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Characterisation of transverse cold cracking in weld metal of a high strength quenched and tempered steel

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ABSTRACT

Hydrogen induced cold cracking in weldments is the most serious problem limiting the use of high strength low alloy steel in the structural industry. Improvement of the weldability of structural steel has significantly reduced the risk of cold cracking in the heat affected zone. Consequently, the avoidance of weld metal cold cracking becomes a major task since the risk of hydrogen induced cold cracking in the HAZ is greatly diminished in modern low carbon steels. Hydrogen induced cold cracking in weld metal is caused by the complex interaction of the diffusible hydrogen content, residual stress and susceptible microstructure. Depending on the joint geometry which induces stresses, cold cracking in weld metal can be of two types with respect to the welding direction: longitudinal cracking which is parallel to the welding direction, and transverse cracking which is perpendicular to the welding direction. Weld deposits for high strength low alloy steels can be susceptible to transverse cold cracking.

The investigation described in this thesis involved the use of the gapped bead-on-plate (G-BOP) test to examine the effect of shielding gas mixture and preheat temperature on weld metal transverse cold cracking behaviour in a flux cored arc welded BIS812EMA quenched and tempered steel. The minimum yield stress of BIS812EMA steel is 690MPa. The consumable used in welding was a low hydrogen flux cored wire.

Literature relevant to the characteristics of weld metal cold cracking has been reviewed, with emphasis on the factors affecting hydrogen induced cold cracking. The process of flux cored arc welding, the weld thermal cycle and the properties of weld metal have been examined. The main types of cold cracking and relevant weldability tests have been discussed.

Two different Ar-CO₂ shielding gas mixtures (Ar-5%CO₂ and Ar-20%CO₂) and five different preheat temperatures (20°C, 40°C, 60°C, 80°C and 100°C) were used at an aim heat input of 1.7 kJ/mm to investigate the effect of welding conditions on the
transverse cold cracking behaviour of the weld metal deposited on BIS812EMA steel. The microstructures, hardness values and fracture surfaces of the weld metals, have been examined in detail using optical and electron microscopy and microhardness testing. The characteristics of non-metallic inclusions in the weld metals have been determined using scanning electron microscopy and automatic image analysis. The continuous cooling transformation diagrams of the base metal and the weld metals have been developed using dilatometry and microstructural observation. Additionally, precipitation hardening by alloy carbides has been studied by tempering of weld metals.

It was found that, below a critical preheat temperature, weld metal transverse cracks developed in the self restrained G-BOP test at least five minutes after welding and initiated below 150°C, confirming that the weld metal transverse cracking was cold cracking.

Shielding with Ar-20%CO₂ mixture resulted in significantly lower susceptibility to transverse cold cracking than shielding with Ar-5%CO₂ mixture. However, the microstructures of the weld metals deposited with the two different Ar-CO₂ shielding gas mixtures were similar and mainly consisted of acicular ferrite and bainitic ferrite. Comparing the two weld metals, the higher oxygen potential of the Ar-20%CO₂ shielding gas mixture caused reduction of hardenability elements such as carbon, boron, manganese and silicon in the weld metal; and the higher oxygen absorption during welding resulted in a higher non-metallic inclusion volume fraction and a slightly larger mean particle size and number of particles per unit volume. These factors led to phase transformation during cooling at higher temperature in weld metal deposited with Ar-20%CO₂, a slightly coarser microstructure with more bainitic ferrite, and a significantly lower hardness than the weld metal deposited with Ar-5%CO₂. In addition, the weld metal deposited with Ar-20%CO₂ shielding gas mixture had a lower diffusible hydrogen level and less nitrogen in solid solution. These characteristics combined with the lower hardness value of the weld metal are considered to be the main
factors contributing to the lower susceptibility to transverse cold cracking in the weld metal deposited with Ar-20%CO₂ compared with Ar-5%CO₂ under the same welding conditions.

Three types of fracture morphologies were found to be associated with the weld metal transverse cold cracking: microvoid coalescence (MVC) fracture, quasi-cleavage (QC) fracture and intergranular (IG) fracture. IG was more prevalent in the weld metal deposited with Ar-5%CO₂ shielding gas mixture while MVC dominated in the weld metal deposited with Ar-20%CO₂. These trends are consistent with the difference in weld metal hardness (strength).

The susceptibility to weld metal transverse cold cracking, based on the G-BOP test, decreased with increasing preheat temperature for both shielding gas mixtures. No cracking was observed for the weld metal deposited with Ar-20%CO₂ shielding gas mixture for a preheat of 40°C, but a preheat of 100°C was required for crack-free welding using Ar-5%CO₂ shielding gas mixture. However, it is concluded that the finer microstructure of the weld metal deposited with Ar-5%CO₂ is capable of both higher strength and lower cold cracking susceptibility than the weld metal deposited with Ar-20%CO₂ provided the hardnesses are similar and below a critical level of about 290 HV.
TABLE OF CONTENTS

ACKNOWLEDGEMENTS i

ABSTRACT ii

TABLE OF CONTENTS v

CHAPTER 1 INTRODUCTION 1

CHAPTER 2 FLUX CORED ARC WELDING 7

2.1 SELF-SHIELDED FLUX CORED ARC WELDING 8

2.2 GAS SHIELDED FLUX CORED ARC WELDING 9

2.3 FLUX CORED WIRE 9

2.4 SHIELDING GAS 10

CHAPTER 3 THE WELD THERMAL CYCLE AND PHASE TRANSFORMATION 12

3.1 THE WELD THERMAL CYCLE 13

3.2 PHASE TRANSFORMATION DURING COOLING 14

3.2.1 Solidification of the Weld Metal 14

3.2.2 Phase Transformations in the Weld Metal 15
3.2.3 Non-Metallic Inclusions in the Weld Metal

3.3 CONTINUOUS COOLING TRANSFORMATION DIAGRAMS

3.3.1 Metallographic Method

3.3.2 Dilatometry

CHAPTER 4 CRACKING IN WELDS AND WELDABILITY TESTING

4.1 THE MAIN TYPES OF COLD CRACKING

4.1.1 Lamellar Tearing

4.1.2 Chevron Cracking

4.1.3 Hydrogen Induced Cold Cracking

4.2 FACTORS AFFECTING COLD CRACKING

4.2.1 Hardenability

4.2.2 Hydrogen

4.2.3 Microstructure

4.2.4 Stresses

4.2.5 Nitrogen

4.2.6 Preheating
4.2.7 Heat Input 41

4.3 WELDABILITY TESTING 41

4.3.1 The Gapped Bead-On-Plate Test 42

4.3.2 The Tensile Restraint Cracking Test 45

4.3.3 The Longitudinal Bead-Tensile Restraint Cracking Test 45

4.4 AIMS OF THE PRESENT INVESTIGATION 47

CHAPTER 5 EXPERIMENTAL METHOD AND MATERIALS 49

5.1 MATERIALS 50

5.2 WELDING PROCEDURES 51

5.2.1 Test Method 51

5.2.2 Welding Parameters 52

5.3 METALLOGRAPHY 54

5.3.1 Optical Metallography and Weld Bead Analysis 54

5.3.2 Hardness 54

5.3.3 Measurement of Width of Columnar Grains 55

5.4 FRACTOGRAPHY 55
5.4.1 Estimation of Cold Crack Percentage

5.4.2 Fracture Surface Examination

5.4.3 Measurement of Microvoid Size on Fracture Surface of Weld Metal

5.4.4 Measurement of the Percentage of Different Fracture Morphologies

5.4.5 Observation of Cross Section of the Fracture Surface

5.5 NON-METALLIC INCLUSION ANALYSIS

5.6 CONSTRUCTION OF CONTINUOUS COOLING TRANSFORMATION DIAGRAMS

5.6.1 Dilatometry

5.6.2 Thermal Cycles

5.6.3 Determination of Ar\textsubscript{1} and Ar\textsubscript{3}

5.6.4 Optical Metallography

5.6.5 Hardness

5.6.6 Measurement of Prior Austenite Grain Size

5.7 TEMPERING OF WELD METALS
CHAPTER 6 RESULTS

6.1 MEASUREMENT OF THERMAL CYCLES IN G-BOP TEST 65

6.2 LONGITUDINAL STRESS DURING SOLIDIFICATION 66

6.3 METALLOGRAPHY 67

6.3.1 Geometry of Weld Bead and Dilution 67

6.3.2 Weld Metal Compositions 68

6.3.3 Optical Microstructure 71

6.3.4 Hardness 72

6.3.5 Columnar Grains 73

6.4 FRACTOGRAPHY 74

6.4.1 Estimation of Percentage Cracking 74

6.4.2 Fracture Morphology 76

6.4.3 Microvoid Size on Fracture Surface of the Weld metals 77

6.4.4 Percentages of Different Fracture Morphologies 78

6.4.5 Microstructures Underlaying the Fracture Surface 79

6.5 NON-METALLIC INCLUSION ANALYSIS 80
6.5.1 Non-Metallic Inclusion Characteristics

6.5.2 Size Distribution of Non-Metallic Inclusions

6.5.3 Alloiying Elements in the Non-Metallic Inclusions

6.6 CONTINUOUS COOLING TRANSFORMATIONS AND MICROSTRUCTURES

6.6.1 Continuous Cooling Transformation (CCT) Diagrams

6.6.2 Microstructures

6.6.3 Hardness

6.6.4 Prior Austenite Grain Size of Dilatometer Samples

6.7 METALLOGRAPHY OF TEMPERED WELD METALS

6.7.1 Hardness

6.7.2 Microstructure

CHAPTER 7 DISCUSSION

7.1 EFFECT OF PREHEAT ON WELDING THERMAL CYCLES

7.2 EFFECT OF PREHEAT ON LONGITUDINAL STRESS AND CRACKING
7.2.1 Longitudinal Stress Development 97

7.2.2 Effect of Preheat 98

7.3 EFFECT OF SHIELDING GAS ON WELD BEAD GEOMETRY 99

7.4 WELD METAL MICROSTRUCTURE ANALYSIS 100

7.4.1 Effect of Preheat Temperature on Microstructure 100

7.4.2 Effect of Shielding Gas on Composition and Microstructure of Weld Metal 101

7.4.3 Columnar Grain Size of the Weld Metals 105

7.4.4 Effect of Non-Metallic Inclusions on Microstructure 106

7.5 THE HARDNESS OF THE WELD METALS 108

7.5.1 Effect of Preheat Temperature 108

7.5.2 Effect of Shielding Gas 109

7.6 WELD METAL TRANSVERSE COLD CRACKING 110

7.6.1 Effect of Preheat on Weld Metal Cold Cracking 111

7.6.2 Effect of Diffusible Hydrogen 113

7.6.3 Effect of Shielding Gas on Weld Metal Cold Cracking 115
7.6.4 Weld Metal Cold Cracking Behaviour in Test Assembly with Load Cells 117

7.6.5 Analysis of Fracture Morphologies 118

7.7 CONTINUOUS COOLING TRANSFORMATIONS 126

7.8 TEMPERED WELD METALS 132

7.9 DETAILED COMPARISON OF THE TWO WELD METALS 133

7.9.1 Microstructural (Property) Differences 134

7.9.2 Diffusible Hydrogen Concentration 136

CHAPTER 8 CONCLUSIONS 137

REFERENCES 143

APPENDICES 160

APPENDIX I Plots of the Thermal Responses for Welding using Ar-20%CO₂ Shielding Gas Mixture at Different Preheat Temperatures in G-BOP Test 161

APPENDIX II Typical EDS Spectra for Non-metallic Inclusions in the Two Weld Metals Deposited at Different Preheat Temperatures 165

PUBLICATIONS 172
CHAPTER 1

INTRODUCTION
The use of welding to join steel has grown rapidly in recent decades, with the growth being marked by many changes. One of these changes has been welding process development, which has seen manual processes decline in favour of more efficient automated processes [Sare 1989]. Another important change has been the improvement in steel quality and properties, such as the development of high strength low alloy steels. However, cracking in weldments, especially hydrogen induced cold cracking, remains one of the most serious problems facing the steel fabrication industry today.

Flux cored arc welding (FCAW) process has the potential to improve productivity due to its high operating efficiency and deposition rate. The important functions of the shielding gas in the FCAW process include protection of the arc and weld pool from contamination by air, provision of a suitable medium for the stable operation of the arc, and control of weld bead geometry and mechanical properties.

High strength low alloy (HSLA) steels are being more widely used in applications such as pipelines, pressure vessels, tanks, and steel structures of all kinds [Savioli 1976]. A large amount of welding is employed in these fields, and it is therefore essential that the steel possesses good weldability.

Weldability has been defined as 'the capacity of a metal to be welded, under fabrication conditions imposed, into a specific, suitable designed product and to perform satisfactorily in the intended service' [A. S. 25-1968; Definitions of Metallurgical Terms 1971]. Basically, there are four extrinsic factors that can be important in determining the weldability of a steel, namely: (1) the steel composition, (2) the joint design, (3) the welding process, and (4) the environment. If one of these factors is unsuitable it may give rise to a cracking problem. Therefore, weldability can also simply be defined in terms of susceptibility to the various types of cracking problems known to be associated with welds [Easterling 1992]. To assess weldability in relation to a cracking problem would require full understanding of the interaction between above variables.
Weld cracking is the most dangerous defect of welding assemblies. Weld cracking can be classified as longitudinal and transverse cracking. Considerable research work has been conducted on the cracking of welds [Di et al. 1994; Hart 1986; Interrante and Stout 1964; Sawhill et al. 1986; Fu et al. 1993; Signes and Howe 1988; Savage et al. 1976; Alam and Dunne 1994] and it has been established that cracking is strongly dependent on the welding technique and the welding parameters. In recent years, the weldability and mechanical properties of structural steel have been significantly improved in the following ways:

(1) the introduction of steels with lower carbon content;

(2) the application of various thermomechanical treatments to enhance strength and toughness by grain size and microstructure control; and

(3) the availability of high quality low hydrogen welding consumables.

Therefore, the risk of HAZ cracking has been reduced and the avoidance of weld metal cracking is becoming increasingly more important. This survey is concerned mainly with hydrogen induced cold cracking in weld metal.

For the last fifty years hydrogen induced cold cracking has been synonymous with the weldability of steels. This type of cracking occurs after the weldment has cooled to below 200°C [Lundin et al. 1990] and is normally delayed for a few hours to a few days while atomic hydrogen diffuses to high tensile stress fields, in which the hydrogen changes to its molecular form at microstructural defects and causes cracking.

The factors that govern the susceptibility of a weldment to hydrogen induced cold cracking can be broadly classified as: (a) presence of hydrogen during welding, (b) a susceptible microstructure, and (c) presence of stresses, residual or applied [Easterling 1992; Yurioka and Suzuki 1990]. All three factors are interrelated: the critical hydrogen concentration necessary for cold cracking can be reduced with increasing stress or in the presence of a more susceptible microstructure.
The hydrogen content in a weldment is dependent on both the hydrogen source and the ability of the weldment to diffuse hydrogen from the weld metal to the HAZ. Diffusible hydrogen level plays a crucial role in initiating cold cracking. For a given weld metal and HAZ hardness-strength combination, the susceptibility of weld metal cold cracking rises with increasing diffusible hydrogen level.

The microstructures of weld metal are largely dependent on the chemical compositions of the consumable and the base metal; the amount of weld metal dilution; the size and distribution of non-metallic inclusions; the prior austenite grain size; and the cooling rate of the weldment. Martensite is very susceptible to cold cracking and avoidance of martensitic structures is predicted through control of cooling rate. Another important microstructure in weld metal is acicular ferrite, which can improve weld metal toughness and strength.

The microstructure of weld metal can be strongly affected by non-metallic inclusions. Inclusions in weld metal can aid nucleation of acicular ferrite within the austenite grains [Abson et al. 1978; Choi and Hill 1978; Ricks et al. 1982; Ferrante and Farrar 1982; Ohkita et al. 1984]; or they may pin the prior austenite grain boundaries, tending to reduce austenite grain size and favour higher temperature transformation products [Harrison and Farrar 1981]. Inclusions also can influence the resistance to hydrogen induced cold cracking by acting as crack nuclei [Tweed and Cohen 1987; Abson and Pargeter 1986; Chin 1969; Van Stone et al. 1985]. The effects of inclusions on the microstructure and toughness of weld metal are mainly dependent on the inclusion type, morphology and size distribution.

The continuous cooling transformation (CCT) diagram relates the composition, cooling rate and austenite grain size of an alloy to its $\gamma$ to $\alpha$ transformation temperature and resultant microstructure. CCT diagrams are widely used for describing the phase transformations during the welding thermal cycle; and diagrams for weld metal have improved understanding of weld metal microstructural development and the influences
of factors, such as chemical composition, oxygen content, welding parameters and prior austenite grain size on the $\gamma$ to $\alpha$ transformation behaviour of weld metal [Abson and Dolby 1978; Kenny et al. 1985; Harrison and Farrar 1987; Kluken et al. 1991; Farrar et al. 1993].

Residual stresses are always present in weldments. These stresses result from non-uniform thermal contractions and transformation stresses which occur as the weld cools under constraint [Davidson 1995]. Residual stresses provide the driving force for hydrogen induced cold cracking. The presence of hydrogen appears to lower the stress level at which cracking occurs.

A common way to reduce the susceptibility to cold cracking in HSLA steels is preheating, which is performed to control the cooling rate and to ensure sufficient removal of hydrogen from the weld metal. Preheat will also allow more time for the thermal relaxation of residual stress by reducing the weld metal cooling rate. Other methods to prevent cold cracking in weld metal include: using a low hydrogen process and consumables; microstructure improvement; careful selection of joint geometry, welding parameters and stress relief treatment; and post weld heat treatment.

Many weldability tests have been developed and can be used to either explain a particular failure or to prevent a similar occurrence under similar circumstances in the field. Growing interest in weld metal cold cracking [McParlan and Graville 1976], initially associated with certain flux cored consumables, led to the development of the gapped bead-on-plate (G-BOP) test. The effects of chemical composition, hydrogen content, cooling rate and preheat on weld metal hydrogen induced cold cracking can be studied using the G-BOP test.

The FCAW process is reviewed in Chapter 2. The weld thermal cycle and phase transformation during cooling are reviewed in Chapter 3. Chapter 4 considers the characteristics of cracking in welds and describes three weldability tests.
The basic aim of the research work presented in this thesis is to study the effects of preheat treatment and shielding gas mixture on the behaviour of weld metal transverse cold cracking in flux cored arc welded BIS812EMA quenched and tempered steel by using the G-BOP test. A more detailed description of the work is given in Section 4.6.
CHAPTER 2

FLUX CORED ARC WELDING
Figure 2.1 Principle of the self-shielded flux cored arc welding process [After Welding Handbook Vol.2 1991]
Flux cored arc welding (FCAW) is defined as a welding process that uses an arc between a continuous filler metal electrode and the weld pool [Welding Handbook Vol. 2 1991]. The process is used with shielding from a flux contained within a tubular electrode. Additional shielding may or may not be required from an externally supplied gas or gas mixture.

The composition, structure and properties of welds produced by the FCAW can be affected by the following metallurgical factors,

(1) the fluxing materials contained in the electrode;
(2) the alloying elements contained in the electrode;
(3) the shielding atmosphere in which the weld is made; and
(4) the composition of the base metal.

The FCAW process combines the productivity of continuous welding with the benefits of having a flux present. Therefore, weld metal can be deposited at higher rates and efficiencies, especially for out-of-position welding, and the welds can be larger and better contoured than those made with solid electrodes.

Two major process variations are offered in flux cored arc welding in the method of shielding to protect the arc and weld pool from atmospheric contamination. The variations are the self-shielding and gas shielding FCAW processes.

2. 1 SELF-SHIELDED FLUX CORED ARC WELDING

In the self-shielded FCAW process, the flux cored wire must provide sufficient shielding through the decomposition and vaporization of core ingredients to protect the molten metal droplets from the atmosphere as they form and transfer across the arc. Many self-shielded electrodes contain substantial amounts of deoxidizing and denitrifying ingredients to cope with the inevitable pick up of nitrogen and oxygen in the welding process. Self-shielded electrodes should also provide arc stabilization, alloy addition and slag control functions. Figure 2.1 shows a schematic illustration of self-
Figure 2.2 Principle of the gas shielded flux cored arc welding process [After Welding Handbook Vol.2 1991]
Seamless tube (may be copper-coated)

Butt seam

Joggle seam

Complex section

(a) Outer sheath
(b) Flux powder

Figure 2.3 Alternative configurations for flux cored wires [After Norrish 1992]
shielded flux cored arc welding process. The self-shielded FCAW process is preferred for field use.

2.2 GAS SHIELDED FLUX CORED ARC WELDING

In the gas shielded FCAW process, an additional applied shielding gas envelopes the arc and the weld pool to protect the molten metal from nitrogen and oxygen in the atmosphere. A schematic illustration of gas shielded flux cored arc welding process is shown in figure 2.2. Either carbon dioxide or argon/carbon dioxide mixtures are commonly used for steel as shielding gas in the FCAW process, the additional shielding gas allows the positional performance, mechanical properties and process tolerance to be improved. Compared with a self-shielded electrode, the ingredients in a gas shielded electrode are for fluxing, deoxidizing, scavenging and alloying additions, rather than for these functions plus the generation of protective vapors. The gas shielded flux cored arc welding process offers high productivity, versatility and good useability, and consequently it is a widely used welding process, which is growing in popularity.

2.3 FLUX CORED WIRE

The flux cored wire is a composite tubular filler metal electrode consisting of a metal outer sheath and a core with a combination of mineral flux and metal powders. The flux core may contain minerals, ferroalloys and materials that provide shielding gases, deoxidizers, slag forming materials, and alloys for the deposit. During welding a substantial slag cover is produced on the face of a weld metal to protect the solidifying weld metal. Flux cored wires have many advantages, such as high deposition rates, alloying addition from the flux core, slag shielding and support, and improved arc stabilization and shielding [Norrish 1992].

Figure 2.3 shows alternative configurations for flux cored wires. Typical finished wire diameters range from 3.2 to 0.8 mm. Three groups of flux cored wires have been developed: (1) plain carbon and alloy steel, which can be subdivided into: (i) rutile gas
shielded, (ii) basic gas shielded, (iii) metal cored, and (iv) self shielded; (2) hard facing alloys; and (3) stainless steel [Norrish 1992].

2.4 SHIELDING GAS

Contamination of the weld metal can result in low strength, low ductility and excessive weld defects, such as porosity and lack of fusion. Therefore, shielding gas plays an important role in the arc welding process. The shielding gases used in the FCAW process have a substantial effect on the form of metal transfer during welding. They have the dual purpose of protecting the arc and weld pool from contamination by air and providing a suitable medium for the stable operation of a sustained low-voltage arc. The efficiency, quality and overall operator acceptance of the welding operation can be affected by the shielding gas.

The basic strength, toughness and corrosion resistance of the weld are produced by the shielding gas interacting with the base metal and with the filler materials. The shielding gas can affect the following factors:

(1) arc and metal transfer characteristics during welding;
(2) penetration, width of fusion and shape of reinforcement;
(3) speed of welding; and
(4) undercutting tendency.

The basic shielding gases are argon, helium, carbon dioxide, oxygen and hydrogen. A mix of gases is generally required to obtain a shielding gas that satisfies the requirements of most material-process combinations. Each basic gas contributes certain characteristics to the performance of the overall mix. Gas mixtures are as follows, argon-oxygen; argon/carbon dioxide; argon/helium; argon/oxygen/carbon dioxide; argon/helium/carbon dioxide and argon/carbon dioxide/hydrogen. The most commonly used shielding gases for FCAW are carbon dioxide and argon/carbon dioxide mixtures [ASM handbook Vol. 6 1993].
In the FCAW process, the higher the percentage of argon in the mixture with CO₂, the higher will be the transfer efficiencies of the deoxidizers contained in the core. The mixture commonly used is 75 percent. argon with 25 percent. carbon dioxide. Weld metal deposited with this shielding gas mixture normally has higher tensile and yield strengths than 100 percent. carbon dioxide shielding gas [ASM handbook Vol. 6 1993].
CHAPTER 3

THE WELD THERMAL CYCLE

AND PHASE

TRANSFORMATION
3.1 THE WELD THERMAL CYCLE

The change of the temperature gradients in the weld pool and the heat affected zone associated with a moving power source is called the weld thermal cycle. This cycle exerts a major effect on the weld quality. Based on previous research results, the following four thermal factors are usually considered during welding [Adams 1958; Rosenthal 1946; Inagaki and Seiguchi 1960; Pugin and Persovskii 1963; Easterling 1992]:

1. Heating rate: The heating rate of welding is, generally, much higher than that of heat treatment, leading to an increase of the phase transformation temperature in steel. Heating rate depends on the welding method, the heat input and the thickness of the plate to be welded.

2. Peak temperature: In addition to chemical composition, the microstructure of steels is related to the peak temperature of heating and the subsequent cooling rate. Peak temperature is important in analyzing the microstructure of the heat affected zone.

3. The dwell time above the phase transformation temperature: In general, this time is short. The longer the dwell time, the more homogenous is the austenite. However, if the temperature is too high (above 1100°C), there is a strong tendency for austenite grain coarsening.

4. Cooling rate: The cooling rate is one of the most important factors influencing the microstructure and mechanical properties of the weld metal and the heat affected zone. For a given welding process, weld geometry and material, the cooling time through the range of 800-500°C (Δt_{85}) is found experimentally to be constant. Therefore, Δt_{85} can effectively represent the cooling time (cooling rate) for a given heat input, material type and weld geometry.
3.2 PHASE TRANSFORMATION DURING COOLING

The final microstructures of the weld metal and heat affected zone will depend on complex interactions between variables such as:

1. the total alloy content;
2. the concentration, chemical composition, and the size distribution of non-metallic inclusions;
3. the solidification microstructure of the weld metal;
4. the phase transformation during cooling; and
5. the prior austenite grain size.

The solidification process and solid state phase transformations play important roles in weld metal cracking, and it is appropriate to examine these processes in more detail.

3.2.1 Solidification of the Weld Metal

In many respects, weld solidification is a fundamentally different process to that of ingot casting; but, on the other hand, it has certain features in common with continuous casting. The solidification characteristics of welding are as follows:

1. the weld pool contains impurities;
2. non-metallic inclusions are formed before and during solidification;
3. dilution of the weld filler alloy occurs;
4. considerable turbulence and, therefore good mixing occurs in the molten metal;
5. the molten metal volume is much smaller than the base metal, particularly for thick plate;
6. the compositions of the molten metal and base metal are similar, unless welding dissimilar materials;
7. the temperature gradient across the melt is very large;
Figure 3.1 A fully developed dendrite. The primary, secondary and tertiary arms all correspond to \(<100>\) growth directions in cubic metals [After Easterling 1992]
Figure 3.2 Schematic illustration of the various types of growth products developed during the solidification of weld metal as a function of alloy composition, growth rate (R) and temperature gradient (T_L) in the melt [After Easterling 1992]
8. welding solidification is naturally a dynamic process, which is strongly affected by welding speed; and

9. in high energy welds or multi-run welds in which the base metal is preheated, temperature gradients and hence solidification behaviour are affected.

The fact that dilution occurs means that the base metal is locally heated above its melting point during fusion welding. The initial solidification of the weld pool occurs epitaxially, as a heterogeneous process. Therefore, the initial grain size of the weld metal is inherited directly from the grain growth zone of the base metal. Epitaxial grain growth occurs by the development of a planar growth front, but with a decrease in the temperature gradient and an increase in constitutional supercooling by segregation, the crystal growth process changes to cellular growth. Cellular microstructures are only stable for a limited range of temperature gradients. When the temperature gradient falls below a certain level, the primary cells become unstable and develop the secondary arms, and then tertiary arms of a dendritic structure [Easterling 1992]. Figure 3.1 shows schematically a growing dendrite. In this case, the dendrite spacing will largely reflect the degree of segregation at the dendritic boundaries, with a finer spacing being associated with less segregation. Figure 3.2 shows schematically the solidification process of the weld metal during a weld run. The weld zone structure is very important in terms of mechanical properties.

Since a finer weld structure gives improved mechanical properties, various methods for refining the weld structure have been developed: magnetic stirring, ultrasonic cavitation and chemical inoculation [Davis and Garland 1975].

3.2.2 Phase Transformation in the Weld Metal

Solid state phase transformation will occur after solidification during the cooling stage. For HSLA steels, phase transformation depends mainly on the cooling rate, alloying elements and non-metallic inclusions.
Cooling rate is the most important factor influencing the transformation products of HSLA steels, and it can be controlled by the heat input, the thickness of plate to be welded, and the welding technique.

The content of alloying elements is controlled by the compositions of both the base metal and the filler material. The final composition of the deposit lies between that of the filler metal and the base metal. The transformation C-curve characteristics of HSLA steels can be effectively controlled by the addition of alloying elements. Carbon, nitrogen, manganese, nickel, copper, and zinc stabilize austenite (austenite formers), whereas zirconium, titanium, phosphorus, vanadium, niobium, wolfram, molybdenum, aluminium, silicon and chromium stabilize ferrite (ferrite formers). The effect of these elements on \( \gamma \) to \( \alpha \) transformation is complicated by the tendency of some elements to exhibit partitioning at the \( \gamma/\alpha \) interface [Honeycombe 1981]. More detailed discussion about the transformation products of HSLA steel and the influence of alloy elements on weld metal microstructure is given in Section 4.3.3.

### 3.2.3 Non-Metallic Inclusions in the Weld Metal

The steel weld deposit usually has a high volume fraction of non-metallic inclusions compared with normal steels. Non-metallic inclusions in weld metal are the result of chemical reactions which take place during solidification. Due to the low oxygen solubility in iron, only very low free oxygen content, which would be in the uncombined state, can be expected. A major amount of the oxygen is present as oxide inclusions resulting from reactions of oxygen with deoxidising elements such as silicon, manganese, titanium and aluminium in the molten weld pool. The solidification rates of welds are extremely fast, and some inclusions do not have enough time to grow in the melt and separate from the liquid steel to the layer of molten slag covering the weld [Liu and Olson 1986], they may remain as a fine dispersion of particles throughout the weld metal. The presence of these second phase particles will affect both the mechanical properties and the phase transformation of the weld metal.
The characteristics of the inclusion population depend on the choice of atmosphere, welding technique, flux, consumables and the subsequent cooling rate [Kiessling 1989]. Microanalysis of weld metal inclusions generally reveals the presence of titanium, manganese, aluminium and silicon [Barritte et al. 1981; Pargeter 1981]. The inclusions may contain many phases, which depend mainly on the flux and metallic additions. The phases may include: Al₂O₃, MnO, TiO₂, Ti(O, N), Ti(O, C), SiO₂ [Dowling et.al 1986]. In general, weld metal inclusions are smaller in size than the inclusions in the base metal, but the inclusion concentration can be high. The size range for weld metal inclusions is about 0.1 µm < D < 3 µm, where D is the diameter of inclusion.

Non-metallic inclusions in weld metal are considered to play two important but opposing roles.

1. Inclusions can affect the decomposition of austenite by aiding nucleation of acicular ferrite within the austenite grains, and also suppressing the formation of brittle bainitic structures. Good weld metal toughness and maximum resistance to the propagation of a fast running cleavage crack can be obtained when a large portion of fine acicular ferrite grains is present in the weld microstructure [Taylor and Farrar 1975]. Nucleation of acicular ferrite is accelerated when there is a combination of large austenite grains and a high density of intragranular inclusions. The inclusions probably nucleate acicular ferrite by acting as an inert substrate and reduce the energy barrier for nucleation [Ricks et al. 1982], and each inclusion can nucleate several grains [Kiessling 1989]. Abson et al. [1978] reported that with an oxygen content of 300 ppm in weld metal, acicular ferrite was the main structural component. Also, Liu and Olson [1986] suggested that the optimum oxygen content of the weld metal should be about 300 ppm, this being the amount in these steels that encourages acicular ferrite at oxide inclusions without loss of toughness due to an excess of inclusion volume fraction.
2. Non-metallic inclusions can easily debond from the base metal or crack under the plastic strain of dislocation pileups and are thus detrimental to toughness [Dieter 1988]. Inclusions may act as both ductile and cleavage fracture initiation sites. Cracking can be related to the different inclusion size distributions and the way in which these inclusions influence the degree of pinning of prior austenite grain boundaries [Cochrane and Keville 1978]. In ductile fracture, inclusions act as microvoid nucleation sites; after nucleation, the voids grow, link up with microvoids growing in an adjacent region, and form a fracture surface [Hill and Passoja 1974]. Kiessling [1989] found that in the ductile fracture regime, the toughness of the weld metal can be detrimentally affected by the inclusions, because they are potential sites for void nucleation. Therefore, the toughness of the weld metal will be reduced with an increase in the number and size of the inclusions. Ductile crack growth will be promoted by an increased size and decreased separation of inclusions [Chin 1969; Van Stone et al. 1985; Abson and Pargeter 1986]. For cleavage fracture, non-metallic inclusions lying within the plastic zone ahead of the crack tip were found to be active as cleavage initiation sites. Once nucleated the crack spread into the surrounding matrix as a brittle crack [McRobie and Knott 1985; Tweed and Knott 1987]. From his experimental work, Potapov [1993] found that under the condition of oxide inclusions particularly located along the ferrite grain boundaries, weld metal brittle fracture susceptibility is increased by a increase in total oxygen content in a weld up to more than 350 ppm. These inclusions block dislocations and serve as crack initial sites as the weld metal is deformed.

The inclusion type and size distribution is more important than the total volume fraction in influencing acicular ferrite nucleation. Titanium containing oxides are considered to be particularly effective in promoting intragranular nucleation of acicular ferrite [Evans, 1986]. Table 3.1 summarizes a number of different compounds which have been reported to nucleate acicular ferrite.

Ahlblom et al. [1983] concluded that high fractions of acicular ferrite are normally seen associated with small diameter inclusions. However, Liu and Olson [1986] indicated
Figure 3.3 Effect of oxygen on the 3-D inclusion size distribution [After Kluken and Grong 1989]
that (1) weld metals with high acicular ferrite content are found to be associated with coarse austenite grains and with a large number of inclusions with diameters larger than 0.2 µm; and (2) high oxygen content weld metals with higher grain boundary ferrite content are found to be associated with fine prior austenite grains and with a large number of inclusions of diameter less than 0.1 µm. Work reported by Dallam et al. [1985] also showed that the mean particle size of inclusions decreased with increasing oxygen content. However, Ferrante and Farrar [1982] and Cochrane et al. [1983] reported that the mean inclusion diameter increased with increasing weld metal oxygen content. They also found that an prior austenite grain size larger than 45 µm tends to produce acicular ferrite. Kluken and Grong [1989] and Babu et al. [1995] subsequently reported that increasing the weld metal oxygen content resulted in displacement of the weld metal inclusion size distribution to larger diameters. Figure 3.3 shows the effect of oxygen on the 3-D inclusion size distribution.

Table 3.1 Summary of different phases reported to nucleate acicular ferrite [Barbaro 1990]

<table>
<thead>
<tr>
<th>METAL</th>
<th>NUCLEANT</th>
<th>REFERENCE</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weld Metal</td>
<td>TiO</td>
<td>Mori et al. 1981; Ohkita et al. 1984</td>
</tr>
<tr>
<td>Weld Metal</td>
<td>TiN</td>
<td>Ito and Nakanishi 1975; North et al. 1979</td>
</tr>
<tr>
<td>Weld Metal</td>
<td>Al inclusions</td>
<td>Bhatti et al. 1984</td>
</tr>
<tr>
<td>Weld Metal</td>
<td>MnS</td>
<td>Bhatti et al. 1984</td>
</tr>
<tr>
<td>Weld Metal</td>
<td>V(CN), VN</td>
<td>Koukaki 1979; North et al. 1979</td>
</tr>
<tr>
<td>Weld Metal</td>
<td>MnSi/TiO</td>
<td>Grong et al. 1986</td>
</tr>
</tbody>
</table>

3.3 CONTINUOUS COOLING TRANSFORMATION DIAGRAMS

During \( \gamma \to \alpha \) transformation, the austenite will decompose to various forms of ferrite, and the final microstructure will depend on the cooling rate. The Continuous Cooling Transformation (CCT) diagram clearly represents the effect of cooling rate on
transformation start and finish temperature, and complementary metallography can be used to establish the microstructure and the hardness. The CCT diagram relates the composition, cooling rate and austenite grain size of the material to its $\gamma$ to $\alpha$ transformation temperature range and the resultant microstructure. In the order of increasing cooling rate, the microstructure commonly found in low carbon low alloy welds can be classified as follows [Zhang and Farrar 1995], primary ferrite and quasi-polygonal ferrite, Widmanstätten side plate ferrite, acicular ferrite, and lath structure (bainite and martensite). In CCT diagrams each transformation product is represented by a separate curve, thus it is possible to form a mixed microstructure. In addition to simulated weld processing, the CCT diagram can be determined subsequent to the following thermomechanical processes: laboratory steel rolling, hot compression testing, continuous cooling torsion testing [Manohar et al. 1996], and normalising. CCT diagrams can be determined by a metallographic method, and by dilatometry and thermal analysis. The first two methods are reviewed in the following sub-sections.

3.3.1 Metallographic Method

The metallographic method is based on the direct metallographic observation of transformation products formed prior to interrupted quenching during cooling at different rates. This practice is the oldest experimental method used to determine the CCT diagrams for steels.

Austenite cooled continuously below the minimum temperature at which it can exist indefinitely as a stable phase does not begin to transform immediately. Faster cooling progressively lowers the transformation start temperature and it then proceeds over a range of temperatures. For a given austenite composition and grain size, the cooling rate determines the temperature range over which the transformation actually takes place and hence the characteristics and mechanical properties of the final structure. This method involves a substantial amount of experimentation but is an informative method for studying continuous cooling transformation.
Figure 3.4 Schematic representation of the heat treating operations involved in the metallographic method for measuring the progress of austenite transformation on continuous cooling at a given rate [After Grange and Kiefer 1941]
Figure 3.5 Expansion - Temperature Curve in steel on cooling
Figure 3.4 is a schematic diagram which shows the method for measuring the progress of austenite transformation during cooling. Specimens are cooled continuously at a predetermined rate from the austenitising temperature to a predetermined temperature (T1, T2, T3, etc.), then, immediately quenched to room temperature. Any untransformed austenite will transform to martensite, which is readily distinguishable from the transformation product formed at higher temperature prior to the final quench [Grange and Kiefer 1941]. Since the phase transformation during continuous cooling proceeds over a wide range of temperatures, the final structure is a mixture of a series of products. In addition, the determination of a CCT diagram requires investigation of several different cooling rate, so a large number of specimens are required by using this technique. Finally, the CCT diagram can be constructed by drawing lines through points which indicate the start and finish of transformation for each cooling curve.

### 3.3.2 Dilatometry

Continuous cooling dilatometry is by far the most widely used method to study the transformation behaviour of steels during continuous heating, cooling and isothermal holding. During heating and cooling, specimen length is continuously monitored and plotted versus temperature and time. The dilatation associated with transformation can be correlated with temperature as shown in Figure 3.5. Many solid materials exhibit structural changes with temperature, and these allotropic or phase changes are usually accompanied by a change in specific volume at the temperature at which transformation takes place. The change in volume is readily detected by monitoring the dimensions of a suitable test piece. This forms the operating principle of dilatometry [Prior 1994]. The main experimental procedure in this method is to record the time-temperature and dilatation-temperature curves during continuous cooling transformation for various cooling rates.

Modern dilatometers can accurately measure the expansion and contraction of samples during heating and cooling. The expansion-temperature curves, also known as
Figure 3.6 Construction to determine transformation temperature for any percentage transformation [After Harrison and Farrar 1989]
transformation curves, are the basis of the development of the CCT diagram, as these curves identify the transformation start and finish temperatures. Figure 3.5 shows an example of the expansion-temperature curve which indicates the transformation temperature range. Although the curves can identify that a phase transformation has taken place, the dilatometer is unable to identify which phase change has occurred. Obviously, a smaller number of specimens is required to construct CCT diagrams by using dilatometry compared with the metallographic method.

Dilatometry is a commonly used method of constructing CCT diagrams applicable to welding where the actual dilatation characteristics are important in the study of transformational stresses and their effect on weldment cracking.

For HSLA steels, transformation during cooling is significant in relation to the mechanical properties of the weld. The weld CCT diagram can show the transformation sequence clearly. Therefore, weld metal CCT diagrams have played an important role in improving the metallurgical understanding of weld metal microstructural development. Basically, the thermal cycles necessary to produce CCT diagrams applicable to welding situations consist of a suitable austenitisation thermal cycle, followed by continuous cooling to room temperature at rates based on appropriate weld cooling rates. The austenitisation thermal cycle depends on the microstructural region of interest and generally involves rapid heating to the peak temperature at which the appropriate weld metal grain size is developed. Temperatures up to 1400°C are needed for weld thermal cycle simulation [Harrison 1983] with a very short austenitizing time. Weld CCT diagrams can be strongly influenced by the composition of the alloy.

For low carbon low alloy weld metals, the microstructure formed from the parent austenite generally consists of different forms of ferrite, but the dilatation/temperature data do not show a clear dilatometric resolution for each phase field. In these cases, quantitative metallography can be used to locate the temperature at which each type of ferrite phase forms. Figure 3.6 shows a construction and to determine the
transformation temperature for any given percentage transformation [Harrison and Farrar 1989].
CHAPTER 4

CRACKING IN WELDS AND
WELDABILITY TESTING
Figure 4.1 Example of lamellar tearing adjacent to the martensitic region of the HAZ in a butt welded structural steel [After Stout 1977]
4.1 THE MAIN TYPES OF COLD CRACKING

Cracks are one of the most harmful types of weld defects because they can greatly reduce the strength of the weldment. Cold cracking and hot cracking are the two major forms of cracking in welds. Hot cracking in steel occurs at high temperatures shortly after weld deposition and at the start of solidification; or in the solid state due to poor hot ductility on weldment heating.

Cold cracking in welds occurs near room temperature and can be delayed for hours or days after cooling. Cold cracking can occur as transverse cold cracking and longitudinal cold cracking. Transverse cracking is perpendicular to the direction of the weld. Weld metal transverse cold cracking normally starts at hydrogen containing defects subject to longitudinal stresses and cracks can extend from the weld metal into the HAZ [ASM Handbook 1993]. Longitudinal cold cracking is parallel to the direction of the weld. They normally start due to stress concentrations at the root and the toe of the weld [Phillips 1968].

Cracks in welds may be induced during service by a deterioration of mechanical properties, or by weldment bearing service conditions beyond the design capability. In this section, the types of cold cracking which are discussed are those cracks attributed to the welding process itself. These cold cracks include: lamellar tearing, chevron cracking and hydrogen induced cold cracking.

4.1.1 Lamellar Tearing

Lamellar tearing is a type of cold cracking that occurs in the base metal beneath the weld. The tearing generally occurs just outside the hardened martensitic/bainitic region of the HAZ [Wingrove 1986], and usually consists of planar areas parallel to the weld fusion boundary. An example of lamellar tearing is shown in Figure 4.1.

The occurrence of lamellar tearing is particularly associated with restrained corners or T joints in thick steel plates, and results from the combination of high localized stress
from weld contraction and poor ductility in the through-thickness direction of the plate. Therefore, lamellar tearing is a hazard in the construction of offshore oil platforms [Takeski 1975], in welded-on attachments to boilers and thick-walled pressure vessels [Goodger 1966; Lancaster 1993].

Susceptibility to lamellar tearing depends on quantity and morphology of non-metallic inclusions, and the magnitude of the induced stresses normal to the plate surface. Non-metallic inclusions, such as oxides, sulfides and silicates, are elongated in the direction of rolling and flattened out parallel to the surface of the plate. These inclusions form planes of weakness, reduce the ductility in the through-thickness direction, and may lead to failure under high stresses in the thickness direction. The magnitude and distribution of induced stresses depend on the design details, plate thickness, the degree of the restraint imposed on the weldment, and the size and orientation of the weld. Hydrogen also has a significant effect on lamellar tearing, because hydrogen is preferentially trapped at inclusions and tends to accelerate the occurrence of lamellar tearing.

Lamellar tearing can be avoided by the following steps,

(1). Using plate material with improved through-thickness properties. A material with high through-thickness ductility can be achieved by limiting oxygen content [ASM Handbook 1993] and sulfur content to low values, reducing the amount of inclusions in the steel.

(2). Sulfide shape control can change the inclusion morphology from stringers to discontinuous globules and therefore the occurrence of lamellar tearing can be reduced.

(3). Reducing the through-thickness stresses by changing the joint design or using lower-strength and high ductility consumables.
Figure 4.2 Longitudinal section of SAW showing chevron cracks [After Lancaster 1993]
Preheating can effectively reduce lamellar tearing by decreasing the severity of hydrogen embrittling effects and reducing the magnitude of residual tensile stresses acting in the through-thickness direction.

Using a 'buttering' technique. It is possible to grind or machine the surface to a level below where tearing is anticipated, and the area can be replaced by laying one or more layers of lower-strength, high ductility weld metal, before the attachment weld is made.

4.1.2 Chevron Cracking

Chevron cracking was first recognized in submerged arc welded structural steel fabrications in 1969 [Thomas 1969; Cotton 1969]. Chevron cracking, also known as 'transverse 45° cracking' or 'staircase cracking' [Mota and Apps 1982; Lancaster 1993; Metals Handbook 1983], can occur in the weld metal, mainly in submerged arc welds [Lancaster 1993; Metals Handbook 1983; Kalev 1976; Satoh and Terasaki 1978], but a small number of cases have been found in manual metal arc welds [Toyoda et al. 1978] and shielded metal arc welds [Metals Handbook 1983], with a wide range of consumables. Figure 4.2 shows chevron cracking in a longitudinal section from a weld. It is obvious that the character of chevron cracking is two sets of small parallel cracks angled about 45° to the weld axis running approximately transverse to it.

Circumstantial evidence suggests that the cause of chevron cracking is probably hydrogen, associated with the use of hygroscopic agglomerated fluxes [Mota and Apps 1982]. Mota and Apps [1982] advanced the hypothesis that chevron cracking in carbon-manganese and low alloy steels consists of two stages. The critical event is the occurrence of vertical components of the 45° cracks due to 'hydrogen embrittlement'. The second stage consists in the linking of the vertical components due to the excessive stress between adjacent crack tips, this may also be assisted by hydrogen. Chevron cracks have been associated with medium strength weld metals. When the strength was
Figure 4.3 Hydrogen induced cold cracking in welds. (1) Typical macroviews of HICC: (a) root crack in a weld metal [After Kihara and Suzuki 1962]; (b) toe crack in a multipass weld [After Suzuki and Yurioka 1982]. (2) Typical fractographic modes of HICC: (a) intergranular; (b) quasicleavage; (c) microvoid coalescence [After Suzuki and Yurioka 1982; Kikuta et al. 1978].
increased above a certain limit, the cracks become oriented perpendicular to the welding direction. Therefore, orientation of the transverse cracks depending on the weld metal composition and strength [Allen et al. 1982]. Chevron cracks appear to be a special form of weld metal hydrogen induced cold cracking in low-medium strength weld metals. The morphology of chevron cracking is not characteristic of normal hydrogen induced cold cracking. On a fine scale the cracks are intercolumnar and joined by fine transcolumnar cracks to produce the macroscopic 45° orientation.

Since transverse cracking is a type of hydrogen induced cold cracking, it can be avoided by using the preventative methods discussed in Section 4.1.3.

4.1.3 Hydrogen Induced Cold Cracking

Hydrogen induced cold cracking (HICC) is the most serious and most widely encountered cracking problem during the fabrication of welded steel structures [Bailey 1989]. This type of cracking generally does not initiate until the weldment has cooled to below 200°C [Lundin et al. 1990; Sawhill 1974; Easterling 1992]. It is also known as cold cracking or delayed cracking because cracking normally occurs after an incubation period to allow hydrogen diffusion to regions of high triaxial stress around crack-like defects [Davidson 1995], which can vary from several minutes, a few hours to a few days. Examples of hydrogen-induced cold cracking are shown in Figure 4.3.

Hydrogen induced cold cracking can occur in the heat affected zone as well as in the weld metal. With improvements in steel compositions and processing, the risk of heat affected zone hydrogen induced cold cracking is being reduced and the avoidance of weld metal hydrogen induced cold cracking is becoming increasingly more important. The occurrence of cold cracking depends on three mutually interactive factors:

(1) a critical concentration of diffusible hydrogen at a potential crack site;

(2) a stress intensity above a critical magnitude;
(3) a susceptible microstructure.

Hydrogen induced cold cracking in steel welds can be classified as follows [Suzuki 1977]:

(1) weld metal cracking or HAZ cracking (from the location of the initiation site);

(2) macrocracking or microcracking (from the crack size);

(3) longitudinal cracking or transverse cracking (from the direction of its propagation relative to the welding direction);

(4) root cracking, toe cracking, heel cracking, or underbead cracking (from the weld location of the crack); and

(5) restraint cracking, or distortion cracking (from the restraint conditions).

Normally, hydrogen in the weld is trapped at sites of high stress intensity. When the concentration reaches a critical level, a small sharp crack can form. Based on a microplasticity mechanism, Beachem [1972] proposed a model for hydrogen induced cracking, in which hydrogen in the lattice ahead of the crack tip aids whatever deformation process the microstructure will permit [North et al. 1990]. Thus, intergranular fracture will occur at the combination of minimum stress intensity and critical hydrogen content at the crack tip region [Li and North 1992]. For any given hydrogen content, the fracture mode will change from ‘brittle’ intergranular fracture to quasi-cleavage and finally transgranular microvoid coalescence fracture with increasing stress intensity factor levels [Beachem 1972].

Hydrogen induced cold cracking in weld metal is normally intergranular and follows the prior austenite grain boundaries of the columnar grains in the as deposited weld metal. The crack may exhibit a variety of fracture surface morphologies depending on the strength of the steel, microstructure, stress intensity factor and concentration of hydrogen at the crack tip [Yurioka and Suzuki 1990].
The growth of cold cracks usually has the following characteristics:

(1) crack growth is a time dependent process;
(2) crack growth exhibits extremely non-uniform features, propagating rapidly between lengthy periods of arrest; and
(3) crack propagation is often of a intergranular nature in both weld metal and HAZ.

Generally, a microstructure susceptible to cold cracking is considered to contain fairly hard grains consisting of martensite or bainite; a coarse prior austenite grain size which increases the amount of segregation at boundaries; slag inclusions, or coarse carbide particles that decorate grain boundaries; and a high dislocation density [Easterling 1992; Coe 1973; Karppi 1978]. Therefore, the higher the strength of a weld the lower will be its resistance to HICC [Bailey et al. 1989]; and the tougher a microstructure, the greater will be its resistance to HICC [Dolby 1977].

The major preventative methods to avoid hydrogen induced cold cracking are as follows.

(1). Preheating, including maintenance of proper interpass temperature. Preheating to reduce the cooling rate is the essential method to prevent the occurrence of hydrogen induced cold cracking. The diffusible hydrogen content of the weld and stress concentrations in the weld can be substantially reduced by preheating and maintaining an interpass temperature.

(2). Heat input control. Hydrogen induced cold cracking can be avoided by using a higher heat input to reduce the cooling rate.

(3). Postweld heat treatment. At high enough temperatures (550 to 600°C) [ASM Handbook 1993], postheating can reduce residual stresses present in the weldment. Microstructural modifications can be induced by postheating, which produces a microstructure less susceptible to hydrogen induced cold cracking. Postheating may
allow hydrogen to escape the weld area, so that the level of hydrogen in the weldment can be reduced below that required to initiate a crack.

(4). Using a low-hydrogen process and electrode wires. Electrode wire is the major source of weld metal hydrogen. Hydrogen pickup can be greatly decreased by: (a) careful selection of suitable low-hydrogen consumables, and baking electrodes or fluxes prior to use and, once dried, taking care in their storage and handling to avoid the uptake of atmospheric moisture; or (b) selecting low-hydrogen processes (such as gas metal arc welding), and using moisture-free gas lines.

(5). Reduced carbon equivalent. For a required level of strength, the steel with the lowest carbon equivalent should be considered. Also, for minimizing the occurrence of hydrogen induced cold cracking, it is necessary to restrict the content of sulfur and phosphorus.

4.2 FACTORS AFFECTING COLD CRACKING

4.2.1 Hardenability

Carbon equivalent can be used as an index for the hardenability of the steel, therefore, the effect of the various alloying elements on the steel transformation behaviour relative to carbon can be ranked. Thus, the purpose of using carbon equivalent is to restrict the composition of steels to help attain good weldability. Also, carbon equivalent can be used for assessing the risk of cracking in welds, as it has been shown that for a given hydrogen level, the critical hardness level at which hydrogen induced cold cracking occurs decreases with decreasing carbon equivalent [Matharu and Hart 1985; Kirkwood 1987].

A carbon equivalent has been adopted by the International Institute of Welding in 1967 as follows [Granjon 1967],
The C\textsubscript{Enw} has been extensively used as an index to describe steel weldability or susceptibility to cold cracking. The C\textsubscript{Enw} normally applies to carbon contents above about 0.18 wt percent. [Easterling 1992].

In order to evaluate the cold cracking tendency of low carbon low alloy steels whose carbon content is less than 0.18 wt percent., Ito and Bessyo [1968] proposed the following carbon equivalent:

\[
P_{em} = C + \frac{Mn + Cu + Cr}{20} + \frac{Si}{30} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B
\]  

(4-2)

In 1983, Yurioka et al. [1983] advanced the following carbon equivalent formula to assess the susceptibility of a wide range of steels to cold cracking:

\[
CE_N = C + A(C) \left( \frac{Si}{24} + \frac{Mn}{6} + \frac{Cu}{15} + \frac{Ni}{20} + \frac{Cr + Mo + V + Nb}{5} + 5B \right)
\]  

(4-3)

where \( A(C) \) is defined as \( A(C) = 0.75 + 0.25 \tanh [20(C - 0.12)] \)

In gas metal arc welding, the transformation behaviour of the weld metal is affected by shielding gas oxygen equivalent. It has been found that the weld metal hardenability \( (P_{cm}) \) decreases with increase shielding gas oxygen equivalent [Ons\o ien et al. 1996]. In order to predicate susceptibility to weld metal cold cracking sufficiently, Ons\o ien et al. [1996] reported a formula which includes the oxygen content in the equation for \( P_{cm} \).

The adjusted \( P_{cm} \), named \( P_{cmo} \), is expressed as:

\[
P_{cmo} = C + \frac{Mn + Cu + Cr}{20} + \frac{Si}{30} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B - \frac{3}{4} \frac{O}{4}
\]  

(4-4)

The \( P_{cmo} \) was developed for welds with oxygen content in the range between 275 and 450 ppm, and \( P_{cmo} \) can be considered as the 'net hardenability' of the weld metal [Ons\o ien et al. 1996].
It is not very satisfactory to only use carbon equivalent to assess the weldability of HSLA steels, since many other factors, such as plate thickness, welding conditions, hydrogen level of electrodes and geometry of the joints, can affect the microstructure and susceptibility to cold cracking.

### 4.2.2 Hydrogen

Hydrogen play a very important role in the cold cracking of weldments. Cold cracking is often called hydrogen induced cracking or hydrogen assisted cracking [Sawhill et al. 1986; Signes and Howe 1988; Savage et al. 1976]. A weld with a lower hardenability level or an ideal microstructure may also be susceptible to cold cracking when the hydrogen concentration is high [Hart 1986]. The dissociation of water vapor and contaminants in the arc environment where the temperature can be 6000°C or higher, can result in hydrogen charging of the molten pool. The solubility of hydrogen decreases with decreasing temperature, and some of it diffuses out of the weldment with cooling. At the same time, some of the hydrogen diffuses from the weld metal to the HAZ. The diffusion of hydrogen will be restricted when the temperature of the weldment reaches a temperature below 100°C [Bibby et al. 1998] because of hydrogen trapping. However free hydrogen remains mobile and can accumulate at stress concentration. When the local hydrogen concentration at a potential crack site reaches a critical level for a given stress intensity factor, cold cracking can initiate. Therefore the sensitivity to cold cracking of a weld joint increases with increasing hydrogen concentration.

The main sources of hydrogen in welding include:

(a) rust, primer, paint and hydrocarbons which may be present on the weld preparation;

(b) lubricating oil and binders used during electrode coating manufacture;

(c) hydrogen deliberately added into the shielding gas to improve welding characteristics;
Figure 4.4 Schematic illustration of hydrogen entering the weld pool during manual metal arc welding [After Hirai et al. 1974]
Figure 4.5 Change in hydrogen content in the weld metal and heat affected zone regions following solidification [After Matsuda et al. 1985]
(d) hydrogen in the electrode core and in the base metal; and

(e) moisture in the flux and electrode coating, on the weld preparation, and from the atmosphere close to the arc.

Figure 4.4 illustrates how the moisture and hydrogen from various sources enter the manual metal arc weld pool. Atmospheric moisture is particularly important in welding applications where only a low hydrogen concentration may be tolerated in a weld. It has been found that moisture in the atmosphere close to the arc has an insignificant influence on the weld hydrogen uptake during gas metal arc welding, because the molten weld pool is shielded with argon and carbon dioxide shielding gases [Coe 1973]. Gedeon [1990] found that, in steel, hydrogen is even more efficiently absorbed if oxygen is added to the shielding gas. Hart [1978] found that low hydrogen fluxes can pick up moisture from the air, resulting in hydrogen pick-up the weld metal. Moisture absorption rates by the flux are greatest in hot humid conditions. The moisture pick-up rate increases sharply at about 60-70 percent relative humidity irrespective of temperature [Davidson 1995]. Salter [1963] showed that hydrogen can increase approximately 3 ml/100g when relative humidity goes from 25 to 72 percent. On the other hand, flux in a cored wire is less susceptible to moisture pick up than manual metal arc electrodes or submerged arc flux because the flux is enclosed within the welding wire and binders are not required.

Hydrogen in a weldment exists in combined (molecular) and diffusible (atomic) forms. The diffusible hydrogen is responsible for cold crack initiation. Typical diffusible hydrogen contents in the weldments may vary from 1 to 40 ml/100g depending on the type of consumable used, the storage conditions and atmospheric humidity [Lundin 1990]. The weld metal hydrogen content decreases continuously with time as temperature decreases during weld solidification, and HAZ hydrogen content increases initially due to hydrogen diffusion from the weld metal into the base metal. Figure 4.5 illustrates the hydrogen distribution in the weld metal and HAZ following solidification. It is apparent that the weld metal hydrogen content may be much higher than that in the
Figure 4.6 Variations in the measured diffusivities of hydrogen in ferritic and austenitic steels [After Coe 1973; Yurioka and Suzuki 1990]
HAZ when the weld cools to the temperature where hydrogen induced cold crack can initiate. Thus, either weld metal or HAZ cold cracking may occur depending on the relative hardness and tensile strength of these regions [Matsuda et al. 1985]. Also, the likelihood of weld metal cold cracking is greater when the diffusible hydrogen content is increased for any given weld metal and HAZ hardness-strength combination [Li and North 1992]. Figure 4.6 illustrates the variation in the diffusivity of hydrogen in austenitic and ferritic steel, where the apparent diffusion coefficient of hydrogen in steel is shown as a function of temperature. This figure indicates that the hydrogen diffusivity in austenite is much lower than that in ferrite at the same temperature. However, the solubility of hydrogen in ferrite is lower than in austenite and thus, with the progress of the austenite-ferrite transformation, hydrogen becomes more and more enriched in residual austenite. Generally, cold cracking is very sensitive to the presence of martensite because the austenite-martensite transformation occurs at the lowest temperatures and originates from the most hydrogen enriched austenite [Easterling 1992].

4.2.3 Microstructure

The weld metal mechanical properties are dependent on the microstructure. Francis et al. [1990] summarized the two primary changes which can occur in the weld metal microstructure due to alteration of the cooling rate. First, coarser microstructures will be produced by decreasing cooling rate. Second, changes in cooling rate will affect the nucleation and growth processes involved in the decomposition of austenite into ferrite plus carbides or martensite. Therefore, the solidification microstructure will depend on the weld metal composition, phase transformation and the weld cooling rate, as well as on the weld metal composition.

According to the International Institute of Welding (IIW) classification, the microstructural features of HSLA steel weld metal can be characterised into a series of major categories: grain boundary ferrite (GBF), Widmanstätten side-plate ferrite
Figure 4.7 Optical micrographs showing typical HSLA steel weld metal microstructures (nital etch). A: grain boundary ferrite; B: Widmanstätten ferrite side plates; C: acicular ferrite; D: ferrite with aligned second phase; and E: martensite [After Grong and Matlock 1986]
[F(SP)], acicular ferrite (AF), ferrite with aligned second phase (bainitic ferrite) [FS(A)] and martensite, which are shown in Figure 4.7. In addition, the ferrite can consist of a non-aligned second phase [FS(NA)]. Normally, the microstructure formed within each single austenite grain after transformation will be a complex mixture of two or more of the above constituents.

When the cooling rate is sufficiently slow, coarse and blocky grain boundary ferrite nucleates and grows along the prior austenite grain boundaries in an elongated or granulated form, often accompanied by Widmanstätten ferrite side-plates which are needle-like ferrite laths which protrude from one or both side of the grain boundary ferrite into the austenite grains [Choi and Hill 1978; Harrison and Farrar 1981]. Grain boundary ferrite can occur in veins throughout the weld metal. At a somewhat faster cooling rate, acicular ferrite will form if appropriate nucleants are available and well distributed. The acicular ferrite is a fine structure with short, multi-variant laths which form intragranularly with typical dimensions in metallographic sections of 5 to 10 \( \mu \text{m} \) length and about 1 \( \mu \text{m} \) thick [Svensson 1994]. The acicular ferrite has been recognised as the most favorable microstructure to display optimum weld metal strength and toughness properties. Therefore, the acicular ferrite is considered to be the structure which is least susceptible to cold cracking. Faster cooling rates or the absence of nuclei for acicular ferrite generally result in a microstructure of ferrite with an aligned second phase. Parallel ferrite laths can grow either directly from the grain boundary or from grain boundary ferrite [Liu et al. 1993]. With greater cooling rates, the austenite will transform to martensite which can exist in the form of packets or individual plates. Martensite strengthens the weld metal but decreases the toughness. As mentioned in Section 4.2.2, the austenite is enriched in hydrogen and with the decomposition of austenite, the ferritic transformation products have a lower solubility for hydrogen and are more susceptible to cold cracking. Consequently, martensite is considered very susceptible to cold cracking since it is formed at the lowest transformation temperatures, originating from the most hydrogen enriched austenite. Overall, the formation of a large
Figure 4.8 Schematic continuous cooling transformation diagram showing the effect of hardenability elements and inclusion formers on weld metal microstructure [After Liu et al. 1993]
Figure 4.9 Schematic diagrams illustrating how the alloying content affects the microstructure in a weld metal. The austenite grains are assumed to have a hexagonal cross-sectional shape. a. With low alloying content, the microstructure is predominantly grain boundary ferrite and Widmanstätten ferrite side plates; b. With higher alloying content, acicular ferrite can nucleate on inclusions (black dots) and occupy a large fraction of the prior austenite grain. [After Bhadeshia et al. 1985]
proportion of grain boundary ferrite, Widmanstätten ferrite side plates or martensite has been found to be detrimental to toughness, because these structures provide preferential easy crack propagation paths, which offer a low resistance to weld metal cleavage fracture [Abson and Pargeter 1986; Grong and Matlock 1986; Farrar and Harrison 1987].

Continuous cooling transformation diagrams can be used in describing the final microstructure of a weld metal, as shown in Figure 4.8. The transformation curves will shift with changes in alloy composition. Hardenability agents such as manganese, nickel, silicon and carbon move the ferrite and bainite transformation curves to longer times, refining the microstructure. Inclusion formers have the opposite effect and move the C-curves to shorter times, promoting the formation of high temperature products.

Alloying elements present in the weld metal are either introduced deliberately through the filler wire and flux or picked up from the base metal as a result of dilution. It is well known that the alloying elements will influence the microstructure of a weld metal. Figure 4.9 shows the development of microstructure affected by the alloying content in a weld metal. In a lean alloy weld metal, a large fraction of the austenite will transform to grain boundary ferrite and Widmanstätten ferrite side-plates, which will dominate the microstructure, as shown in Figure 4.9a. In a higher alloy content weld metal, acicular ferrite tends to nucleate on non-metallic inclusions, lowering the growth rate of both grain boundary ferrite and side plate, as shown in Figure 4.9b. Saggese et al. [1983] and Grong and Matlock [1986] indicated that high volume fractions of acicular ferrite are always achieved when sufficient titanium is added either through the filler wire or the flux. Zhang and Farrar [1997] carried out a systematic investigation on the microstructure and toughness of C-Mn-Ni low alloy weld metals. It was found that both manganese and nickel promoted acicular ferrite at the expense of grain boundary ferrite and Widmanstätten side-plate ferrite. Manganese has a stronger potential for suppressing the formation of sideplate structure than nickel. However, nickel can affect the microstructure through changing the weld metal columnar grain size. Zhang and
Farrar [1997] showed that the addition of nickel initially refined the prior austenite grain size but further additions caused it to rapidly coarsen, and some martensite was formed at high nickel content. Hart [1986] indicated that for C-Mn-Mo weld metals, the amounts of grain boundary ferrite and ferrite with aligned second phase were reduced, and acicular ferrite was promoted and refined with increasing molybdenum content.

On the other hand, Munning Schmidt-Van Derr Burg et al. [1985] and Svensson and Gretoft [1990] reported that a fine microstructure may not be always beneficial to weld metal toughness when the deposits are overalloyed. The positive effects of fine acicular ferrite structure can be offset by a simultaneous increase in hardness and strength, or the presence of martensite and segregation structures in the weld metal. Therefore, to obtain the best mechanical properties and to avoid weld metal hydrogen induced cold cracking through development of an acicular ferrite microstructure by alloying, care should be taken to ensure a sufficiently high proportion of acicular ferrite, a low volume fraction of martensite and low levels of microsegregation.

4.2.4 Stresses

The residual stresses in welds also play an important role in cold cracking. Residual stresses supply the driving force for hydrogen induced cold cracking. In all welded structures, residual stresses are produced in areas around the weld. On a macroscopic scale, residual stresses are also produced near regions which are martensitic [Masubuchi 1980], because austenite-martensite transformation occurs at relatively low temperature, and the associated transformation expansion generates plastic strains and elastic strains up to the yield point, which increases strongly as the temperature falls. During welding, there are three kinds of stresses:

(1) *Thermal stress induced by the welding thermal cycle.* Because a weldment is heated locally by the welding heat source, the weldment temperature distribution is non-uniform and changes as the welding progresses. Differential stresses develop as a result of contraction during cooling.
(2) **Structural stress (phase transformation stress) induced by phase transformation.** For austenite to ferrite transformation, the lower the $\gamma$ to $\alpha$ transformation temperature the larger is the dilatation in transforming from the f.c.c. to the b.c.c. lattice, and hence the larger the stress change [Easterling 1992]. The increase in phase volume introduced by $\gamma$ to $\alpha$ transformation should reduce the tensile stress in welds because of shrinkage during cooling, thus leading to a reduced sensitivity to cold cracking. However, the non-uniformity of the stress distribution will ensure that the counteractive effect is limited.

(3) **Restraint stress.** This type of stress develops during cooling and continues to exist when the weld has cooled to ambient temperature. It arises because of working load, welding geometry or welding technique.

Yurioka and Suzuki [1990] indicated that the incidence of hydrogen induced cold cracking is determined by the local stress acting at a particular site rather than the average stress acting on a weld. A high local stress can originate from stress concentration at the tip of a notch, such as the weld root and weld toe. It is therefore important to prevent stress concentration. The residual stress acting on a weld is a function of weld size, joint geometry, fit up, external restraint and the yield strength of the base metal and weld metal. The stress level can be reduced by careful selection of joint geometry, welding parameters and stress relief treatment. These factors can contribute to mitigation of cold cracking in a weld joint.

### 4.2.5 Nitrogen

Nitrogen can be absorbed in the weld pool during arc welding and the saturation level for nitrogen is close to the equilibrium solubility at one atm pressure and a liquid metal temperature of 1600°C: 0.045 percent by mass [Easterling 1992]. The amount dissolved from a 50%$\text{N}_2$-50%$\text{O}_2$ arc atmosphere rises to 0.16 wt. percent, so that air is a particularly unfavorable atmosphere for the arc welding of steel. Nitrogen is damaging
in two ways: by causing porosity; and by embrittling the weld deposit. Porosity appears when the nitrogen content exceeds about 0.045 wt. percent. Nitrogen in the form of FeN has a severely embrittling effect on weld metal. Furthermore, nitrogen in a carbon or carbon-manganese steel may be responsible for strain age embrittlement. If nitrogen bearing steel is subject to plastic strain and simultaneously or subsequently heated at a temperature of about 200°C, the toughness is reduced [Lancaster 1993].

4.2.6 Preheating

Preheating controls the cooling rate of the weld metal through the transformation temperature range. Preheating of joints to be welded is an extremely effective method for reducing (1) the cooling rate of the weld and HAZ, (2) the magnitudes of distortion and residual shrinkage stresses, and (3) the arc energy input required to deposit a given weld. The first two factors are essential to prevent cracking in hardenable steels. The third is often necessary for welding of thick sections of highly conductive metals. The main purpose of preheating is to slow down the cooling rate. A slower cooling rate gives hydrogen more time to diffuse out of the weldment and prevents the formation of susceptible microstructures, such as low temperature transformation products [Adams 1958]. The cooling time from 300 to 100°C, $\Delta t_{3-1}$, also plays an important role in cold cracking because, and for the same initial hydrogen level, the cold cracking tendency varies significantly with the cooling time, $\Delta t_{3-1}$ [Chakravarti and Bala 1989].

The second purpose of preheating is to reduce distortion and residual stress. As weld metal cools through the austenitic range, the steel has little strength and good plasticity. Therefore, the weld metal and HAZ deform plastically to accommodate the change in dimensions imposed by shrinkage. Upon cooling to room temperature, residual stresses build up because of continuous shrinkage, but can be reduced by preheat treatment.
4.2.7 Heat Input

The heat input is one of the most important variables in the welding process since it governs heating rates, cooling rates and weld pool size. Generally, cooling rate decreases with increasing heat input, and a high heat input can prevent the formation of susceptible microstructures containing low temperature transformation products. Thus a high heat input can lead to a reduction in susceptibility to cold cracking. On the other hand, very low cooling rate will result in a longer time spent above the grain coarsening temperature, leading to a coarse, low toughness transformed structure. It is therefore necessary to seek a heat input that gives the optimum combination of grain size and cooling rate.

4.3 WELDABILITY TESTING

Granjon [1954] collated and classified the weldability tests that had been used up to the early 1950s. It was mentioned that 60 tests had been used to determine various aspects of weldability at that time. In the past 40 years, many additional weldability tests have been developed. These developments are based mainly on the desirability of small scale laboratory tests to evaluate the susceptibility of the base metal and weld metal to cold cracking.

The major reasons that cold cracking is studied by using weldability tests are to determine the dominant variables affecting cold cracking, to compare the relative susceptibilities of steels to cold cracking, and to develop procedures to prevent cold cracking.

In principle, weldability tests for evaluating cracking susceptibility can be mainly classified as one of two types.

1. Self restraint tests. This type of test does not have an externally applied stress, but has an undefined internal stress level due to shrinkage, thermal contraction and transformation stresses.
Figure 4.10 A diagram of the standard G-BOP test [After McParlan and Graville 1976]
2. **External restraint tests.** The main characteristic of this type of test is that a load is externally applied to the test weld.

Three weldability tests for evaluating cold cracking susceptibility are discussed in this section, including the Gapped Bead-On-Plate (G-BOP) test, Tensile Restraint Cracking (TRC) test and Longitudinal Bead-Tensile Restraint Cracking (LB-TRC) test. The G-BOP test is a self restraint tests, whereas the TRC and LB-TRC tests are external restraint tests.

The transverse weld metal cracking tests which are of particular interest to the current investigation are the G-BOP and LB-TRC tests.

**4.3.1 The Gapped Bead-On-Plate Test**

The Gapped Bead-On-Plate (G-BOP) test was developed by Graville and McParlan [Graville and McParlan 1974; McParlan and Graville 1976] at the time when there was growing interest in weld metal cold cracking. It is perhaps the oldest self restraint test and is a simple, inexpensive test for evaluating the relative sensitivity to transverse hydrogen-induced cold cracking in weld metal.

Transverse cracks often occur when welding over the small gap created when two backing bars are butted together without welding. Welding over a transverse gap to promote weld metal cracking has been recognized as a basis for a transverse cracking test and has been formalized as the G-BOP test [Graville and McParlan 1974; Chakravarti and Bala 1989; Vuik and Graville 1993].

In this test, a bead is deposited across a gap between two steel blocks. Due to the longitudinal contraction stresses, a transverse crack is forms in the weld metal. The standard G-BOP test blocks used for depositing the weld beads are shown in Figure 4.10. The test consists of two blocks, one of which has a machined recess. In addition to the recess, the mating faces of the two blocks are machined, The blocks are clamped together to prevent rotation and a bead is deposited along the top surface over the gap.
Figure 4.11 Buttering of G-BOP test to reduce weld metal dilution [After Hart 1986]
After welding, the blocks are left in the clamps for a minimum of 24 hours [Graville and McParlan 1974]. To examine for cracking, the welded blocks are removed from the clamp and the weld is heated to a dull red heat in the vicinity of the gap, allowed to cool to room temperature and then broken open. Any crack formed would show as a blue discoloration because of oxide layer formation [McParlan and Graville 1976; Chakravarti and Bala 1989]. The thickness of the blocks originally selected was 50 mm, and therefore the weld cools rapidly through the transformation range; and moderate preheats have a large effect on the low temperature cooling rate.

The composition of the base metal can have a significant effect on weld metal cracking in the G-BOP test. Dilution is typically 40-60 percent [Graville 1995] depending on the process used. In order to minimize dilution effects, a mild or low carbon structural steel should be used as the base metal, and the base blocks should be taken from a single heat of steel for a series of tests [Graville and McParlan 1974]. On the other hand, Hart [1986] minimized base metal effects by buttering the blocks with the relevant weld metal before depositing the test welds. The buttering of the G-BOP test piece is illustrated in Figure 4.11. This procedure reduces dilution to about 10 percent [Hart 1986].

The G-BOP test can be used to assess the cold cracking susceptibility of different combinations of electrode wires and flux or gas used for weld deposits based on two criteria: cracking percentage at room temperature and a 10 percent crack preheat temperature (CPT) [Chakravarti and Bala 1989]. The test can be used to study the effects of dilution from the base metal on weld metal cold cracking susceptibility, and it is able to determine the influences on weld metal cold cracking of micro-alloy elements such as niobium and vanadium picked up from the base metal [Graville 1995]. The G-BOP test also can be used to study the effects of preheat and cooling rate on weld metal cold cracking. The test can be used to examine quantitatively the effect of the major variables such as hydrogen level, chemical composition, stress and microstructure on hydrogen-induced cold cracking in the weld metal. Chakravarti and Bala [1989] indicated that correlation of weld metal composition, cooling time $\Delta t_{5/1}$ and diffusible
hydrogen level to the 10 percent CPT could be useful for determining those electrodes least susceptible to cold cracking.

Using the G-BOP test, Chakravarti and Bala [1989] found that the percentage of cold cracking decreased with increasing preheat temperature or increasing cooling time $\Delta t_{3/1}$, and about 90 percent of the diffusible hydrogen can be evolved in three days.

Hart [1986] studied the effects of composition and microstructure on weld metal hydrogen induced cold crack for C-Mn based weld metals using the G-BOP test, and come to the following major conclusions,

(1) the susceptibility to cracking was primarily controlled by weld metal hardness at a higher hydrogen level of 10 ml/100g, with the risk of cracking increasing with increasing hardness; and

(2) the susceptibility to cracking principally depended on microstructural type rather than hardness at lower hydrogen levels of less than 5 ml/100g.

McParlan and Graville [1976] reported that the influence of hydrogen content on the preheat to avoid weld metal cold cracking in the G-BOP test can be rationalized on the basis of hydrogen diffusion out of the weldment before it cools to room temperature. The preheat to avoid cold cracking in a higher alloy weld metal is less sensitive to changes in hydrogen content within the practical range than a less alloyed weld metal.

Lazor and Graville [1983] investigated the effects of alloy elements, such as niobium, vanadium, molybdenum and nickel, on susceptibility to weld metal transverse cracking using the G-BOP test. They found that the susceptibility to cracking increased significantly when weld metal contained 0.06 weight percent Nb, whereas contents of 0.05 weight percent V, 0.2 weight percent Mo and 1.2 weight percent Ni in weld metal had minimal influence. They also reported that there was no relation between weld metal hardness and crack susceptibility.
Figure 4.12 G-BOP test instrumented to measure stress [After McParlan and Graville 1976]
The longitudinal stress across the gap in the G-BOP test can be studied using an instrumented version of the test, as shown in Figure 4.12. McParlan and Graville [1976] reported that the longitudinal stress was controlled by the restraint and the transformation temperature of the weld metal in the G-BOP test. In general, longitudinal stress decreased with decreasing transformation temperature caused by an increase in alloy content. However, if the restraint is very high, the longitudinal stress will be determined by the yield characteristics of the weld metal.

4.3.2 The Tensile Restraint Cracking Test

The Tensile Restraint Cracking (TRC) test was developed by Suzuki et al. [1963] for evaluating cold cracking in the HAZ.

The test involves the butt welding of a pair of steel plate specimens. Immediately after cooling, the weld specimen is subjected to a constant tensile load for a period of time until root cracking occurs. When a specimen is subjected to a certain stress a crack initiates at a certain time and growth occurs at a time dependent rate. As the stress level decreases, the time for crack nucleation and propagation increases. Below a certain level of stress, cracking does not occur, and this level of stress is termed the critical stress. As the preheat temperature increases and the diffusible hydrogen content of the weld metal decreases, the critical stress becomes greater until it finally reaches the yield stress of the weld metal [Suzuki et al. 1963; Inagarki et al. 1965].

4.3.3 The Longitudinal Bead-Tensile Restraint Cracking Test

Matsuda et al. [1979] specially developed the Longitudinal Bead-Tensile Restraint Cracking (LB-TRC) test by modifying the original TRC test [Suzuki et al. 1963] for testing the susceptibility of higher strength steel to weld metal transverse hydrogen-induced cold cracking.
Figure 4.13 LB-TRC test specimen and testing method [After Matsuda et al. 1979]
A diagram of the LB-TRC test is shown in Figure 4.13. The test consists of two plates containing a slot for deposit of the test weld. A U-groove is machined in the middle of the thickness of the abutting plates and supporting rollers are placed under the specimen in order to prevent bending of the joint and eccentric stress during testing. The two plates are lightly butted together without a gap before welding. After welding, a constant load is applied parallel to the weld line. A predetermined load is applied at a predetermined temperature or time [Matsuda et al. 1979] after the completion of welding, and is maintained constant until cracking occurs. The maximum stress level (critical stress) that can be applied without cracking is characteristic of the susceptibility for weld metal hydrogen-induced cold cracking [Graville and Vuik 1993].

Cracking in the LB-TRC test initiates at the root of the weld bead, and propagates mainly along the columnar crystals in the weld metal and finishes in a shear type fracture at the bead surface.

The effect on the cold cracking susceptibility of preheating temperature, hydrogen level and composition of base metal and electrode can be well evaluated in the LB-TRC test.

Matsuda et al. [1979] considered that the critical stress and fracture time increase with increase of preheating temperature in the LB-TRC test, and the test is one of the most suitable methods for investigating the effect of preheating on critical stress and for comparison of the cold cracking susceptibility for different high strength weld metals.

In the LB-TRC test, the difference in the critical stress seems to be caused by the difference in diffusible hydrogen content, so the test can be used to compare the cold cracking susceptibility of weld metals obtained with different electrodes.

Compared with the original TRC test, advantages of the LB-TRC test are a smaller specimen size and a lower testing load.
4.4 AIMS OF THE PRESENT INVESTIGATION

The arc welding process is still considered to be the most effective joining method for joining of steels in the structural industries. However, severe thermal cycling in the welding process can make steel welds susceptible to cracking. Sensitivity to hydrogen induced cold cracking is often considered to be the most important factor in the weldability of high strength steels. Developments in steel design have greatly improved the crack resistance of the heat affected zone by making the steel less susceptible to hydrogen cracking, but similar development has not take place significantly for weld consumables. Therefore, the weld metal often becomes the site where cracking initiates. Håkansson and Dixon [1996] found that even slight relaxation in the welding procedure can lead to extensive transverse hydrogen induced cold cracking in welds. Detection of sub-surface transverse cracking is extremely difficult. Radiography has been found to be incapable of detecting such cracks [Dixon and Taylor 1996]. Ultrasonic testing can be used to detect transverse cracks, but it is quite sensitive to the orientation of the crack, the presence of other reflectors close to the crack and the skill of the operator. Consequently, the essential method to minimize the risk of hydrogen induced cold cracking in weld metal is to develop safe welding procedures. Cracking in the weld metal can occur as transverse cracks which are perpendicular to the welding direction, and as longitudinal cracks which are parallel to the weld direction. Although longitudinal cold cracking has been investigated in detail, relatively little work has been reported on the investigation of weld metal transverse cold cracking. The main aim of this work was to investigate the sensitivity of a welded high strength quenched and tempered steel to transverse cold cracking and to define the conditions for crack-free welding.

In this work, the gapped bead-on-plate (G-BOP) test has been used for testing transverse weld metal hydrogen induced cold cracking in a welded submarine steel. The main variables studied are as follows,
1. the effect of preheat temperature on cracking behaviour,

2. the effect of shielding gas mixture on cracking behaviour, and

3. the effect of non-metallic inclusions on microstructural transformation and cracking behaviour.

The continuous cooling transformation diagrams of the base metal and the weld metals have been developed to describe phase transformation during the welding thermal cycle and to improve the understanding of weld metal microstructural development and the influences of various microstructural factors on transverse cold cracking susceptibility.

The experimental techniques are described in Chapter 5. The experimental results are presented in Chapter 6. Chapter 7 consists of the discussion of the experimental work and the conclusions of the investigation are given in Chapter 8.
CHAPTER 5

EXPERIMENTAL METHODS AND MATERIALS
5.1 MATERIALS

The steel used for the research was BIS812EMA submarine steel supplied by the Defence Science and Technology Organisation (DSTO), Australia. BIS812EMA steel is a 690 MPa. yield stress high strength low alloy (HSLA) steel with small additions of potent micro-alloying elements such as copper, titanium, niobium, boron and vanadium to achieve an extra boost in strength. The chemical composition of BIS812EMA steel is given in Table 5.1, along with the calculated carbon equivalent (CE_N) and P_cm parameter. The typical mechanical properties of BIS812EMA steel are shown in Table 5.2. BIS812EMA steel is made in rolled form by BHP steel, Slab & Plate Products Division, Port Kembla and heat treated by Bisalloy Steels, Unanderra using first a spray quench from a temperature of about 920°C and then a temper at about 590°C [Dixon and Taylor 1997].

Table 5.1 Chemical composition (wt%) and carbon equivalent of BIS812EMA steel

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>V</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.13</td>
<td>0.24</td>
<td>0.93</td>
<td>0.011</td>
<td>0.002</td>
<td>0.48</td>
<td>1.28</td>
<td>0.39</td>
<td>0.02</td>
<td>0.01</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Cu</th>
<th>Al</th>
<th>Nb</th>
<th>B</th>
<th>N</th>
<th>Ca</th>
<th>O</th>
<th>Fe</th>
<th>CE_N</th>
<th>P_cm</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.21</td>
<td>0.07</td>
<td>0.01</td>
<td>0.0066</td>
<td>-</td>
<td>≤3 ppm</td>
<td>0.009</td>
<td>bal.</td>
<td>0.495</td>
<td>0.301</td>
</tr>
</tbody>
</table>

Since the carbon content of the base metal is less than 0.18 weight percent, the carbon equivalent was calculated based on the formula for estimating the effect of carbon and the other alloying elements on crack sensitivity, as recommended by Yurioka et al. [1983]:

\[
CE_N = C + A(C) \left( \frac{Si}{24} + \frac{Mn}{6} + \frac{Cu}{15} + \frac{Ni}{20} + \frac{Cr+Mo+V+Nb}{5} + 5B \right)
\]  

(5-1)

where A(C) is defined as A(C) = 0.75 + 0.25 tanh [20(C - 0.12)]
The $P_{cm}$ parameter of the base metal is calculated using a formula advanced by Ito and Bessyo [1968], equation (5.2). This equation is suggested to be more suitable for quenched and tempered steels [Dixon and Taylor 1997].

$$P_{cm} = C + \frac{Mn + Cu + Cr}{20} + \frac{Si}{30} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B \quad (5.2)$$

Table 5.2 Typical mechanical properties of BIS812EMA steel

<table>
<thead>
<tr>
<th>0.2% Proof Stress (MPa)</th>
<th>UTS (MPa)</th>
<th>Elongation (A5)%</th>
<th>Yield Stress (MPa)</th>
<th>Charpy Impact Energy (J) at -18°C</th>
<th>Charpy Impact Energy (J) at -60°C</th>
</tr>
</thead>
<tbody>
<tr>
<td>754</td>
<td>845</td>
<td>18</td>
<td>690</td>
<td>90</td>
<td>75</td>
</tr>
</tbody>
</table>

The consumable used to join BIS812EMA plate was a 1.2 mm diameter low hydrogen flux cored welding electrode, Dual Shield II120-M2. The flux cored wire was always baked in a furnace at 40°C before welding to maintain dryness.

5.2 WELDING PROCEDURES

5.2.1 Test Method

The self restrained G-BOP test was chosen to investigate the transverse hydrogen induced cold cracking in the weld metal, because of the simplicity and low cost. The G-BOP test has two main advantages: it always gives some response in the form of a crack from 2 percent up to 100 percent [Hannerz and Xu 1993], and it is similar to many real weld situations, where there is a gap.

Two kinds of G-BOP test methods were used in the current work. The standard G-BOP test method was used for most of the weldments. An instrumented G-BOP test method was used for some tests, in which two load cells were incorporated into the set-up to measure the reaction force during welding, so that the onset of cracking could be monitored, and the longitudinal stress across the gap could be determined. Pt - Pt/Rh, R type thermocouple was plunged into the test blocks under the weld bead to record the
Figure 5.1 Schematic diagram of standard G-BOP test blocks and the location of thermocouple in the test blocks. Section A - A' shows that the thermocouple tip is capped with a metal plug. All dimensions in mm
Figure 5.2 Schematic diagram of instrumented G-BOP test blocks and the location of the weld bead. All dimensions in mm
Figure 5.3 Photograph showing the instrumented G-BOP test blocks clamped in the jig holding the two blocks in place for the deposition of weld
Figure 5.4 Photograph showing the instrumented G-BOP test assembly after welding
temperature history. A schematic diagram of the standard G-BOP test blocks and the location of the thermocouple in the test blocks are shown in Figure 5.1. The instrumented G-BOP test blocks and the location of the weld bead are shown schematically in Figure 5.2. The photograph in Figure 5.3 reveals the G-BOP test blocks with the two load cells incorporated, clamped in the jig which held the two blocks in place for the deposition of the weld. Figure 5.4 shows an instrumented G-BOP test blocks after welding.

5.2.2 Welding Parameters

The welding technique used to deposit the welds was gas shielded flux cored arc welding (FCAW). The deposition of test welds was carried out in a workshop at the Aeronautical and Maritime Research Laboratory, Ship Structures and Materials Division, DSTO, Melbourne. Automatic feeding of filler wire was used for the welding process. The equipment allowed welding current and voltage to be set to desired values. The welding travel speed was controlled by a trailer. During the welding process, the load cells and thermocouple were connected to a computer, in order that the reaction force and temperature profile could be monitored and recorded.

The aim heat input was 1.7 kJ/mm, but the heat inputs obtained during welding varied from 1.4 to 1.7 kJ/mm, as shown in Table 5.3. The letter code indicates whether the sample was deposited with Ar-20%CO₂ or Ar-5%CO₂ shielding gas mixture and the last two digits are in increasing order of the preheat temperature.

Test blocks were preheated by an underlay electric blanket, which is the pink coloured component in Figure 5.3. The preheat temperatures were measured on the top surface of the test blocks using a contact thermometer. Five different preheat temperatures were employed in the tests as follows,

20°C (no preheating), 40°C, 60°C, 80°C and 100°C.
### Table 5.3 Welding parameters

<table>
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<tr>
<th>Sample Code</th>
<th>Preheat Temperature, °C</th>
<th>Shielding Gas</th>
<th>Heat Input, kJ/mm</th>
<th>Load Cell</th>
<th>Thermocouple</th>
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</tr>
<tr>
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<tr>
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<td>1.55</td>
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<td>No</td>
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<tr>
<td>5-10</td>
<td>100</td>
<td>5%CO₂</td>
<td>1.58</td>
<td>No</td>
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</tr>
</tbody>
</table>

Carbon dioxide is a non-flammable gas and because its heat of formation is high it is a stable compound which can be mixed with argon to provide a gas shield around the arc to protect the molten metal from contamination by the atmosphere. Two kinds of shielding gas mixtures were used in the tests:

1. $19.8 \pm 0.2\%$ carbon dioxide, balance argon ($20\%$CO₂), and

2. $5.06 \pm 0.1\%$ carbon dioxide, balance argon ($5\%$CO₂).
Figure 5.5 Schematic illustration of the measurement of hardness in radial directions
The welded G-BOP test samples were allowed to stand for at least 24 hours before being released from the clamps, then the samples were broken open.

5.3 METALLOGRAPHY

All the specimens investigated for transverse cracking propensity were sectioned in the transverse direction of the test welds within 10 mm of the break opened fracture surface. The samples were cold mounted in epoxy resin. The mounted specimens were finished by mechanical polishing on a 1 µm diamond pad and by etching in 2.5% Nital.

5.3.1 Optical Metallography and Weld Bead Analysis

Macrography of the weld bead was recorded by a stereo macroscopy. Investigation of the effects of preheat temperatures and shielding gases on the microstructure of weldments were carried out by a Nikon optical microscope. After capturing the macro-image of the weld bead, image analysis was used to measure the areas of the weld metal and HAZ for each specimen. Dilution of the weld bead was also measured by using image analysis after redrawing the original joint geometry on the image.

The degree of dilution normally depends on the type of joint, the edge preparation and the process used. The percentage dilution, D is defined as [DuPont and Marden 1996]:

\[ Pct\ D = \frac{\text{Melted cross sectional area of the base metal}}{\text{Cross sectional area of filler metal}} \times 100 \]  

(5-3)

5.3.2 Hardness

Hardness measurements were taken by using a LECO M-400-H1 microhardness testing machine with a 200 g load. The hardness was measured in radial directions as shown in Figure 5.5. The distance between indentations was 0.5 mm in each direction. In order to calculate the average hardness value of the weld metal, HAZ and base metal under different welding conditions, twenty indentations were performed on the weld metal and
Figure 5.6 Schematic illustration of the sampling positions for the measurement of width of columnar grains in weld metal
HAZ for each specimen. Six indentations were placed on the base metal for each specimen.

5.3.3 Measurement of Width of Columnar Grains

In order to establish any difference in the size of columnar grains in the weld metals deposited with the two kinds of shielding gas mixtures at different preheat temperatures, the width of the columnar grains was determined for the two weld metals.

After the weld bead analysis and general microstructure examination, specimens were re-polished firstly on a 6 µm pad and finished on a 1 µm pad to remove the 2.5% Nital etch. The columnar grain boundaries were revealed by etching in picric acid at 50 to 60°C for about two minutes. The picric acid solution contains 100 ml of saturated picric acid, 10 drops of detergent and 10 drops of concentrated HCl. The measurement was carried out on microphotographs of each specimen, which were recorded using a Nikon optical microscope. Two photos were taken for each specimen at the same position as shown in Figure 5.6. The magnification was 110 times. Then three lines, five centimeters long, were drawn perpendicular to the columnar grains at random on each photomicrograph, and the number of columnar grains intersected along each line were counted. From these counts, the average widths of the columnar grains for the two weld metals were calculated for each preheat temperature.

5.4 FRACTOGRAPHY

5.4.1 Estimation of Cold Crack Percentage

After being broken open, the fracture surface of the weldments were immediately coated with Lacomite to protect the surface from oxidation and rust. The fracture surfaces were ultrasonically cleaned in acetone before examination. Estimation of the crack percentage was conducted by image analysis after capturing a macro-image of the fracture surface
of the cracked sample. The cracked area and total weld metal area were calculated by using the software package NIH Image 1.58.

\[
\text{Crack percentage} = \frac{\text{Cracked area}}{\text{Weld metal area}} \times 100 \tag{5-4}
\]

5.4.2 Fracture Surface Examination

Fracture surfaces of cracked samples of weld metals deposited with the two kinds of shielding gases at different preheat temperatures were examined using a Leica S440 scanning electron microscope (SEM).

5.4.3 Measurement of Microvoid Size on Fracture Surface of Weld Metal

The microvoid size on the fracture surface of cracked samples was determined by measuring the diameter of dimples on SEM fractographs for each sample. A total of twenty-five dimples were chosen at random from each sample for measurement, and the average microvoid size on the fracture surface was calculated for each of the weld metals deposited with different shielding gases at different preheat temperatures.

5.4.4 Measurement of the Percentage of Different Fracture Morphologies

The percentages of different fracture morphologies were measured on the fracture surfaces of weld metals deposited with 5%CO₂ and 20%CO₂ shielding gases respectively at 20°C and weld metal deposited with 5%CO₂ at 80°C preheat. The measurement was carried out using image analysis after capturing low magnification SEM images which showed the detailed features of the fracture surface. The fracture surfaces of the two weld metals contained several different morphologies: dimples, quasi-cleavage and intergranular fractures.

5.4.5 Observation of Cross Section of the Fracture Surface

In order to ascertain the microstructure corresponding to the fracture surface, the
Figure 5.7 Schematic diagrams of nickel electroplating method for fracture surface of weld metal
Figure 5.8 Schematic illustration of cross section of nickel plated fracture surface of weld metal.
fracture surface samples of two weld metals deposited with 5%CO₂ and 20%CO₂ shielding gases respectively at 20°C were electroplated with nickel. A schematic diagram for nickel electroplating is shown in Figure 5.7. Firstly, the fracture surface of the sample was electrolytically cleaned at about 90°C for two minutes as shown in Figure 5.7a, and then the cleaned samples were electroplated with nickel at about 30 to 40°C for 8 to 10 hours as shown in Figure 5.7b.

The nickel plated samples were cut to establish a longitudinal cross section at about 90° to the fracture surface of the weld metal. Figure 5.8 shows a schematic illustration of a longitudinal cross section of the fracture surface. The cross sectioned samples were cold mounted in epoxy resin and prepared by grinding and mechanical polishing to a 1 µm finish, then etching with 2.5% Nital solution to identify the phases present in the microstructure. The microstructures corresponding to the cross-sections of the weld metal fracture surfaces were observed by using both the Nikon optical microscope and the Leica S440 scanning electron microscope.

In order to determine the relationship between the weld metal fracture surface and the orientation of the columnar grains, the longitudinal cross section samples were repolished on a 6 µm and then a 1 µm diamond pads to remove the 2.5% Nital etch and then etched with picric acid at 50 to 60°C for about two minutes after examination of the corresponding microstructures. Then the cross sectioned samples were observed using a Nikon optical microscope.

5.5 NON-METALLIC INCLUSION ANALYSIS

Quantitative measurements of non-metallic inclusion sizes can be used to characterise the inclusion population in the weld metal and to assess their effects on the microstructure and mechanical properties of the weld metal.
The samples for the analysis of non-metallic inclusions in the weld metal were finished by mechanical polishing on a 1 µm diamond pad.

The non-metallic inclusions in the weld metals were examined by using a Leica S440 scanning electron microscope under an accelerating voltage of 20 kV. The magnification of the SEM image was 5000 times. Over 350 inclusions, in total, were included in the SEM image used for the analysis of each sample. The size distributions of non-metallic inclusions in the weld metal were determined by image analysis, using the software package NIH Image 1.58. Energy dispersive spectroscopy (EDS) analysis was used to determine the alloying elements present in the non-metallic inclusions.

The important inclusion characteristics are the number of particles \( N \); the volume fraction \( V_v \); the arithmetic mean two dimensional and three dimensional particle diameters, \( \bar{d}_a \) and \( \bar{d}_v \); the area fraction of particles \( A_a \); the number of particles per unit area \( N_a \); the number of particle per unit volume \( N_v \); the mean particle centre to centre volume spacing \( \lambda_v \); the total particle surface area per unit volume \( S_v \); and the mean area of the particles \( \bar{A} \). The area of particles defined as \( A \) can be obtained by using NIH Image 1.58 directly, other characteristics were calculated by computer according the following basic stereometric relationships [ASTM 1989; Underwood 1970; Kluken et al. 1988],

\[
V_v = A_a = \frac{\text{Area of the detected features}}{\text{Field area}} \quad (5-5)
\]

\[
d_a = 2 \sqrt{\frac{\bar{A}}{\pi}} \quad (5-6)
\]

\[
\bar{d}_a = \frac{\Sigma d_a}{N} \quad (5-7)
\]

\[
\bar{d}_v = \pi \frac{d_a}{2} \quad (5-8)
\]
Figure 5.9 Testing machine and control system of Theta Industries Dilatronic® dilatometer
Figure 5.10 Dimensions of hollow cylindrical sample for dilatometry tests. All dimensions in millimeters
\[ N_V = 6 \frac{V_V}{\pi d_V^3} \]  \hspace{1cm} (5-9) 

\[ N_s = N_V d_V \]  \hspace{1cm} (5-10) 

\[ \lambda_v = 0.554 \sqrt[3]{\frac{1}{V_V}} \]  \hspace{1cm} (5-11) 

\[ S_v = \pi N_V (d_V)^2 \]  \hspace{1cm} (5-12) 

and

\[ \bar{A} = \frac{A_{A\kappa}}{N_a} \]  \hspace{1cm} (5-13)

5.6 CONSTRUCTION OF CONTINUOUS COOLING TRANSFORMATION DIAGRAMS

In the present work, the CCT diagrams for the two weld metals and the base metal were determined using dilatometry and microstructural observation. The weld metal \( \gamma \) to \( \alpha \) transformation start and finish temperatures (\( A_{\gamma} \) and \( A_{\alpha} \)) under various cooling rates were determined, the resultant microstructures were examined and the hardness was measured for all dilatometer samples.

5.6.1 Dilatometry

A computer aided high speed heating and quenching Theta Industries Dilatronic® dilatometer was used to measure the variation in one dimension (linear thermal expansion) of a sample with changing temperature. The dilatometer was positioned horizontally and was designed so as to measure expansion in the direction of the long axis. The photograph in Figure 5.9 shows the testing machine and control system of the dilatometry equipment.

Hollow cylindrical samples which were cut from the all weld metals and the base metal were provided by DSTO. Sample dimensions are as shown in Figure 5.10. An R type
Figure 5.11 Specimen holder and quenching device of Theta Industries Dilatronic®
dilatometer
Figure 5.12 Schematic diagram of thermal cycles
Figure 5.13 Schematic illustration of method to determine the temperatures $Ar_3$ and $Ar_1$
A thermocouple was spot welded onto the sample outer surface in a central position with respect to length to monitor the temperature during the thermal cycle.

Figure 5.11 shows the quenching device of the Theta dilatometer. The sample was placed between two sample holders made of quartz. To prevent oxidization, the test was performed under a vacuum of about \(10^{-4}\) torr. The sample was heated by induction via a high-frequency AC coil, which is approximately 40 mm long and 15 mm in internal diameter. The sample was located directly in the centre of this coil during heat treatment. The cooling medium used in the test was high purity helium which has high heat capacity.

Samples were austenitised at 1250°C for 13 seconds to produce large average prior austenite grain sizes and then cooled from the peak temperature to room temperature at varying constant cooling rates in the range of 1°C/sec. - 380°C/sec. from 800 to 500°C by employing helium gas. The phase transformations that took place were examined dilatometrically, metallographically and the constituents present in each sample were determined.

5.6.2 Thermal Cycles

The thermal cycles were as shown in Figure 5.12. To reproduce a similar thermal cycle to actual welding, linear cooling was used from 1250 to 900°C, and from 900°C to room temperature exponential cooling was adopted to provide a thermal profile which is closely similar to the actual welding thermal cycle.

5.6.3 Determination of \(\text{Ar}_3\) and \(\text{Ar}_1\)

For all samples that were heat treated in the dilatometer a plot of expansion versus temperature and a plot of temperature versus time were produced. A method that has been used to determine the transformation start and finish temperatures (\(\text{Ar}_3\) and \(\text{Ar}_1\)) is shown in Figure 5.13. In this method, two points were selected on the straight section
Figure 5.14 Mounting method for each dilatometer sample
of the expansion versus temperature curve just below and just above the transformation regime, and then tangents were drawn through these points on either side of the transformation. The point at which the curve deviated from linearity was determined to be the temperature at which the transformation starts or finishes. The temperature value at this point was determined by the analysis program and was recorded. Using the computer analysis program, this point is located by identifying where there is a 2.5 percent deviation in the expansion value between the extended straight line and the actual curve of expansion versus temperature. The transformation start and finish temperatures reported for each sample are mean values of four measurements. Finally, the CCT diagrams were constructed by plotting temperature versus time.

5.6.4 Optical MetalloKraphy

All samples were cold mounted in epoxy resin with the longitudinal axis positioned parallel to the polished face as shown in Figure 5.14, and then finished by mechanically polishing on a 1 µm diamond pad and by etching in 2.5% Nital solution to reveal the general microstructure. The microstructural constituents corresponding to different transformation temperatures at varying cooling rates were identified by using a Leica optical microscopy.

5.6.5 Hardness

The hardness of all samples was measured using a LECO M-400-H1 microhardness testing machine with a 200 g load. Seven indentations were taken per sample along the length of the sample at the centre of the polished surface. The distance between indentations was 1 mm. A typical hardness value was obtained by averaging the measured results.

5.6.6 Measurement of Prior Austenite Grain Size

The linear intercept austenite grain sizes of the weld metals deposited with 5%CO₂ and 20%CO₂ shielding gases and base metal (BIS812EMA steel) were determined by using
delineation of the prior austenite grain boundaries by ferrite allotriomorphs. The measurements were made for dilatometer samples of the all weld metals and base metal reheated to 1250°C for thirteen seconds. A Nikon optical microscope was used with the aid of a scaled eye piece.

The prior austenite grain boundaries were revealed by etching in picric acid solution at 50 to 60°C for about two minutes. The final prior austenite grain sizes of the weld metals and base metal were then calculated using the average grain size of twenty-five measured grains for each sample.

### 5.7 TEMPERING OF WELD METALS

To study precipitation hardening by alloy carbides, samples of the two weld metals deposited with 5%CO₂ and 20%CO₂ shielding gases respectively without preheat (20°C) were heated to 550°C and tempered for various periods of time as follows, 0.25 hour, 0.5 hour, 1 hour, 2 hours and 72 hours.

All samples were cold mounted in epoxy resin and finished by mechanical polishing on a 1 µm diamond pad, and then by etching in 2.5% Nital to reveal the microstructure.

The microstructures were examined by using a Nikon optical microscope. A JEOL 2000 FX transmission electron microscope (TEM) were employed in an attempt to reveal carbide precipitates in the two weld metals without tempering and tempered at 550°C for 0.5 hour. Thin foils were prepared for transmission electron microscopy examination. First, disc samples of 3 mm in diameter were mechanically thinned to about 100 µm in thickness by grinding on silicon carbide paper of grade 1200. Then they were electropolished using a Struers twin jet polisher (TENUPOL-2) in a solution of 10% perchloric acid and 90% glacial acetic acid at room temperature. The electropolishing voltage was 110 V.
Hardness measurement for all the samples was carried out using a LECO M-400-H microhardness tester with a 200 g load. The data consisted of twenty indentations for each sample, and the distance between indentations was 0.5 mm. Finally, the average hardness was calculated for each sample.
CHAPTER 6

RESULTS
Figure 6.1 Cooling times $\Delta t_{8/5}$ and $\Delta t_{8/3}$ for different preheat temperatures obtained in G-BOP test welds using Ar-20%CO$_2$ shielding gas mixture.
Figure 6.2 Cooling time $\Delta t_{3/1.5}$ for different preheat temperatures obtained in G-BOP test welds using Ar-20\%CO$_2$ shielding gas mixture
The results of experimental work which are described in Chapter 5 are presented in this Chapter. To clarify the effects of shielding gas and preheat on weld metal transverse hydrogen induced cold cracking, the welding thermal cycles and reaction forces; and the weld metal hardness, composition, microstructures and non-metallic inclusions were systematically analyzed.

6.1 MEASUREMENT OF THERMAL CYCLES IN G-BOP TEST

The microstructure and mechanical properties of the weld metal are highly dependent on the welding thermal cycle, therefore, it is necessary to measure the actual welding thermal cycles, particularly cooling cycles which will affect the final microstructure and hydrogen content in the weld metal.

The temperature profile from an embedded thermocouple was monitored during several tests, in which the weld metals were deposited at different preheat temperatures. After welding, the cooling rates were determined from the recorded data. The cooling rate was characterised by cooling times $\Delta t_{8/5}$, $\Delta t_{8/3}$ and $\Delta t_{3/1.5}$. According to the literature, susceptibility to cold cracking is governed by the cooling time $\Delta t_{3/1}$ for a constant hydrogen level in the weld metal [Chakravarti and Bala 1989]. However, in this work, the cooling time $\Delta t_{3/1}$ for weld metals deposited at different preheat temperatures could not be measured due to a limitation in data recorded during the tests. Therefore, cooling time $\Delta t_{3/1.5}$ was determined instead of $\Delta t_{3/1}$ in the present work. The measured values of $\Delta t_{8/5}$, $\Delta t_{8/3}$ and $\Delta t_{3/1.5}$ are listed in Table 6.1.

The plots of $\Delta t_{8/5}$ and $\Delta t_{8/3}$ for different preheat temperatures are shown in Figure 6.1, and the plot of $\Delta t_{3/1.5}$ for different preheat temperatures is shown in Figure 6.2. These two figures show the expected trend of decreasing cooling rate with increasing preheat temperature. The figures also show that the cooling rate was significantly decreased for
Figure 6.3 Longitudinal stress across the gap obtained in G-BOP test welds
a higher preheat temperature. The plots for the thermal responses of welding at different preheat temperatures are attached in Appendix I.

Table 6.1 Cooling times $\Delta t_{8/5}$, $\Delta t_{8/3}$ and $\Delta t_{3/1.5}$ and peak temperatures during welding

<table>
<thead>
<tr>
<th>Sample Code</th>
<th>Preheat Temperature, °C</th>
<th>Heat Input, kJ/mm</th>
<th>Peak Temperature, °C</th>
<th>$\Delta t_{8/5}$, Second</th>
<th>$\Delta t_{8/3}$, Second</th>
<th>$\Delta t_{3/1.5}$, Second</th>
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</thead>
<tbody>
<tr>
<td>20-01</td>
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<td>1.72</td>
<td>1015.3</td>
<td>8.03</td>
<td>19.11</td>
<td>26.78</td>
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<td>40</td>
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<td>1277.7</td>
<td>9.75</td>
<td>24.34</td>
<td>211.74</td>
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6.2 LONGITUDINAL STRESS DURING SOLIDIFICATION

Determination of longitudinal stress across the gap during weld metal solidification is helpful for the investigation of weld metal transverse cold cracking behaviour. In this work, the main purpose of measuring stress across the gap during weld deposition is to determine the time and temperature of initiation of weld metal transverse cracking after welding. The exact magnitude of the restraint stress developed in the G-BOP test has not been determined.

Records of load against time and load against temperature during weld metal cooling were made for four tests. The pre-load and the maximum load during the tests are listed in Table 6.2. The longitudinal stress across the gap during weld metal solidification was calculated using the measured data of reaction force across the gap and the weld metal area for the samples listed in Table 6.2. The stress evolution with time is shown in Figure 6.3. There is little obvious change in stress with changing preheat temperature.
Figure 6.4 Longitudinal stress evolution as a function of cooling temperature in G-BOP test welds using Ar-20%CO₂ shielding gas mixture
1. Sample 5-05: Preheat temp.: 40°C, 100% cracked, (with load cells)

Fracture surface  
Polished surface

2. Sample 5-09: Preheat temp.: 80°C, 49% cracked, (without load cell)

Fracture surface  
Polished surface

Figure 6.5 Typical macroviews of weld bead shapes of G-BOP tests deposited with Ar-5%CO₂ shielding gas mixture
1. Sample 20-01: Preheat temp.: 20°C, 100% cracked, (with load cells)

Fracture surface  Polished surface

2. Sample 20-02: Preheat temp.: 20°C, 69% cracked, (without load cell)

Fracture surface  Polished surface

1mm

Figure 6.6 Typical macroviews of weld bead shapes of G-BOP tests deposited with Ar-20%CO₂ shielding gas mixture
The weld metal cracking occurred at least five minutes after welding as shown in Figure 6.3.

Table 6.2 The actual load during the welding tests

<table>
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<tr>
<th>Sample Code</th>
<th>Shielding Gas</th>
<th>Preheat Temperature, °C</th>
<th>Pre-load, kN</th>
<th>Maximum Load, kN</th>
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<tr>
<td>20-04</td>
<td>20%CO₂</td>
<td>40</td>
<td>3.70</td>
<td>23.11</td>
</tr>
<tr>
<td>20-07</td>
<td>20%CO₂</td>
<td>60</td>
<td>3.02</td>
<td>21.03</td>
</tr>
<tr>
<td>5-05</td>
<td>5%CO₂</td>
<td>40</td>
<td>1.81</td>
<td>21.78</td>
</tr>
</tbody>
</table>

Figure 6.4 shows the stress evolution with cooling temperature in the tests. The stress dipped over an intermediate temperature range (750 to 500°C for 60°C preheat; 500 to 250°C for 20°C and 40°C preheat) before rising steeply. The dip is probably associated with transformation of austenite and the rapid stress rise following the dip is due to the rising strength of the ferrite and increasing contractional stress. The stress decreased sharply when cracking occurred. It was found that all cracks occurred below 150°C. It is also clear that the development of stress on cooling from 500 to 200°C occurred more gradually and at a higher temperature for the higher preheat weld.

6.3 METALLOGRAPHY

6.3.1 Geometry of Weld Bead and Dilution

Cracking in the weld metal is associated with stress concentration in the weld deposit area [Li et al. 1995] and the geometry of the weld bead can influence the susceptibility to weld metal cold cracking, particularly the stage of crack initiation.

Typical macrophotographs of G-BOP test samples are shown in Figures 6.5 and 6.6, which illustrate the shapes of the weld beads at fracture sites and at polished transverse
Figure 6.7 Areas of weld metal and HAZ deposited with (a) 5% CO$_2$ and (b) 20% CO$_2$ shielding gas mixtures in G-BOP tests.
Figure 6.8 Dilution of weld metal with base metal of G-BOP test samples
cross sections taken from within 10 mm of the fracture surface. The typical weld beads for 5%CO₂ shielding gas are shown in Figure 6.5 and for 20%CO₂ shielding gas in Figure 6.6. The weld metal cold cracking is revealed clearly on the fracture surface.

The areas of the weld metal and HAZ for G-BOP test samples were measured and the results are listed in Table 6.3 and plotted in Figure 6.7(a) for 5%CO₂ shielding and in Figure 6.7(b) for 20%CO₂ shielding gas. The data show that the weld metal area for 20%CO₂ shielding gas was generally larger than that for 5%CO₂ shielding gas at the same preheat temperature. This behaviour could be a contributing factor in the observation that weld metal deposited with 20%CO₂ shielding gas was less susceptible to cold cracking.

The final deposit weld metal composition will depend on the individual compositions of the base metal and the flux cored wire; the welding conditions (current, voltage, speed and shielding gas) which determine reactions in the arc and the extent of dilution. The dilution of weld metal with base metal was also measured as listed in Table 6.3, and the measured results are plotted in Figure 6.8. It was found that at the same preheat temperature, the percent dilution for 20%CO₂ shielding gas was larger than for 5%CO₂ shielding gas when the preheat temperatures were ≥ 60°C. For other preheat temperatures (20°C to 40°C) the dilutions with 20%CO₂ and 5%CO₂ shielding gases were similar.

### 6.3.2 Weld Metal Compositions

The chemical compositions of the 5%CO₂ and 20%CO₂ all weld metals were analysed by DSTO, and the results are shown in Table 6.4, along with the calculated carbon equivalents, which provide a comparison of relative weld metal hardenability. Considering the effect of oxygen content on the hardenability of the weld metals, carbon equivalent was calculated using the equation proposed by Onsøien et al. [1996]:
In order to understand the influences of shielding gas and dilution on the compositions of the two weld metals, the chemical compositions of the deposited weld metals using 5%CO₂ and 20%CO₂ shielding gases were analysed by Central Laboratory, BHP Flat Products, Wollongong. The compositions of the two deposited weld metals are shown in Table 6.5, along with the calculated carbon equivalent $P_{cno}$.

$$P_{cno} = C + \frac{Mn + Cu + Cr}{20} + \frac{Si}{30} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B - \frac{3}{4}O$$  \hspace{1cm} (6-1)

### Table 6.3 Weld bead and HAZ areas and the percentage of dilution

<table>
<thead>
<tr>
<th>Sample Code</th>
<th>Shielding Gas</th>
<th>Preheat Temperature, °C</th>
<th>Heat Input, kJ/mm</th>
<th>Weld Metal Area, mm²</th>
<th>HAZ Area, mm²</th>
<th>Dilution, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>20-01</td>
<td>20%CO₂</td>
<td>20</td>
<td>1.72</td>
<td>36.29</td>
<td>48.79</td>
<td>19.9</td>
</tr>
<tr>
<td>20-02</td>
<td>20%CO₂</td>
<td>20</td>
<td>1.60</td>
<td>40.87</td>
<td>45.00</td>
<td>18.0</td>
</tr>
<tr>
<td>20-03</td>
<td>20%CO₂</td>
<td>20</td>
<td>1.65</td>
<td>41.08</td>
<td>42.21</td>
<td>18.3</td>
</tr>
<tr>
<td>20-04</td>
<td>20%CO₂</td>
<td>40</td>
<td>1.68</td>
<td>35.89</td>
<td>51.57</td>
<td>21.5</td>
</tr>
<tr>
<td>20-05</td>
<td>20%CO₂</td>
<td>40</td>
<td>1.40</td>
<td>41.88</td>
<td>50.92</td>
<td>25.1</td>
</tr>
<tr>
<td>20-06</td>
<td>20%CO₂</td>
<td>60</td>
<td>1.54</td>
<td>40.14</td>
<td>47.34</td>
<td>23.9</td>
</tr>
<tr>
<td>20-07</td>
<td>20%CO₂</td>
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<td>1.52</td>
<td>39.07</td>
<td>53.18</td>
<td>24.9</td>
</tr>
<tr>
<td>20-08</td>
<td>20%CO₂</td>
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<td>1.56</td>
<td>36.14</td>
<td>49.27</td>
<td>23.9</td>
</tr>
<tr>
<td>20-09</td>
<td>20%CO₂</td>
<td>100</td>
<td>1.57</td>
<td>36.70</td>
<td>61.84</td>
<td>24.6</td>
</tr>
<tr>
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<td>1.54</td>
<td>34.28</td>
<td>65.97</td>
<td>19.0</td>
</tr>
<tr>
<td>5-02</td>
<td>5%CO₂</td>
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<td>38.47</td>
<td>64.12</td>
<td>20.3</td>
</tr>
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<td>5-03</td>
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<td>1.40</td>
<td>37.92</td>
<td>62.46</td>
<td>20.8</td>
</tr>
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<td>5-04</td>
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<td>39.23</td>
<td>64.63</td>
<td>22.7</td>
</tr>
<tr>
<td>5-05</td>
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<td>1.57</td>
<td>30.90</td>
<td>46.57</td>
<td>24.3</td>
</tr>
<tr>
<td>5-06</td>
<td>5%CO₂</td>
<td>60</td>
<td>1.55</td>
<td>28.36</td>
<td>40.75</td>
<td>21.6</td>
</tr>
<tr>
<td>5-07</td>
<td>5%CO₂</td>
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<td>35.32</td>
<td>53.92</td>
<td>17.4</td>
</tr>
<tr>
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<td>5%CO₂</td>
<td>80</td>
<td>1.75</td>
<td>38.50</td>
<td>52.02</td>
<td>22.0</td>
</tr>
<tr>
<td>5-09</td>
<td>5%CO₂</td>
<td>80</td>
<td>1.55</td>
<td>35.06</td>
<td>47.39</td>
<td>18.1</td>
</tr>
<tr>
<td>5-10</td>
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<td>1.58</td>
<td>33.52</td>
<td>51.78</td>
<td>20.8</td>
</tr>
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### Table 6.4 Chemical compositions (wt%) and carbon equivalent of all weld metal samples

<table>
<thead>
<tr>
<th>Shielding Gas</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>V</th>
<th>Ti</th>
<th>Cu</th>
<th>Al</th>
<th>Nb</th>
<th>B</th>
<th>N</th>
<th>Co</th>
<th>O</th>
<th>P$_{cmo}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>5%CO$_2$</td>
<td>0.02</td>
<td>0.28</td>
<td>1.85</td>
<td>0.007</td>
<td>0.005</td>
<td>0.03</td>
<td>2.59</td>
<td>0.01</td>
<td>0.02</td>
<td>0.03</td>
<td>0.03</td>
<td>0.007</td>
<td>0.01</td>
<td>&lt;0.001</td>
<td>0.0089</td>
<td>&lt;0.002</td>
<td>0.039</td>
<td>0.146</td>
</tr>
<tr>
<td>20%CO$_2$</td>
<td>0.02</td>
<td>0.18</td>
<td>1.61</td>
<td>0.007</td>
<td>0.005</td>
<td>0.03</td>
<td>2.54</td>
<td>0.01</td>
<td>0.02</td>
<td>0.02</td>
<td>0.03</td>
<td>0.006</td>
<td>0.01</td>
<td>&lt;0.001</td>
<td>0.0040</td>
<td>&lt;0.002</td>
<td>0.046</td>
<td>0.125</td>
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</tbody>
</table>

### Table 6.5 Chemical compositions (wt%) and carbon equivalent of diluted weld metal samples

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<tr>
<th>Shielding Gas</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>V</th>
<th>Ti</th>
<th>Cu</th>
<th>Al</th>
<th>Nb</th>
<th>B</th>
<th>N</th>
<th>Co</th>
<th>O</th>
<th>P$_{cmo}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>5%CO$_2$</td>
<td>0.070</td>
<td>0.33</td>
<td>1.87</td>
<td>0.010</td>
<td>0.005</td>
<td>0.12</td>
<td>2.35</td>
<td>0.072</td>
<td>0.021</td>
<td>0.033</td>
<td>0.058</td>
<td>0.008</td>
<td>0.018</td>
<td>0.0018</td>
<td>0.0083</td>
<td></td>
<td>0.036</td>
<td>0.212</td>
</tr>
<tr>
<td>20%CO$_2$</td>
<td>0.055</td>
<td>0.24</td>
<td>1.64</td>
<td>0.009</td>
<td>0.004</td>
<td>0.11</td>
<td>2.25</td>
<td>0.086</td>
<td>0.019</td>
<td>0.023</td>
<td>0.049</td>
<td>0.005</td>
<td>0.014</td>
<td>0.0009</td>
<td>0.0042</td>
<td></td>
<td>0.050</td>
<td>0.165</td>
</tr>
</tbody>
</table>
Figure 6.9 Microstructures of weld metals deposited with 5%CO₂ shielding gas at different preheat temperatures.
Figure 6.10 Microstructures of weld metals deposited with 20%CO₂ shielding gas at different preheat temperatures
The level of diffusible hydrogen in the welds deposited using the two shielding gas mixtures were measured at DSTO for the all weld deposits and the results are shown in Table 6.6.

Table 6.6 Level of diffusible hydrogen in the deposited metal

<table>
<thead>
<tr>
<th>Shielding Gas</th>
<th>Diffusible Hydrogen (ml/100g deposited metal)</th>
</tr>
</thead>
<tbody>
<tr>
<td>5%CO₂</td>
<td>8.6</td>
</tr>
<tr>
<td>20%CO₂</td>
<td>6.4</td>
</tr>
</tbody>
</table>

6.3.3 Optical Microstructure

The weld pool solidification structure can be characterized by its microstructure and its macrostructure. The microstructure depends on chemical composition and cooling rate. The mechanical properties of the weld metal strongly depend on the solidification microstructure. A susceptible microstructure is one of the important factors in the formation of transverse weld metal cold cracking. Once a cold crack occurs, the course and extent of the crack path in the weld metal can be determined by the microstructure in combination with the direction of stress [Li et al. 1995].

The microstructures of the G-BOP test samples were examined in detail. Weld metals deposited with 5%CO₂ and 20%CO₂ shielding gases at different preheat temperatures are shown in Figure 6.9 and Figure 6.10, respectively.

It was observed that the microstructures were mainly acicular ferrite and bainitic ferrite (ferrite with aligned second phase). In some cases, grain boundary ferrite with Widmänstatten ferrite side plates were observed. Comparing the micrographs in Figures 6.9 and 6.10, it can be concluded that the weld metal deposited with 20%CO₂ shielding gas contained more bainitic ferrite. For deposits produced by the same shielding gas, there were few obvious differences in microstructure for different preheat temperatures, but the ferrite plate thickness appeared to increase as preheat temperature was raised.
Figure 6.11 Microstructures of the grain coarsened HAZ for 5%CO₂ shielding gas at different preheat temperatures
Figure 6.12 Microstructures of the grain coarsened HAZ for 20%CO₂ shielding gas at different preheat temperatures
Figure 6.13 Hardness profiles for weld metal and HAZ with 5%CO₂ shielding gas mixture
Figure 6.14 Hardness profiles for weld metal and HAZ with 20% CO₂ shielding gas mixture
Figure 6.15 Average hardness values and standard deviations for the two weld metals, a. deposited with 5%CO₂ shielding gas; b. deposited with 20%CO₂ shielding gas
Figure 6.16 Average hardness values and standard deviations for heat affected zones, a. with 5%CO₂ shielding gas; b. with 20%CO₂ shielding gas
Figure 6.17 Comparison of average hardness values for the two steel weld metals at different preheat temperatures
Figure 6.18 Comparison of average hardness values for heat affected zones at different preheat temperatures
Figures 6.11 and 6.12 respectively show the microstructures of the grain coarsened area of the HAZ of the base metal for 5\%CO_2 and 20\%CO_2 shielding gas at different preheat temperatures. It is obvious that the grain coarsened area is substantially martensitic.

### 6.3.4 Hardness

It is well known that cold cracking in weld metal is strongly related to the hardness value [Pussegoda and Tyson 1981].

Microhardness profiles in radial directions shown in Figure 6.13 were determined for samples at different preheat temperatures with 5\%CO_2 shielding gas, and Figure 6.14 shows the corresponding traverses for 20\%CO_2 shielding gas. In all samples the hardness of the grain coarsened HAZ, which showed a martensitic microstructure, was found to be higher than for the weld metal, and the hardness of the weld metal near the fusion boundary was found to be higher than the other regions of the weld metal.

The average values of microhardness of the weld metal and HAZ have been calculated and are listed together with their standard deviations in Table 6.7. The average hardness values and standard deviations as a function of preheat temperature are plotted in Figure 6.15 for the two weld metals and in Figure 6.16 for the HAZs. For each weldment sample, there was no significant change in hardness throughout the weld metal except for the area near the fusion line. The standard deviation of weld metal hardness values was less than twenty points. However, the weld HAZ basically consists of three sub-zones: grain coarsened; grain refined and intercritical zones, and the hardness values varied significantly from zone to zone, resulting in the large standard deviations for HAZ hardness values: most standard deviations were larger than thirty-five points.

A comparison of the average hardness values for different preheat temperatures is shown in Figure 6.17 for the weld metals and in Figure 6.18 for the HAZs. It was found that the average values of weld metal hardness with 5\%CO_2 shielding gas were higher than for 20\%CO_2 shielding gas at the same preheat temperature, particularly
Figure 6.19 Features of columnar grains (or cells) in weld metal deposited with 5% CO₂ shielding gas at different preheat temperatures.
Figure 6.20 Features of columnar grains (or cells) in weld metal deposited with 20% CO₂ shielding gas at different preheat temperatures
Figure 6.21 Average width of the columnar grains (or cells) for the two steel weld metals at different preheat temperatures.
when the preheat temperature was lower than 60°C. In contrast and as expected, the average values of hardness of the HAZ for 5%CO₂ and 20%CO₂ shielding gases were similar at the same preheat temperature.

Table 6.7 Average microhardness values and standard deviations for weld metals and HAZs

<table>
<thead>
<tr>
<th>Preheat Temperature, °C</th>
<th>5%CO₂</th>
<th>20%CO₂</th>
<th>5%CO₂</th>
<th>20%CO₂</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
<td>330.3±18.1</td>
<td>291.7±12.4</td>
<td>394.0±36.8</td>
<td>396.5±39.3</td>
</tr>
<tr>
<td>40</td>
<td>307.8±13.3</td>
<td>291.2±14.8</td>
<td>399.3±37.7</td>
<td>398.0±33.6</td>
</tr>
<tr>
<td>60</td>
<td>314.0±10.7</td>
<td>283.9±16.5</td>
<td>376.4±40.0</td>
<td>376.6±38.0</td>
</tr>
<tr>
<td>80</td>
<td>300.5±12.4</td>
<td>282.2±14.1</td>
<td>370.0±36.3</td>
<td>375.9±40.0</td>
</tr>
<tr>
<td>100</td>
<td>291.1±19.2</td>
<td>286.8±9.0</td>
<td>372.5±44.9</td>
<td>385.8±42.7</td>
</tr>
</tbody>
</table>

6.3.5 Columnar Grains

The development of solidification structures within a weld is highly dependent on the solidification conditions, which are a function of welding parameters, such as heat input, consumable type and shielding gas. The solidification process of the welds investigated in this study is characterised by columnar growth. In order to assess the effect of shielding gas and preheat on solidification, the average widths of columnar grains in the different weld metals have been measured in this work.

The features of the columnar grain in weld metals deposited with 5%CO₂ and 20%CO₂ shielding gases at different preheat temperatures are shown in Figures 6.19 and 6.20.

The average width of the columnar grains together with their standard deviations for the two weld metals are listed in Table 6.8 as a function of preheat temperature. Figure 6.21 illustrates the average widths of the columnar grains for the two weld metals obtained at different preheat temperatures.
Table 6.8 Average widths of columnar grains and standard deviations in weld metals

<table>
<thead>
<tr>
<th>Preheat temperature, °C</th>
<th>Average width and standard deviation, µm</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>5% CO₂ shielding gas</td>
</tr>
<tr>
<td>20</td>
<td>29.04±2.93</td>
</tr>
<tr>
<td>40</td>
<td>32.71±2.91</td>
</tr>
<tr>
<td>60</td>
<td>32.88±2.74</td>
</tr>
<tr>
<td>80</td>
<td>33.01±2.25</td>
</tr>
<tr>
<td>100</td>
<td>32.97±2.38</td>
</tr>
</tbody>
</table>

It was found that there was a little difference in the average widths of columnar grains between the two weld metals. The average width of the columnar grains in the weld metal deposited with 5%CO₂ shielding gas was smaller than for 20%CO₂ shielding gas at 20°C, but for higher preheat temperatures, the average width of columnar grains in the weld metal deposited with 5%CO₂ shielding gas was larger than for 20%CO₂ shielding gas at the same preheat temperature. On the other hand, for 5%CO₂ shielding gas, the average widths of the columnar grains were slightly increased as the preheat temperature was increased from 20 to 40°C, and the average widths of the columnar grains were very similar between 40 and 100°C. For 20%CO₂ shielding gas, the average width of the columnar grains was relatively insensitive to preheat temperature (about 30 µm).

6.4 FRACTOGRAPHY

6.4.1 Estimation of Percentage Cracking

The estimation of cold cracking percentage in weld metal is very important for the investigation of the effects of welding parameters, such as preheat temperature and shielding gas on cold cracking behaviour.
Figure 6.22 The crack percentages for the two steel weld metals from G-BOP tests without load cell in the test assemblies.
The measured cracking percentages in the two weld metals are summarised in Table 6.9.

Table 6.9 Cold cracking percentage of the weld metals

<table>
<thead>
<tr>
<th>Sample Code</th>
<th>Shielding Gas</th>
<th>Preheat Temperature, °C</th>
<th>Heat Input, kJ/mm</th>
<th>Load Cell</th>
<th>Crack Percentage, %</th>
</tr>
</thead>
<tbody>
<tr>
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<td>20%CO₂</td>
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<td>1.72</td>
<td>yes</td>
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</tr>
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<td>20%CO₂</td>
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<td>1.65</td>
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<td>20%CO₂</td>
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<td>0</td>
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<td>5%CO₂</td>
<td>40</td>
<td>1.57</td>
<td>yes</td>
<td>100</td>
</tr>
<tr>
<td>5-06</td>
<td>5%CO₂</td>
<td>60</td>
<td>1.55</td>
<td>no</td>
<td>37</td>
</tr>
<tr>
<td>5-07</td>
<td>5%CO₂</td>
<td>60</td>
<td>1.53</td>
<td>no</td>
<td>77</td>
</tr>
<tr>
<td>5-08</td>
<td>5%CO₂</td>
<td>80</td>
<td>1.75</td>
<td>no</td>
<td>34</td>
</tr>
<tr>
<td>5-09</td>
<td>5%CO₂</td>
<td>80</td>
<td>1.55</td>
<td>no</td>
<td>49</td>
</tr>
<tr>
<td>5-10</td>
<td>5%CO₂</td>
<td>100</td>
<td>1.58</td>
<td>no</td>
<td>0</td>
</tr>
</tbody>
</table>

Figure 6.22 illustrates the cold cracking percentage for samples obtained without load cells. The results clearly indicate that the carbon dioxide content in the shielding gas has a strong effect on cold cracking, with 20%CO₂ shielding gas significantly reducing the susceptibility to cold cracking. There was no cold cracking in weld metal deposited with 20%CO₂ shielding gas at 40°C preheat and above. The results also show that the susceptibility to cold cracking decreases with increasing preheat temperature, as is
Figure 6.23 The crack percentages for the two steel weld metals in G-BOP tests with load cells in the test assemblies.
Figure 6.25 Low magnification micrographs of fracture regions of cold cracked G-BOP test samples in weld metals deposited with (a) 5%CO₂ and (b) 20%CO₂ shielding gases. Identified zones are described in the text.
Figure 6.26 Typical fracture morphologies of cold cracked surfaces of weld metal deposited with 5%CO₂ shielding gas. Zone A: QC; zone B: QC and MVC; zone C: IG and MVC; and zone D: MVC
Figure 6.27 Typical fracture morphologies of cold cracked surfaces of weld metal deposited with 20%CO₂ shielding gas. Zone A(a): MVC; zone A(b): QC and MVC; zone B: MVC; and zone C: IG and MVC
expected. Figure 6.23 illustrates the crack percentage in test samples obtained from test assemblies with load cells. The results also show that the susceptibility to cold cracking decreases with increasing preheat temperature.

Figure 6.24 compares the crack percentage in weld metal for test samples obtained with and without load cells at the same preheat temperature. It is obvious that the crack percentage for test samples with load cells is higher than that without load cells in the test assembly.

6.4.2 Fracture Morphology

Fracture morphology analysis greatly assists the study of the fine fractographic feature and the detailed fracture characteristics of the crack path in weld metal.

Low magnification SEM micrographs of the cold cracked fracture surface of the weld metals deposited with 5%CO₂ and 20%CO₂ shielding gases are shown in Figure 6.25a and 6.25b respectively. Figure 6.25a is a fractograph for a sample of 5%CO₂ and 20°C preheat which cracked completely through the weld cross section. The cracked fracture surface of the weld metal in this case consists of five zones (see Figure 6.25a): zone A is the initial cracking site; zone B is a mixed ductile and quasi-cleavage fractures zone; zone C is characterised by fracture along columnar grain boundaries; zone D is a ductile fracture zone and zone E is the shear lip. These zones were typical of the 5%CO₂ shielding gas welds. A fractograph for the conditions, 20%CO₂ shielding gas and 20°C preheat, is shown in Figure 6.25b for a sample which cracked partially (69%) through the weld cross section. It was found that the cracked fracture surface of the weld metal in this case consisted of three zones (see Figure 6.25b): zone A is the initial cracking site; zone B is a ductile fracture zone; and zone C contains small dark island regions consisting of intergranular fracture. Zone D on Figure 6.25b is the un-cracked area.

The detailed fracture morphologies of different zones in the weld metal deposited with 5%CO₂ and 20%CO₂ shielding gases are shown in Figures 6.26 and 6.27.
Figure 6.28 Typical fracture morphologies: comparison of dimple sizes for fracture surfaces
For weld metal deposited with 5% CO₂ shielding gas, the fracture surface near the initial crack site mainly consisted of quasi-cleavage (QC) fracture, see Figure 6.26 (zone A). Fracture occurred by a mixture of MVC and QC fracture in zone B, see Figure 6.26 (zone B). The fracture morphology changed to intergranular (IG) fracture together with limited MVC fracture when the crack followed the ferrite veins along the columnar grain boundaries, see Figure 6.26 (zone C). Fracture occurred by MVC in zone D, in which cavity nucleating non-metallic inclusions can be observed clearly in Figure 6.26 (zone D).

For weld metal deposited with 20% CO₂ shielding gas, the fracture surface near the initial crack site consisted of MVC, see Figure 6.27 [zone A(a)], or a combination of MVC and QC fracture, see Figure 6.27 [zone A(b)]. Most of the cracked fracture surface showed MVC fracture in zone B, nucleated by non-metallic inclusions, see Figure 6.27 (zone B). The black island regions (zone C) showed a mixture of IG and limited MVC fracture, see Figure 6.27 (zone C: island regions).

### 6.4.3 Microvoid Size on Fracture Surface of the Weld Metals

The results for microvoid size measurements can be used to ascertain the effect of welding parameters, especially shielding gas and preheat temperature on the features of ductile fracture surfaces of the weld metals.

Figure 6.28 shows the microvoid sizes on the fracture surfaces of cracked samples deposited under different welding conditions. All photographs were taken using the same magnification. There are some differences in the dimple size, which could be related to both local heterogeneity in inclusion size and spacing and some systematic trends in inclusion characteristics with preheat.

The average microvoid sizes and their standard deviations on cracked fracture surface for the two steel weld metals for different preheat temperatures are listed in Table 6.10.
Figure 6.29 Average microvoid sizes on the fracture surfaces for the two steel weld metals as a function of the preheat temperature.
Figure 6.30 Percentages of different fracture morphologies, a. weld metal deposited with 5%CO₂ shielding gas at 20°C preheat; and b. weld metal deposited with 20%CO₂ shielding gas at 20°C preheat.
Table 6.10 Average microvoid sizes and standard deviations on cracked fracture surfaces for weld metals

<table>
<thead>
<tr>
<th>Preheat temperature, °C</th>
<th>Average microvoid size and standard deviation, µm</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>5% CO₂ shielding gas</td>
</tr>
<tr>
<td>20</td>
<td>2.38 ± 0.84</td>
</tr>
<tr>
<td>40</td>
<td>2.19 ± 0.61</td>
</tr>
<tr>
<td>60</td>
<td>2.49 ± 1.00</td>
</tr>
<tr>
<td>80</td>
<td>2.66 ± 0.79</td>
</tr>
</tbody>
</table>

Figure 6.29 illustrates the average microvoid size for the two weld metals obtained at different preheat temperatures. It was found that the average microvoid size did not vary significantly with preheat. The average microvoid size decreased slightly when the preheat temperature was increased from 20 to 40°C, and then the average size increased slightly with increasing preheat temperature. However, systematic trends in inclusion characteristics with preheat temperature were not obvious. In addition, the standard deviations indicated that the microvoid sizes were not homogeneous.

6.4.4 Percentage of Different Fracture Morphologies

The fracture surface of a weld can reflect its toughness: a low toughness weld will usually show a brittle fracture surface, such as intergranular and cleavage fracture modes, whereas a high toughness weld will show a ductile fracture surface characterised by microvoids. Therefore, quantitative analysis of the fracture morphologies may clarify the susceptibility of the weld metal to transverse cold cracking.

Figure 6.30 illustrates the percentage of different fracture morphologies for weld metals deposited with 5% CO₂ shielding gas (Figure 6.30a) and 20% CO₂ shielding gas (Figure 6.30b) at 20°C preheat. It is clear that the crack morphologies of the weld metals are significantly different. The percentage of ductile fracture of the weld metal deposited
Figure 6.31 Typical microstructures associated with ductile fracture obtained in the two weld metals, a&b: SEM micrographs of cross sections of nickel plated crack surfaces; c&d: optical micrographs of cross sections of nickel plated crack surfaces
Figure 6.32 Typical microstructures associated with cleavage fracture obtained in the two weld metals, a&b: SEM micrographs of cross sections of nickel plated crack surfaces; c&d: optical micrographs of cross sections of nickel plated crack surfaces
with 20%CO₂ shielding gas is much higher than for 5%CO₂ shielding gas. Therefore, it can be deduced that the toughness of the weld metal deposited with 20%CO₂ shielding gas is higher.

6.4.5 Microstructures Underlying the Fracture Surface

Crack propagation in a weld is highly dependent on the microstructure. It is generally considered that a crack propagates more easily in brittle structures, such as low temperature transformation products and columnar grain boundaries. Therefore, systematic investigation of the microstructures underlying the fracture surface can be helpful in understanding weld metal transverse cold crack initiation and propagation.

The microstructures in cross sections perpendicular to the cracked fracture surfaces in weld metal deposited with 5%CO₂ and 20%CO₂ shielding gases at 20°C preheat were examined by both optical microscopy and SEM. Typical microstructures underlying the ductile fracture and cleavage fracture surfaces are shown in Figures 6.31 and 6.32. There appeared to be little difference in the microstructures associated with the MVC and QC modes of fracture. The ductile fracture surface appeared rough and some sections of dimples can be observed. The cleavage fracture surface appeared relatively smooth.

Optical micrographs of structures underlying the fracture surfaces are shown in Figure 6.33 for the weld metal deposited with 5%CO₂ shielding gas at 20°C preheat and in Figure 6.34 for the weld metal deposited with 20%CO₂ shielding at 20°C preheat.

The corresponding columnar grain features of the cross sections in the weld metals deposited with 5%CO₂ and 20%CO₂ shielding gases are shown in Figure 6.35 and Figure 6.36, respectively. It is obvious that the weld metal crack followed the columnar grain boundaries (region A in Figure 6.35 and A and A’ in Figure 6.36) in the early stages of cracking, and propagated substantially parallel to the columnar grain
Figure 6.37 Appearance of typical non-metallic inclusions at the two different oxygen levels.

a. low oxygen weld metal, 390 ppm (deposited with 5%CO₂ shielding gas); b. high oxygen weld metal, 490 ppm (deposited with 20%CO₂ shielding gas)
boundaries (D-E in Figure 6.35 and C and E in Figure 6.36). However, in some regions, the weld metal cracked across the columnar grains (B and C in Figure 6.35 and B and D in Figure 6.36), producing a step-like crack path. The early crack path in the weld metal deposited with 20%CO$_2$ shielding gas (Figure 6.36) was particularly tortuous, showing frequent sharp steps before establishing a flatter fracture profile. The more chaotic alignment of the columnar grains close to the root appears to be associated with this effect.

6.5 NON-METALLIC INCLUSION ANALYSIS

Non-metallic inclusions can strongly influence the microstructure and toughness of the weld metal, so that either an increase or a decrease in the resistance to hydrogen induced cold cracking. These effects are mainly dependent on the nature and distribution of the inclusions. Therefore, it is very important to analyse the characteristics of non-metallic inclusions in weld metals.

The inclusion characteristics are quite sensitive to the oxygen content and the deoxidizing element concentrations. Inclusion formation depends strongly on the concentrations of dissolved oxygen and deoxidizing elements, and on the weld cooling rate.

6.5.1 Non-Metallic Inclusion Characteristics

Figure 6.37 shows typical images of non-metallic inclusions for the weld metals deposited with the two shielding gases. The non-metallic inclusions were globular in shape and the images showed the trend that the higher oxygen weld metal (deposited with 20%CO$_2$) contained larger inclusions (Figure 6.37b) than low oxygen weld metal (deposited with 5%CO$_2$) (Figure 6.37a).

The important non-metallic inclusion characteristics have been calculated from the basic stereometric relationships given by Equations 5-5 to 5-13 in Chapter 5. The results are summarized in Table 6.11.
<table>
<thead>
<tr>
<th>Preheat Temperature °C</th>
<th>Shielding gas</th>
<th>N</th>
<th>$V_v = \frac{N}{A_A}$, %</th>
<th>$d_1$, μm</th>
<th>$d_2$, μm</th>
<th>$N_s$, 1/μm²</th>
<th>$N_v$, 1/μm³</th>
<th>$\lambda_v$, μm</th>
<th>$S_v$, μm¹</th>
<th>$\overline{A}$, μm²</th>
<th>$P_{&lt;0.3\mu m}$, %</th>
<th>$P_{&lt;0.5\mu m}$, %</th>
<th>$P_{&gt;1\mu m}$, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
<td>5%CO₂</td>
<td>385</td>
<td>0.183</td>
<td>0.412</td>
<td>0.647</td>
<td>0.0083</td>
<td>0.0129</td>
<td>2.362</td>
<td>0.0170</td>
<td>0.220</td>
<td>40.3</td>
<td>69.6</td>
<td>3.1</td>
</tr>
<tr>
<td>20</td>
<td>20%CO₂</td>
<td>651</td>
<td>0.287</td>
<td>0.413</td>
<td>0.649</td>
<td>0.0130</td>
<td>0.0201</td>
<td>2.038</td>
<td>0.0266</td>
<td>0.221</td>
<td>36.4</td>
<td>72.4</td>
<td>2.9</td>
</tr>
<tr>
<td>40</td>
<td>5%CO₂</td>
<td>545</td>
<td>0.239</td>
<td>0.392</td>
<td>0.616</td>
<td>0.0120</td>
<td>0.0195</td>
<td>2.058</td>
<td>0.0232</td>
<td>0.199</td>
<td>39.8</td>
<td>77.8</td>
<td>2.9</td>
</tr>
<tr>
<td>40</td>
<td>20%CO₂</td>
<td>395</td>
<td>0.214</td>
<td>0.440</td>
<td>0.691</td>
<td>0.0086</td>
<td>0.0124</td>
<td>2.394</td>
<td>0.0186</td>
<td>0.249</td>
<td>34.2</td>
<td>67.6</td>
<td>5.4</td>
</tr>
<tr>
<td>60</td>
<td>5%CO₂</td>
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<td>0.239</td>
<td>0.471</td>
<td>0.740</td>
<td>0.0084</td>
<td>0.0113</td>
<td>2.469</td>
<td>0.0194</td>
<td>0.285</td>
<td>28.1</td>
<td>63.7</td>
<td>6.0</td>
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<tr>
<td>60</td>
<td>20%CO₂</td>
<td>460</td>
<td>0.292</td>
<td>0.473</td>
<td>0.743</td>
<td>0.0101</td>
<td>0.0136</td>
<td>2.321</td>
<td>0.0236</td>
<td>0.289</td>
<td>28.9</td>
<td>65.2</td>
<td>6.1</td>
</tr>
<tr>
<td>80</td>
<td>5%CO₂</td>
<td>373</td>
<td>0.180</td>
<td>0.410</td>
<td>0.644</td>
<td>0.0083</td>
<td>0.0129</td>
<td>2.362</td>
<td>0.0168</td>
<td>0.217</td>
<td>39.4</td>
<td>73.5</td>
<td>4.5</td>
</tr>
<tr>
<td>80</td>
<td>20%CO₂</td>
<td>533</td>
<td>0.289</td>
<td>0.444</td>
<td>0.697</td>
<td>0.0114</td>
<td>0.0163</td>
<td>2.185</td>
<td>0.0249</td>
<td>0.254</td>
<td>33.6</td>
<td>68.1</td>
<td>4.9</td>
</tr>
<tr>
<td>100</td>
<td>5%CO₂</td>
<td>673</td>
<td>0.281</td>
<td>0.403</td>
<td>0.633</td>
<td>0.0134</td>
<td>0.0212</td>
<td>2.002</td>
<td>0.0267</td>
<td>0.210</td>
<td>40.0</td>
<td>75.2</td>
<td>3.7</td>
</tr>
<tr>
<td>100</td>
<td>20%CO₂</td>
<td>595</td>
<td>0.379</td>
<td>0.485</td>
<td>0.762</td>
<td>0.0125</td>
<td>0.0164</td>
<td>2.181</td>
<td>0.0299</td>
<td>0.303</td>
<td>32.1</td>
<td>65.2</td>
<td>9.4</td>
</tr>
</tbody>
</table>

$N$: the number of particles  
$V_v$: the volume fraction  
$d_1$: the average two dimensional particle diameter  
$N_s$: the number per unit area of particles  
$\lambda_v$: the mean particle centre to centre volume spacing  
$A_A$: the area fraction of particles  
$d_2$: the average three dimensional particle diameter  
$N_v$: the number per unit volume of particles  
$S_v$: the total particles surface area per unit volume  
$\overline{A}$: the average area of the particles  

Table 6.11 The important characteristics of non metallic inclusions
Figure 6.38 The volume fractions of non-metallic inclusions in the two steel weld metals as a function of preheat temperature
Figure 6.39 The total particle surface area per unit volume in the two steel weld metals as a function of preheat temperature
Figure 6.40 Two dimensional mean particle diameters of the two steel weld metal inclusions as a function of preheat temperature
Figure 6.41 Mean particle areas of the two steel weld metal inclusions as a function of preheat temperature
Figure 6.42 The number per unit area of particles in the two steel weld metals as a function of preheat temperature.
Figure 6.43 The mean particle centre to centre volume spacing in the two steel weld metals as a function of preheat temperature
Figure 6.38 shows the volume fraction ($V_v$) of non-metallic inclusions in the weld metal deposited with 5%CO$_2$ and 20%CO$_2$ shielding gases at different preheat temperatures, and Figure 6.39 shows the total particle surface area per unit volume ($S_v$). It was found that $V_v$ and $S_v$ were generally higher for the higher carbon dioxide concentration in the shielding gas, consistent with an increased oxidation potential. According to the chemical composition of deposited weld metal showed in Table 6.5, the oxidisable elements manganese and silicon were also in lower concentration in the weld metal deposited with 20%CO$_2$ shielding gas.

Figure 6.40 shows the two dimensional mean particle diameter of non-metallic inclusions and Figure 6.41 shows the mean particle area of inclusions in the two weld metals deposited at different preheat temperatures. It is clear that the mean particle size is higher for a higher oxygen content in the weld metal.

The size of non-metallic inclusions may be affected by cooling rate at higher temperatures (above 800°C), with the size increasing with decreasing cooling rate because of the increased time for coarsening to take place. In this work, the preheat temperature was from 40 to 100°C, and the high temperature cooling rate is not expected to be strongly affected by the preheat temperature. This conclusion is consistent with the absence of a correlation between the non-metallic inclusion characteristics and preheat temperature in the present work.

Figure 6.42 shows the number per unit area of particles ($N_a$) of non-metallic inclusions and Figure 6.43 shows the mean particle centre to centre volume spacing ($\lambda_v$) of inclusions in weld metals deposited with the two different shielding gases at different preheat temperatures. There was no obvious effect of carbon dioxide concentration in the shielding gas on these non-metallic inclusion characteristics.
Figure 6.44 Mean particle centre to centre volume spacing $\lambda_v$, $\mu$m as a function of the average microvoid size on the fracture surface of cracked steel weld metals
Figure 6.45 The size distribution histograms of inclusions in weld metal deposited with 5% CO₂ shielding gas at different preheat temperatures
Figure 6.46 The size distribution histograms of inclusions in weld metal deposited with 20%CO₂ shielding gas at different preheat temperatures.
Figure 6.44 illustrates the mean particle centre to centre volume spacing $\lambda_v$ against the average microvoid size on the fracture surface of cracked weld metals. The plot indicates that the average microvoid size was quite consistent with the values of $\lambda_v$. Therefore, it can be considered that most of the microvoids were nucleated at the non-metallic inclusions in the weld metals, and that void size related closely to mean particle spacing.

### 6.5.2 Size Distribution of Non-Metallic Inclusions

The results from the determination of non-metallic inclusions size distribution in the two weld metals deposited at different preheat temperatures are presented graphically in Figures 6.45 and 6.46, based on an analysis of total number of selected particles for each sample. It can be seen that the size distribution is roughly the same for both weld metals at different preheat temperatures. It is evident that the distribution histograms are neither normal nor symmetrical. Therefore, the inclusion size distribution histograms showed a non-uniform distribution, with a skew towards larger particle sizes. The peak of the size distribution typically fell between 0.2 to 0.5 $\mu$m. Combined with the observation on the inclusion locations and sizes, it was found that the non-metallic inclusion distribution was non-uniform in the weld metal. The reason of the non-uniform inclusion distribution is that epitaxial grain growth pushes some particles ahead of the moving interface while the advancing interfaces can passively trap other particles, the relatively large inclusions would position themselves preferentially [Uhlmann et al. 1964; Bolling and Cissé 1971].

The number percentage of inclusion (NPI) has been introduced to describe the size distribution of non-metallic inclusions in more detail. NPI is defined by the following equation [Zhu 1995]:

\[ NPI = \text{Number of Inclusions} / \text{Total Number of Particles} \]
Figure 6.47 The NPI for two dimensional particle diameter less than or equal to 0.3 µm in the two steel weld metals as a function of preheat temperature.
Figure 6.48 The NPI for two dimensional particle diameter larger than 1 µm in the two steel weld metals as a function of preheat temperature
Where $\Sigma N'_i$ is the sum of inclusions with diameters equal to and less or larger than a given size, and $\Sigma N_i$ is the total number of inclusions in the field examined. Data for $P_{<0.3\mu m}$, $P_{\leq 0.5\mu m}$ and $P_{>1\mu m}$ for non-metallic inclusions in the two weld metals deposited at different preheat temperatures are listed in Table 6.11. In most cases, more than 65 percent of the inclusions were less than or equal to 0.5 $\mu m$ in diameter; and less than 6 percent of the inclusions are larger than 1$\mu m$ in diameter.

Figures 6.47 and 6.48 illustrate the NPI for smaller ($d_{\leq 0.3}$ $\mu m$) and larger ($d_{>1}$ $\mu m$) non-metallic inclusions in weld metals deposited at different preheat temperatures. It is clear that for weld metal deposited with 20%$CO_2$ shielding gas which had a higher oxygen potential, the NPI for smaller inclusions was generally lower and the NPI for larger inclusions was generally higher.

The inclusion distributions did not show any correlation with the preheat temperature, probably because the preheats were not high enough to influence the high temperature cooling rate.

6.5.3 Alloying Elements in the Non-Metallic Inclusions

The alloying elements present in the non-metallic inclusions of the weld metals were examined using EDS. Typical EDS spectra for inclusions in the weld metals deposited at different preheat temperatures are attached in Appendix II. The qualitative EDS analysis showed that there was little difference in the chemical compositions of the non-metallic inclusions for the range of welding conditions used. The spectra indicated that all the inclusions were rich in manganese (Mn) and titanium (Ti), and contained a certain amount of silicon (Si) and aluminium (Al). It was also found that the inclusions in the
Figure 6.49 Continuous cooling transformation (CCT) diagram of steel weld metal deposited with 5%\(\text{CO}_2\) shielding gas. Samples austenitised at 1250°C for 13 seconds before cooling at the indicated values of \(\Delta t_85\).
weld metal deposited with 5%CO₂ shielding gas contained a small amount of potassium (K), possibly coming from the flux.

6.6 CONTINUOUS COOLING TRANSFORMATIONS AND MICROSTRUCTURES

A continuous cooling transformation (CCT) diagrams can clearly represent the influence of cooling rate on the γ to α transformation start and finish temperatures, the resultant microstructure and the hardness. The steel weld metal CCT diagrams can play a important role in understanding the development of the weld metal microstructure and the effects of chemical composition, welding parameters and prior austenite grain size on the γ to α transformation of the weld metal.

Following the nomenclature of Bainite Committee of the Iron and Steel Institute of Japan [Araki 1992], the regions of CCT diagram were labelled as follows, α_q: quasi-polygonal ferrite; α_p: granular bainite; α_p⁰: bainitic ferrite; A_f: acicular ferrite; α': martensite and P: pearlite.

6.6.1 Continuous Cooling Transformation (CCT) Diagrams

The CCT diagrams were characterised by multilayer transformation curves.

Figure 6.49 shows the CCT diagram for the weld metal deposited with 5%CO₂ shielding gas. During quenching, the transformation start temperature (Ar₃) decreased gently with increasing cooling rate, and the transformation finish temperature (Ar₁) decreased more sharply with increasing cooling rate (800-500°C) from 7.6 to 29.2°C/sec, then decreased slightly with increasing cooling rate. The bulk of transformation during quenching occurred in the temperature range between 570 to 310°C. In this temperature range, for low carbon, low alloy weld metal, the transformed microstructure usually consists of quasi-polygonal ferrite (coarse), granular bainite,
Figure 6.50 Continuous cooling transformation (CCT) diagram of steel weld metal deposited with 20%CO₂ shielding gas. Samples austenitised at 1250°C for 13 seconds before cooling at the indicated values of $\Delta t_{8/5}$. 
Figure 6.51 Continuous cooling transformation (CCT) diagram of base metal (BIS812EMA steel). Samples austenitised at 1250°C for 13 seconds before cooling at the indicated values of $\Delta t_{85}$.
bainitic ferrite, acicular ferrite and lath martensite (very fine). In general, the microstructural features become finer with increasing cooling rate. When the cooling rate was higher than 126.5°C/sec, the microstructure consisted of mainly acicular ferrite with a small amount of martensite. When the cooling rate was between 126.5 and 12.6°C/sec, the microstructure contained acicular ferrite, bainitic ferrite, granular bainite or quasi-polygonal ferrite, with the amount of acicular ferrite and bainitic ferrite decreasing with decreasing cooling rate. When the cooling rate between 800-500°C was in the range 12.6 and 9.5°C/sec, the microstructural features were bainitic ferrite, granular bainite and quasi-polygonal ferrite. When the cooling rate was lower than 9.5°C/sec, the corresponding microstructure was mainly granular bainite and quasi-polygonal ferrite, with a small amount of pearlite.

Figure 6.50 shows the CCT diagram for the weld metal deposited with 20%CO₂ shielding gas. The transformation start temperature (Ar₃) decreased slightly with increasing cooling rate during quenching. The transformation finish temperature (Ar₁) also decreased with increasing cooling rate, with three distinct stages during quenching. First, the Ar₁ decreased slightly with increasing cooling rate from 7.6 to 42.2°C/sec (800 to 500°C), then the Ar₁ decreased sharply when the cooling rate increased from 42.2 to 75.9°C/sec (800 to 500°C), after that, the Ar₁ decreased very slightly with increasing cooling rate. The majority of transformation during cooling occurred in the temperature range 615 to 360°C. Over the range of cooling rates, microstructures included quasi-polygonal ferrite, granular bainite, bainitic ferrite and acicular ferrite. The amounts of acicular ferrite and bainitic ferrite increased with increasing cooling rate, and the amount of quasi-polygonal ferrite and granular bainite decreased with increasing cooling rate.

Figure 6.51 shows the CCT diagram for the base metal (BIS812EMA). During quenching, the transformation start and finish temperatures (Ar₃ and Ar₁) decreased slightly with increasing cooling rate. Transformation occurred over the temperature
range 585 to 275°C. For cooling rates varied from 7.6 to 378.7°C/sec. (800 to 500°C),
the corresponding microstructure was mainly martensite. When the cooling rate was
lower than 34.5°C/sec, some bainitic ferrite can be observed, with the amount
increasing with decreasing cooling rate. A slow cooling rate of 1.3°C/sec. (800 to
500°C) was also employed so that equilibrium transformation condition would be
approached. The transformation temperature rose significantly (see Table 6.9), and the
corresponding microstructure was mainly bainitic. The results indicate that the base
metal has a very high hardenability.

The $M_s$ temperature is the most important parameter characterizing the martensitic
transformation in steels. For reference, the $M_s$ temperature of the base metal was
calculated using Andrew’s equation [Andrew 1965],

$$M_s (°C) = 539 - 423C - 30.4Mn - 12.1Cr - 17.7Ni - 7.5Mo \quad (6-3)$$

and for the two steel weld metals, the $M_s$ temperatures were calculated using the
equation developed by Olson et al. [1996]:

$$M_s (°C) = 521 - 350C - 13.6Cr - 16.6Ni - 25.1Mn - 30.1Si - 20.4Mo
-40Al - 1.07Cr.Ni + 21.9(Cr+0.73Mo)C \quad (6-4)$$

Table 6.12 shows the cooling rates and corresponding $Ar_3$ and $Ar_1$ temperatures for the
two steel weld metals and the base metal, and the calculated $M_s$ temperatures are also
shown in this table.

6.6.2 Microstructures

The microstructures of the two weld metals and the base metal were examined in
samples reheated in the dilatometer to 1250°C and held for 13 seconds, then
continuously cooled at various rates.
Figure 6.52 Typical microstructures of dilatometer samples obtained in steel weld metal deposited with 5%CO₂ shielding gas at different cooling rates. Samples austenitised at 1250°C for 13 seconds before cooling.
Figure 6.52 (continued), Typical microstructures of dilatometer samples obtained in steel weld metal deposited with 5%CO₂ shielding gas at different cooling rates. Samples austenitised at 1250°C for 13 seconds before cooling.
Table 6.12 CCT diagram data for base metal and weld metals: reference cooling rate, \( \text{Ar}_1 \) and \( \text{Ar}_3 \), and \( \text{Ms} \) (calculated)

<table>
<thead>
<tr>
<th>Reference Cooling Time ((\Delta t_{g/s})) (Second)</th>
<th>Reference Cooling Rate ((800-500^\circ C)) ((^\circ C/sec.))</th>
<th>Base Metal (B1S812EMA)</th>
<th>Weld Metal ((5%\text{CO}_2\text{ shielding gas}))</th>
<th>Weld Metal ((20%\text{CO}_2\text{ shielding gas}))</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>(\text{Ar}_{12}, ^\circ C)</td>
<td>(\text{Ar}_{14}, ^\circ C)</td>
<td>(\text{Ar}_{12}, ^\circ C)</td>
</tr>
<tr>
<td>1</td>
<td>378.7</td>
<td>430.3</td>
<td>277.0</td>
<td>517.3</td>
</tr>
<tr>
<td>3</td>
<td>126.5</td>
<td>432.5</td>
<td>286.5</td>
<td>522.0</td>
</tr>
<tr>
<td>5</td>
<td>75.9</td>
<td>435.1</td>
<td>302.0</td>
<td>534.9</td>
</tr>
<tr>
<td>7</td>
<td>54.2</td>
<td>436.5</td>
<td>299.4</td>
<td>537.9</td>
</tr>
<tr>
<td>9</td>
<td>34.5</td>
<td>438.9</td>
<td>312.9</td>
<td>535.1</td>
</tr>
<tr>
<td>13</td>
<td>29.2</td>
<td>447.6</td>
<td>332.2</td>
<td>540.4</td>
</tr>
<tr>
<td>15</td>
<td>25.3</td>
<td>459.2</td>
<td>338.7</td>
<td>550.9</td>
</tr>
<tr>
<td>20</td>
<td>18.9</td>
<td>458.6</td>
<td>339.7</td>
<td>565.5</td>
</tr>
<tr>
<td>25</td>
<td>15.1</td>
<td>467.4</td>
<td>356.4</td>
<td>560.5</td>
</tr>
<tr>
<td>30</td>
<td>12.6</td>
<td>487.2</td>
<td>361.6</td>
<td>565.1</td>
</tr>
<tr>
<td>40</td>
<td>9.5</td>
<td>507.2</td>
<td>374.2</td>
<td>566.8</td>
</tr>
<tr>
<td>50</td>
<td>7.6</td>
<td>581.3</td>
<td>461.1</td>
<td></td>
</tr>
<tr>
<td>300</td>
<td>1.3</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

\(\text{Ms} = 424.34^\circ C\) \(\text{Ms} = 415.19^\circ C\) \(\text{Ms} = 425.09^\circ C\)

Figure 6.52 shows the microstructures of dilatometer samples of weld metal deposited with 5\%\text{CO}_2\text{ shielding gas}. It is clear that the microstructure was mainly acicular ferrite and some martensite can be observed for a cooling rate \((800-500^\circ C)\) of 378\(^\circ C\)/sec. For a cooling rate of 126.5\(^\circ C\)/sec, acicular ferrite, bainitic ferrite, granular bainite and quasi-polygonal ferrite can be observed. The amount of acicular ferrite decreased with decreasing cooling rate, and acicular ferrite was difficult to detect when the cooling rate was 12.6\(^\circ C\)/sec. For a cooling rate of 9.5\(^\circ C\)/sec \((800-500^\circ C)\), the microstructure mainly consisted of granular bainite and quasi-polygonal ferrite. A small amount of pearlite can be observed for a cooling rate of 7.6\(^\circ C\)/sec.
Figure 6.53 Typical microstructures of dilatometer samples obtained in steel weld metal deposited with 20%CO₂ shielding gas at different cooling rates. Samples austenitised at 1250°C for 13 seconds before cooling.
Figure 6.53 (continued), Typical microstructures of dilatometer samples obtained in steel weld metal deposited with 20% CO₂ shielding gas at different cooling rates. Samples austenitised at 1250°C for 13 seconds before cooling.
Figure 6.54 Typical microstructures of dilatometer samples obtained in BIS812EMA steel at different cooling rates. Samples austenitised at 1250°C for 13 seconds before cooling.
800-500°C: 18.9°C/s, ($\Delta t_{0.5} = 20s$)

800-500°C: 15.1°C/s, ($\Delta t_{0.5} = 25s$)

800-500°C: 12.6°C/s, ($\Delta t_{0.5} = 30s$)

800-500°C: 9.5°C/s, ($\Delta t_{0.5} = 40s$)

800-500°C: 7.6°C/s, ($\Delta t_{0.5} = 50s$)

800-500°C: 1.3°C/s, ($\Delta t_{0.5} = 300s$)

Figure 6.54 (continued), Typical microstructures of dilatometer samples obtained in BIS812EMA steel at different cooling rates. Samples austenitised at 1250°C for 13 seconds before cooling.
Figure 6.55 Average hardness values and standard deviations for dilatometer samples obtained in steel weld metal deposited with 5% CO₂ shielding gas at different cooling rates. Samples austenitised at 1250°C for 13 seconds before cooling.
Figure 6.56 Average hardness values and standard deviations for dilatometer samples obtained in steel weld metal deposited with 20\%CO\textsubscript{2} shielding gas at different cooling rates. Samples austenitised at 1250°C for 13 seconds before cooling.
Figure 6.57 Average hardness values and standard deviations for dilatometer samples obtained in BIS812EMA steel at different cooling rates. Samples austenitised at 1250°C for 13 seconds before cooling.
Figure 6.53 shows the microstructures obtained from dilatometer samples of the weld metal deposited with 20%CO₂ shielding gas. It is obvious that quasi-polygonal ferrite, granular bainite, bainitic ferrite and acicular ferrite are present in samples quenched at varying cooling rates from 7.6 to 126.5°C/sec (800 to 500°C). However, the amount of acicular ferrite decreased with decreasing cooling rate.

Figure 6.54 shows the microstructures of the base metal obtained from the application of continuous cooling transformation at varying cooling rates. When the cooling rate was higher than 34.5°C/sec (800-500°C), the microstructure was fully martensitic, and for a cooling rate lower than 34.5°C/sec, the microstructure consisted of martensite and bainitic ferrite, and the amount of bainitic ferrite increased with decreasing cooling rate. At the slow cooling rate of 1.3°C/sec (800-500°C), the microstructure mainly consisted of granular bainite, bainitic ferrite and quasi-polygonal ferrite.

6.6.3 Hardness

Table 6.13 lists the average hardness values and their standard deviations for dilatometer samples of the two weld metals and the base metal for different cooling times. The data are plotted in Figure 6.55 for weld metal deposited with 5%CO₂ shielding gas; in Figure 6.56 for the weld metal deposited with 20%CO₂ shielding gas and in Figure 6.57 for the base metal. These figures show the expected trend of decreasing hardness with decreasing cooling rate.

6.6.4 Prior Austenite Grain Size of Dilatometer Samples

The prior austenite grain size can strongly affect the type and size of the constituents formed on γ to α phase transformation, thus influencing the mechanical properties of the weld metal. It is therefore worthwhile to study the prior austenite grain size for the different weld simulated samples and the BIS812EMA steel sample.
Table 6.13 Average hardness values and standard deviations for dilatometer samples

<table>
<thead>
<tr>
<th>Reference Cooling Rate, °C/sec. (800-500°C)</th>
<th>Reference Cooling Time Δₜ₈/₅₅ Sec.</th>
<th>Base Metal (BIS812EMA)</th>
<th>Weld Metal (5%CO₂ shielding gas)</th>
<th>Weld Metal (20%CO₂ shielding gas)</th>
</tr>
</thead>
<tbody>
<tr>
<td>378.7</td>
<td>1</td>
<td>431.9±9.8</td>
<td>342.2±8.5</td>
<td></td>
</tr>
<tr>
<td>126.5</td>
<td>3</td>
<td>431.6±15.6</td>
<td>342.2±8.5</td>
<td>256.7±8.1</td>
</tr>
<tr>
<td>75.9</td>
<td>5</td>
<td>415.7±12.0</td>
<td>317.8±8.0</td>
<td>246.3±5.8</td>
</tr>
<tr>
<td>54.2</td>
<td>7</td>
<td>429.1±7.0</td>
<td>283.5±11.3</td>
<td>231.4±7.5</td>
</tr>
<tr>
<td>42.2</td>
<td>9</td>
<td>406.1±15.2</td>
<td>296.0±14.4</td>
<td></td>
</tr>
<tr>
<td>34.5</td>
<td>11</td>
<td>402.1±12.6</td>
<td>295.5±8.0</td>
<td>243.8±8.6</td>
</tr>
<tr>
<td>29.2</td>
<td>13</td>
<td>409.3±12.8</td>
<td>263.6±6.5</td>
<td>250.3±7.5</td>
</tr>
<tr>
<td>25.3</td>
<td>15</td>
<td>400.9±14.0</td>
<td>277.0±6.9</td>
<td></td>
</tr>
<tr>
<td>18.9</td>
<td>20</td>
<td>391.6±10.4</td>
<td>263.3±9.6</td>
<td>220.7±8.2</td>
</tr>
<tr>
<td>15.1</td>
<td>25</td>
<td>369.1±11.3</td>
<td>251.5±8.1</td>
<td>223.8±6.3</td>
</tr>
<tr>
<td>12.6</td>
<td>30</td>
<td>343.8±18.2</td>
<td>242.7±5.4</td>
<td>225.1±6.8</td>
</tr>
<tr>
<td>9.5</td>
<td>40</td>
<td>262.6±19.9</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

The average prior austenite grain sizes in the weld and base metal samples reheated in the dilatometer (1250°C, 13 seconds) were:

1. 92.8 ± 34.3 µm, weld metal deposited with 5%CO₂ shielding gas,
2. 85.5.4 ± 21.2 µm, weld metal deposited with 20%CO₂ shielding gas, and
3. 111.3 ± 40.2 µm, BIS812EMA quenched and tempered steel

It is clear that the average prior austenite grain size of the reheated weld metal deposited with 5%CO₂ shielding gas is slightly larger than the weld metal deposited with 20%CO₂ shielding gas. The grain sizes of the two weld metals are smaller than the base metal.
Figure 6.58 Average hardness values and standard deviations for the two steel weld metals deposited at 20°C preheat and tempered at 550°C for various period of time.
because the weld metal contains non-metallic inclusions which can pin the prior austenite grain boundaries and impede grain growth. The prior austenite grain sizes are inhomogeneous; therefore, there is a wide variation in grain size in all samples. In addition, the large average prior austenite grain size in the base metal contributes to its hardenability, resulting in a high hardness.

6.7 METALLOGRAPHY OF TEMPERED WELD METALS

Tempering is a process which involves precipitation of various carbides of iron and alloying elements from a supersaturated solid solution [Kurdjumow 1960; Bain and Paxton 1966]. Fine alloy carbides can produce a hardness increase at a higher tempering temperature. This hardness increase is referred to as secondary hardening. The alloy weld metals deposited in this work have the potential to show secondary hardening, since the rapid cooling rate, after welding is likely to produce a super saturated acicular ferritic or bainitic structure.

6.7.1 Hardness

Average hardness values and their standard deviations for the two weld metals deposited at 20°C preheat and tempered at 550°C for various period of time are listed in Table 6.14. The average hardness values and standard deviations for the two tempered weld metals as a function of tempering time are plotted in Figure 6.58. It is obvious that the average hardness value dropped sharply when both weld metals were tempered at 550°C for 0.25 hour. When the weld metals were tempered for 0.5 hour, the average hardness values increased slightly, implying that alloy carbides precipitated in a finely dispersed form resulting in dispersion strengthening (secondary hardening). After tempering from 1 to 72 hours, the average hardness values fell, indicating particle coarsening and dislocation recovery. It is found that the hardness of weld metal deposited with 5%CO₂ shielding gas decreased more than for 20%CO₂ shielding gas when the weld metals were tempered for 2 and 72 hours.
Figure 6.59 Typical microstructures of weld metal (deposited at 20°C preheat with 5%\text{CO}_2 shielding gas), tempered at 550°C for various period of times
Figure 6.60 Typical microstructures of weld metal (deposited at 20°C preheat with 20%CO₂ shielding gas), tempered at 550°C for various period of times.
Figure 6.61 Typical TEM micrographs showing the presence of bainitic structure in the two steel weld metals deposited at 80°C preheat.
Figure 6.62 Typical TEM micrographs showing dislocations in the two steel weld metals deposited at 80°C preheat.
Table 6.14 Average hardness values and standard deviations of tempered weld metals

<table>
<thead>
<tr>
<th>Tempering Time, Weld Metal (5%CO₂)</th>
<th>Weld Metal (20%CO₂)</th>
</tr>
</thead>
<tbody>
<tr>
<td>untempered</td>
<td>330.3±18.1</td>
</tr>
<tr>
<td>0.25</td>
<td>287.1±13.3</td>
</tr>
<tr>
<td>0.5</td>
<td>294.6±12.5</td>
</tr>
<tr>
<td>1</td>
<td>289.3±15.2</td>
</tr>
<tr>
<td>2</td>
<td>289.4±10.7</td>
</tr>
<tr>
<td>72</td>
<td>280.0±10.1</td>
</tr>
</tbody>
</table>

6.7.2 Microstructure

Microstructures of tempered weld metals were examined by both optical and transmission electron microscopy (TEM). Figures 6.59 and 6.60 show the microstructures of weld metals deposited with 5%CO₂ and 20%CO₂ shielding gases tempered at 550°C for various period of times. The microstructures of both tempered weld metals appeared to be similar to the original deposited weld metals, because microstructural changes only occur on a fine scale which can not be resolved by optical microscopy.

In an attempt to determine if carbide precipitated during weld cooling, thin foils of the two weld metals deposited at 80°C preheat were examined by TEM. Figures 6.61 and 6.62 show typical TEM micrographs of the two steel weld metals. The presence of bainitic structure is shown in Figure 6.61, and it is clear in Figure 6.62 that the dislocation density is very high. However, carbide precipitates were not observed.

According to the plots of average hardness values of the tempered weld metals (see Figure 6.58), secondary hardening occurred when the weld metals were tempered for 0.5 hour. Therefore, samples of the two weld metals tempered for 0.5 hour were examined by TEM. Typical TEM micrographs of the two tempered steel weld metals are
Figure 6.63 Typical TEM micrographs showing alloy carbides in steel weld metal with 5% CO₂ shielding gas (deposited at 20°C preheat), tempered at 550°C for 0.5 hour.
Figure 6.64 Typical TEM micrographs showing alloy carbides in steel weld metal with 20% CO₂ shielding gas (deposited at 20°C preheat), tempered at 550°C for 0.5 hour.
Figure 6.65 Diffraction pattern from carbide in the steel weld metal (deposited at 20°C preheat), tempered at 550°C for thirty minutes
shown in Figure 6.63 for 5%CO₂ shielding gas and Figure 6.64 for 20%CO₂ shielding gas. It was found that alloy carbides precipitated in the bainitic matrix. The carbide particles nucleated preferentially at bainitic lath boundaries (Figure 6.64a) or at lath packet boundaries (Figure 6.63a) and to a lesser extent with the grains (Figures 6.63b and 6.64b). The carbide particles in the two tempered weld metals were typically present in a low volume fraction. The carbides were often plate-shaped but also showed irregular shapes. Most of the carbide precipitates were fine and generally less than 100 nm in size. Figure 6.65 shows a typical diffraction pattern from carbide in tempered weld metal. The results of electron diffraction analysis indicated that most of carbide precipitates in the two tempered weld metals were MC type alloy carbides, where M could consist of titanium, niobium and/or vanadium. Fe₃C carbides were also identified in the tempered weld metals.
CHAPTER 7

DISCUSSION
Based on the G-BOP test results in Chapter 6 for weld metals deposited with various preheat temperatures and two different kinds of shielding gas mixtures, the discussion presented in this Chapter is focused on the following aspects:

1. the effects of preheat treatment and shielding gas mixture on the characteristics of the weld metals;
2. the effects of preheat treatment and shielding gas mixture on the behaviour of weld metal transverse cold cracking;
3. the fracture morphology of weld metal transverse cold cracking;
4. the continuous cooling transformation characteristics of weld metals and base metal; and
5. the tempered weld metals.

7.1 EFFECT OF PREHEAT ON WELDING THERMAL CYCLES

It is well known that the preheat treatment will affect the welding thermal cycle, particularly the cooling rate. According to the CCT diagram data in Table 6.9, the phase transformation temperature ranges during cooling occurred between 567°C (Ar$_3$) and 312°C (Ar$_1$) for weld metal deposited with 5%CO$_2$ shielding gas, and between 613°C (Ar$_3$) and 364°C (Ar$_1$) for weld metal deposited with 20%CO$_2$ shielding gas. Therefore, the conventional cooling time $\Delta t_{8/5}$ does not accurately reflect the effect of cooling rate on phase transformation for the weld metals investigated. $\Delta t_{8/3}$, which is more appropriate, was also determined as a function of preheat temperature in this work (see Figure 6.1). The trend clearly shows that cooling rate decreased with increasing preheat temperature. Obviously, the preheat can be used to prevent the formation of a crack susceptible microstructure, because a higher temperature product, such as acicular ferrite, is relatively ductile and crack resistant, and a lower temperature product, such as martensite, is more crack sensitive. A prolonged $\Delta t_{8/3}$ also contributes to a greater
Figure 7.1 Hydrogen diffusion coefficient as a function of temperature in bcc iron
amount of hydrogen effusion from the weld metal, resulting in a lower hydrogen content.

Figure 7.1 shows the hydrogen diffusion coefficient \( (D_h) \), which describes the diffusivity of hydrogen as a function of temperature in bcc iron. \( D_h \) is calculated by an Arrhenius equation [Askeland 1996],

\[
D_h = D_0 \exp\left(-\frac{Q}{RT}\right)
\]  

(7-1)

where \( Q \) is the activation energy (J.mol\(^{-1}\)), \( R \) is the gas constant (8.314 J.mol\(^{-1}\).K\(^{-1}\)), \( T \) is the absolute temperature (K), and \( D_0 \) is a constant for a given diffusion system.

Using appropriate values for \( Q \) and \( D_0 \), it can be shown that the hydrogen diffusion coefficient in ferrite at 150°C is equal to the carbon diffusion coefficient in austenite at about 1460°C. This demonstrates the high mobility of hydrogen atoms in iron, even at low temperatures. However, equation (7-1) does not take account of hydrogen trapping at a temperature below about 200°C which results in lower apparent diffusion coefficients [Yurioka and Suzuki 1990; Steigerwald et al. 1959]. Nevertheless, hydrogen diffusion is still relatively fast at low temperatures, and residual diffusible hydrogen can cause weld metal embrittlement problems after the temperature of the weldment cools below 100°C [Bibby et al. 1998]. Many researchers have recognized that hydrogen induced cold cracking is a function of the cooling rate at low temperatures and the cooling time between 300°C to 100°C (\( \Delta t_{3/1} \)) has been frequently used for investigating hydrogen induced cold cracking. \( \Delta t_{3/1} \) is considered more significant in determining the extent of hydrogen effusion than \( \Delta t_{8/15} \) [Chakravarit and Bala 1989].

Due to a data collecting limitation, \( \Delta t_{3/1.5} \) was measured in this work (see Table 6.1 and Figure 6.2). It is obvious that the cooling rate between 300°C to 150°C decreased with
increasing preheat temperature, especially when the preheat was increased from 80 to 100°C. Consequently, the cooling rate in the lower temperature range is retarded with preheat, allowing hydrogen atoms sufficient time to escape from the weldment by diffusion. The resistance to weld metal hydrogen induced cold cracking would therefore be expected to be raised. Comparing Figures 6.1 and 6.2, it is evident that the effect of preheat on $\Delta t_{3/1.5}$ is much greater than on $\Delta t_{8/5}$ and $\Delta t_{8/3}$.

7.2 EFFECT OF PREHEAT ON LONGITUDINAL STRESS AND CRACKING

7.2.1 Longitudinal Stress Development

The plots of longitudinal stress across the gap during weld metal solidification clearly indicate the time and temperature of initiation of weld metal transverse cracking (see Figure 6.3: longitudinal stress against time; and Figure 6.4: longitudinal stress against temperature). All cracks occurred at least five minutes after welding and initiated below 200°C. It can be concluded that the weld metal transverse cracking was definitely cold cracking, since the occurrence of cracking was delayed until the temperature was below 200°C [Yurioka and Suzuki 1990; Sawhill et al. 1974].

The weld metal contracted during cooling after solidification causing an increase in longitudinal stress. However, an expansion occurred during transformation of the weld metal. The expansion caused a transient lowering of the stress which is evident in the stress troughs between about 700 and 300°C in Figure 6.4. After transformation, the stress increased with decreasing temperature in an approximately linear manner. When the stress reached a value of about 500-600 MPa, the longitudinal stress in the weld metal across the gap appeared to relax with an approximately linear decrease with time after peak stress (see Figure 6.3). This relaxation is probably due to plastic yielding. The sharp vertical decreases in stress at critical times (Figure 6.3) correspond to
Figure 7.2 The duration of longitudinal stress relaxation as a function of preheat temperature in G-BOP tests.
initiation and intermittent propagation of weld metal cracks. The decreasing stress level was related to the percentage of cracking. Figure 6.3 shows that the longitudinal stress sharply decreased to a level near the pre-load stress when 100 percent cold cracking occurred. When partial cracking occurred, the stress only decreased to a certain level and then maintained that level with increasing cooling time. The final percentages of cold cracking are given in Table 6.9.

7.2.2 Effect of Preheat

Preheat has the following effects on longitudinal stress and cracking:

(a) After the stress reached the peak value, the duration of stress relaxation increased with increasing preheat temperature, as shown in Figure 7.2. Therefore, after welding, the occurrence of weld metal cold cracking is delayed by increasing the preheat temperature. This effect arises because the slower cooling rate allows more time at higher temperature for relaxation of the restraint stress to occur by plastic flow after the yield point is reached. At the same time hydrogen has more time to diffuse out of the weld metal in the low temperature range above the critical temperature. Hence the susceptibility to weld metal hydrogen induced cold cracking decreases with increasing preheat, because of reduction of diffusible hydrogen content and relaxation of the restraint stress.

(b) The curves of longitudinal stress against temperature in Figure 6.4 indicate that there is a trend towards increased longitudinal stress peak sharpness with decreasing preheat. This behaviour is probably due to the faster cooling at lower preheat leading to a rapid rise in stress at lower temperature.

(c) The curves of longitudinal stress against time in Figure 6.3 clearly show that the cold crack percentage decreases with increasing preheat, because of the reduced susceptibility of the weld metal microstructure and a lower diffusible hydrogen content resulting from more complete hydrogen effusion from the weld metal.
Figure 7.3 The areas of weld metal deposited with the two different shielding gas mixtures as a function of preheat temperature.
7.3 EFFECT OF SHIELDING GAS ON WELD BEAD GEOMETRY

The shielding gas in the arc welding processes primarily provides a suitable medium for the stabilization of the arc and prevents weld pool contamination from the atmosphere. Secondly, the shielding gas has an effect on the weld bead geometry. Small variations in the composition of the gas mixture can have a significant influence on arc/metal transfer characteristics and weld bead shape [Lucas 1997]. Most studies of the effect of shielding gas on weld bead geometry are concerned with the influence of oxygen additions on a shielding gas such as argon. Stenbacka and Persson [1989] indicated that when the CO₂ content in the shielding gas is equal to or less than 25 percent, the oxygen potential (OP) of the shielding gas can be estimated using the following formula,

\[ \text{OP} = O_2 + \frac{1}{2}CO_2 \]  

Therefore, the shielding gas with 20%CO₂ has a higher oxygen potential than that with 5%CO₂. The greater oxidizing potential of 20%CO₂ shielding gas mixture is evident in the lower recoveries of the oxidizable elements, such as manganese and silicon in the weld metal (Tables 6.4 and 6.5).

Figure 7.3 compares the weld metal areas of deposits using the two shielding gases. It was found that most of the welds deposited with 20%CO₂ shielding gas had a larger cross sectional area than those deposited with 5%CO₂ shielding gas. The average weld metal areas (including all preheat conditions) were 35.16 ± 3.37 mm² for 5%CO₂ shielding gas and 38.67 ± 2.28 mm² for 20%CO₂ shielding gas. The average widths of the weld bead were 13.75 ± 0.97 mm for 5%CO₂ shielding gas and 14.25 ± 0.57 mm for 20%CO₂. The average depths of the weld bead were 4.05 ± 0.13 mm for 5%CO₂ and 4.28 ± 0.31 mm for 20%CO₂. These results are consistent with experimental observations reported by Walsh and Savage [1986] and show that the oxygen potential of the shielding gas has a potent effect on the weld pool geometry,
increasing the pool depth, width, and volume when deposited at the same nominal heat input.

This influence of oxygen on the weld pool is due to the following factors.

1. The addition of oxygen to a shielding gas can reduce the surface tension of the metal [Jönsson et al. 1995], thereby improving the wetting conditions of the workpiece [Lucas 1992]; and

2. The addition of carbon dioxide or oxygen to a shielding gas can cause an increase in the anode drop and increase the arc energy transfer to the weld pool. Therefore, the weld metal area and the total fused area can be increased at an equivalent current [Norrish 1992; Walsh and Savage 1986].

The dilution of the weld bead deposited with 20% CO₂ shielding gas was larger than or similar to that deposited with 5% CO₂ shielding gas at the same preheat temperature, as shown in Figure 6.8. The average dilution of the weld bead was 20.70% ± 2.03% for 5% CO₂ shielding gas and 22.23% ± 2.71% for 20% CO₂ shielding gas. This result is consistent with the observation that the addition of CO₂ to argon can improve base metal fusion and increase weld bead penetration [Lucas 1997]. Therefore, increasing CO₂ concentration in the shielding gas is expected to increase weld bead dilution and therefore, the extent of alloy pick up from the base metal.

7.4 WELD METAL MICROSTRUCTURAL ANALYSIS

7.4.1 Effect of Preheat Temperature on Microstructure

The microstructure of the weld metal is highly dependent on the cooling rate and the chemical composition. The preheat treatment will influence the cooling rate of the weld metal. However, phase transformation during cooling occurred in the range of about 600°C to 300°C (refer to CCT diagram data in Table 6.9); and Table 6.1 shows that there were only small changes in the cooling time Δtₜ/₃ with varying preheat
temperature. The small cooling time differences are not likely to affect significantly the transformation products formed by decomposition of austenite on cooling. Therefore, the microstructure at the level of optical microscopy remained almost constant with changing preheat temperature, as shown in Figures 6.9 and 6.10, except for general coarsening of the ferrite plate structure with increasing preheat temperature. It can be concluded that preheating from 20 to 100°C did not have significant influence on the microstructures of the weld metals.

7.4.2 Effect of Shielding Gas on Composition and Microstructure of Weld Metal

The weld metal composition is controlled by the consumable, the base metal and chemical reactions occurring in the weld pool at elevated temperatures. It is thereby affected by the choice of flux cored wire and shielding gas mixture, the base metal chemical composition and the operational conditions. The shielding gases used in this investigation were argon rich mixtures containing 5%CO₂ and 20%CO₂. Since CO₂ is an oxidizing agent at high temperature, it will react with oxidisable elements, particularly carbon, boron, manganese and silicon, in the molten weld pool. The chemical compositions and carbon equivalents of the base metal, the all weld metals and deposited weld metals are shown in Table 7.1. The elemental compositions of the diluted weld metals (X_{DW}%), shown in Table 7.1d, were calculated by using the Rule of Mixtures:

\[
X_{DW}\% = X_{BM} \cdot \frac{\text{%dilution}}{100} + X_{WM} \cdot \left( \frac{100 - \text{%dilution}}{100} \right)
\]  

(7-3)

were \(X_{BM}\) is the composition of element X in the base metal (Table 7.1a) and \(X_{WM}\) is the composition of the element in the all weld metal sample (Table 7.1b). The compositions in Table 7.1d represent the mean values based on the calculated compositions of all of the deposited weld metals, using the individual percent dilutions of each sample.
Table 7.1 Comparison of chemical compositions and carbon equivalents of BIS812EMA steel and the two weld metals

(a) Compositions (wt%) and carbon equivalents of BIS812EMA steel

|   | C  | Si | Mn | P  | S  | Cr | Ni | Mo | V  | Ti | Cu | Al | Nb | B  | N  | Ca | O  | Fe | CE_N | P_{cm} |
|---|----|----|----|----|----|----|----|----|----|----|----|----|----|----|----|----|----|-----|-------|
|   | 0.13 | 0.24 | 0.93 | 0.011 | 0.002 | 0.48 | 1.28 | 0.39 | 0.02 | 0.01 | 0.21 | 0.07 | 0.01 | 0.0066 | - | ≤3 ppm | 0.009 | bal. | 0.495 | 0.301 |

(b) Compositions (wt%) and carbon equivalents of all weld metal samples

| Shielding Gas | C  | Si | Mn | P  | S  | Cr | Ni | Mo | V  | Ti | Cu | Al | Nb | B  | N  | Co | O  | P_{cmo} |
|---------------|----|----|----|----|----|----|----|----|----|----|----|----|----|----|----|----|-------|
| 5%CO₂        | 0.02 | 0.28 | 1.85 | 0.007 | 0.005 | 0.03 | 2.59 | 0.01 | 0.02 | 0.03 | 0.03 | 0.007 | 0.01 | <0.001 | 0.0089 | <0.002 | 0.039 | 0.146 |
| 20%CO₂       | 0.02 | 0.18 | 1.61 | 0.007 | 0.005 | 0.03 | 2.54 | 0.01 | 0.02 | 0.02 | 0.03 | 0.006 | 0.01 | <0.001 | 0.0040 | <0.002 | 0.046 | 0.125 |

(c) Compositions (wt%) and carbon equivalents of diluted weld metal samples

| Shielding Gas | C  | Si | Mn | P  | S  | Cr | Ni | Mo | V  | Ti | Cu | Al | Nb | B  | N  | Co | O  | P_{cmo} |
|---------------|----|----|----|----|----|----|----|----|----|----|----|----|----|----|----|----|-------|
| 5%CO₂        | 0.070 | 0.33 | 1.87 | 0.010 | 0.005 | 0.12 | 2.35 | 0.072 | 0.021 | 0.033 | 0.058 | 0.008 | 0.018 | 0.0018 | 0.0083 | 0.036 | 0.212 |
| 20%CO₂       | 0.055 | 0.24 | 1.64 | 0.009 | 0.004 | 0.11 | 2.25 | 0.086 | 0.019 | 0.023 | 0.049 | 0.005 | 0.014 | 0.0009 | 0.0042 | 0.050 | 0.165 |

(d) Calculated compositions* (wt%) and carbon equivalents of diluted weld metal samples (mean values)

| Shielding Gas | C  | Si | Mn | P  | S  | Cr | Ni | Mo | V  | Ti | Cu | Al | Nb | B  | N  | Co | O  | P_{cmo} |
|---------------|----|----|----|----|----|----|----|----|----|----|----|----|----|----|----|----|-------|
| 5%CO₂        | 0.04 | 0.27 | 1.66 | 0.008 | 0.004 | 0.12 | 2.32 | 0.09 | 0.02 | 0.03 | 0.07 | 0.02 | 0.01 | <0.0022 | 0.0071 | <0.002 | 0.033 | 0.199 |
| 20%CO₂       | 0.04 | 0.19 | 1.46 | 0.008 | 0.004 | 0.13 | 2.26 | 0.09 | 0.02 | 0.02 | 0.07 | 0.02 | 0.01 | <0.0022 | 0.0031 | <0.002 | 0.038 | 0.186 |

* Using Rule of Mixtures
Comparing Table 7.1b with Table 7.1c, it is evident that the contents of some alloying elements: carbon, chromium, molybdenum copper, niobium and boron, were significantly higher in the deposited weld metals, indicating that these elements were picked up from the base metal during weld processing due to the dilution. The two most significant changes from the all weld composition are the molybdenum and boron contents, from low levels to those which are significant in terms of alloy hardenability.

Comparing the measured and calculated compositions of the deposited weld metals (Tables 7.1c and 7.1d), the actual measured values of carbon, manganese and silicon are significantly higher than the calculated values. These differences underline the danger of using estimated weld metal compositions based on dilution and a simple rule of mixtures. Such a method is frequently used (for example, see Yurioka et al. [1983]) particularly to estimate weld metal carbon contents. Differences in conditions of deposition of all weld metal samples and the actual diluted weld deposits can evidently lead to different recoveries of carbon, manganese, silicon and other elements.

Since the dilutions of most samples deposited with 5%CO₂ and 20%CO₂ shielding gases were similar, the differences in composition of weld metals deposited using these two shielding gases must be due to differences in shielding gas oxidation potential. Significant differences occur in the levels of carbon, silicon, manganese, boron, nitrogen and oxygen (Table 7.1c). During deposition of the bead-on-plate welds, the contents of the oxidisable elements are lower for the shielding gas with higher oxygen potential (20%CO₂). Boron and carbon, in particular, are readily oxidised and removed as oxides from the weld pool during the weld processing. Consequently, as shown in Table 7.1c, the contents of carbon and boron are markedly lower in weld metal deposited with 20%CO₂ shielding gas. Some of the oxides are trapped in the weld metal leading to a higher oxygen content for the 20%CO₂ shielding gas.

The effect of oxygen (oxides) in reducing hardenability is taken into account by the \( P_{\text{emo}} \) carbon equivalent [Onsøien et al. 1996]. Table 7.1c clearly shows the difference in
chemical compositions and the carbon equivalents of the two deposited weld metals. The weld metal deposited with 5%CO₂ shielding gas contains higher carbon, boron, manganese, silicon, titanium and nitrogen, and lower oxygen. The value of the $P_{cmo}$ parameter is $0.212$ for the weld metal deposited with 5%CO₂ shielding gas and $0.165$ for 20%CO₂ shielding gas.

Soon after solidification the carbon will mostly be in solution in austenite, although alloy carbides, such as niobium, vanadium or titanium carbides, can precipitate from austenite on cooling. The fast cooling rate following weld deposition ($\Delta t_{85} = 8s$) limits the amount of precipitation. On transformation to ferrite (acicular or bainitic ferrite), carbon will be rejected into remnant austenite pools which can transform to alloy carbides or MA islands. The degree of supersaturation of the ferrite with carbon will depend on the temperature of transformation of austenite to ferritic product. If martensite forms, the carbon will essentially be retained in supersaturated solid solution. The boron in the weld metal can be soluble and/or present as boron nitrides. However, the strong nitride formers, titanium and aluminum, can be expected to fix much of the nitrogen in the weld metal. Therefore, it can be assumed that most of the boron in the weld metals is in solution. The manganese and silicon will either be in solid solution or compounded in the non-metallic inclusions entrapped within the weld metal. The effect of inclusions is discussed in the following sub-section.

It is well known that small changes in carbon and boron contents may have a large effect on the austenite to ferrite decomposition kinetics [Pickering 1978]. Carbon, manganese and nitrogen are strong austenite stabilizers and they tend to inhibit austenite to ferrite transformation during cooling, resulting in formation of ferrite products at lower temperatures [Easterling 1992]. Boron retained in solid solution in weld metals may segregate to the austenite grain boundaries, reducing the contribution of the boundary’s interfacial energy to ferrite nucleation [Hsien et al. 1996; Bang and Ahn 1997], with the result that the formation of grain boundary ferrite is inhibited and acicular ferrite formation is promoted and the product is refined. Therefore, higher
contents of carbon, boron, manganese and nitrogen in weld metals will enhance the hardenability.

Prior austenite grain size also influences hardenability. Although Section 6.6.4 indicates that the average prior austenite grain size of the weld metal deposited with 5\%CO\textsubscript{2} shielding gas is larger than the weld metal deposited with 20\%CO\textsubscript{2}, this result is for reheated weld metal in dilatometer samples. The higher volume fraction of non-metallic inclusions for the weld metal deposited with 20\%CO\textsubscript{2} shielding gas probably accounts for this grain size difference. The prior austenite grain sizes in the deposited bead-on-plate welds were not measured but the similarity in the average widths of the columnar grain in the weld metals deposited with 5\%CO\textsubscript{2} and 20\%CO\textsubscript{2} shielding gases (Section 6.3.5) make it unlikely that there is a large enough difference in austenite grain size to influence the hardenability.

Figures 6.11 and 6.12 show that the microstructures of the weld metals deposited with the two different shielding gases are similar and mainly consist of acicular ferrite and bainitic ferrite. However, in comparing with the weld metal deposited with 5\%CO\textsubscript{2} shielding gas, the higher oxygen concentration for 20\%CO\textsubscript{2} shielding gas caused loss of carbon, boron, manganese and silicon by oxidation reactions. The lower carbon and alloying element contents lowered the hardenability of the weld metal deposited with 20\%CO\textsubscript{2} shielding gas, resulting in phase transformation at higher temperatures during cooling and a slightly coarser microstructure. This microstructure is likely to have a reduced susceptibility to hydrogen induced cold cracking because of its lower hardness. Further, the higher transformation temperature means that hydrogen effusion is likely to be more efficient because of the higher diffusibility of hydrogen in ferrite compared to austenite.

7.4.3 Columnar Grain Size of the Weld Metals

It is well established that initial weld metal solidification during cooling takes place epitaxially, where the partly melted base metal at the fusion boundary acts as seed
crystals for the weld metal columnar grains [Grong and Matlock 1986]. Therefore, the columnar grain size in the weld metal is closely related to the HAZ prior austenite grain size [Davies and Garland 1975; Savage et al. 1976]. The columnar grain size can also be influenced by other factors, such as the alloy element and oxygen contents in the weld pool and thus the non-metallic inclusion population and size distribution. The variations in the columnar grain size in the weld metal will affect the transformation temperature and the final microstructure [Farrar et al. 1993]. The average width of columnar grains in both weld metals deposited at different preheat temperatures were measured in this work, and the results are shown in Table 6.8. It is evident from this table that the preheat did not influence the columnar grain size. The average width of the columnar grains was $32.12 \pm 1.54 \, \mu m$ in weld metal deposited with 5\%CO$_2$ shielding gas and $30.67 \pm 0.31 \, \mu m$ in weld metal for 20\%CO$_2$ shielding gas. The widths of the columnar grains in the two weld metals were very similar. Consequently, it can be concluded that the shielding gas did not have a significant effect on the columnar grain size in the weld metals, and therefore, the columnar grain size did not make a significant contribution to the final weld metal microstructure.

### 7.4.4 Effect of Non-Metallic Inclusions on Microstructure

Adding carbon dioxide to the shielding gas gives a significant source of oxygen absorption into the weld pool. Most of the oxygen will dissipate into the surroundings, but some oxygen will combine with alloying elements in the weld pool to form oxides. During solidification, some of the oxides are entrapped as inclusions in the weld metal. The inclusions are considered to play an important role in determining the final weld metal microstructure.

Manganese, silicon and titanium are strong deoxidizers which will react with oxygen during welding to form MnO, SiO$_2$ and TiO etc. [Pollard and Milner 1971; Grong and Christensen 1983; Francis et al. 1990]. Some of the oxides are entrapped within the solidified weld pool as finely distributed inclusion particles. As indicated in Section 7.3,
20%\text{CO}_2\ shielding\ gas\ has\ a\ higher\ oxygen\ potential,\ and\ thus\ more\ oxides\ are\ formed\ during\ the\ weld\ metal\ solidification.\ The\ characteristics\ of\ the\ non-metallic\ inclusions\ in\ the\ two\ weld\ metals\ were\ systematically\ analysed\ by\ using\ image\ analysis\ in\ the\ current\ work,\ and\ the\ results\ are\ presented\ in\ Section\ 6.5.\ Since\ the\ non-metallic\ inclusion\ characteristics\ can\ be\ affected\ by\ cooling\ rate\ at\ higher\ temperature\ and\ the\ higher\ temperature\ cooling\ rate\ is\ not\ strongly\ influenced\ by\ preheating\ from\ 20^\circ\text{C}\ to\ 100^\circ\text{C},\ it\ is\ understandable\ that\ the\ preheat\ treatment\ did\ not\ have\ a\ significant\ influence\ on\ the\ characteristics\ of\ the\ non-metallic\ inclusions.\ However,\ the\ different\ CO_2\ contents\ in\ the\ shielding\ gas\ mixtures\ are\ expected\ to\ change\ the\ characteristics\ of\ the\ inclusions.\ The\ overall\ average\ values\ of\ the\ non-metallic\ inclusion\ characteristics\ in\ weld\ metals\ deposited\ with\ the\ two\ shielding\ gases\ (including\ all\ preheat\ conditions)\ are\ listed\ in\ Table\ 7.2.\ The\ inclusions\ in\ the\ weld\ metal\ deposited\ with\ 20%\text{CO}_2\ shielding\ gas\ mixture\ have\ a\ higher\ volume\ fraction\ and\ a\ slightly\ larger\ mean\ particle\ size\ and\ number\ of\ particles\ per\ unit\ volume\ due\ to\ higher\ oxygen\ absorption\ in\ the\ molten\ pool\ during\ welding.\ The\ larger\ number\ of\ particles\ per\ unit\ volume\ probably\ contributes\ to\ higher\ transformation\ temperature\ during\ cooling,\ due\ to\ a\ larger\ population\ of\ heterogeneous\ nucleation\ sites.

Table 7.2 Average values of non-metallic inclusion characteristics

<table>
<thead>
<tr>
<th>Shielding gas</th>
<th>$V_v$, %</th>
<th>$\bar{d}_v$, $\mu$m</th>
<th>$\bar{d}_v$, $\mu$m</th>
<th>$N_v$, $1/\mu$m$^2$</th>
<th>$N_v'$, $1/\mu$m$^3$</th>
<th>$\lambda_v$, $\mu$m</th>
<th>$S_v$, $\mu$m$^{-1}$</th>
<th>$\bar{A}$, $\mu$m$^2$</th>
</tr>
</thead>
<tbody>
<tr>
<td>5%\text{CO}_2</td>
<td>0.224</td>
<td>0.418</td>
<td>0.656</td>
<td>0.0101</td>
<td>0.0156</td>
<td>2.251</td>
<td>0.0206</td>
<td>0.226</td>
</tr>
<tr>
<td>20%\text{CO}_2</td>
<td>0.292</td>
<td>0.451</td>
<td>0.708</td>
<td>0.0111</td>
<td>0.0158</td>
<td>2.224</td>
<td>0.0247</td>
<td>0.263</td>
</tr>
</tbody>
</table>

As is well known, fine inclusions, particularly titanium bearing inclusions, can provide nucleation sites for acicular ferrite and contribute to the increase in the ferrite transformation kinetics of solidified weld metal during cooling. The EDS spectra attached in Appendix II indicate that all the inclusions were rich in manganese and titanium and contained certain amounts of silicon and aluminium. Normally, the
proportion of acicular ferrite increases with increasing manganese, silicon and titanium contents in the weld metal [Evans 1980, 1986, 1992]. This effect could contribute to the finer acicular ferrite structure in the weld metal deposited with 5%CO₂ shielding gas. Another contributing factor could be greater nucleating efficacy because of the generally finer size of the non-metallic inclusions. However, the non-metallic inclusion characteristics of both weld metals are similar and it can be concluded that the non-metallic inclusions only make a small contribution to the weld metal microstructures.

7.5 THE HARDNESS OF THE WELD METALS

The hardness of the weld metal can be readily measured and, to a certain extent, it reflects the mechanical properties of the weld metal. Basically, the hardness of the steel weld metal will be dependent on the chemical composition and microstructure of the weld metal.

7.5.1 Effect of Preheat Temperature

The preheat treatment does not contribute significantly to the chemical composition of the weld metal, but it can modify the cooling rate, and affect the steel weld metal microstructure and hardness. If the weldment cools rapidly during solidification, a hard weld metal microstructure is likely to form. If the cooling rate is low, a softer microstructure will be formed. Thus, the weld metal hardness can be used as an indicator of weld metal microstructure. Phase transformation during cooling in the two weld metals investigated occurred between 600°C and 300°C (refer to CCT diagram data in Table 6.12). The cooling rate over this temperature range decreased with increasing preheat temperature (refer to Figure 6.1), thereby resulting in a decrease in the average hardness values of the weld metal with increasing preheat temperature (refer to Figure 6.17).
7.5.2 Effect of Shielding Gas

Comparison of average hardness values of the two weld metals using different shielding gas mixtures (Figure 6.15) indicates that the weld metal deposited with 5%CO₂ shielding gas had a higher hardness than that deposited with 20%CO₂ shielding gas at the same preheat temperature.

In the gas shielded FCAW process, the shielding gas oxygen equivalent plays an important role in controlling the pyrometallurgical reactions that occur during welding. Therefore, due to the difference in CO₂ content, the two shielding gas mixtures used in this work have contributed significantly to the chemical compositions of the weld metal. Table 7.1c shows the chemical compositions of both deposited weld metals investigated in the present work. It is clear that the weld metal using 5%CO₂ shielding gas contained less oxygen and more carbon, boron, silicon and manganese than that using 20%CO₂ shielding gas.

Carbon is the most important alloying element to influence the hardenability of steel. It is well known that carbon, manganese and silicon contribute solid solution hardening and lower the transformation temperature of austenite to ferrite during weld metal cooling. On the other hand, the hardness of weld metal can be increased by carbide precipitation. Therefore, the hardness of weld metal is increased with increasing carbon content. Boron is a very powerful promoter of hardenability on a weight percentage basis when the amount of boron is below approximately 0.010 weight percent [Linnert, 1994]. The capacity of boron to promote hardenability of weld metal lies in its ability to segregate to the prior austenite grain boundaries and to retard the formation of proeutectoid ferrite [Ohkita et al. 1984; Grong and Matlock 1986]. The higher boron content in the weld metal deposited with 5%CO₂ shielding gas is expected to contribute to the higher hardenability and hardness of the weld metal. It has been reported that changes in oxygen content in the weld metal by using different shielding gas mixtures has a significant influence on the silicon and manganese recovery [Lathabai and Stout...
1985], leading to changes in weld metal hardness. The manganese and silicon in solid solution will increase the hardenability of the weld metal [Evans 1986; Grong and Matlock 1986]. Manganese and silicon are strong non-metallic inclusion formers in weld metal. However, not all of the manganese and silicon in the weld metal are in the form of inclusions. In the present work, the weld metal deposited with 5%CO₂ shielding gas had higher manganese and silicon contents but the inclusion volume fraction was lower. Therefore, more manganese and silicon would be expected to be in solid solution increasing the hardenability of the weld metal.

The weld metal carbon equivalent can be taken as a hardenability index, which ranks the effect of the various alloying elements on the austenite to ferrite transformation behaviour relative to carbon. The deposited weld metal using 5%CO₂ shielding gas had much higher carbon equivalent (P<sub>emo</sub> = 0.212) than the deposited weld metal using 20%CO₂ shielding gas (P<sub>emo</sub> = 0.165), consistent with the observation that the weld metal deposited with 5%CO₂ shielding gas had a higher hardness.

7.6 WELD METAL TRANSVERSE COLD CRACKING

Weld metal transverse cold cracking occurs perpendicular to the axis of the weld deposit. In general, cracking occurs at temperatures below 200°C and the onset is delayed for minutes or hours after welding. The time delay depends on the type of steel, the magnitude of the welding stress and the hydrogen content in the weld metal. Weld metal transverse cold cracking is caused by the complex interaction of the diffusible hydrogen supply, tensile residual stress concentration and susceptible microstructure.

The transverse cracking in the weld metals studied in this work initiated from the weld root in the gap between the test blocks. The cracking was confirmed to be cold cracking since it occurred below 150°C and was delayed for at least five minutes after welding (refer to Figures 6.3 and 6.4). The microstructures of the weld metals investigated did not appear to be a major factor in the transverse cold cracking. The microstructures
observed in the two weld metals predominantly consisted of acicular ferrite (refer to Figures 6.9 and 6.10). However, a sufficiently high concentration of hydrogen can also induce cold cracking in acicular ferrite in spite of its relatively high toughness [Alam et al. 1999].

The characteristics of the transverse cold cracking in the weld metals deposited with the two shielding gas mixtures at different preheat temperature were investigated. The results can be summarised as follows.

(1) The susceptibility to cold cracking decreased with increasing preheat temperature in both weld metals.

(2) The weld metal deposited with 20%CO₂ shielding gas was less susceptible to cold cracking than that deposited with 5%CO₂ shielding gas at the same preheat temperature.

(3) There were three types of fracture morphologies associated with the cold cracking: microvoid coalescence (MVC) fracture, quasi-cleavage (QC) fracture and intergranular (IG) fracture.

7.6.1 Effect of Preheat on Weld Metal Cold Cracking

Preheat treatment can play an important role in reducing weld metal cold cracking in high strength steel welding since preheat is normally performed to control the cooling rate during weld metal solidification and solid state transformation, and to ensure sufficient removal of hydrogen from the weld metal.

In this work, the preheat treatment exerted a marked influence on the susceptibility to weld metal cold cracking, as summarized in Section 6.4.1. The cold crack percentage decreased with increasing preheat temperature. For weld metal deposited with 20%CO₂ shielding gas, crack-free welding was obtained at 40°C preheat. For weld metal deposited with 5%CO₂ shielding gas, the preheat temperature needed for crack-free welding was 100°C.
It is well known that the main cause of weld metal cold cracking is the diffusible hydrogen generated under the welding arc at high temperatures. Restricted entry of hydrogen and/or effusion of hydrogen from the weld metal is necessary to avoid cold cracking. Referring to the results in Section 6.1, increasing preheat temperature decreases the cooling rate during weld solidification, especially in the lower temperature range. Therefore, preheat firstly offers the benefit of reducing susceptibility to cold cracking by enhancing hydrogen diffusion; the hydrogen has a longer time to diffuse out of weld metal before a critical temperature is reached.

Another factor important to weld metal cold cracking is a susceptible microstructure, which is frequently identified as martensite or bainite formed at a faster cooling rate through the transformation range. Preheat treatment can alter metallurgical transformation to produce a microstructure of lower hardness which is less susceptible to cold cracking. However, as discussed in Section 7.4.1, preheat from 20 to 100°C did not have a significant effect on weld metal microstructure, at least on the scale of optical microscopy, but the average hardness values of the weld metals decreased with increasing preheat temperature. The hardness decrease arises from microstructural changes which usually cannot be resolved by optical microscopy, such as dislocation annihilation and rearrangement (recovery) and precipitate coarsening.

The third main factor for weld metal cold cracking is residual stress concentration in the weld metal. During the welding process, elastic and plastic deformation may occur due to non-uniform heating and thermal contraction, leading to distortion upon cooling. As the weldment cools weld metal will shrink more than the base metal which has only been heated by a small amount. The generated stresses result in elastic and plastic deformation and residual stress. However, the base metal will expand if it is preheated. During weld solidification, the temperature difference between weld metal and base metal will be reduced with increasing preheat temperature, which leads to a lower thermal contraction difference between weld metal and base metal, thereby reducing the extent of elastic and plastic deformation and residual stress. Two phenomena will tend
to counteract the stress due to thermal contraction: the transformation of austenite into ferrite which gives a volume expansion; and stress relaxation due to plastic strain and recovery. The former effect occurs over the transformation range (> about 350°C in this case); and the latter occurs more readily at higher temperature because of thermally activated dislocation motion and annihilation.

7.6.2 Effect of Diffusible Hydrogen

Hydrogen is a significant contaminant in welds because it can cause embrittlement and cold cracking. In the flux cored arc welding (FCAW) process, weld metal hydrogen can be considered normally to arise from the following sources [Yurioka and Suzuki 1990],

(a) moisture in the flux;
(b) moisture in the shielding gas;
(c) moisture in the atmosphere;
(d) hydrogen in the cored wire steel; and
(e) organic substances in the flux.

The quantity of hydrogen dissolved in the weld depends on chemical composition of the weld metal; partial pressure of hydrogen in the arc atmosphere; the transfer time of a molten metal droplet from the electrode to the molten pool; size of the droplet; and the temperature to which the metal has been heated during the welding process [Mikula 1994]. In the metallurgy of welding, the hydrogen present in weld metal can be divided into two types [Smialkowski 1961]:

(1) diffusible hydrogen which is dissolved in the weld metal in an atomic or ionic form and which will evolve from the weld metal at ambient or slightly higher temperatures; and

(2) residual hydrogen which includes the hydrogen contained in the micro-pores present in the body of the weld metal or on the surfaces between the matrix and non-metallic
inclusions and the hydrogen which is present in irreversible traps in strongly bonded form. This hydrogen will not migrate and effuse from the weld metal at lower temperatures.

The diffusible hydrogen contents of the investigated weld metals are shown in Table 6.6. The weld metal deposited with 5%CO₂ shielding gas contained more diffusible hydrogen (8.6 ml/100g) than that deposited with 20%CO₂ shielding gas (6.4 ml/100g). These values were obtained from all weld metal deposits and it is assumed that a similar diffusible hydrogen ratio prevails in the diluted weld metals. The analysis of cold crack percentage for the two weld metals (Section 6.5.1) indicated that the weld metal deposited with 20%CO₂ shielding gas had lower susceptibility to transverse cold cracking, consistent with its lower diffusible hydrogen content.

The diffusible hydrogen content in the weld metal is a key factor for determining the susceptibility to weld metal hydrogen induced cold cracking. The weld metal that contained more diffusible hydrogen had lower resistance to cold cracking at the same preheat temperature, and it required a higher preheat temperature to prevent the occurrence of cold cracking.

Several models have been proposed for hydrogen induced cold cracking. One of the main models is the decohesion model. This model was proposed by Troiano [1962] and developed by Oriani [1970]. The model assumes that due to gradients of elastic stress, hydrogen diffusion results in a build up of hydrogen atoms in the crystal lattice where a tensile stress concentration exists. A tensile stress concentrated in the front of a crack tip attracts a relatively fast flow of highly mobile hydrogen atoms from other parts of the weld metal [Machnienko et al. 1986]. The micromechanism of hydrogen embrittlement can be interpreted as hydrogen causing a reduction in the mutual attraction forces between the iron atoms in the crystal lattice of the weld metal. The study of Markworth et al. [1973] suggested that the Fe-H attractive force is stronger than the Fe-Fe cohesive forces in the α-iron lattice, leading to strong hydrogen atom bonding to Fe atoms, and
Figure 7.4 Weld metal transverse cold crack percentages as a function of the average hardness values of the two steel weld metals.
increasing the Fe-Fe interatomic distance and generating a local elastic stress. Consequently, hydrogen atoms dissolved in an α-iron lattice deform the structure and reduce the cohesive energy of the crystal lattice of the metal. The reduction in binding energy can result in decohesion at or just ahead of the crack tip. Depending on the diffusible hydrogen content in the stress concentration zone, the resistance to cold cracking will be reduced in these regions.

Hydrogen induced cold cracking propagates in a discontinuous manner. Each time a crack propagates, an incubation period is needed for hydrogen to build up ahead of the new crack tip position by diffusion from surrounding regions [Vasudevan et al. 1981].

### 7.6.3 Effect of Shielding Gas on Weld Metal Cold Cracking

As discussed in Section 7.5.2, the shielding gas mixtures contributed significantly to the chemical compositions of the weld metals, and thereby influenced their hardness. The weld metal deposited with 5%CO₂ shielding gas had a higher average hardness than that deposited with 20%CO₂ shielding gas at the same preheat temperature. It has been established that the hydrogen induced cold cracking susceptibility of a weld metal microstructure increases with increasing weld metal hardness [Hart and Watkinson 1975; Düren and Niederhoff 1988]. For a diffusible hydrogen content between 5ml/100g and 10ml/100g in the deposited weld metal, the hardness plays a primary role in control of the resistance to weld metal cold cracking [Hart 1986]. Figure 7.4 shows the cold crack percentage as a function of the average value of weld metal hardness for the present work. The susceptibility to weld metal cold cracking increased with increasing weld metal hardness, and no cracking was observed for a weld metal hardness below about 291 HV0.2. This condition was met by the 20%CO₂ welds by a preheat of 40°C, but required 100°C preheat for the 5%CO₂ welds.

It should be noted that the weld metal deposited with 20%CO₂ shielding gas was lower in hardness than that deposited with 5%CO₂ at the same preheat temperature and that, in
general, 20%CO₂ shielding gas produced lower hardness weld metal. The maximum hardness for the weld metal deposited with 20%CO₂ was about 292 HV0.2 (obtained for 20°C preheat) and this was closely similar to the lowest hardness (about 291 HV0.2) of the weld metal deposited with 5%CO₂ (obtained for 100°C preheat). However, at this hardness level, the weld metal deposited with 5%CO₂ showed no cracking but the weld metal deposited with 20%CO₂ was about 70 percent cracked. The implication is that at a similar hardness the microstructure of weld metal deposited with 5%CO₂ is less crack susceptible and that the superior cracking resistance of weld metal deposited with 20%CO₂ at a given preheat temperature is due primarily to its lower hardness. Comparison of the microstructures of the two steel weld metals, Figures 6.9 and 6.10, reveals that the weld metal deposited with 5%CO₂ had a finer structure with a higher proportion of acicular ferrite - a structure which would normally be expected to show high crack resistance at a given hardness level. Therefore, provided the hardness of weld metal deposited with 5%CO₂ shielding gas does not exceed 290 HV the transverse cold cracking resistance is expected to be superior to weld metal of the same hardness deposited with 20%CO₂.

The different CO₂ contents of the shielding gas mixtures had an effect on the weld bead geometry as discussed in Section 7.3. Larger weld bead cross-sectional area was obtained for steel weld metal deposited with 20%CO₂ shielding gas. A larger weld metal cross-sectional area serves to reduce the contraction stress.

It should also be noted that the nitrogen content in the steel weld metal deposited with 5%CO₂ shielding gas was twice as high as the 20%CO₂ shielded weld metal. Generally, total exclusion of atmospheric contamination is impossible by using a shielding gas during FCAW welding. Nitrogen can be picked up from the air during welding via ineffective shielding, ventilation conditions and jetting arc phenomena [Gonzalez and Torga 1991]. In this work, it is possible that nitrogen was reduced in the case of the
20%CO₂ shielding gas because of more complete reaction between nitrogen and oxygen in the arc atmosphere.

The diffusion rate of nitrogen in the weld metal rapidly decreases as the weld solidifies, and the remaining nitrogen occurs in the weld metal normally in two states: as a solute species or precipitated as a nitride. Both states have a detrimental influence on the weld metal mechanical properties [Verhagen et al. 1970]. In the present work, the nitrogen content in the weld metals was relatively low (less than 85 ppm), but the strong nitride formers, such as titanium and aluminium, would be expected to tie up much of the nitrogen leaving a small concentration in solid solution in the weld metal. However, since titanium and aluminium were significant components of the non-metallic inclusions it is unlikely that all of the nitrogen will be fixed by these elements. For example, for titanium to combine stoichiometrically as TiN with all of the 85 ppm nitrogen in the weld metal deposited with 5%CO₂ shielding gas would require 0.03 weight percent titanium, which is the total titanium content of the weld metal. It is therefore likely that some solute nitrogen will be present, especially in the weld metal deposited with 5%CO₂. Nitride particles can act as crack initiators during local yielding of weld metal [Gonzalez and Torga 1991], and the precipitated nitrides can increase the hardness of the weld metal. Nitrides can also reduce the weld metal toughness [Smith and Johannes 1968; Krivenko et al. 1967]. Nitrogen in solid solution is also considered to decrease toughness and ductility of weld metal [Leonard 1977; den Ouden 1968; Gonzalez and Torga 1991]. Consequently, in comparison with the steel weld metal obtained using 20%CO₂ shielding gas, the higher nitrogen content in the weld metal deposited with 5%CO₂ could be another factor that contributes to the higher susceptibility to cold cracking.

7.6.4 Weld Metal Cold Cracking Behaviour in Test Assembly with Load Cells

Weld metal crack percentages were also estimated from samples which came from the
weldments made with load cells in the test assemblies. One phenomenon discovered during the tests, was that the cracks occurred in all samples with load cells attached to the test assemblies, and the crack percentage for weld metals obtained with load cells was higher than that without load cells at the same preheat temperature. The comparison of these results is shown in Figure 6.24. A possible explanation is that use of load cells during welding reduced the stability and stiffness of the test assembly, and changed the magnitude and direction of the restraint force.

7.6.5 Analysis of Fracture Morphologies

In the present research, the fracture surfaces of the weld metal subjected to transverse cold cracking consisted of three types of fracture morphologies: microvoid coalescence (MVC) fracture, quasi-cleavage (QC) fracture and intergranular (IG) fracture. These observations are consistent with the Beachem [1972] model which suggests that hydrogen facilitates every fracture mode which is allowed by the local microstructure [Vuik and Graville 1993]. It is well known that the type of fracture morphology is highly dependent on the stress intensity, hydrogen concentration and microstructure of the weld metal [Beachem 1972]. Gedeon and Eagar [1990] confirmed that the prevalence of the IG fracture mode of cold cracking decreases with increasing stress level and hydrogen content, and that high hydrogen content at low stress intensity suppresses MVC fracture and IG fracture becomes the dominant fracture mode. Beachem [1972] indicated that IG fracture occurs at the minimum combination of stress intensity and critical hydrogen level at the crack tip. At higher stress intensity, the fracture mode shifted from IG to the QC and MVC modes which require more plastic flow.

The G-BOP test is a self-restrained cracking test. Therefore, the local stress acting at a stress concentration where a crack can initiate plays a more significant role in hydrogen induced cold cracking than does the mean residual stress. Also, the fracture morphology of hydrogen induced cold cracks can be influenced by the diffusible hydrogen present in
the weld metal and the strength of the weld metal [Wang and Apps 1991]. These factors are related in turn to the microstructure of the weld metal. In the G-BOP test, it is obvious that the stresses resulting from the weld thermal cycle under self-restraint conditions are concentrated and increase to a triaxial state at the root of the weld bead in the gap region. The diffusible hydrogen is likely to accumulate near this gap region because of the higher local stress intensity at the weld root. Therefore, in the present work, weld metal transverse cold cracking always initiated initially from the root of the weld bead.

The fracture mode of hydrogen assisted cold crack in low carbon low alloy weld metal is dependent on the strength of the weld metal as illustrated below [Wang and Apps 1991]:

- Intergranular (IG) fracture along the columnar grain boundaries
- Quasi cleavage (QC) and intergranular (IG) fracture
- Quasi cleavage (QC) fracture
- Microvoid coalescence (MVC) fracture

The results of the measurements of hardness and the calculation of the carbon equivalent $P_{emo}$ for the two weld metals indicate that the strength of the weld metal deposited with 5%CO$_2$ shielding gas is higher than that deposited with 20%CO$_2$ shielding gas. The percentages of the different fracture morphologies for the two weld metals deposited at 20°C preheat are shown in Figure 6.30. The weld metal deposited with 20%CO$_2$ shielding gas was only partially cracked and contained more ductile fracture (MVC fracture).

The results given in Section 6.4.2 can be summarised as follows,
1. For the weld metal deposited with 5%CO₂ shielding gas, the fracture morphology near the initial crack site was mainly quasi-cleavage (QC) fracture. The transverse crack then propagated through the weld metal with the sequence of fracture modes: a mixture of QC and MVC fracture → a mixture of IG with limited MVC fracture → MVC fracture.

2. For the weld metal deposited with 20%CO₂ shielding gas, the fracture morphologies near the initial crack site consisted of MVC fracture or a combination of MVC and QC fracture. Then the crack propagated mainly in the form of MVC fracture except that a mixture of IG fracture with limited MVC occurring in isolated island regions until the crack eventually stopped.

In the QC fracture mode the characteristic striation-like markings can be recognized. However, it is difficult to identify whether any characteristics of pure cleavage fracture exist. Thus quasi-cleavage fracture (QC) has been assumed to dominate [Kikuta et al. 1978]. In hydrogen embrittlement, QC is a brittle fracture mode and is considered to be cracking on a glide plane, with decohesion occurring in conjunction with hydrogen accumulation around dislocations [Yurioka and Suzuki 1990]. Dislocation pileups at structural defects will produce an oversaturation of hydrogen, which will cause local stress concentration and embrittlement. Therefore, hydrogen transport by dislocations to the site of crack initiation is a necessary part of the embrittlement process [Graville 1968]. Since weld metal transverse cold cracking initiated at least five minutes after welding in the present work, the concentration of hydrogen at dislocations would have increased, finally resulting in the occurrence of QC fracture.

MVC fracture is a ductile fracture mode which is caused by microvoid nucleation and coalescence at non-metallic inclusions, resulting in a fine dimple pattern. High local stress intensity can generate a large plastic zone at the hydrogen concentrated crack tip. This plastic zone can include numerous non-metallic inclusions acting as microvoid nuclei. The microvoids grow by a process of plastic deformation until coalescence.
occurs. Therefore, cold cracking under high stress intensity in the presence of inclusions is typically in the form of MVC.

In the weld metals deposited with different shielding gas mixtures, the chemical compositions and mechanical properties were different. In each weld metal, the chemical compositions and microstructures were not homogeneous and the hydrogen distribution was non-uniform. The fracture modes of transverse cold cracking will differ for different compositions and microstructures. Therefore, the crack can propagate in the form of IG fracture, QC fracture, MVC fracture or a combination of MVC and QC fractures depending on the local restraint stress, microstructure and hydrogen distribution.

Many researchers have suggested that hydrogen can aid plastic processes, and thereby the dislocations can move more easily in the presence of hydrogen because of a lower lattice interaction stress [Beachem 1972; Matsumoto et al. 1981; Tabate and Birnbaum 1983, 1984; Robertson and Birnbaum 1984]. When the hydrogen accumulates at the stress concentrated crack tip, hydrogen can interact with dislocations and enhance their movement, resulting in localized plasticity near the crack tip. Thus, cracking occurring under medium stress intensity in a weld metal region of higher local strength is in the form of QC. On the other hand, MVC absorbs more energy than QC or IG since significant plastic deformation takes place. If the hydrogen concentration in the vicinity of the crack is high, its distribution can still be important in determining the fracture mode. For distribution of hydrogen over a relatively large volume of weld metal undergoing plastic deformation around the stress concentrated crack tip, void formation is expected to be promoted, and coalescence would be easier. On the other hand, local high concentration of hydrogen at traps ahead of the crack tip could promote localized QC. Consequently, the lower the degree of the localization of hydrogen the greater the chance for cracking in the form of MVC rather than QC [Perng and Altstetter 1987].
Figure 7.5 Schematic diagrams based on the Beachem diagram [beachem 1972] showing interrelationships between stress intensity factor, crack tip hydrogen content and cracking mode. (a) weld metal deposited with 5\%CO_2 shielding gas; and (b) weld metal deposited with 20\%CO_2 shielding gas. The upper dashed line represents the stress intensity required to produce crack propagation in the absence of hydrogen. Thin line sections of the curve indicate periods when cracking has not initiated or does not progress.
When a crack propagates, the local stress intensity at the crack tip decreases and the plastic zone can decrease. Therefore, further crack growth can change from MVC and QC fractures to the lower plastic deformation process of intergranular (IG) fracture. IG fracture shows a relatively plane surface as the crack follows the ferrite veins along the columnar grain boundaries. IG cracking generally occurs under low stress intensity. Cracks grown by IG fracture are associated with a period of time for diffusion and build-up of hydrogen at a grain boundary to increase the local plasticity to aid crack nucleation. IG is normally considered to be a brittle fracture mode but regions of MVC can occur along intergranular boundaries. In either case IG is caused by an accumulation of hydrogen at the grain boundary, either reducing cohesion between atoms so that interatomic bond separation can be achieved [McMahon et al. 1981]; or facilitating plastic flow and microvoid formation. IG is also a relatively slow fracture mode compared with the MVC and QC modes [Oriani and Josephic 1972].

The influence of hydrogen and stress intensity at the crack tip on crack fracture modes are illustrated schematically in Figure 7.5a for the weld metal deposited with 5% CO₂ shielding gas at 20°C preheat, and in Figure 7.5b for the weld metal deposited with 20% CO₂ shielding gas at 20°C preheat. These diagrams are modified forms of the Beachem [1972] diagram. Since the time and temperature variables and microstructural variability are not included, these diagrams are highly schematic. The cold cracking fracture process can consist of mixed modes depending on the changes in the local stress intensity and hydrogen concentration at the crack tip. In Figures 7.5a and 7.5b, a mixed MVC and QC zone has been added to the Beachem [1972] diagram in an attempt to rationalize the cracking sequence observed in the present work.

For both Figures 7.5a and 7.5b, point X indicates the combination of stress intensity and hydrogen concentration just after welding. As time passes and the temperature falls hydrogen accumulates at sites of potentially high triaxial stress intensity and the stress intensity rises because of thermal contraction against the restrained weld test.
configuration. Point O arbitrarily indicates the hydrogen and stress intensity prevailing at 200°C and four minutes after welding.

In Figure 7.5a, as the temperature falls below the critical temperature (assumed to about 120°C), the hydrogen concentration and stress intensity reach the critical levels and a crack in the weld metal deposited with 5%CO₂ shielding gas initiates in the form of QC fracture (point A). The crack propagates then by a mixed fractures of QC and MVC depending on the local stress intensity and the hydrogen distribution near the crack tip (region B-B’ in Figure 7.5a). A possible explanation for the occurrence of limited MVC fracture in region B-B’ is that the local stress intensity ahead of the crack tip is high enough to maintain a large plastic zone, promoting MVC nucleated by non-metallic inclusions. As the crack grows, the stress intensity is reduced and decreases to a level which gives rise to a smaller plastic zone [Alam et al. 1997]. It is well established that hydrogen induced cracking is discontinuous [Steigerwald et al. 1959; Vasudevan et al. 1981] and cracking could be arrested in the time taken for the hydrogen to accumulate and stress intensity to fall from B’ to C. At C cracking restarts with the crack mode changing to IG fracture (at lower stress intensity) along the columnar grain boundaries with little sign of deformation. The hydrogen concentration rises prior to the onset of IG because hydrogen accumulated at the grain boundaries causes a rise in hydrogen at the crack tip. Finally, fracture occurs in the form of MVC (region D in Figure 7.5a). A possible reason for this change is that the remaining uncracked area decreases substantially with the progress of cracking, particularly approaching the top of the weld bead, and thus the stress and the local stress intensity at the crack tip may increase to a high level during the last stage of unstable crack propagation. In addition, a concentration of non-metallic inclusions in this upper region could provide sites for hydrogen accumulation [Olson et al. 1996], and thus the conditions are appropriate locally (high stress intensity and hydrogen concentration) for MVC cracking to occur (point D).
Figure 7.6 Schematic representation of fracture profile of hydrogen embrittlement in the HAZ of high strength steel [After Kikuta et al. 1978]
As discussed previously, the weld metal deposited with 20%CO₂ shielding gas had a lower hardenability and a higher volume fraction of non-metallic inclusions. It is expected that the inclusions are favourable sites for hydrogen trapping in the weld metal. In Figure 7.5b, five minutes after welding, the temperature falls below 200°C (point O). As the temperature continues to fall and the hydrogen concentration and stress intensity reach the critical levels, cracking is initiated by MVC or a combination of MVC and QC depending on the local stress concentration, hydrogen localization and the local microstructure (point A). Because of the lower strength and the higher non-metallic inclusions content, the plastic zone ahead of the crack tip is larger and the crack grows mainly by MVC (region B-B’ in Figure 7.5b). Eventually, the stress intensity starts to fall because of stress relief by cracking. As the crack propagates, part of the atomic hydrogen that diffuses into the triaxially stressed region is lost as molecular hydrogen [Okuda et al. 1987; Savage et al. 1976]. Also, hydrogen diffuses out of the weld metal over time resulting in decreasing amounts of hydrogen which are available for diffusion to the triaxially stressed regions. Cracking will no longer be possible once the hydrogen concentration has fallen below point C in Figure 7.5b - the critical level for the local stress at the crack tip.

Small islands of IG fracture were observed within the MVC region for the weld metal deposited with 20%CO₂ (zone 3, Figure 6.25b). It is likely that a local microstructural effect is exerted on the cracking mode due to the crack encountering an easy crack path along prior austenite columnar grain boundaries, causing a localized departure from the general fracture mode of MVC.

According to Kikuta et al. [1978], the hydrogen embrittlement fracture path in the bainitic or martensitic HAZ of high strength steels can be classified into one of four types: translath (TL) fracture, interlath (IL) fracture, intercolony (IC) fracture and intergranular (IG) fracture. In actual hydrogen induced cold cracking, the four types of crack path can occur as a mixture. A schematic representation of the fracture profile of hydrogen embrittlement cracking is illustrated in Figure 7.6. Similar cracking
Figure 7.7 Typical columnar grain features underlaying the fracture surface of weld metal deposited with 20%CO₂ shielding gas showing the transverse cold cracking fracture path.
Figure 7.8 Comparison of percentages of different fracture morphologies in weld metal deposited with 5%CO₂ shielding gas at different preheat temperatures, a. 20°C preheat; and b. 80°C preheat
characteristics are likely to apply to the present case of transverse cold cracking of weld metal. In this case however, the more irregular structure of acicular ferrite dominates rather than lath packets and the crack path is related to the prior solidification structure - either intercellular boundaries arising from cellular-dendritic growth or the intercolumnar (intergranular) boundaries between the prior austenite grains. Over much of the fracture surface the crack path was approximately intercolumnar (IC or IG) or intercellular (ICell), Figures 6.35 and 6.36. Segments of transcolumnar (TC) fracture were also present giving a step-like appearance, see Figure 7.7. The presence of crack segments parallel to intercolumnar or intercellular boundaries is evident in Figures 6.35 and 6.36 and suggests that boundary segregation, both in composition and inclusions, lowers cracking resistance. When these boundaries deviate from a direction perpendicular to the direction of the tensile stress, the crack path can change to a transcolumnar direction producing a step before again propagating parallel to the intercolumnar boundaries. In this way, the general direction of cracking is maintained perpendicular to the tensile stress direction. Despite the presence of cracking along prior boundaries, all three modes of cracking (MVC, QC and IG) were associated with this crack path, depending on position and weld metal type. In the steel weld metal deposited with 20%CO₂ shielding gas at 20°C preheat, MVC dominated whilst IG was more prevalent in the weld metal deposited with 5%CO₂ shielding gas at 20°C preheat.

As discussed previously, the resistance to weld metal transverse cold cracking increased with increasing preheat temperature because higher preheat allows more significant hydrogen effusion from the weld metal and lowers the weld metal hardness. It was also found that the proportions of the fracture modes changed with increasing preheat temperature. Figure 7.8 compares the percentages of the different fracture modes in weld metal deposited with 5%CO₂ shielding gas at 20°C preheat (Figure 7.8a) and 80°C preheat (Figure 7.8b). The average hardness values of weld metal deposited with 5%CO₂ shielding gas at 20°C and 80°C preheat temperatures were 330 HV0.2 and 301
HV0.2, respectively. The weld metal deposited at 80°C preheat was only partially cracked and MVC fracture dominated instead of IG. For weld metal deposited at 20°C preheat, Figure 7.8a, IG was more extensive than MVC, indicating that the toughness of the weld metal increased with increasing preheat temperature along with decreasing steel weld metal hardness. These results show that the hardness level is a good indicator of susceptibility to transverse cold cracking, consistent with cracking hierarchy proposed by Wang and Apps [1991]. In this hierarchy, increasing strength shifts the fracture mode from MVC towards IG. Conversely, decreasing the strength (or hardness) favours fracture by MVC. However, initiation of MVC requires high hydrogen and/or high restraint according to the Beachem [1972] analysis and therefore the probability of cold cracking will decrease as the strength (hardness) of the weld metal decreases and the ductility increases. The average hardness value of the weld metal deposited with 5%CO₂ shielding gas at 80°C preheat was still higher and the proportion of MVC fracture mode was smaller compared with the weld metal deposited with 20%CO₂ at 20°C preheat (see Table 6.7 and Figure 6.30b). However, the toughness of the weld metal deposited with 5%CO₂ shielding gas was significantly improved by preheating. Moreover, as discussed in Section 7.6.3, the weld metal deposited with 5%CO₂ shielding gas has a lower susceptibility to transverse cold cracking than that deposited with 20%CO₂ shielding gas at the same strength (hardness) level provided the hardness can be controlled by preheat to a value less than the critical level (about 290 HV).

7.7 CONTINUOUS COOLING TRANSFORMATIONS

It well known that the continuous cooling transformation (CCT) diagrams produced for welding situations can provide a valuable means of examining and understanding weld metal behaviour [Harrison et al. 1981; Ito et al. 1982; Harrison and Farrar 1987]. CCT diagrams can be used to describe microstructural developments occurring in weld metal.
Figure 7.9 Comparison of the profiles of the CCT diagrams for the two weld metals and BIS812EMA steel.
They also can be used for investigations into effects of transformation behaviour on steel weld metal toughness and the influence of individual alloying elements, prior austenite grain size and inclusion content on weld metal transformation behaviour.

The CCT diagram relates the chemical composition, cooling rate and prior austenite grain size of the material to its $\gamma$ to $\alpha$ transformation temperature range and the resultant microstructure and hardness. In the present work, the CCT diagrams for two steel weld metals (deposited with 5%CO$_2$ and 20%CO$_2$ shielding gas mixtures, respectively) and base metal (BIS812EMA quenched and tempered steel) have been determined. The $\gamma$ to $\alpha$ transformation start temperature ($Ar_3$) and finish temperature ($Ar_1$) decreased with increasing cooling rate, as would be expected since the start of $\gamma$ to $\alpha$ transformation shifts to lower temperatures and longer times with increasing cooling rate during continuous cooling [Krauss 1980].

Figure 7.9 shows the profiles of the three CCT diagrams. It is clear that the $Ar_3$ and $Ar_1$ temperatures for dilatometer samples of steel weld metal deposited with 20%CO$_2$ shielding gas were higher than those for weld metal deposited with 5%CO$_2$ shielding gas at the same cooling rate during continuous cooling transformation. The $Ar_3$ and $Ar_1$ temperatures for both of the steel weld metals were also higher than those values for the base metal at the same cooling rate. For example, for a cooling time $\Delta t_{815}$ of about nine seconds (very close to the cooling rate measured during the real welding process), the $Ar_3$ and $Ar_1$ for samples of the three different materials are as follows (refer to Table 6.12):

- weld metal using 20%CO$_2$ shielding gas: 603°C ($Ar_3$) and 443°C ($Ar_1$);
- weld metal using 5%CO$_2$ shielding gas: 538°C ($Ar_3$) and 338°C ($Ar_1$); and
- base metal: 437°C ($Ar_3$) and 299°C ($Ar_1$).
Basically, there are four major factors influencing the CCT diagram [Liu, 1983]:

1. Composition of alloy investigated;
2. Prior austenite grain size;
3. Austenitization temperature;

Theoretically, the CCT diagram reflects the stability and decomposition kinetics of austenite. Alloying elements play important roles in determining phase transformation during cooling. Interstitial elements in solution, such as carbon and nitrogen, will strongly stabilize the austenite, and reduce the kinetics of pearlitic and bainitic transformation, which are controlled by diffusion. Other alloying elements, such as manganese and nickel will also retard the phase transformation during cooling, and molybdenum and chromium displace the ferritic and pearlitic transformations to longer times and extend the bainitic transformation [Harrison and Farrar 1989]. Therefore, increasing the content of these alloying elements will decrease the rates of the pearlitic and bainitic transformations, and the onset of the transformation is generally displaced to longer times and lower temperatures.

The effect of prior austenite grain size has relatively less effect on the decomposition of super-cooled austenite. It is known that grain boundaries act as strong nucleation sites. Grain boundaries have a higher energy than the surrounding structure and there is a larger driving force for nucleation of diffusional products at these sites [Honeycombe 1981]. An increase in austenite grain size results in a decrease in the total area of grain boundaries per unit volume, and hence the number of nucleation sites is reduced. Thus, the pearlitic and bainitic transformation temperatures decrease with increasing prior austenite grain size.

For most steels, high temperature austenitizing increases the grain size, dissolves carbides and non-metallic inclusions and homogenizes austenite, thus suppressing
diffusional transformation, and resulting in lower phase transformation temperatures
during cooling.

Carbides and non-metallic inclusions can suppress austenite grain growth during
heating and act as nucleation sites [Ito et al. 1982] when pearlitic and bainitic
transformations occur during cooling. Therefore, they will reduce the stability of
austenite and enhance phase transformation during cooling.

The weld metal continuous cooling transformation characteristics can affected by the
oxygen content, which is related to the volume fraction of non-metallic inclusions in the
weld metal. Normally, the volume fraction of inclusions decreases with decreasing
oxygen content in the weld metal. Decreasing the volume fraction of inclusions will
reduce the interaction of inclusions with austenite grain boundaries and allow larger
austenite grains to become established. The increased grain size will depress the \( \gamma \) to \( \alpha \) transformation temperature due to the change in the grain boundary area [Harrison and
Farrar 1981]. On the other hand, intragranular oxides can nucleate ferrite and raise \( \text{Ar}_3 \)[Dunne 1999].

The samples used to determine the CCT diagrams of the weld metals were made from
the two kinds of all weld metal deposits. Comparison between the compositions of the
two all weld metals in Table 7.1b indicated that the weld metal using 20%CO₂ shielding
gas mixture had higher oxygen content and lower manganese, silicon and nitrogen
content. On the other hand, prior austenite grain sizes measurements of the dilatometer
weld metal samples showed that the weld metal deposited with 20%CO₂ shielding gas
had smaller average austenite grain size (refer to Section 6.6.4). Considering the
analysis of factors affecting the CCT diagram, the \( \gamma \) to \( \alpha \) transformation temperatures
during continuous cooling will increase with decreasing manganese and nitrogen
content in weld metal; and the prior austenite grain size will decrease with increasing
weld metal oxygen content and therefore the \( \gamma \) to \( \alpha \) transformation temperatures will be
Figure 7.10 Comparison of microstructures for actual weld samples and dilatometer samples

a. Actual weld samples: $\Delta t_{85} = 8.3$ Seconds; $20^\circ$C preheating

b. Dilatometer samples: $\Delta t_{85} = 9$ Seconds

50$\mu$m
depressed. The combination of these effects resulted in the observation that \( \gamma \) to \( \alpha \) transformation of weld metal using 20%\( \text{CO}_2 \) shielding gas mixture occurred at a higher temperature during continuous cooling. The microstructures of dilatometer samples for the two weld metals were quite similar (refer to Figures 6.51 and 6.52), and at the same cooling rate, the microstructures of weld metal deposited with 5%\( \text{CO}_2 \) was slightly finer. The steel weld metal deposited with 5%\( \text{CO}_2 \) shielding gas had a higher content of hardenability elements such as manganese and silicon, and transformed from the austenite at lower temperature during continuous cooling. The hardness measurements for the dilatometer samples reflected this fact.

During the real welding process, the measured cooling time \( \Delta t_{q/5} \) was about eight seconds for weld metal deposited at 20\( ^\circ \)C preheat. Therefore, the microstructure of dilatometer samples quenched at a cooling time \( \Delta t_{q/5} \) of about nine seconds is appropriate to compare with the microstructure of the actual bead on plate weld in the G-BOP test. This comparison is shown in Figure 7.10. For the two weld metals, the microstructures of actual weld and dilatometer samples were similar and mainly contained acicular ferrite and bainitic ferrite. However, a small amount of grain boundary ferrite with Widmanstatten ferrite side plates was observed in the actual weld metals, and the microstructures were finer than the all weld metal dilatometer samples. Small amounts of granular bainite and quasi-polygonal ferrite were present in the dilatometer samples. The slight difference in microstructures are due to the different contents of some alloying elements in the actual weld metals and dilatometer samples. Comparing Table 7.1b and Table 7.1c, it is clear that actual weld metals contain more carbon, chromium and molybdenum because of dilution with the BIS812EMA steel base. Carbon is a strong austenite stabilizer and lowers the transformation temperature of austenite to ferrite. Chromium and molybdenum inhibit the austenite to ferrite transformation and may only slightly lower the transformation temperature [Paxton 1981]. Therefore, the microstructure of dilatometer samples consisted of more higher
Figure 7.11 Comparison of average hardness values of dilatometer samples obtained in the two weld metals and base metal cooled at various rates. The broken vertical lines represent the range of $\Delta t_{85}$ times measured for the G-BOP test welds. (Samples austenitised at 1250°C for 13 seconds before cooling)
temperature products, such as granular bainite and quasi-polygonal ferrite. Figure 7.10 indicates that the microstructures of the HAZ in the actual weldment and the dilatometer sample of the base metal were very similar, with microstructures consisting substantially of lath martensite.

A comparison of the average hardness values for the dilatometer samples of the all weld metals and the base metal as a function of cooling time (800-500°C) is shown in Figure 7.11. It was found that the average hardness value of weld metal deposited with 5%CO₂ shielding gas was higher than for 20%CO₂ shielding gas at the same cooling rate, and the average hardness of the base metal was much higher than for the weld metals at same cooling rate. The average hardness of the base metal decreased significantly at slow cooling rates.

In comparison with the weld metals, the base metal (BIS812EMA quenched and tempered steel) contained much higher carbon, boron, chromium and molybdenum, and it also contained other alloying elements such as manganese, nickel, and silicon. The base metal had a higher hardenability, with a carbon equivalent $C_e$ of 0.495 and $P_{cm}$ parameter of 0.301, hence promoting the formation of lower temperature products during continuous cooling [Porter and Easterling 1992]. Lower temperature products: martensite and bainitic ferrite; were obtained for a range of cooling rates (refer to Figure 6.53). The much higher average hardness values compared to the two weld metals shown in Figure 7.11 also indicate that the base metal had significantly higher hardenability and strength.

The $\Delta t_{9/5}$ values recorded in Figure 6.1 for actual welds were in the range of eight to ten seconds depending on the preheat temperature. This range is marked on Figure 7.11. A value of $\Delta t_{9/5}$ of eight seconds corresponds to welding without preheat and the average weld metal hardness values of 330 and 292 HV0.2 for the bead on plate welds deposited with 5%CO₂ and 20%CO₂ shielding gases can be compared with average values of about 302 and 240 HV0.2 for the dilatometer samples of the two all weld
metals at $\Delta t_{85}$ equal to eight seconds (Figure 7.11). Tables 7.1b and 7.1c indicate that the amounts of hardening elements, such as carbon, chromium and molybdenum are lower in all weld metals compared with the actual weld metals. Therefore, the lower hardness values for samples subjected to simulated weld thermal cycling in the dilatometer is probably from the lower hardenabilities arising from the leaner composition and a finer $\gamma$ grain size than in the actual welds. This effect was more marked for the lower hardenability weld metal deposited with 20%CO$_2$ shielding gas. However, the effect of the dilatometer heat treatment on the hardness of the base metal was minor, about 420 HV0.2 for the actual weld in the grain coarsened zone ($\Delta t_{85} = 8$ s) and 423 HV0.2 for the simulated cycle ($\Delta t_{85} = 8$ s), indicating that the base metal hardenability is so high that prior $\gamma$ grain size differences have little effect on the hardness of the microstructure produced for $\Delta t_{85}$ value close to eight seconds.

7.8 TEMPERED WELD METALS

The process of coherent / semi-coherent precipitation from a supersaturated solid solution normally increases strength and hardness, with some reduction of ductility [Irvine 1966]. Alloy carbides can be formed during tempering and precipitation of alloy carbides in quenched (martensitic) structures can result in secondary hardening. In alloy steels, alloy carbides can be produced during tempering in the range 500 to 600°C [Semiatin and Stutz 1986], and the carbides can produce a significant or a slight hardening effect over a range of tempering time.

In the present work, the two weld metals deposited at 20°C preheat were tempered at 550°C for various periods of time. A secondary hardening peak was observed after tempering for 0.5 hour (Figure 6.58) and alloy carbide precipitation was confirmed by TEM (Figures 6.63 and 6.64). Both weld metals contained strong carbide forming
Figure 7.12 Incremental hardness $\Delta H$ above fully tempered level (72 hours) of the two weld metals (deposited at 20°C preheat and then tempered at 550°C) as a function of tempering time
elements such as vanadium, titanium and niobium. A tempering temperature of 550°C is high enough to allow sufficient diffusion of the carbide forming elements, and therefore fine alloy carbides, such as VC, TiC and NbC, can be precipitated during tempering. The diffusion of the carbide forming elements is slower than carbon diffusion [Metals Handbook 1990], and thus the alloy carbides persist as a fine dispersion of coherent / semi-coherent particles and resist coarsening. The difference between the atomic volume of fcc (vanadium, titanium, niobium) carbides and the bcc ferrite matrix is relatively high, leading to elastic distortion of the ferrite lattice and the precipitates (coherency strain in the ferrite matrix [Irvine and Pickering 1960]). Therefore, the alloy carbides tend to counteract the normal loss of hardness due to tempering, resulting in secondary hardening.

If it is assumed that the weld metals tempered at 550°C for 72 hours were fully recovered, the incremental hardness $\Delta H$ above the fully tempered level for the two weld metals as a function of tempering time can be estimated, as shown in Figure 7.12. $\Delta H$ values of weld metal deposited with 5%CO$_2$ shielding gas were larger than for 20%CO$_2$ shielding gas, indicating that more solute was retained in solid solution during cooling, and more alloy carbides were formed during tempering, thus producing stronger secondary hardening. This result reinforces the conclusion that the hardenability of weld metal deposited with 5%CO$_2$ shielding gas was higher than for 20%CO$_2$ shielding gas, thus contributing to a higher susceptibility to transverse cold cracking.

7.9 DETAILED COMPARISON OF THE TWO WELD METALS

Susceptibility to transverse hydrogen induced cold cracking depends on microstructure, hydrogen content and stress concentration. In the comparative G-BOP tests conducted the stress concentration at the root of the gap is essentially constant, so the other factors will determine the cracking susceptibility.
Figure 7.13 Schematic diagram showing estimated contributions of ferrite grain size, dislocation density, precipitation and solid solution to hardness (strengthening) in the two weld metals after (a) welding with no preheat; (b) welding with no preheat and tempering at 550°C for 0.5h; and (c) welding with a preheat of 100°C.
7.9.1 Microstructural (Property) Differences

In comparison with the weld metal deposited with 20%CO₂ shielding gas, the weld metal deposited with 5%CO₂ shielding gas had more hardening elements, such as carbon, boron, manganese and silicon in solid solution, and it had higher carbon equivalent parameters. Therefore, this weld metal was higher in hardness than that with 20%CO₂ at the same preheat temperature. The higher hardness value of the weld metal was a major factor contributing to a higher susceptibility to transverse cold cracking in weld metal deposited with 5%CO₂ shielding gas. Due primarily to the lower transformation temperature range, the weld metal deposited with 5%CO₂ shielding gas had a fine microstructure with a higher proportion of acicular ferrite compared with the weld metal deposited with 20%CO₂ (see Figures 6.9 and 6.10). It has been proposed that a microstructure which contains a high proportion of acicular ferrite promotes weld metal strength and toughness [Taylor and Farrar 1975; Zhang and Farrar 1997] and should contribute to a less crack susceptible weld metal. However, an increase in hardness can offset the positive influence of fine acicular ferrite structure on the mechanical properties of the weld metal [Svensson and Gretoft 1990]. It was found in the present work (see Figure 7.4) that the critical hardness value for crack-free welds was about 290 HV0.2 for both shielding gas mixtures, and the weld metal deposited with 5%CO₂ shielding gas had a higher crack resistance at a similar hardness level than that deposited with 20%CO₂ shielding gas. Comparing the two steel weld metals, it can be concluded that if the hardness value can be controlled to about 290 HV0.2 or less (by using preheating), the weld metal deposited with 5%CO₂ shielding gas should have better mechanical properties (higher strength and good toughness) and should be less susceptible to transverse cold cracking because of a more refined microstructure.

A summary of the probable strengthening (and hardening) contributions of various microstructural factors for the two weld metals after different heat treatment conditions is shown schematically in Figure 7.13. Hardness data used to scale Figure 7.13 are
given in Table 7.3, and the basis of the proposed strengthening contributions are as follows.

Table 7.3 Average hardness values (HV0.2) of the two steel weld metals after the indicated heat treatment

<table>
<thead>
<tr>
<th>Heat Treatment</th>
<th>5% CO₂ Shielding Gas</th>
<th>20% CO₂ Shielding Gas</th>
</tr>
</thead>
<tbody>
<tr>
<td>As welded (20°C preheat)</td>
<td>330</td>
<td>292</td>
</tr>
<tr>
<td>As welded (20°C preheat) then tempered at 550°C for 0.5h</td>
<td>295</td>
<td>255</td>
</tr>
<tr>
<td>As welded (100°C preheat)</td>
<td>291</td>
<td>287</td>
</tr>
</tbody>
</table>

Weld metal deposited with 5% CO₂ shielding gas and no preheat is expected to show a strong grain size strengthening effect, comparatively high solid solution strengthening and significant dislocation strengthening because of the low transformation temperature. In contrast, precipitation strengthening will initially be low because of the low transformation temperature. The weld metal deposited with 20% CO₂ was structurally coarser so the grain size effect is expected to be lower as is dislocation strengthening. The higher transformation temperatures allow more complete precipitation on cooling with a complementary decrease in solid solution strengthening.

The superposition of a tempering treatment (550°C, 0.5 hr) resulted in an overall strength decrease which was retarded by secondary hardening (Figure 6.58). This treatment is expected to result in loss of dislocation strengthening as well as enhanced precipitation at the expense of solid solution strengthening with little change in grain size. The grain size effect serves to substantially maintain the hardness (strength) differential (38 to 40 points HV0.2) between the weld metals deposited with the two shielding gases, Figure 7.13(b).
For a preheat of 100°C instead of no preheat, the hardness and strength levels of the two types of weld metals were quite similar (see Table 7.3) despite the different compositions and hardenabilities. The grain size differential remained and a similar trend is expected in dislocation strengthening because of the hardenability difference, leading to a larger dislocation strengthening effect in the weld metal deposited with 5%CO₂. However, a greater precipitation effect is expected for the weld metal deposited with 20%CO₂ because of the higher transformation temperature, partially reducing the strength differential [see Figure 7.13(c)].

7.9.2 Diffusible Hydrogen Concentration

Hydrogen measurements in the all weld test samples showed a higher concentration (8.6 ml/100g) in the weld metal deposited with 5%CO₂ shielding gas compared to the weld metal deposited with 20%CO₂ (6.4 ml/100g). This difference can be related directly to the higher oxygen potential of the 20%CO₂ shielding gas which removes hydrogen from the plasma and the molten pool. Although hydrogen was not measured in the diluted G-BOP welds, a similar ratio is assumed for welding without preheat, increasing the likelihood of hydrogen induced cold cracking in the weld metal deposited with 5%CO₂. With increasing preheat the diffusible hydrogen content should decrease because of effusion and the differential between the concentrations in the two weld metals is likely to decrease. Diffusible hydrogen will be a significant factor in transverse hydrogen induced cold cracking, but for a preheat of 100°C the concentration would be expected to be relatively low (2-3 ml/100g) and similar for both weld metals.
CHAPTER 8

CONCLUSIONS
The self restrained G-BOP test was successfully employed in this research work to evaluate the effect of shielding gas mixture and preheat temperature on the susceptibility of the quenched and tempered steel BIS812EMA to weld metal transverse cold cracking for a nominal heat input of 1.7 kJ/mm. Detailed investigations have been carried out on the fracture morphologies of the G-BOP test samples and the phase transformation behaviour, microstructure, non-metallic inclusion distributions and the hardness of the weld metals. The following conclusions can be drawn.

1. The susceptibility to cold cracking was remarkably reduced by employing argon with 20%CO₂ shielding gas, instead of argon with 5%CO₂ shielding gas, at the same preheat temperature and welding conditions, indicating that the shielding gas mixture has a significant influence on weld metal transverse cold cracking behaviour.

2. Argon with 20%CO₂ shielding gas has higher oxygen potential than argon with 5%CO₂ shielding gas, resulting in loss of hardenability elements such as carbon, boron, manganese, silicon and nitrogen due to oxidation; as well as a higher non-metallic inclusion volume fraction and slightly larger mean particle size and number of particles per unit volume. These changes lead to phase transformation during cooling at higher temperatures and produce a steel weld metal of significantly lower hardness. The higher oxygen potential of 20%CO₂ shielding gas also increases the arc energy, improves the wetting conditions, and results in a larger bead volume and dilution.

3. Shielding gas influences the hydrogen and nitrogen contents in the weld metal and thus the susceptibility to cold cracking. Compositional analysis showed that the weld metal obtained using the shielding gas with the higher oxidizing potential, Ar-20%CO₂, had a lower diffusible hydrogen level than that for Ar-5%CO₂ shielding gas. As well as an effect due to oxidizing potential, the higher CO₂ shielding gas mixture would be expected to result in a higher arc efficiency and thus a lower cooling rate. Effusion of hydrogen from the weld metal is thereby promoted, and the cold cracking resistance is increased.
In addition, the weld metal deposited with 20\%CO₂ had less nitrogen in solid solution, which would also be expected to favour to a lower susceptibility to cold cracking than the weld metal deposited with 5\%CO₂ under the same conditions.

4. The susceptibility to cold cracking decreases strongly with increasing preheat temperature. Crack-free welding was obtained at a preheat temperature of 40°C using Ar-20\%CO₂ shielding gas, and 100°C using Ar-5\%CO₂ shielding gas. Therefore, preheat treatment is an important method for preventing weld metal transverse cold cracking.

5. The preheat treatment affects the welding thermal cycle, especially the cooling rate. It was evident that cooling rate decreased with increasing preheat temperature, suppressing the formation of a cold-crack susceptible microstructure and allowing hydrogen more time to diffuse out of the weld metal. However, the preheat temperature range investigated, from 20 to 100°C, is not high enough to significantly influence the microstructure of the weld metal, although the average hardness values of the weld metal decreased with increasing preheat temperature due to sub- (optical) microscopic changes such as dislocation recovery and precipitate coarsening. Preheat also allows more effective relaxation of stresses developed because of thermal contraction and phase transformation.

6. The microstructures of both weld metals consisted mainly of acicular ferrite and bainitic ferrite. Grain boundary ferrite with Widmänstatten ferrite side plates was also observed in some case. The weld metal produced using Ar with 20\%CO₂ shielding gas contained more bainitic ferrite. The steel weld metal deposited with 5\%CO₂ shielding gas had a finer microstructure with a higher proportion of acicular ferrite.

7. Weld metal transverse cold cracking in the G-BOP test always initiates from the root of the steel weld bead. The fracture surfaces of the cracks in the present case consisted of three fracture modes: microvoid coalescence (MVC) fracture, quasi-cleavage (QC)
fracture and intergranular (IG) fracture. At 20°C preheat, the steel weld metal deposited with Ar-20%CO₂ shielding gas was characterised by a higher proportion of MVC fracture, and IG fracture was more prevalent in the weld metal deposited using Ar-5%CO₂, consistent with the trend expected on the basis weld metal strength. The crack path was mainly intercolumnar or intercellular, and segments of transcolumnar fracture were also present, giving a step-like crack appearance.

8. The hardness value of weld metal deposited with 5%CO₂ shielding gas was significantly higher than that using 20%CO₂ under all equivalent welding conditions and this difference influenced the fracture characteristics, promoting IG fracture instead of MVC at lower preheat. However, at higher preheat (80°C), the hardness value markedly decreased and the favoured fracture mode changed from IG to MVC in the weld metal deposited with 5%CO₂.

9. The peak of the size distribution histograms for the non-metallic inclusions in weld metals typically fell between 0.2 and 0.5 microns. The non-metallic inclusions in the weld metals mainly consisted of complex oxides of manganese, titanium, silicon and aluminium.

10. The CCT diagrams indicate that the γ to α transformation temperature ranges for the two weld metals and the base metal were significantly different: 20%CO₂ shielding gas: 603 and 443°C; 5%CO₂ shielding gas: 538 and 338°C; and BIS812EMA steel: 437 and 299°C. These results are consistent with the trends in CEm, Pcm and Pcmo.

11. The G-BOP tests indicate that argon with 20%CO₂ shielding gas is a satisfactory gas mixture to use in the flux cored arc welding process to prevent transverse cold cracking of weld metal deposited on BIS812EMA steel, provided a minimum preheat of 40°C is used. The greater resistance to transverse cold cracking of weld metal deposited
with Ar-20%CO₂ shielding gas compared to Ar-5%CO₂ shielding gas is due to the combined effects of shielding gas mixture and preheat temperature on the phase transformation behaviour, the microstructure and its hardness, the characteristics of the non-metallic inclusions in the weld metal, and the fracture properties.

12. The weld metal deposited with 5%CO₂ shielding gas had a finer microstructure with a higher proportion of acicular ferrite. Therefore, it can be expected that, as observed, the weld metal deposited with 5%CO₂ shielding gas will have better transverse cold cracking resistance than that using 20%CO₂ at the same hardness level. However, a higher preheat is required to allow more hydrogen effusion from the weld metal and to control the hardness to a value below the critical level (about 290 HV).

13. No single factor was responsible for the higher transverse cold cracking resistance of the weld metal deposited with Ar-20%CO₂ shielding gas mixture, but rather a series of factors contributed to this result. The superior performance over the weld metal deposited with Ar-5%CO₂ shielding gas mixture for the same welding conditions is due, at a process level, to a higher shielding gas oxidising potential, a higher effective heat input and a larger weld bead. These effects result in less highly alloyed weld metal, a higher volume fraction of non-metallic inclusions and a slower cooling rate. The structural consequences are transformation of austenite at a higher temperature to coarser and lower hardness products, reduced solute hydrogen and nitrogen contents and lower levels of the solid solution strengtheners, carbon, boron, manganese and silicon. The opposite trends apply to the weld metal deposited with Ar-5%CO₂, but the transformation of austenite at lower temperature (due to the higher hardenability/carbon equivalent) produces a finer structure with a greater proportion of acicular ferrite. At a similar hardness level, this structure showed a higher resistance to transverse cold cracking than weld metal deposited with 20%CO₂ shielding gas.

Although some of these conclusions are general, conclusions about cracking susceptibility apply to the specific G-BOP test performance of welds produced under
the defined flux cored arc welding conditions. The relative susceptibility to transverse hydrogen induced cold cracking was established, but since the G-BOP test is based on more severe stress concentration conditions than are likely to occur in practice, the results do not offer direct guidance to the likelihood of transverse cracking under practical welding conditions.
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APPENDICES
APPENDIX I Plots of the Thermal Responses for Welding Using Ar-20%CO₂ shielding gas mixture at Different Preheat Temperatures in G-BOP Test

Sample. 20-01

Heat Input: 1.722kj/mm
TPeak : 1015.35°C
Preheat Temp.: 20°C
Δt8/5 : 8.028Sec.

Sample. 20-04

Heat Input: 1.682kj/mm
TPeak : 1117.64°C
Preheat Temp.: 40°C
Δt8/5 : 8.328Sec.
Sample. 20-05

Heat Input: 1.403 kJ/mm
TPeak: 1428.38°C
Preheat Temp.: 40°C
Δt8/5: 8.004 Sec.

Sample. 20-06

Heat Input: 1.538 kJ/mm
TPeak: 1205.74°C
Preheat Temp.: 60°C
Δt8/5: 8.612 Sec.
Sample. 20-07
Heat Input: 1.52kj/mm
$T_{\text{Peak}}$: 1178.4°C
Preheat Temp.: 60°C
$\Delta t/5$: 8.40Sec.

Sample. 20-08
Heat Input: 1.562kj/mm
$T_{\text{Peak}}$: 1292.62°C
Preheat Temp.: 80°C
$\Delta t/5$: 7.773Sec.
Sample. 20-09

Heat Input: 1.565 kJ/mm

Peak: 1277.71°C

Preheat Temp.: 100°C

Δt/5: 9.75 Sec.
APPENDIX II Typical EDS Spectra for Non-metallic Inclusions in the Two Weld Metals Deposited at Different Preheat Temperatures

1. Shielding gas mixture: 5%CO₂ + Ar, Preheat temperature: 20°C
2. Shielding gas mixture: 5% CO₂ + Ar. Preheat temperature: 40°C
3. Shielding gas mixture: 5% CO₂ + Ar. Preheat temperature: 60°C
4. Shielding gas mixture: 5\% \text{CO}_2 + \text{Ar}. Preheat temperature: 80^\circ\text{C}

Full Scale = 1290

Full Scale = 1141

Full Scale = 1982

Full Scale = 1340
5. Shielding gas mixture: 20%CO₂ + Ar. Preheat temperature: 20°C
6. Shielding gas mixture: 20% CO\textsubscript{2} + Ar. Preheat temperature: 40°C
7. Shielding gas mixture: 20% CO₂ + Ar. Preheat temperature: 60°C


