, in the current study, trends with regards to the welding current couldn’t be established with acceptable \( R^2 \) values due to excessive scatter in the data of the samples selected.

![Total Volume (%)](image)

**Figure 4.9.6** Total area of inclusions per microscopic view at a magnification of 1300X as a function of heat input for inclusions observed in the weld metal of welds deposited without preheat and at a preheat temperature of 80°C.

It is generally agreed that the presence of non-metallic inclusions in steel reduces the ductility and toughness (when ignoring their influence on the nucleation of acicular ferrite) [5]. The higher volume fraction of inclusions measured at lower heat input levels is therefore likely to promote embrittlement of the weld metal. Non-metallic inclusions are also generally harder than steel, and when present in high volume fractions in fine dispersions, these particles may contribute to the overall hardness of the weld metal [3]. The higher number of small inclusions at low heat input levels, as shown in Figures 4.9.3 to 4.9.6, may therefore contribute to increasing the hardness and reducing the weld metal toughness. Although the higher hardness values measured in the weld metal for welds deposited at low heat levels, shown in Figure 4.7.7(a), can be attributed to the relative coarseness and types of transformation products present, a higher inclusion count and a finer NMI distribution may have contributed to the trends observed.

Non-metallic inclusions also serve as effective sites for hydrogen accumulation, referred to as hydrogen traps. Traps include precipitates, non-metallic inclusions, dislocations, grain boundaries and phase boundaries between transformation products, amongst others. Iron phases with different crystal structures can also be regarded as hydrogen traps depending on the size and nature of the interstitial lattice sites occupied by hydrogen between the iron atoms [3,9].

The ability of any microstructural feature to trap hydrogen will fall between two extremes known as reversible and irreversible trapping. The ease with which hydrogen is released during reheating or during the intended service of the steel to re-enter the steel matrix and contribute to diffusible hydrogen levels can be quantified by means of the hydrogen binding or trapping energy. As the trapping energy increases, hydrogen...
trapping becomes less reversible and ever higher temperatures are needed for the trap to release the hydrogen back into solution [3,9].

The resistance of weld metal to WMHACC is associated with the permeability coefficient, $J$, (or the ease with which hydrogen penetrates and spreads through the microstructure), the effective diffusivity, $D_{eff}$, (or the diffusion rate), and the apparent solubility, $C_{app}$, of hydrogen [7,10,11]. It has been reported [12] that, with increasing numbers of hydrogen trapping sites, associated decreases were observed in both the apparent diffusion rate and permeability coefficient, as well as an increase in the hydrogen solubility in the microstructure. Huang et al. [10] reported that the presence of inclusions slowed the hydrogen diffusivity and demonstrated that higher numbers of inclusions, areas covered, and volume percentages increased the permanent trapping efficiency of hydrogen. HIC resistance therefore tends to decrease with an increase in the number of inclusions, the area covered, and their volume fraction.

Lower heat inputs therefore not only slow down hydrogen diffusion by reducing the weld pool temperature, but also indirectly by increasing the amount, area and volume percentage of inclusions, as shown in Figures 4.9.3 to 4.9.6, which in turn increases the amount of trapped hydrogen. A lower hydrogen diffusion rate, reduced permeability and a higher solubility in the weld metal promote more extensive WMHACC at low heat input levels. This effect is especially pronounced when the geometry of the weld beads deposited at low welding currents favours WMHACC.

It was shown in Figure 4.6.9 that, as the geometry of the welds improved at higher welding current levels, lower heat inputs were required to achieve a predicted WMHACC probability of 50%. Using the equations for the trend lines shown in Figures 4.6.9 and 4.6.13, the number of inclusions, the inclusion size, the area covered by the inclusions and the volume percentage of inclusions could be calculated along this probability boundary for welds deposited without preheat and at a preheat temperature of 80°C. These calculated values are shown graphically in Figures 4.9.7 and 4.9.8.
Figure 4.9.7 Non-metallic inclusion characteristics required to ensure a 50% probability of WMHACC for welds deposited without preheat: (a) average number of inclusions counted per microscopic view at a magnification of 1300X, (b) the average size of inclusions, (c) the total area covered by inclusions within the microscopic view, and (d) the volume percentage of inclusions in the weld metal.

As discussed earlier, WMHACC resistance tends to decrease with an increase in the number of inclusions, the surface area covered by inclusions and the inclusion volume fraction. The average inclusion size required to ensure a 50% probability of cracking, shown in Figure 4.9.7(b), decreases with increasing welding current, which implies that finer distributions of inclusions can be sustained at higher welding current levels. As expected, the graphs level off at current levels above 130 A as the geometry of the welds improve. With an improvement in geometry yielding higher resistance to WMHACC, the negative effects of a slower apparent hydrogen diffusion rate, lower permeability and a higher hydrogen solubility due to hydrogen trapping by inclusions, as well as the increase in hardness and reduction of toughness brought about by the presence of inclusions are, to some extent, compensated for. The attributes of inclusions along the boundary of a 50% predicted probability of WMHACC for welds deposited at a preheat temperature of 80°C show a similar trend, although less drastic compared to welds deposited without preheat.
4.9.2 The chemistry of non-metallic inclusions

In this section, the chemical compositions of the non-metallic inclusions observed in the welds are discussed. The volume fraction of NMI’s present in weld metal is directly related to the oxygen and sulphur contents of the liquid weld metal, but also depends on those elements added to welding consumables with a high affinity for O and S such as Mn, Si, Al, Ti and Ca, amongst others [13].

Figures 4.9.9 to 4.9.12 contain collections of high-resolution SEM micrographs of inclusions with images taken in both SE and BSE modes, coupled with EDS maps of the elements present in the inclusions. These figures represent a large range of heat inputs and preheat temperatures in order to determine if the chemistry of the NMI’s remains consistent. Figures 4.9.9 to 4.9.11 are images of inclusions in weld metal deposited without preheat and at heat input levels of 0.4 kJ/mm, 0.7 kJ/mm and 0.9 kJ/mm, respectively, while the weld shown in Figure 4.9.12 was deposited at a preheat temperature of 80°C and a heat input of 1.0 kJ/mm.
Figure 4.9.9 EDS maps of elements in non-metallic inclusions observed in the weld metal of a weld deposited without preheat at a heat input of 0.4 kJ/mm.

These images indicate that the inclusions are mainly spherical oxides of the (Mn-Si-Al-Ti) type. The inclusions appear to be largely free of iron with the main constituents being Mn, Si and O, with some Ti mainly partitioned to the edges. Tables 3.1.1 and 3.1.3 in Section 3.1 showed that the sulphur contents of both the weld metal and base material were around 0.01%, which is regarded as fairly low. The weld metal is largely free of sulphides, although small percentages of sulphur were measured in a number of smaller inclusions. EDS maps for sulphur, included in Figures 4.9.9 and 4.9.12, confirm the absence of sulphur in the larger NMI’s observed.
Figure 4.9.10 EDS maps of elements in non-metallic inclusions found in the weld metal of a weld deposited without preheat at a heat input of 0.7 kJ/mm.
Figure 4.9.11 EDS maps of elements in non-metallic inclusions found in the weld metal of a weld deposited without preheat at a heat input of 0.9 kJ/mm.

As shown in Figure 4.9.9, nitrogen is also present within the inclusions, but it appears to be mostly associated with titanium. These particles were found at the edges of the oxide inclusions and are most likely nitrides of the form TiN, although it has been noted [4,14] that nitrides in steel could also assume the form of complex (Ti)(C,N) type carbonitrides. Dong et al. [14] showed that nitrides are often incorporated into the structure of oxides or grow around the outer edges of these inclusions. Ti and N are, however, also present to a smaller extent throughout the rest of the inclusion structure. Titanium shows a strong presence at the edges...
of inclusions in Figures 4.9.10 to 4.9.12, although nitrogen maps are not included in Figures 4.9.10 and 4.9.11. In Figure 4.9.12, a high nitrogen concentration is evident throughout the larger inclusion in the micrograph. The high stability of TiN in steels at elevated temperatures suggests that these particles may have served as nucleation sites for the oxide inclusions during cooling.

![Micrograph of inclusions](image1)

![EDS maps of elements](image2)

**Figure 4.9.12** EDS maps of elements in non-metallic inclusions found in the weld metal of a weld deposited at a heat input of 1.0 kJ/mm and at a preheat temperature of 80°C.

A high concentration of aluminium is not evident in the inclusion shown in Figure 4.9.9, but higher amounts were detected in the inclusions shown in Figures 4.9.10 to 4.9.12. In all these figures, aluminium seems to be
incorporated in the Mn-Si-O oxide structure, which together with low levels of titanium form complex Mn-Si-Ti-Al-O type oxide inclusions. The aluminium and titanium contents of the parent material and undiluted weld metal shown in Tables 3.1.1 and 3.1.3 were 0.039% and 0.012%, and <0.006% and 0.01%, respectively. The aluminium measured in the inclusions is therefore most likely present due to dilution effects between the weld and base metals.

Figures 4.9.13 to 4.9.16 show selected EDS spectra measured for non-metallic inclusions, as well as the associated high resolution BSE images. Since EDS analysis is done by analysing the characteristic X-rays emitted by back-scattered electrons that penetrate below the substrate surface, the interaction volume of the incident electron beam below the surface of the substrate must also be considered when analysing the chemical composition obtained in this way. Each inclusion examined is numbered on the BSE micrographs and corresponding numbers are given on the spectrum belonging to that specific inclusion. Numbers were assigned in the order in which the analyses were performed. The quadrant in the top corner of each spectrum gives the elements present in each inclusion, while Tables 4.9.1 to 4.9.4 summarise the results from each figure.

Figures 4.9.13 and 4.9.14 display the EDS analyses of inclusions observed at low magnifications in welds deposited without preheat at heat input levels of 0.4 kJ/mm and 1.0 kJ/mm, respectively. Summaries of the measured inclusion compositions relevant to Figures 4.9.13 and 4.9.14 are given in Tables 4.9.1 and 4.9.2. Figure 4.9.15 and Table 4.9.3 show the chemical compositions measured in three different locations within a single inclusion at a high magnification in weld metal deposited without preheat at a heat input of 1.0 kJ/mm. Figure 4.9.16 and Table 4.9.4 refer to a low magnification EDS analysis of inclusions in a weld deposited at a heat input level of 0.7 kJ/mm and a preheat temperature of 80°C. These analyses were selected for discussion as they are representative of welds deposited using a wide range of heat inputs and preheat temperatures.

As discussed earlier, the compositions summarised in this section are not absolute since the interaction volume of the incident electron beam with the subsurface area of the substrate needs to be taken into consideration when EDS point analyses are performed on such small particles. It is also not clear in each case how much of the inclusion is included in the analysis since the subsurface volume of each inclusion depends on the location on the diameter of the sphere where the surface was cut. Only general trends will therefore be discussed in the rest of this section.
Figure 4.9.13 EDS point analyses of non-metallic inclusions observed in the weld metal of a weld deposited preheat-free at a heat input of 0.4 kJ/mm.
Figure 4.9.14 EDS point analyses of non-metallic inclusions found in the weld metal of a weld deposited without preheat at a heat input of 1.0 kJ/mm.
Figure 4.9.15 EDS point analyses of non-metallic inclusions found in the weld metal of a weld deposited without preheat at a heat input of 1.0 kJ/mm.
**Figure 4.9.16** EDS point analyses of non-metallic inclusions found in the weld metal of a weld deposited at a heat input of 0.7 kJ/mm and at a preheat temperature of 80°C.

**Table 4.9.1** EDS analyses (wt%) of inclusions found in weld metal deposited without preheat at a heat input of 0.4 kJ/mm, corresponding to the spectra shown in Figure 4.9.13.

<table>
<thead>
<tr>
<th>Spectrum</th>
<th>Fe</th>
<th>O</th>
<th>Mn</th>
<th>Si</th>
<th>Ti</th>
<th>Al</th>
<th>S</th>
<th>Cr</th>
</tr>
</thead>
<tbody>
<tr>
<td>2</td>
<td>59.8</td>
<td>13.2</td>
<td>16.4</td>
<td>6.3</td>
<td>1.9</td>
<td>1.8</td>
<td>0.4</td>
<td>0.3</td>
</tr>
<tr>
<td>5</td>
<td>31.1</td>
<td>23.2</td>
<td>28.9</td>
<td>10.0</td>
<td>5.0</td>
<td>1.2</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>18</td>
<td>28.8</td>
<td>25.9</td>
<td>26.3</td>
<td>10.2</td>
<td>4.6</td>
<td>3.8</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>19</td>
<td>49.5</td>
<td>18.2</td>
<td>18.6</td>
<td>7.9</td>
<td>2.6</td>
<td>-</td>
<td>0.3</td>
<td>-</td>
</tr>
<tr>
<td>21</td>
<td>69.0</td>
<td>11.6</td>
<td>11.2</td>
<td>4.9</td>
<td>1.4</td>
<td>1.6</td>
<td>0.3</td>
<td>-</td>
</tr>
<tr>
<td>25</td>
<td>81.9</td>
<td>6.9</td>
<td>6.8</td>
<td>2.9</td>
<td>0.7</td>
<td>0.6</td>
<td>0.2</td>
<td>-</td>
</tr>
</tbody>
</table>
Table 4.9.2 EDS analyses (wt%) of inclusions found in weld metal deposited without preheat at a heat input of 1.0 kJ/mm, corresponding to the spectra shown in Figure 4.9.14.

<table>
<thead>
<tr>
<th>Spectrum</th>
<th>Fe</th>
<th>O</th>
<th>Mn</th>
<th>Si</th>
<th>Ti</th>
<th>Al</th>
<th>S</th>
<th>C</th>
</tr>
</thead>
<tbody>
<tr>
<td>Spectrum 5</td>
<td>52.4</td>
<td>15.1</td>
<td>17.2</td>
<td>7.3</td>
<td>2.1</td>
<td>1.6</td>
<td>0.3</td>
<td>3.9</td>
</tr>
<tr>
<td>Spectrum 15</td>
<td>23.5</td>
<td>24.7</td>
<td>28.9</td>
<td>11.2</td>
<td>5.1</td>
<td>2.6</td>
<td>0.2</td>
<td>3.8</td>
</tr>
<tr>
<td>Spectrum 19</td>
<td>65.4</td>
<td>10.1</td>
<td>12.8</td>
<td>4.8</td>
<td>1.1</td>
<td>0.9</td>
<td>0.2</td>
<td>4.7</td>
</tr>
<tr>
<td>Spectrum 23</td>
<td>25.1</td>
<td>25.0</td>
<td>27.7</td>
<td>11.0</td>
<td>5.1</td>
<td>1.8</td>
<td>0.2</td>
<td>4.2</td>
</tr>
<tr>
<td>Spectrum 27</td>
<td>64.0</td>
<td>10.0</td>
<td>15.0</td>
<td>5.7</td>
<td>1.7</td>
<td>0.9</td>
<td>0.2</td>
<td>2.5</td>
</tr>
<tr>
<td>Spectrum 31</td>
<td>59.0</td>
<td>14.1</td>
<td>16.0</td>
<td>6.3</td>
<td>3.3</td>
<td>1.2</td>
<td>0.2</td>
<td>-</td>
</tr>
</tbody>
</table>

Table 4.9.3 EDS analyses of inclusions found in weld metal deposited without preheat at a heat input of 1.0 kJ/mm, corresponding to the spectra shown in Figure 4.9.15.

<table>
<thead>
<tr>
<th>Spectrum</th>
<th>Fe</th>
<th>O</th>
<th>Mn</th>
<th>Si</th>
<th>Ti</th>
<th>Al</th>
<th>S</th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>Spectrum 1</td>
<td>22.4</td>
<td>27.7</td>
<td>30.0</td>
<td>12.4</td>
<td>5.9</td>
<td>1.6</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>Spectrum 2</td>
<td>12.2</td>
<td>29.8</td>
<td>33.7</td>
<td>6.9</td>
<td>15.8</td>
<td>1.1</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>Spectrum 3</td>
<td>29.3</td>
<td>24.5</td>
<td>24.7</td>
<td>10.3</td>
<td>9.5</td>
<td>1.4</td>
<td>0.3</td>
<td></td>
</tr>
<tr>
<td>Spectrum 8</td>
<td>98.7</td>
<td>-</td>
<td>1.0</td>
<td>0.3</td>
<td>-</td>
<td>-</td>
<td>-</td>
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</tr>
</tbody>
</table>

Table 4.9.4 EDS analyses of inclusions found in weld metal deposited at a heat input of 0.7 kJ/mm and at a preheat temperature of 80°C, corresponding to the spectra shown in Figure 4.9.16.

<table>
<thead>
<tr>
<th>Spectrum</th>
<th>Fe</th>
<th>O</th>
<th>Mn</th>
<th>Si</th>
<th>Ti</th>
<th>Al</th>
<th>S</th>
<th>Ca</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>Spectrum 1</td>
<td>35.6</td>
<td>23.0</td>
<td>24.7</td>
<td>10.0</td>
<td>3.5</td>
<td>2.4</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Spectrum 3</td>
<td>34.8</td>
<td>19.8</td>
<td>30.0</td>
<td>8.7</td>
<td>3.4</td>
<td>2.2</td>
<td>1.0</td>
<td>0.2</td>
<td>-</td>
</tr>
<tr>
<td>Spectrum 4</td>
<td>34.4</td>
<td>18.4</td>
<td>31.0</td>
<td>7.0</td>
<td>8.6</td>
<td>0.6</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Spectrum 5</td>
<td>36.9</td>
<td>21.9</td>
<td>25.4</td>
<td>10.3</td>
<td>3.2</td>
<td>2.2</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Spectrum 10</td>
<td>70.0</td>
<td>11.4</td>
<td>10.4</td>
<td>5.1</td>
<td>1.3</td>
<td>1.0</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Spectrum 11</td>
<td>71.6</td>
<td>10.8</td>
<td>10.4</td>
<td>5.0</td>
<td>1.1</td>
<td>1.1</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Spectrum 16</td>
<td>67.6</td>
<td>11.6</td>
<td>12.4</td>
<td>5.2</td>
<td>1.1</td>
<td>1.1</td>
<td>-</td>
<td>0.9</td>
<td>-</td>
</tr>
<tr>
<td>Spectrum 21</td>
<td>73.3</td>
<td>10.2</td>
<td>9.5</td>
<td>4.6</td>
<td>0.9</td>
<td>1.0</td>
<td>-</td>
<td>-</td>
<td>0.5</td>
</tr>
<tr>
<td>Spectrum 23</td>
<td>73.9</td>
<td>9.6</td>
<td>10.3</td>
<td>4.2</td>
<td>0.8</td>
<td>1.0</td>
<td>0.2</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Spectrum 24</td>
<td>59.4</td>
<td>13.3</td>
<td>17.9</td>
<td>5.7</td>
<td>1.8</td>
<td>1.2</td>
<td>0.6</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Spectrum 25</td>
<td>56.3</td>
<td>12.3</td>
<td>19.9</td>
<td>6.8</td>
<td>3.4</td>
<td>-</td>
<td>0.2</td>
<td>1.0</td>
<td>-</td>
</tr>
<tr>
<td>Spectrum 27</td>
<td>62.4</td>
<td>13.1</td>
<td>15.5</td>
<td>6.2</td>
<td>1.5</td>
<td>1.2</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

All of the inclusions shown in Tables 4.9.1 to 4.9.4 contain oxygen, and those in Table 4.9.2 contain carbon. SEM-EDS techniques cannot accurately measure the amounts of light elements with atomic numbers (Z) below 11, such as O (Z = 8) and C (Z = 6). These elements do not emit a large number of X-rays and when they do, the wavelengths of the X-rays are longer than those of heavier elements. These X-rays are therefore easily absorbed within the surrounding material and are, consequently, not easily detected and their analysed percentages inaccurate. The amounts of oxygen and carbon measured are tabulated here merely to indicate their presence within the inclusions. The tabulated amounts of carbon and oxygen in the spectra also altered the final amounts of other elements present. The chemical compositions shown should therefore be used merely as a comparative tool.
As discussed earlier, acicular ferrite nucleation is dependent on the inclusion size distribution and volume fraction in the weld metal. The chemistry of inclusions is, however, also important. All the inclusions studied in this section contained high concentrations of manganese. When manganese, an austenite-stabilising element, segregates preferentially to inclusions, acicular ferrite nucleation is favoured in the Mn-depleted matrix. Acicular ferrite will also form with greater ease due to the consequent decrease in hardenability [2,7]. In the absence of oxide inclusions, fewer nucleation sites are available for the nucleation of acicular ferrite and the matrix has a higher dissolved manganese content, stabilising high temperature austenite over a wider temperature range. Under these conditions the formation of bainite is encouraged [15]. This effect is important in the mechanism of acicular ferrite formation although it has been shown that an optimum manganese content in the matrix favours acicular ferrite over allotriomorphic ferrite formation in low-alloy C-Mn weld metal [16,17].

Similar trends with regard to the chemical compositions of the inclusions are evident in Tables 4.9.1 to 4.9.4. The inclusions analysed shown, in decreasing order, the presence of Fe, Mn, Si, Al, Ti, in some cases S, and in rare cases Cr, Ca, and Mo. The maps in Figures 4.9.9 to 4.9.12, however, showed low concentrations of iron compared to other elements such as Mn and Si. This can most likely be attributed to inclusion of a large portion of the surrounding metal matrix in the interaction volume. In Figure 4.9.3, three spectrums are shown for a single inclusion, two of which (spectrums 1 and 3) were measured towards the edge and one (spectrum 2) in the centre of the inclusion. The interaction volume of spectrum 2, being situated in the centre of the inclusion, is expected to encapsulate less of the surrounding matrix, and the values in Table 4.9.3 show a lower concentration of iron in this spectrum. This suggests that at least some of the iron in the surrounding steel matrix is included in the spectra due to interaction volume effects.

Low concentrations of sulphur were detected in most of the inclusions shown in Figures 4.9.13 and 4.9.14, whereas only a small number of the inclusions analysed in Figure 4.9.16 contained sulphur. This does not appear to be an interaction volume effect, since no sulphur was detected in the steel matrix (shown as spectrum 8 in Figure 4.9.15). This suggests that sulphur segregates to the smaller non-metallic inclusions in the weld metal. It has been noted by several authors that spherical oxide inclusions have a significant influence on the hydrogen-induced cracking (HIC) behaviour of steel. Xue et al. [18], Jin et al. [19], Gan et al. [20], and Ohaeri [3] found that, in the absence of large external stresses, HIC initiated in the steel around Si- and Al-enriched oxide inclusions. Du et al. [21] observed crack initiation at globular oxide inclusions and Dong [14] showed that the inclusions observed on hydrogen-induced crack surfaces were mostly oxides containing Fe, Mn, Al and Ca.

On the other hand, Mohtadi-Bonab et al. [6] found that HIC does not initiate at oxide inclusions. These authors reported that crack initiation generally appeared to be associated with MnS particles and carbonitrides, but they observed that spherical oxides did play a role in crack propagation. A similar observation was made by Huang et al. [10].

In the current study, crack initiation was not observed to occur at non-metallic inclusions, but rather at the wagon tracks at the toes of the welds. Cracks propagated towards other stress concentrators near the root of the welds, such as incomplete penetration at lower welding currents. The oxide inclusions did, however, play a significant role in crack propagation and in determining the fracture mode, as will be discussed subsequently.
Figure 4.9.17 High-resolution SEM micrographs of non-metallic inclusions of various sizes observed in welds deposited at various heat input levels, without preheat and at a preheat temperature of 80°C.
4.9.3 The role of non-metallic oxide inclusions in WMHACC

Figure 4.9.17 contains a collection of high-resolution SEM micrographs of non-metallic inclusions observed in weld metal deposited at a range of heat input levels. These figures show significant microvoid formation between the oxide inclusions and the metal matrix surrounding it. It is evident from these micrographs that the interface is completely incoherent.

Authors who observed hydrogen-related damage associated with spherical oxide inclusions [3,6,10,14,18-21] all maintained that incoherency between the surfaces of oxide inclusions and the steel matrix plays a significant role in determining both hydrogen-induced crack initiation and crack propagation.

These, and other authors [7,9,22] established that the microvoids created by incoherency of the interfaces are effective hydrogen sinks and act as trapping sites for hydrogen. These traps are regarded as irreversible due to hydrogen recombination within the microvoids to form molecular hydrogen (H₂), which cannot diffuse into the matrix since dissociation of molecular hydrogen is required before re-entry is possible. The build-up of H₂ increases the pressure within the microvoids, inducing stress in the surrounding metal. If the local stress exceeds the yield strength of the steel, plastic deformation can occur around inclusions.

If plastic deformation occurs, more dislocations are introduced into the structure, creating a higher concentration of reversible hydrogen traps [23]. Both reversible and irreversible hydrogen traps are considered detrimental to the HIC resistance. Zhang et al. [23] showed that the total hydrogen-induced stress is additive to external stresses and follows a linear relationship with the hydrogen concentration. In addition, hydrogen generally diffuses in the direction of local stress concentrations and is attracted to the inclusions. A higher number of inclusions in the weld metal raises the likelihood that weld metal hydrogen-assisted cold cracks initiate at and propagate through NMI clusters.

As stated before, spherical inclusions produce a relatively mild stress concentration in the surrounding matrix, but the interfaces between the inclusions and the matrix are not necessarily of a similar spherical form. Figure 4.9.18 contains a collection of high-resolution SEM micrographs of inclusions that were lost during the polishing process. It is evident that in the positions previously occupied by inclusions, roughly cubic voids remain in the matrix. The microvoids around intact inclusions also appear to have cubic shapes. The reason for the cubical voids is unclear, but it is most likely a crystallographic effect since the crystal lattice of the ferritic matrix is body centred cubic.

The stress concentrations at the corners and edges of these cubic voids are higher than those around spherical voids, especially in the presence of hydrogen pressure build-up within the voids. The presence of these stress concentrations is likely to intensify the stress caused by the internal hydrogen pressure and increase the risk of WMHACC. If not crack initiation, then certainly the energy required for crack propagation would be lowered, detracting from the apparent toughness of the acicular ferrite-containing microstructure.
Figure 4.9.18 High-resolution FEG-SEM micrographs of cubic voids surrounding incoherent spherical non-metallic inclusions.

Propagating cracks select the easiest pathway through the metal, and incoherent oxide inclusions therefore facilitate crack growth, especially if found in dense clusters. A propagating crack may even change direction in order for its path to intersect the largest number of non-metallic inclusions possible. Such clusters of inclusions were observed within the weld metal during this investigation, as shown in Figure 4.9.19.

Figures 4.9.19(a), (b), (c), and (f) are BSE images of the microstructure at low magnifications, showing areas with high densities of inclusion clusters. Not only are the size, area and volume fraction important when inclusions are studied in the context of the WMHACC behaviour of the welds, but the distribution of these inclusions also plays a role. Figures 4.9.19(c) and (d) are SE images at high magnifications of inclusions in close proximity. Such closely spaced inclusions within inclusion clusters are expected to ease crack propagation. Due to the factors discussed above, these clusters are likely to decrease the toughness of the steel and its resistance to WMHACC. This will be discussed in more detail in Section 4.9.10.
Figure 4.9.19 High-resolution FEG-SEM micrographs of non-metallic inclusion clusters observed in the weld metal of welds deposited without preheat at heat input levels of (a) 0.5 kJ/mm (BSE), (b) 0.5 kJ/mm (BSE), (c) 1.0 kJ/mm (SE), (d) 1.0 kJ/mm (SE), and at a preheat temperature of 80°C at heat input levels of (e) 0.4 kJ/mm (BSE), and (d) 0.4 kJ/mm (BSE).
Figure 4.9.20 contains a collection of high-resolution SEM micrographs of the propagation paths of microcracks ahead of advancing macrocrack tips. These microcracks were observed ahead of arrested macrocracks that did not extend all the way through the weld metal. Each microcrack was photographed in both SE and BSE imaging modes in order to delineate its path more clearly, except for (a) and (b) where the non-metallic inclusions were lost from the matrix during fracture.

It is evident from Figure 4.9.20 that non-metallic inclusions are present within the microcracks and that these cracks advanced between inclusions where their propagation would have required the lowest amount of energy. The matrix would essentially have fractured along the weakest link in the microstructure. It is unclear whether such microcracks are always present at the head of an advancing macrocrack tip or if they formed during the last expenditure of fracture energy. Nevertheless, the microcracks followed paths that linked the microvoids around the inclusions where undoubtedly stress concentrations would be found due to the shape of the voids and hydrogen pressure within. Similar observations were made by Kurji [24] around non-metallic inclusions in weld metal deposited by E6010 electrodes under similar welding conditions.

Most of the WMHACC fractures observed in Section 4.3 of the current study occurred at the lower heat input levels investigated. These welds displayed the highest inclusion numbers, areas and volume fractions. Although no cracks were observed to initiate at inclusions, the presence of NMI’s would certainly have contributed to lowering of the WMHACC resistance.

In addition, Zorc et al. [25] linked the presence of non-metallic inclusion clusters near the centre of the weld with solidification crack initiation and propagation. Steel weld metals that showed a high number and volume percentage of oxide inclusions showed a higher incidence of solidification cracking when compared to weld metals with low inclusion contents, especially when inclusions near the weld centreline were found in dense lines and clusters, such as those shown in Figure 4.9.19. Cracks could initiate at either non-metallic inclusions near low-melting-point eutectic phases found at the weld centreline or in the eutectics themselves, and propagate through inclusion clusters and along brittle phases at the weld centreline. This could explain the absence of clear trends in Section 4.5 where solidification cracking was observed at both low heat input levels (where inclusion numbers and volume percentages were high) and at high heat input levels (where segregation of elements would be the highest).
Figure 4.9.20 High-resolution FEG-SEM micrographs of WMHACC cracks propagating via non-metallic inclusions in weld metal deposited without preheat at heat input levels of (a)-(b) 0.8 kJ/mm (welding current 120 A, welding speed 225 mm/min), and (c)-(f) 1.0 kJ/mm (welding current 120 A, welding speed 173 mm/min).

REFERENCES


4.10 FRACTURE SURFACES

Figure 4.8.44 showed that microcracks at the head of arrested macrocracks tend to grow along hard and brittle microstructural constituents such as carbides and harder microphases, while Figure 4.9.20 demonstrated that microcracks propagate between non-metallic inclusions. The observation of cracks preferentially advancing through brittle transformation products, banded cementite structures, inclusion clusters and generally weaker junctures of the microstructure could explain the typical step-wise nature of the WMHACC cracks observed in Section 4.2. These microstructural components also act as reversible or irreversible hydrogen traps, further reducing their fracture toughness. The crack consequently changes direction in the presence of such microstructural inhomogeneity to lower its propagation energy [1].

During the course of this investigation, hydrogen cracks in the weld metal were observed to initiate exclusively at the wagon tracks at the toes of the welds. Cracks propagated downward through the weld metal to stress concentrations located near the root of the weld. The direction of crack growth was, for the main part, through the weld metal with a few exceptions deviating through the heat-affected zones for no apparent reason other than the geometry of the weld beads and the presence of a harder microstructure in the high-temperature heat-affected zone.

The fracture surfaces of welds showed topographical changes from a mixed-mode ductile fracture with a dimpled quasi-cleavage appearance near the site of crack initiation at the wagon tracks, to a brittle cleavage fracture mode closer to the root of the weld.

Figure 4.10.1 contains a collection of high-resolution SEM photomicrographs of the fracture surfaces of the weld metal near the crack initiation site in the welds, deposited at heat input levels of 0.4 kJ/mm (a) to (c), and 0.8 kJ/mm (d) to (f). These fracture surfaces are dominated by fine equiaxed dimples, typically seen on ductile fracture surfaces (Dieter 1988), interspersed with cleavage-type facets.
There are no observable differences in the fracture surfaces between weld metal deposited at low and high heat input levels nor was any relationship observed with welding current, and it is assumed that due to the similarities in microstructures and the presence of high volume fractions of inclusions, similar mechanisms were active during crack propagation in all cases.

The dimples observed Figure 4.10.1 originated at non-metallic inclusions within the weld metal, as confirmed by the higher magnification micrographs shown in Figure 4.10.2. The dimples evident in Figures 4.10.1 and 4.10.2 are indicative of a ductile mode of fracture. In this mode, dimples originate at non-metallic inclusions by means of a process of particle/matrix interface decohesion. As illustrated in Figure 4.10.3, this is the first step in the nucleation of microvoids around hard particles such as oxide inclusions. As shown in Figure 4.9.17, the oxide inclusions have incoherent interfaces with the surrounding weld metal matrix and consequently microvoids already exist around these particles. During ductile fracture of the weld metal the
energy required for interface decohesion is therefore low and crack propagation resumes by the growth and eventual coalescence of such microvoids [2,3].

![Micrograph Images](image1.png)  ![Micrograph Images](image2.png)

Figure 4.10.2 High-resolution FEG-SEM micrographs of non-metallic inclusions situated in the dimples observed in ductile fracture surfaces for weld metal deposited without preheat at heat input levels of (a) to (c) 0.4 kJ/mm, and (d) 0.8 kJ/mm.

![Schematic Diagram](image3.png)

Figure 4.10.3 Schematic illustration of microvoid coalescence around incoherent non-metallic inclusions [3].
The mode of fracture illustrated in Figure 4.10.3 is facilitated by the presence of high concentrations of hydrogen in the weld metal. As discussed in Section 2.4.1 of the literature review, hydrogen is attracted to areas of high stress concentration such as the tip of an advancing crack. According to the HELP model of hydrogen embrittlement, the high localised diffusible hydrogen concentration at the crack tip accelerates dislocation motion.

The model of hydrogen-enhanced localised plasticity (HELP) sets forth that the energy barriers to dislocation motion are reduced by substitutionally dissolved hydrogen, which enhances dislocation motion by reducing the elastic interactions between dislocations and increasing their mobility. A higher hydrogen concentration therefore facilitates rapid plastic deformation around particles during the growth and coalescence of microvoids at the tip of an advancing crack.

Beachem [4] suggested that the diffusible hydrogen content dissolved at the crack tip, the stress intensity and the operative fracture mode in steel are interrelated. Beachem’s diagram illustrating this interrelationship is given in Figure 4.10.4, which shows that at increasing amounts of hydrogen, quasi-cleavage (QC) and eventually microvoid coalescence are favoured over intragranular type fractures. At low hydrogen contents or low stress intensities, crack propagation therefore proceeds by means of intragranular fracture with little to no plastic deformation. As the hydrogen concentration increases, higher levels of plastic deformation precedes the crack tip, which fits in with the HELP model of hydrogen-assisted cracking by enhanced localised plasticity.

![Diagram](image.png)

**Figure 4.10.4** Beachem’s diagram representing the interrelationship between dissolved hydrogen, stress intensity and fracture mode typical in steels, with $C_{IG}$, $C_{QC}$, and $C_{MVC}$ being the critical concentrations of H affecting a transition in fracture mode for a given stress intensity ($IG = $ Intergranular fracture, $QC = $ Quasi Cleavage, $MVC = $ Micro-Void Quaescence) [4].

Beachem further postulated that hydrogen merely facilitates the fracture mechanism that would have caused failure in the absence of hydrogen, similar in effect to lowering the stress intensity required for such a failure to occur. This is also evident in Figure 4.10.4 as the decrease in stress intensity required for intragranular ($C_{IG}$), quasi-cleavage ($C_{QC}$) and microvoid coalescence ($C_{MVC}$) type fractures with an increase in hydrogen content. Therefore, in the presence of enough hydrogen, any microstructure, even one with a high inherent toughness such as acicular ferrite, can become susceptible to hydrogen-induced cracking.
In the fractographs shown in Figure 4.10.2, more densely packed inclusions are visible in the weld metal than were evident in the BSE micrographs displayed in Figure 4.9.2 at the same magnifications. This is due to the topographical nature of the plane of observation. In a polished and etched micrograph, a flat surface is viewed, and taking into consideration the size distribution of inclusions, their position relative to the surface plane, and the distances between inclusions, there is a lower probability that the polished plane will intersect a large number of inclusions compared to a fracture surface. A growing crack will tend to minimise its energy of propagation while simultaneously maximising its kinetic energy, and in order to do this, it propagates between microstructural features with low toughness and high stress concentrations such as inclusions, exposing them to the surface. This is illustrated in Figure 4.10.5 [3].

![Diagram](image)

**Figure 4.10.5** Schematic illustrating the difference in the observational plane between a (a) polished surface and (b) a fracture surface [3].

Dimple formation on ductile and quasi-cleavage fracture surfaces only pertains to relatively small non-metallic inclusions. For comparatively large inclusions, cleavage facets form during fracture due to the higher stress concentration caused by the larger particle and the high H$_2$ pressure that can build up in the microvoids at the incoherent particle/matrix interface. As stated before, the stress caused by such a pressure increase in the microvoids is additive to any externally applied constraint stresses, leading to brittle cleavage facets opening up around large inclusions.

Figure 4.10.6 contains a collection of high-resolution SEM micrographs showing cleavage facets around large inclusions found in the weld metal for welds deposited over the entire heat input range tested. The river markings on the surfaces of the facets indicate the direction of fracture propagation and can be traced to the point of origin [5]. In each case, these markings lead to inclusions much larger than those scattered in the surrounding weld metal. Rather than microvoid growth and coalescence, the cleavage facets opened ahead of the advancing crack tip due to the reduced toughness brought about by the presence of such large inclusions.

The role of non-metallic inclusions in crack propagation is further highlighted by the high-resolution SEM micrographs given in Figure 4.10.7, which show areas where advancing cracks changed direction when dense inclusion clusters were encountered. Careful examination of the river markings on the cleavage facets in these images indicates a point of origin away from the dimpled portions of the fracture surfaces. This suggests that the cleavage facets extended through the microstructure towards the inclusion clusters where
they changed direction, as indicated by the shift in the focus of the electron microscope with dimples appearing blurry further away from the cleavage facet-dimple interface.

Figure 4.10.6 High-resolution FEG-SEM micrographs of cleavage facets around non-metallic inclusions in weld metal deposited without preheat at heat input levels of (a) to (b) 0.4 kJ/mm, and (c) to (d) 0.8 kJ/mm.

Such a change in direction would constitute a reduction in the crack propagation energy when inclusion clusters are encountered, but also a lowering of the kinetic energy of crack propagation as dimple formation is accompanied by extensive plastic deformation ahead of the advancing crack tip.

Another explanation for this phenomenon could be the inhomogeneity of the microstructure, as a crack propagating through islands of bainite is likely to show cleavage-type fracture. Once the growing crack encounters areas where acicular ferrite predominates, dimple formation would again be favoured and the crack would extend between inclusions.

Figure 4.10.8 shows high-resolution SEM micrographs of the remainder of the fracture surfaces closer to the weld root where crack propagation transitioned from quasi-cleavage to pure cleavage fracture. The fracture surfaces show extensive river markings and the shear ridges and steps typical of cleavage fracture [5]. It is
evident that the kinetic energy of crack propagation increased as the intact portion of the weld throat decreased with the advancing crack tip. Rapid and final brittle failure occurred in this portion of the weld metal.

**Figure 4.10.7** High-resolution FEG-SEM micrographs of sharp transitions between regions of ductile fracture and cleavage fracture in weld metal deposited without preheat at heat input levels of (a) 0.8 kJ/mm, and (b) to (c) 1.0 kJ/mm.
Figure 4.10.8 High-resolution FEG-SEM micrographs of fracture surfaces transitioning into cleavage closer to the root of the weld for welds deposited without preheat at heat input levels of (a) 0.4 kJ/mm (welding current 120 A, welding speed 432 mm/min), (b) to (c) 0.4 kJ/mm (welding current 140 A, welding speed 536 mm/min), (d) to (f) 0.8 kJ/mm (welding current 130 A, welding speed 212 mm/min), and (g) to (h) 1.0 kJ/mm (welding current 120 A, welding speed 173 mm/min).

REFERENCES


4.10 RESIDUAL STRESS MEASUREMENTS

As discussed in Section 2.4 of the literature review, the three requirements for hydrogen-assisted cold cracking to occur are a high enough hydrogen content, a susceptible microstructure, and the presence of a tensile stress. Thus far, the factors determining the susceptibility of the microstructure to WMHACC have been discussed in detail. These factors include the cooling rate, which influences not only the microstructure, but also the diffusible hydrogen concentration remaining in the weld metal after cooling, the weld metal and heat-affected zone hardenesses, the transformation products in the weld microstructures, and the presence of non-metallic inclusions. However, WMHACC cannot occur in the absence of a tensile stress. This section of the investigation focuses on the residual stresses that develop in the root pass of MWIC test pieces after welding with E6010 electrodes.

Residual stresses are self-equilibrating and elastic in nature and remain in the welded joint even when the component is at equilibrium and all other sources of stress and heat are removed. Residual stress is additive to external stress and often undetectable until failure occurs at applied stresses well below the yield strength of the material [1]. In the presence of diffusible hydrogen, the stress required to initiate and propagate cracks is reduced due to embrittlement, placing a greater emphasis on the residual stresses that develop during welding.

Post-weld residual stress in MWIC test pieces is a direct result of the restraint induced by the welds anchoring the X70 plate to the thick T-piece underneath it. These welds prevent the weld metal from freely contracting on cooling. The large thermal gradients between the weld metal and surrounding base metal lead to the development of a tensile residual stress, which is at a maximum in the centre of the weld bead. The contracting weld metal causes an opposing elastic strain in the adjacent base metal, which, over a certain distance, equilibrates to zero. If the stress in the weld metal exceeds the yield strength at any given temperature on cooling, non-uniform plastic deformation occurs within the weld metal, reducing the residual stress to the value of the yield strength at that temperature [2].

The residual stress in the weld is affected by many factors, including the base metal and weld metal yield strengths, the restraint intensity, the heat input and the parent metal plate thickness [3,4]. Lower yield strengths in the weld and base metals facilitates plastic deformation, relieving a large portion of the residual stress. As the heat input increases, the deposition rate also increases and additional contraction of the weld metal may cause a higher stress. More heat is also placed into the joint, causing steeper thermal gradients and resulting in a larger increase in stress. The base plate thickness is important since it impedes free contraction of the weld metal, i.e. the thicker the plate, the higher the restraint imposed on the joint. The presence of stress concentrations in the weld metal further exacerbates the effect of stress on crack initiation and [5].

The total stress at the six o’clock position in a pipeline girth weld during the lifting operation is a combination of the restraint intensity and the externally applied lifting stress. According to Alam et al. [5] this total stress is most closely simulated by a restraint length of 100 mm during WIC testing, based on the empirical equations given in Section 2.8 of the literature review (Equations 2.8.1 to 2.8.3). The same authors concluded that the higher stress intensity in an MWIC test piece with a restraint length of 25 mm could be viewed as simulating the most extreme stress concentrations in actual pipeline girth welds, or alternatively, as providing a large inherent safety factor.
Measuring the residual stress in an MWIC test piece with a restraint length of 25 mm is therefore relevant to the pipeline industry. At the very least, it gives a measure of the level of stress that resulted in the crack/no crack boundaries derived in Sections 4.3 and 4.4, and is related to the cooling times and hardness values in the test specimens used to generate the boundaries. At lower stress intensities, these boundaries will certainly shift, with cracks occurring at different heat input levels.

In the MWIC test pieces, the most critical design feature influencing the restraint intensity, and hence, the residual stress, is the restraint length, which is the region directly adjacent to the welded joint and the only portion of the X70 plate material free from the anchor welds. North et al. [6] and Alipooramirabad et al. [4] used finite element modelling to show that the level of residual stress in MWIC welded joints decreases with a decrease in the restraint intensity (in other words, an increase in restraint length from 25 mm to 100 mm, and finally to 150 mm). The latter authors confirmed this with residual stress measurements performed using neutron diffraction techniques. Longer restraint lengths also resulted in slower cooling rates, which facilitated more hydrogen effusion from the weld beads during cooling.

The same authors [4] reported that, for plate thicknesses ranging from 12 mm to 20 mm, the maximum stress in the transverse direction across the welds stabilised when the restraint length (and hence the restraint intensity) fell below 50 mm. No change in these plateau stress levels was observed with further changes in heat input.

During the course of this investigation, the plate thickness and restraint length were kept constant at 10 mm and 25 mm, respectively, and the effects of varying heat inputs and preheat temperatures on the residual stress were investigated. The results given here therefore pertain only to plate thicknesses up to 10 mm, an MWIC restraint length of 25 mm and single batches of E6010 electrodes and X70 plate.

A steady-state neutron source was used in the determination of the residual stress. There are several advantages to using neutron diffraction over other techniques such as X-ray diffraction, hole drilling and magnetic techniques. Neutron beams are non-destructive, penetrate deeper into thicker sections than X-rays, and can be used to determine bulk macroscopic stresses over larger areas of the material if enough measurement points are selected [1].

As described in Section 3.6, the diffraction angle of the neutron beam changes with a change in the lattice spacing between diffraction planes in grains in the steel with their lattice planes oriented favourably to the incident beam. Equation 3.6.1 was used to determine the lattice spacings of both strained and unstrained weld metal as a function of the diffraction angle \( \theta \) (Bragg angle) and Equations 3.6.2 to 3.6.7 were then used to calculate the residual stresses based on the difference in lattice spacing between strained test pieces and unstrained test coupons.

Residual stress can only be measured in uncracked MWIC test pieces since the initiation of a crack relieves the stress that caused it. Combinations of parameters unlikely to result in WMHACC were therefore selected from Figures 4.4.5, 4.4.10 and 4.4.15. The effect of heat input was investigated for welds deposited at a constant preheat temperature of 50°C by welding at heat input levels of 0.5 kJ/mm, 0.7 kJ/mm, and 1.0 kJ/mm. A constant heat input of 0.7 kJ/mm was then selected to investigate the effect of preheat by welding without preheat and at preheat temperatures of 50°C and 80°C.

Stresses were measured at locations 1 mm apart over a gauge volume of 2x2x2 mm³, transversely aligned across the welds. At each of these individual locations, the stress was measured in the transverse,
longitudinal and normal directions. These measurements were taken at offsets of 1.6 mm and 3.0 mm from the underside of the joint. In this section, all residual stresses are presented as a function of distance from the weld centreline. A watermarked and calibrated micrograph of the weld macrostructure was superimposed on the graphs representing residual stress in order to illustrate the exact location within the weld geometry where the Bragg angle measurements were taken for calculation of the stress. In Figures 4.11.1 to 4.11.5 the error bars show uncertainties in the measured stress values, calculated using Equations 3.6.11 to 3.6.13.

Figures 4.11.1(a) to (c) show the residual stresses measured in a weld deposited at a heat input of 0.5 kJ/mm and a preheat temperature of 50°C at offsets of (a) 1.6 mm and (b) 3.0 mm from the underside of the joint. Figure 4.11.1(c) shows the Von Mises stress, which is the resultant of the triaxial stress state, calculated using Equation 3.6.14.

Figure 4.11.1(a) shows the typical trend observed in all welds tested, in this case for an offset of 1.6 mm from the underside of the weld. The stress in the transverse, longitudinal and normal directions vary with changes in the weld bead geometry. This trend is most evident in the transverse and longitudinal directions where a peak stress of about 740 MPa was reached in the centre of the weld bead in the transverse direction. As the throat thickness of the weld deposit decreases towards the wagon tracks at the toes of the weld, the stress shows an associated decrease in both of these areas. It then rises to another peak a short distance into the base metal after which it gradually decreases to equilibrate further into the base plate.

At an offset of 3 mm (Figure 4.11.1(b)), a somewhat similar trend is observed, with a maximum stress of 660 MPa in the transverse direction, 80 MPa lower than the peak stress measured at an offset of 1.6 mm. This could be due to non-uniform plastic deformation occurring closer to the stress concentration created by the wagon track near the 3 mm offset position. The highest longitudinal stresses were measured in the base material, with maximum values of 599 MPa at an offset of 1.6 mm and 533 MPa at an offset of 3 mm. The higher longitudinal stresses measured in the base material can be attributed to the higher level of restraint caused by the greater plate thickness in this area (as compared to the weld throat thickness).

The peak in the centre of the weld is due to this region of the weld being the last to cool. It experiences the highest elastic strain due to the constraint provided by the material that has already solidified and contracted around it. As the throat thickness of the weld decreases towards the wagon tracks, a smaller cross-sectional area is available to resist the increasing stress on cooling and the residual stress in this portion of the weld exceeds the yield strength sooner than any other region. Localised plastic deformation decreases the residual stress to a level equal to the yield strength of the weld at the temperature where deformation occurred. A second stress peak is observed in the base material in a position that coincides with the top edge of the single-V weld preparation where the base material plate again assumes its full thickness. The contracting weld bead exerts an elastic tensile stress on the surrounding plate and the highest level of restraint is achieved at the point where the base material closest to weld bead is at its thickest. The maximum residual stress in the base material is therefore achieved at this point.

The residual stresses measured during the course of this investigation are consistently high, brought about in part by the short restraint length used in the MWIC samples. The stresses in the normal direction are much lower than those in the transverse and longitudinal directions since there is less material and therefore less constraint in this direction. In the transverse direction, free contraction is impeded by the anchor welds and the base material, whereas in the longitudinal direction contraction is hindered by the weld bead itself, which extends from the X70 plate material into the run-on/run-off tabs to the sides of the T-pieces.
Figure 4.11.1 Residual stress measurements at offsets of (a) 1.6 mm and (b) 3.0 mm from the underside of the joint, and (c) the resultant Von Mises stress for both offset levels, for a weld deposited at a heat input of 0.5 kJ/mm and a preheat temperature of 50°C.
The Von Mises criterion stipulates that plastic yielding in the presence of a triaxial stress state will only occur if the resultant Von Mises stress exceeds the yield strength of the material [7]. The individual stress components in any of the three directions can therefore exceed the yield strength of the material, but yielding will not occur as long as the resultant stress remains below the yield strength. The resultant residual stress cannot exceed the yield strength of the material since yielding lowers the stress to the level of the yield strength. Even though the individual stresses measured in the transverse and longitudinal directions are high, the resultant Von Mises stresses are expected to reach maximum values coinciding with the yield strength of both the weld metal and that of the X70 plate material in all cases.

Figure 4.11.1(c) shows the Von Mises stresses for the triaxial stress states at both offsets. The same trends are evident as those observed for the individual stress components, with the resultant stress varying with bead geometry. The minimum yield strength of weld metal deposited using E6010 electrodes is about 410 MPa, and that of X70 plate around 480 MPa. The values for the Von Mises stress shown in Figure 4.11.1(c) seem to fall within this range, with the peak value at the weld centreline at an offset of 1.6 mm recorded as about 484 MPa. It should be noted that the yield strengths of the parent material and the weld metal can be higher than the values given above, as these are quoted as minimum values. Slight variations in yield strength are also expected as a function of heat input and dilution.

Figures 4.11.2 and 4.11.3 show the residual stress distribution in welds deposited at a preheat temperature of 50°C, but at higher heat inputs of 0.7 kJ/mm and 1.0 kJ/mm. The same trends as those observed earlier are evident at both heat input levels. Maximum transverse stresses were measured as 680 MPa and 646 MPa for offsets of 1.6 mm and 3.0 mm, respectively. Corresponding maximum longitudinal stresses were recorded as 555 MPa and 559 MPa for offsets of 1.6 mm and 3.0 mm. In Figure 4.11.2, the longitudinal stresses reach their maximum values in the weld metal, as opposed to in the base plate as shown in Figure 4.11.1. Strong peaks were, however, observed in the base metal corresponding to the edge of the single-V weld preparation where full plate thickness is achieved adjacent to the joint. The strong weld metal peak in the longitudinal direction shown in Figure 4.11.2 can be attributed to the larger weld bead deposited at a higher heat input, resulting in higher restraint against free contraction. The difference between the maximum stress levels in the transverse direction measured at offsets of 1.6 mm and 3.0 mm is less pronounced in the thicker weld bead shown in Figure 4.11.2. The thicker weld bead creates more constraint, resulting in less yielding closer to the top surface of the weld.

In Figure 4.11.3, the peak transverse stresses in the centre of the weld reach values of 676 MPa and 787 MPa at offsets of 1.6 mm and 3.0 mm, respectively, but in this weld, the maximum residual stresses were measured in the base metal. The peak stress at 3 mm offset (735 MPa) is about 154 MPa higher than the peak stress at 3 mm offset (889 MPa). This difference is higher than that observed in Figures 4.11.1 and 4.11.2, suggesting that less deformation took place due to the larger radius on the wagon track rendering it a less severe stress concentrator. The longitudinal residual stresses in Figure 4.11.3 reach maximum values of 564 MPa and 633 MPa in the base metal at offsets of 1.6 and 3.0 mm, respectively.

The Von Mises stresses in both figures appear to peak at values below 500 MPa close to the centre of the welds. These values correspond well with the expected yield strengths of the weld metal and base material.
Figure 4.11.2 Residual stress measurements at offsets of (a) 1.6 mm and (b) 3.0 mm from the underside of the joint, and (c) the resultant Von Mises stress at both offset levels, for a weld deposited at a heat input of 0.7 kJ/mm and a preheat temperature of 50°C.
Figure 4.11.3 Residual stress measurements at offsets of (a) 1.6 mm and (b) 3.0 mm from the underside of the joint, and (c) the resultant Von Mises stress for both offset levels, in a weld deposited at a heat input of 1.0 kJ/mm and a preheat temperature of 50°C.
The heat input does not appear to affect the maximum stress values in any direction in any significant way. The maximum transverse stresses at the centre of the weld bead decreased from 740 MPa to 680 MPa across the heat input range tested for at an offset of 1.6 mm, and increased from 660 MPa to 787 MPa for an offset of 3.0 mm. These differences are most likely not statistically relevant. The effect of heat input on residual stress is discussed in more detail later.

The Von Mises stresses are contained within the minimum yield strength range of around 410 MPa to 480 MPa, with peak values of 471 MPa at an offset of 1.6 mm and 455 MPa at an offset of 3.0 mm in Figures 4.11.2 and 4.11.3.

Figures 4.11.4 and 4.11.5 show the residual stresses in welds deposited at a constant heat input of 0.7 kJ/mm, without preheat and at a preheat temperature of 80°C, respectively. In both figures, the maximum transverse and longitudinal stresses are found in the base metal in a position corresponding to the location where full plate thickness is achieved at the edge of the single-V joint preparation. In Figure 4.11.4, representing welds deposited without preheat, the peak stresses in the transverse direction were measured as 613 MPa and 649 MPa at offsets of 1.6 mm and 3.0 mm, respectively. The maximum stresses in the weld beads were measured as 602 MPa and 576 MPa for the two offsets, with the differences between the peak stresses for the two offsets less pronounced than that observed in Figures 4.11.1 to 4.11.3. The longitudinal stresses in Figure 4.11.4 peaked at 497 MPa and 527 MPa for offsets of 1.6 mm and 3.0 mm.

In Figure 4.11.5, considerably higher stresses were measured in the parent material than in the weld metal in all three directions. The transverse stresses peaked at 689 MPa and 614 MPa in the base metal, with maximum values of 521 MPa and 422 MPa in the centre of the weld. The longitudinal stresses reached maximum values of 609 MPa and 546 MPa at the respective offsets.

Even though the differences between the maximum stresses observed in Figure 4.11.5 are small, a distinct difference is evident between the residual stress distributions measured for the two offsets. There is a sharp decrease in the residual stress corresponding to the wagon track on the right. The throat thickness in this part of the weld is less than that on the left. With the 3.0 mm offset located closer to the top surface of the weld, and hence closer to the stress concentration caused by the wagon track, localised plastic deformation was assumed to be confined to this part of the weld since it offered less resistance due to the smaller cross-sectional area, decreasing the local stress accordingly. More plastic deformation occurred at the wagon track on the right, resulting in the characteristic drop in residual stress not being as pronounced on the left side of the weld bead.

The Von Mises stresses in Figures 4.11.4 and 4.11.5 are again well within expected ranges. The maximum stresses are 439 MPa at an offset of 1.6 mm, and 456 MPa at an offset of 3 mm. The calculated Von Mises stresses correlated well with the expected weld metal and parent material yield strengths, with little variation between the measured values.

The residual stress levels did not vary significantly between the two offset levels. The results suggest that the peak stresses observed in the welds were affected by the weld bead geometry due to localised plastic deformation in parts of the weld metal. The heat input and preheat temperature did not appear to affect the peak stresses recorded in any significant way. Similar values were found in welds deposited at low heat input levels and lower preheat temperatures than in welds deposited at high heat input levels and higher preheat temperatures.
Figure 4.11.4 Residual stress measurements at offsets of (a) 1.6 mm and (b) 3.0 mm from the underside of the weld, and (c) the resultant Von Mises stress for both offset levels, in a weld deposited without preheat at a heat input of 0.7 kJ/mm.
Figure 4.11.5 Residual stress measurements at offsets of (a) 1.6 mm and (b) 3.0 mm from the underside of the joint, and (c) the resultant Von Mises stress for both offset levels, in a weld deposited at a heat input of 0.7 kJ/mm and a preheat temperature of 80°C.
Figures 4.11.6 and 4.11.7 shows detailed comparisons between the residual stresses measured across weld beads deposited at heat input levels of 0.5 kJ/mm, 0.7 kJ/mm and 1.0 kJ/mm at a constant preheat temperature of 50°C. Figure 4.11.6 shows the residual stresses at an offset of 1.6 mm and Figure 4.11.7 at 3.0 mm.

![Graphs showing residual stresses](image1)

**Figure 4.11.6** Comparison of residual stresses as a function of increasing heat input levels of 0.5 kJ/mm, 0.7 kJ/mm and 1.0 kJ/mm for a constant preheat temperature of 50°C, measured at an offset of 1.6 mm from the underside of the joint.

No clear trend is observed with regard to a change in the residual stress level as a function of heat input. The only real difference visible between the stresses measured at the two offsets is that in Figure 4.11.6 (at an offset of 1.6 mm), the transverse stresses reach values of zero at a distance of around 7 to 10 mm from the weld centreline before equilibrating further into the base metal, while the longitudinal stresses reaches zero at a distance of about 17 to 20 mm from the weld centreline. At an offset of 3.0 mm, the situation reverses with the transverse stresses reaching zero at a distance of more than 17 mm from the weld centreline, while the longitudinal stresses reach zero at a distance of 8 to 9 mm from the weld centreline. The Von Mises stresses at both offsets show similar trends, equilibrating much further into the base metal.
Figure 4.11.7 Comparison of residual stresses as a function of increasing heat input levels of 0.5 kJ/mm, 0.7 kJ/mm and 1.0 kJ/mm for a constant preheat temperature of 50°C, measured at an offset of 3.0 mm from the underside of the joint.

Figures 4.11.8 and 4.11.9 compare the residual stresses measured in welds deposited at constant heat input, but using different preheat temperatures (no preheat, 50°C and 80°C), at offsets of 1.6 mm and 3.0 mm. Again, there appears to be no clear trend for the magnitude of the residual stress as a function of the preheat temperature. All welds display similar stress values with little difference between the residual stress distributions for the two offsets.

These results are in agreement with the findings of Alipooramirabad et al. [4] who, as mentioned previously, showed that the residual stress in MWIC test pieces become independent of heat input at restraint lengths below 50 mm for plate thicknesses between 12 to 20 mm. Although 10 mm thick plates were used for all welds in the current investigation, the conclusion drawn by Alipooramirabad et al. seem to hold for a restraint length of 25 mm, with heat input having little effect on the residual stress levels measured in the welds. Like heat input, preheat also did not affect the magnitude of the measured residual stresses to any significant extent. This suggests that the high restraint introduced during MWIC testing at lower restraint lengths was the dominant factor determining the residual stress levels in the welds tested.
**Figure 4.11.8** Comparison of residual stresses resulting from increasing preheating temperature (no preheat, 50°C and 80°C), for a constant heat input of 0.7 kJ/mm measured at an offset of 1.6 mm from the underside of the joint.
Figure 4.11.9 Comparison of residual stresses resulting from increasing preheating temperature (no preheat, 50°C and 80°C), for a constant heat input of 0.7 kJ/mm measured at an offset of 3.0 mm from the underside of the joint.

The triaxial stress state that is active in a material governs its cracking response. A crack tends to initiate and grow most readily when it is oriented perpendicular to the line of action of the stress acting on it. All cracks observed in the weld metal during this investigation initiated at wagon tracks and propagated longitudinally along the welds. The transverse component of the triaxial stress is therefore primarily responsible for the extensive WMHACC observed in the weld metal of the MWIC test pieces in this study.

From Figures 4.11.8 and 4.11.9, the transverse stress in the weld metal corresponding to the region of the wagon tracks varied from 250 to 472 MPa at both offset heights, depending on the geometry of the weld bead and the extent of localised plastic deformation in the weld metal on cooling. These values range from well below the room temperature yield strength of the weld metal, to very close to the yield strength. If the minimum yield strengths of the parent material and the weld metal are within the range of 410 to 480 MPa at room temperature, then in the presence of hydrogen, very little resistance is offered against WMHACC at these operative transverse residual stress levels as it is well established that WMHACC can occur at stress levels well below the yield strength.

In Figures 4.4.6, 4.4.11 and 4.4.16, in which the lines predicting WMHACC probabilities of 10%, 50% and 90% were superimposed on the original crack/no crack data sets, a large amount of scatter was evident between the lines of 10% and 90% probability of WMHACC. This scatter could, in part, be due to fluctuations in the transverse component of the residual stress, which varies as a function of the variables discussed above.

It can therefore be concluded that the magnitude of the peak tensile residual stress in the weld metal is approximately the same for all the MWIC samples tested. The high level of restraint, and hence the residual stress, therefore plays a significant role in promoting WMHACC, but the trends observed in the crack/no crack boundaries shown in Figures 4.3.1, 4.3.3, 4.3.5, 4.3.6, 4.4.5, 4.4.10 and 4.4.15 are more likely attributable to the cooling rate and resultant the contributory factors of bead geometry, hydrogen content, hardness, metallurgical transformation products and non-metallic inclusion content in the weld metal.

At lower restraint intensities, such as at MWIC restraint lengths higher than 50 mm, it is likely that higher cooling rates could be accommodated by the weld metal before the onset of WMHACC, resulting in the
crack/no crack boundaries shifting to lower heat input levels. A higher hydrogen concentration, and more brittle transformation products with higher hardness and higher inclusion levels, can most likely be tolerated within the weld metal as a consequence of a lower restraint intensity.

Therefore, the values of these variables along the 50% probability of WMHACC line, shown in each section of this study for the contributory factors discussed above, pertain to the stress levels shown in this section. If the combination of residual stress and the maximum lifting stress during the welding of X70 pipelines with E6010 electrodes is better simulated by MWIC test pieces with a restraint length of 100 mm, as suggested by Alam et al. [5], then the results of this investigation can be viewed as conservative and as providing a safeguard against the most severe weld defects in the weld.

REFERENCES


CHAPTER 5
CONCLUSIONS AND RECOMMENDATIONS

This chapter summarises the most important conclusions drawn during the course of this investigation, comments on the implications of this project for the Australian pipeline industry and provides some recommendations for future work.

5.1 CRACK/NO CRACK BOUNDARIES

- The WMHACC behaviour of root beads in X70 linepipe steel welded with E6010 consumables appears to be dependent on the heat input, the preheat temperature and the welding current used to deposit the beads. Although higher preheat temperatures and heat inputs reduce the likelihood of WMHACC, cracks were observed at all preheat temperatures evaluated (i.e. at no preheat, 50°C and 80°C) and throughout the entire heat input range tested (0.3 to 1.0 kJ/mm). Welding at higher heat inputs or preheat temperatures therefore does not necessarily safeguard welds against cracking.

- When the incidence of WMHACC was examined simply on the basis of heat input and preheat temperature, only a small portion of the welding parameter matrix tested could be regarded as safe, with significant scatter observed in the unsafe portion. The risk of WMHACC could be more accurately estimated when its dependence on individual welding parameters was isolated. The welding current affects the arc force, which in turn affects the weld bead geometry. Shallow penetration is achieved at low welding currents, introducing an additional stress concentration at the weld root that increases the risk of WMHACC. With an increase in preheat temperature from no preheat to 50°C, there is an average increase in the minimum allowable welding speed required to prevent WMHACC of about 155 mm/min at each current level and with an increase in preheat temperature from 50°C to 80°C, there is an additional average increase in the allowable minimum welding speed of 273 mm/min at each welding current.

- Increasing preheat temperature and heat input affected the WMHACC behaviour in two significant ways: first, by reducing the cooling rate and allowing more time for hydrogen to effuse from the weld metal and second, by increasing the depth of penetration and reducing the severity of the stress concentration caused by incomplete penetration at the root of the weld.

5.2 MODELLING OF THE CRACK/NO CRACK BOUNDARIES: LOGISTIC REGRESSION ANALYSIS

- Bivariate correlation analysis showed that increasing the welding current has a significantly stronger effect on preventing WMHACC than reducing the welding speed.

- Statistical models of the predicted probability of WMHACC as a function of the welding current and welding speed were developed and are given by the following equations for welds deposited without preheat, and at preheat temperatures of 50°C and 80°C:

\[
\text{Probability}_{\text{cracking (RT)}} = \frac{e^{27.174 - 0.242 \text{(Current)} + 0.017 \text{(Speed)}}}{1 + e^{27.174 - 0.242 \text{(Current)} + 0.017 \text{(Speed)}}}
\]

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Probability_{cracking (50^\circ C)} = \frac{e^{70.652 - 0.656 (\text{Current}) + 0.037 (\text{Speed})}}{1 + e^{70.652 - 0.656 (\text{Current}) + 0.037 (\text{Speed})}} \\
Probability_{cracking (80^\circ)} = \frac{e^{26.247 - 0.328 (\text{Current}) + 0.03 (\text{Speed})}}{1 + e^{26.247 - 0.328 (\text{Current}) + 0.03 (\text{Speed})}}

- A comprehensive model for the predicted probability of WMHACC as a function of the welding current, travel speed and the temperature (preheat or ambient atmospheric) was also determined and is given by the following equation:

Probability_{cracking} = \frac{e^{37.781 - 0.311 (\text{Current}) + 0.021 (\text{Speed}) - 0.111 (\text{Temperature})}}{1 + e^{37.781 - 0.311 (\text{Current}) + 0.021 (\text{Speed}) - 0.111 (\text{Temperature})}}

These equations can be used to assess the predicted probability of WMHACC as a function of welding parameters during the development of welding procedure specifications, and allow for changes to be made to welding parameters to ensure that the probability of WMHACC remains within acceptable limits.

### 5.3 SOLIDIFICATION CRACKING

No correlation was found between the welding parameters and solidification cracking in the current study. Cracking appeared to occur at random throughout the parameter matrix tested. The conclusion can therefore be drawn that the experimental setup used in this investigation was not appropriate for assessing solidification crack susceptibility.

It is therefore recommended that future work focuses on establishing trends between solidification cracking and welding parameters, preferably using a test designed to evaluate this type of defect. If MWIC test pieces are used in a future study, the restraint length should be varied along with the welding parameters to determine the thresholds where constraint, or stress intensity, becomes an overriding factor in determining the relationship between solidification cracking and welding parameters.

The effect of variations in chemistry between E6010 electrodes from different suppliers also needs to be prioritised in future studies, especially with regard to elements such as sulphur, phosphorus and carbon.

### 5.4 WELD METAL COOLING RATE

- Mathematical models were developed for the two critical weld metal cooling times (i.e \( \Delta t_{8/5} \) and \( \Delta t_{11} \)) in MWIC test specimens as a function of the heat input at each of the three preheat conditions. The following three equations estimate the \( \Delta t_{8/5} \) cooling time (seconds), i.e. the time it takes for the weld metal to cool from 800\(^\circ\)C to 500\(^\circ\)C, as a function of the heat input (kJ/mm) for preheat-free welds, and for preheat temperatures of 50\(^\circ\)C and 80\(^\circ\)C:

\[
\Delta t_{8/5} (\text{RT}) = 9.4393x - 1.9411 \\
\Delta t_{8/5} (50^\circ \text{C}) = 9.4564x - 1.3434 \\
\Delta t_{8/5} (80^\circ \text{C}) = 9.9045x - 0.4705
\]

The \( \Delta t_{8/5} \) time provides an indication of the cooling rate through the critical temperature range where the solid state transformation of austenite to its transformation products occurs on cooling.
• Similar equations were developed to model the \( \Delta t_{8/1} \) cooling interval, i.e. the time required for the weld metal to cool from 800°C to 100°C:
\[
\Delta t_{8/1} (RT) = 77.517x + 6.0305 \\
\Delta t_{8/1} (50^\circ C) = 107.1x + 16.949 \\
\Delta t_{8/1} (80^\circ C) = 188.16x + 48.618
\]
The \( \Delta t_{8/1} \) time determines the amount of hydrogen effusion from the welds.

• Higher heat inputs and/or preheat temperatures increase the time required to cool through both temperature intervals. At lower welding currents, the heat input, and hence \( \Delta t_{8/5} \) and \( \Delta t_{8/1} \), required to maintain a predicted WMHACC probability of 50% decreased exponentially with an increase in welding current (and consequently depth of penetration). At welding currents above 130 A, where the stress concentration created by incomplete penetration became less severe, the heat input and preheat temperature required to maintain a 50% probability of failure decreased less rapidly and eventually stabilised.

• An increase in heat input also resulted in a thicker weld throat, which decreased the risk of WMHACC. Weld metal with a larger cross-sectional area can more effectively resist stresses imposed on the welds.

• The lowest heat input (i.e. the shortest cooling interval) required to maintain a 50% probability of cracking was obtained at welding currents of 160 A. Similar trends were observed for all factors contributing to the occurrence of WMHACC that are affected by heat input and cooling rate, such as the weld metal and heat-affected zone hardnesses, the weld metal throat thickness, and the non-metallic inclusions present in the weld metal.

5.5 WELD METAL AND HEAT-AFFECTED ZONE HARDNESS

• The hardness of both the weld metal and the heat-affected zone decreased linearly with an increase in heat input and preheat temperature. This can be attributed to minor changes in the weld metal and heat-affected zone microstructures and an increase in the overall coarseness of the microstructures.

• The difference in hardness between weld metal deposited at the minimum heat input of 0.3 kJ/mm and the maximum heat input of 1.0 kJ/mm was consistently around 55 HV for welds deposit at all three preheats. For both the weld metal and the heat-affected zones there was a decrease in hardness of about 7 HV for an incremental increase in heat input of 0.1 kJ/mm.

• An average difference in hardness between welds deposited without preheat and at a preheat temperature of 50°C was about 16 HV, and between welds deposited at preheat temperatures of 50°C and 80°C about 10 HV.

5.6 WELD METAL AND HEAT-AFFECTED ZONE MICROSTRUCTURES

• The weld metal and heat-affected zone microstructures were not affected in any significant way by changes in heat input or preheat temperature.
Two forms of primary ferrite were observed in the weld metal, namely grain boundary ferrite (PF(GB)), referred to as allotriomorphic ferrite, and primary idiomorphic ferrite (PF(I)) nucleated intragranularly on non-metallic inclusions. Other forms of ferrite observed in the weld metal included acicular ferrite (AF), Widmanstätten ferrite and upper bainite, both designated simply as either ferrite side-plates with an aligned second phase (FS(A)) or ferrite side-plates with a non-aligned second phase (FS(NA)) wherever their unambiguous identification were complicated by their similar appearance. Ferrite-carbide aggregates (FC) were also observed and were tentatively identified as degenerate pearlite, granular bainite and lower bainite.

In the high-temperature heat-affected zone microstructures upper bainite dominated. Lower bainite, ferrite-carbide aggregate and primary ferrite were also observed.

In both the weld metal and heat-affected zone microstructures a gradual coarsening was observed in both the prior austenite grain size and the transformation products with an increase in heat input and/or preheat temperature. A gradual increase in reconstructive (diffusion-controlled) transformation products was observed with a decrease in cooling rate (i.e. increasing heat input and/or preheat temperature) at the expense of displacive transformation products. This merely constituted a shift in the balance between the two types of transformation products as both types continued to be observed in all welds investigated.

Microcracks were observed to propagate through hard and brittle carbon-enriched microphases and continuous cementite seams, which facilitated crack propagation to a greater extent than softer ferritic transformation products.

The microstructures observed were consistent with the hardness measurements.

5.7 **NON-METALLIC INCLUSIONS**

- Non-metallic inclusions in the weld metal were observed to be spherical, incoherent oxide particles of Mn-Si-Al-Ti-O type. Some iron was also present within the inclusions. Nitrogen was often observed as part of the oxide, but most often associated with titanium (in the form of titanium nitride particles at the edges of the oxide inclusions). No sulphide inclusions were observed within the weld metal.

- The number of non-metallic inclusion, the total area covered and the inclusion volume percentage decreased, while their average size increased, with increasing heat input.

- Non-metallic inclusions reduces the ductility of weld metal, acts as irreversible hydrogen trapping sites, and can influence the hardness. Observations of microcracks ahead of macrocrack tips showed that these cracks tend to propagate between non-metallic inclusions.

5.8 **FRACUTRE SURFACES**

- In the current study, hydrogen cracks in the weld metal were observed to initiate exclusively at the wagon tracks at the toes of the weld metal. Cracks propagated downward through the weld metal to stress concentrations located near the root of the weld. The direction of crack growth was, for the main part, through the weld metal with a few exceptions deviating through the heat-affected zones.
No observable differences were found between the fracture surfaces of welds deposited at low and high heat input levels, and it is assumed that due to the similarities in microstructures and the high volume fractions of inclusions, similar mechanisms were active during crack propagation in at both heat input extremes.

Near the site of crack initiation at the wagon tracks, the fracture surfaces were composed of fine dimples interspersed with cleavage-type facets, which is typical of a quasi-cleavage mixed mode fracture. Towards the root of the weld metal the fracture surface appearance changed to a cleavage brittle mode of fracture, showing steps and shear ridges accompanied by extensive river markings.

5.9 RESIDUAL STRESS

Similar trends in residual stress distribution were observed in all the weld metals tested. A peak in the stress was recorded in the centre of the welds followed by a rapid decrease in stress in the vicinity of the wagon tracks. Another peak in residual stress was observed in the base metal corresponding to the edge of the single-V joint preparation where the base metal returns to its full thickness. The lower stress associated with the wagon tracks can be ascribed to a lower resistance to yielding on cooling due to the smaller cross-sectional area of the weld metal in this region. The residual stress distributions in all three directions (transverse, longitudinal and normal) therefore show a strong correlation with weld bead geometry.

The magnitude of the residual stress was not a found to be a strong function of the heat input or preheat temperature, and showed very little variance between the offset levels of 1.6 mm and 3.0 mm.

The transverse residual stress component is responsible for initiating and propagating weld metal hydrogen-assisted cold cracks in the weld metal. The measured transverse stresses corresponding to the region of the wagon tracks where cracks initiated varied from around 250 MPa to 472 MPa for both offset heights, with the peak stress depending on the geometry of the weld bead and the extent of localised plastic deformation in the weld metal on cooling.

SUMMARY OF CONCLUSIONS

The study focused on the effects of welding parameters on the WMHACC behaviour of the root-bead weld metal deposited on X70 linepipe steel using E6010 electrodes. The parameters affected the heat input and root bead geometry directly. The cooling rate of the weld metal is primarily dependent on the heat input of welding and the preheat temperature, and the cooling rate in turn affects a variety of factors which can either contribute to- or minimise the occurrence of WMHACC.

The most prominent effect of increasing the cooling rate is the lowering of the amount of diffusible hydrogen which is the main cause of WMHACC. A lowering of the cooling rate also results in a microstructure more resistant to WMHACC and alters the physical attributes of non-metallic inclusions (NMI’s). Higher heat inputs lowered the amount of NMI’s, their total area covered and volume percentage. Although in the current study crack initiation was not observed at NMI’s, they did serve to assist crack propagation, and therefore, fewer NMI’s at higher heat inputs contributed to a lowering of the incidence of WMHACC.