Heat affected zone structure and properties of a welded copper bearing HSLA steel

Xiaping Lin

University of Wollongong
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HEAT AFFECTED ZONE STRUCTURE AND PROPERTIES OF A WELDED COPPER BEARING HSLA STEEL

A thesis submitted in fulfilment of the requirements for the award of the degree of

DOCTOR OF PHILOSOPHY

from

THE UNIVERSITY OF WOLLONGONG

by

XIAPING LIN, B. Sc., M. Sc.

DEPARTMENT OF MATERIALS ENGINEERING

1991
DECLARATION

The work submitted in this thesis has been carried out by the candidate whilst enrolled as a full-time postgraduate student at the Department of Materials Engineering of the University of Wollongong. The results obtained from this study and the conclusions drawn are those of the candidate, except where otherwise stated.

The work contained in this thesis has not been submitted for a degree to any other university or similar institution.

XIAPING LIN
ACKNOWLEDGEMENTS

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setting up and modifying the weld simulator equipment, and assistance with the weld simulation experiments, as well as for his encouragement and suggestions throughout the present work.

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The research work was sponsored by the Australian Welding Research Association (AWRA) and by the Australian Government Industry Research and Development Board via a research grant for a project on 'High Productivity Welding'.
ABSTRACT

A weld joint produced by a fusion welding process consists of the weld metal and a heat affected zone (HAZ). The macro- and microstructures and mechanical properties of these regions determine the weldability of a material. Although the composition of the weld metal can be varied by the choice of filler metal and the extent of dilution, the HAZ has a composition which is essentially the same as the base plate and is an identifiable region because of the structural changes induced by the weld thermal cycle. The HAZ is important because of its potential to develop structures which adversely affect the properties of the joint.

In the present work, the structures and properties of the HAZs produced by various welding processes have been investigated for a commercial structural plate steel. The structure and property gradients across the HAZ have been examined to determine the critical region of the HAZ which governs the properties of the whole HAZ. The effects on the HAZ of weld process type (bead-on-plate submerged arc [BOP SA], bead-on-plate flux cored arc [BOP FCA], and four wire submerged arc [4 wire SA]); and the welding parameters of heat input, welding speed, multi-passes and postweld heat treatment have been analysed in order to assess the weldability of the steel. The mechanical properties of toughness, tensile properties and hardness are of major concern in the present investigation.

The steel investigated was a low carbon, copper-bearing, precipitation hardening steel (HSLA 80) which has been recently developed by BHP Steel, SPPD, Port Kembla, Australia. It is based on a modified ASTM A710 steel chemical composition and is produced by a thermomechanical control rolling process (TMCP). The high yield strength of this steel (80ksi or 550Mpa for plate ≤25mm) is achieved by copper precipitation hardening through an aging heat treatment at 550°C for 1/2 hour after rolling. The reason
for the development of this steel was to produce an 80ksi grade steel which can be welded more easily and can be more economically produced than quenched and tempered HY 80 type steels and thus can be qualified as a replacement in various structural applications. The weldability of this new steel is thus of critical importance in proving its advantages over HY 80.

This thesis reports a detailed investigation of the structure and properties of the weld HAZ, which is widely regarded as a critical region in terms of the weldability of a steel.

As a result of the microstructural gradient which develops across the HAZ, it is difficult to carry out Charpy impact toughness testing on a particular microstructural region of the HAZ. In order to facilitate mechanical testing, especially impact testing, thermal simulation experiments have been conducted to reproduce in "bulk form" structures similar to those of different sub-regions of the actual HAZ. A comparison has been made of the results obtained from actual and simulated HAZs.

The effect of multi-pass welding and postweld heat treatment on the HAZ structure has been simulated to assess the response of the HAZ to a series of thermal cycles. A partial $\gamma \rightarrow \alpha$ continuous cooling transformation diagram for the grain coarsened HAZ region under weld thermal cycle conditions was also obtained by analysing the cooling curves associated with thermal cycles simulating those experienced in the HAZ during welding under different conditions.

The microstructure of the grain coarsened HAZ region for both actual and simulated welds generally consisted of ferrite in the form of grain boundary allotriomorphs, Widmanstatten sideplates and laths, together with martensite-austenite (MA) islands. The dominant constituents were lath ferrite and MA islands. Low carbon lath martensite was also found in the HAZ of some low heat input welds, particularly BOP FCA welds.
A general problem in welding precipitation hardening steels is that the HAZ thermal cycle can destroy the precipitation hardening and reduce the hardness locally to below the level of the base plate. Such a softened HAZ was observed in the present steel for the BOP SA, BOP FCA and the 4 wire SA welds; as well as for the simulated HAZs. The loss of precipitation hardening was found to be due to solution of copper on re-austenitising and the resulting supersaturation of ferrite on cooling. For intercritical heating, a significant part of the softening was due to rapid overaging of copper precipitate particles in the untransformed ferrite.

It was found that the HSLA 80 steel showed a good overall toughness in the HAZ for the welding conditions investigated. The toughness of this type of steel in the hot rolled and aged condition is due to its low carbon content (0.055%) and a fine grained structure. In addition, the low carbon equivalent (0.41), relative to the strength, ensured that HAZ toughness generally exceeded the minimum requirements for HY 80 and was similar to that of the base plate. Of the various HAZ sub-zones, the grain coarsened region (GCHAZ) near the fusion line exhibited the lowest toughness and highest hardness values and, therefore, this region is likely to govern the overall HAZ toughness. The heat input did not appear to have a major effect on HAZ toughness, despite the observation that HAZ structural refinement and an increase of HAZ hardness occurred with decreasing heat input.

It was established by simulated multiple weld thermal cycles that multi-pass welding generally refines the HAZ structure and improves the toughness of the HAZ. However, it was found that a second weld thermal cycle to a subcritical peak temperature, consistent with a high heat input, could markedly increase the hardness of an original grain coarsened HAZ region produced by a low heat input, because of precipitation of copper from supersaturated ferrite. This combination of thermal cycles appears to have the potential to reduce the toughness in this local region.
Strengths similar to that of the base plate were obtained from transverse tensile tests on weld joints produced at heat inputs of 2.5 and 5kJ/mm by 4 wire submerged arc welding, despite the softening which occurred in the HAZ. However, for a high heat input of 10kJ/mm, significant degradation of weld strength occurred because of the wide softened HAZ. Varying the welding speed of 4 wire SAW showed little effect on HAZ structure, toughness, hardness and tensile properties.

Postweld heat treatment of the GCHAZ region at 550°C for 1 hour significantly reduced its toughness. This embrittlement was attributed to precipitation hardening by copper which resulted in a considerable increase in hardness. However, postweld heat treatment at 450°C and 650°C were found to improve the toughness and reduce the hardness of the GCHAZ.

Investigation of the γ→α transformation temperature of the grain coarsened HAZ region under simulated welding conditions showed that during the cooling part of a weld thermal cycle, austenite begins to transform at temperature between 600-650°C for the equivalent heat input range of 1.9-4.9kJ/mm. A lower transformation temperature was associated with a lower heat input.

Comparison with a reference steel indicated that the copper and nickel additions to the HSLA 80 suppressed the HAZ transformation temperature. The major associated microstructural change was the predominance of a nondiffusional second constituent (martensite-austenite islands) rather than a diffusional one (pearlite and/or bainite).

The research investigation makes two main contributions to knowledge in the field of the physical metallurgy of ferrous alloy welding. The first is the provision of detailed data on the structure and properties of the HAZ of a modified A710 type precipitation hardening steel for welding by flux cored arc and submerged arc processes under various
conditions. This characterisation of the structure and properties has allowed definition of welding conditions leading to satisfactory strength and toughness in the HAZ.

The second contribution is a general finding concerning limitations of the Rosenthal analysis of heat transfer during welding which is based on the assumption of a moving point heat source. The implication of this analysis, and a widely accepted view, is that a constant heat input dictates a constant HAZ cooling rate and hence structure. However, structure and properties have been observed to vary in a small but significant way with position around the fusion line of a single weld bead at a given heat input and between welding processes at the same nominal heat input. In both cases, variations in weld bead shape affect the local heat transfer conditions and hence the cooling rate.
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CHAPTER 1

INTRODUCTION
Materials for naval and offshore constructions must meet a variety of requirements principally associated with loadings, environment and life-cycle maintenance. The fracture safety of these structures is addressed mainly through the use of tough structural alloys. The current steels available for construction of naval vessels are mainly HTS, HY 80 and HY 100. The high tensile steel (HTS) is a carbon-manganese steel with 52 ksi minimum yield strength. The HY 80 and HY 100 are high yield steels with minimum yield strength of 80 and 100 ksi respectively.

Development of HY 80 steel began after World War II as a high-strength steel to replace the HTS steel (1). The HY 80 steel was actually developed from a 1894-vintage Krupp armour steel (2) that relied on nickel and chromium for strength and toughness. The toughness and strength achieved in the HY-series steels is obtained with the use of relatively high amounts of alloying elements combined with a quench and temper (Q&T) heat treatment.

The HY-series steels have martensitic structure, which usually shows high hardenability and low crack resistance. Welding of these steels thus requires the use of stringent welding process controls and specially designed filler materials to retain adequate properties in the as-welded condition. Additionally, preheat prior to welding and interpass temperature control is needed when using these steels to avoid and minimize cracking in weldments. These controls and requirements are still in place today. Unfortunately, they increase the cost of welded structure considerably.

To reduce cost, an easily weldable steel with similar strength and toughness properties was needed to replace the HY80 steel.

High strength low alloy (HSLA) C-Mn steels with carbon levels below 0.15 per cent and fine grained ferritic microstructures were considered by the U.S. Navy as the best candidate steels to replace HY80 steel (1). The use of microalloying and
thermomechanical processing results in fine grain size, reduced C level and therefore, enhanced strength and toughness. HSLA steels can be produced with similar strength and toughness to Q&T steels and inherently good weldability can be achieved due to their low carbon contents. The U.S. Navy reported (1) that estimates of the reduction in cost of welded ship structures are in the range of $0.4 to $0.9 per pound, which projects to a total savings of $0.5 to $2 billion over the next two decades.

Of the commercially available HSLA steels, ASTM A710 Grade A steel, which was originally used by the offshore industry, immediately meets the property goals without requiring any alloy development or modification. The successful certification program of 80 ksi HSLA steel, based on the A710 steel system occurred in early 1984 to replace the HY 80 steel in surface ship hull structural application (1). In the last few years, application of HSLA 80 steel has been increased as a substitute for HY80 in cruiser deck, bulkhead and hull applications (3).

The chemical composition of A710 steel Grade A is listed in Table 1.1. High yield strength of this steel is attributed to Cu precipitation hardening in a ferrite matrix while still retaining good toughness. Because of the low carbon content and ferritic structure, this steel is extremely weldable without the use of preheat and many stringent process controls required for HY80. In addition, the presence of about 1% Cu contributes to good formability (4) and excellent corrosion resistance (5). Furthermore, A710 steel has a reported high resistance to fatigue crack growth (6).

The ASTM A710 Grade A steel can be produced in three classes (shown in Table 1.1). In each case, precipitation hardening is used to achieve the required strength level. With the minimum required yield strength beyond 80 ksi (550 MPa), only classes 1 and 3 are relevant in developing the HSLA 80 steel. Although yield strengths of both class 1 and class 3 are above 80 ksi, it is reported (1) that the Charpy toughness of class 1 plate with 16 and 19mm thickness is well below that of the class 3 and does not meet the required
Table 1.1 Chemical composition range and processing conditions for A710 Grade A steel (ASTM Designation: A710)(ref.7)

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Ni</th>
</tr>
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<tbody>
<tr>
<td>≤0.07</td>
<td>0.4-0.7</td>
<td>≤0.4</td>
<td>≤0.025</td>
<td>≤0.025</td>
<td>0.7-1.0</td>
</tr>
<tr>
<td>Cu</td>
<td>Cr</td>
<td>Mo</td>
<td>Nb</td>
<td>CEw</td>
<td></td>
</tr>
<tr>
<td>1.0-1.3</td>
<td>0.6-0.9</td>
<td>0.15-0.25</td>
<td>≥0.02</td>
<td></td>
<td>0.4-0.57</td>
</tr>
</tbody>
</table>

Grade and Class | Conditions
--- | ---
Grade A, Class 1 | as-rolled and precipitation heat treated
Grade A, Class 2 | normalized and precipitation heat treated
Grade A, Class 3 | quenched and precipitation heat treated
high and low-temperature toughness goal (48J at -84°C and 81J at -18°C) for 80 ksi steel for surface ship construction. Therefore, the HSLA 80 steel developed in recent years is produced mainly through the quenching and aging (Q&A) route, corresponding to the ASTM A710, class 3 steel.

Research work has demonstrated (3,8) that at least equivalent performance with the HY 80 steel system can be achieved for the Q&A HSLA 80 steel. The low carbon content (≤0.07%) increases the weldability and toughness properties, as shown by many researchers (8,9). Currently, the quench and aged HSLA 80 steel is finding increasing applications in the ship building industry (3).

The HSLA 80 steel can also be produced by controlled-rolling (CR) process with a modified A710 chemical composition. The CR process takes advantage of the recrystallization kinetics of the steel resulting from both deformation during rolling and microalloying. By using vanadium or niobium carbonitrides to suppress austenite grain growth while retarding austenite recrystallization during finishing, a fine ferrite gain size can be achieved through thermomechanically controlled processing (TMCP). Enhanced strength and toughness can be obtained as a result of fine grain size. It is claimed (1) that the controlled-rolled steels usually have better low temperature toughness and weldability strength than those produced by the quenched and aged processing route. However, unless specially designed plate mill facilities are used, it is difficult to produce the controlled-rolled plate with thickness over 13mm whilst retaining the high yield strength. The thickness limitation is based on plate mill capacity to introduce sufficient strain at low rolling temperatures.

More recently, following the establishment of strict TMCP practices for thick plate (10, 11), a Cu age-hardened 80 ksi yield strength grade steel has been developed (12) by BHP Steel, Australia, based on a modified A710 composition and utilizing TMCP technology.
Although ASTM A710 includes an as-rolled and aged version (class 1), the alloy design is more suitable for the quench and aging process route (class 3), due to the high quench hardenability provided by the Ni, Cr and Mo additions. As a result, class 1 (control rolled) A710 steels exhibit mechanical properties inferior to class 3 (Q&A) A710 steel as mentioned previously.

To facilitate the TMCP and aging process route, a modified leaner alloy design was developed by BHP and used for modification of A710 steel. Chemical compositions of both conventional and modified A710 are listed in Table 1.2. The latter steel can be produced more cheaply because fewer expensive alloying elements are used.

Yield strengths in excess of 80 ksi (550 MPa) can be achieved for the thinner plate (≤25mm) and 73 ksi (500 MPa) for thicker plate (>25mm). The thinner plate can therefore be classified as HSLA 80 steel or more appropriately as CR HSLA 80 to differentiate it from Q&A HSLA 80 steel.

CR HSLA 80 plates of various thicknesses exhibit excellent Charpy toughness (12). Figure 1.1 shows the Partial Charpy toughness transition curves for 20mm plate. The circular symbol shows the high-and low-Charpy energy requirement for certification of HSLA 80 steel for surface ship construction (1). It shows that a Charpy energy comfortably in excess of the requirement can be achieved for the 20mm plate.

As a newly developed steel, little work has been reported on the weldability of the CR HSLA 80 steel. Although researchers have demonstrated (13,14,15,16) that excellent weldability can be achieved for the Q&A A710 steel, investigation of weldability is still necessary for the CR HSLA 80 steel, due to the difference in chemical composition and the thermomechanical processing for the two types of steels. Furthermore, weld joint mechanical properties and microstructures under different welding conditions are also needed to understand the effect of welding conditions on the quality of the weld joint.
Table 1.2  Comparison of chemical compositions of (A) conventional A710 steel and (B) the modified A710 (CR HSLA 80) steel (refs.7 and 12)

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>≤0.07</td>
<td>0.4-0.7</td>
<td>≤0.4</td>
<td>≤0.025</td>
<td>≤0.025</td>
<td>0.7-1.0</td>
</tr>
<tr>
<td>B</td>
<td>0.05</td>
<td>1.4</td>
<td>0.25</td>
<td>0.012</td>
<td>0.003</td>
<td>0.85</td>
</tr>
<tr>
<td>Steel</td>
<td>Cu</td>
<td>Cr</td>
<td>Mo</td>
<td>Nb</td>
<td>Ti</td>
<td>CEnw</td>
</tr>
<tr>
<td>A</td>
<td>1.0-1.3</td>
<td>0.6-0.9</td>
<td>0.15-0.25</td>
<td>≥0.02</td>
<td>-</td>
<td>0.4-0.57</td>
</tr>
<tr>
<td>B</td>
<td>1.1</td>
<td>-</td>
<td>-</td>
<td>0.02</td>
<td>0.013</td>
<td>0.41</td>
</tr>
</tbody>
</table>
Fig. 1.1 Charpy V-notch impact toughness versus test temperature for 20mm CR HSLA 80 steel plate (Data from BHP SPPD). The circular symbols show the high- and low-energy requirements for certification of HSLA 80 steel for surface ship construction (ref.1).
The weldability of steel is a complex property since it covers both the sensitivity to weld cracking and the toughness required by service conditions and test temperatures. Easterling (17) defined "good weldability" as a function of several factors, which include:

1. type of welding process;
2. environment;
3. alloy composition; and
4. joint design and size.

Being decisive to the steel's weldability, unsuitability of any of these factors may cause cracking problems. In simple terms, weldability is often defined as susceptibility of steel to various types of cracking problems associated with welds. For a fixed environment and joint configuration, the weldability of a particular steel (fixed alloy composition) is only dependent on the type of welding process.

A weld joint consists of weld metal and heat affected zone (HAZ). For the fusion welding process, the weld bead is the fusion zone where dilution of the deposited metal (wire) by melted base plate occurs. The HAZ is usually referred to the volume of base metal, which is structurally affected by the weld thermal cycle. The present investigation mainly concerns the HAZ of the weld joint.

The important weldability problems associated with the HAZ in arc welding of structural steels are hydrogen cracking (or cold cracking) and HAZ toughness. Hydrogen cracking occurs when all the following factors are present:

1. a sufficient hydrogen concentration;
2. a susceptible microstructure;
3. an applied tensile stress.
For a given level of weld hydrogen and joint restraint, the risk of cracking depends on the microstructure and hardness of the heat-affected zone, as well as its susceptibility.

Formation of unfavorable brittle constituents such as Widmanstatten ferrite sideplate, coarse bainite packets and martensite-austenite islands were reported to result in the embrittlement of HAZ of steel (18,19,20) and to increase the susceptibility of cracking.

It is well known that high hardness in the HAZ is related to the high hardenability and therefore, increasing harness increases the susceptibility to cracking in the weld HAZ. Since 1940, 350 HV has been the limiting hardness value in various welding specifications (21,22,23) to obtain good weldability. Since HAZ hardness is proportional to the carbon equivalent (CE) of the steel, CE must be limited to keep the HAZ hardness below the critical level. Although CE formulae for assessing the risk of hydrogen cracking in the HAZ has been modified over time (24), it is recommended that CE should be kept as low as possible to achieve good weldability.

Fracture toughness is the mechanical property directly reflecting crack resistance and HAZ toughness is the most important mechanical property when weldability is concerned. Because of the critical nature of naval and offshore structures, highlighted by tragic accidents, a thorough evaluation of the toughness behaviour of weldments is required.

HAZ microstructure characterization, hardness measurement and toughness assessment have been carried out in the present work, with the objective of gaining a better understanding of the welding characteristics and the optimum welding conditions for the newly developed CR HSLA 80 steel. The welding process and conditions were varied. Moreover, in order to assess the toughness of the HAZ in detail, it is important to locate region(s) of the HAZ which show a low toughness level. Weld thermal cycle simulation was used to simulated weld HAZ structures and to elucidate property differences
associated with the microstructural gradient across the HAZ.
CHAPTER 2

THE HEAT AFFECTED ZONE
2.1 INTRODUCTION

The HAZ is the volume of base material which has been heated to various peak temperatures and cooled again rapidly during fusion welding. Figure 2.1 shows a schematic diagram of the HAZ structure as a function of the temperature reached. Depending on the peak temperature, the HAZ can be divided into a number of sub-regions (Fig. 2.1). Each sub-region has its own distinct microstructure, and therefore, possesses its own mechanical properties. The structure of the HAZ is determined by the welding conditions, prior thermal and mechanical history and more importantly, the chemical composition of the material.

With the use of welded steel constructions in a large variety of applications, it became apparent that some steels used in the past showed extreme susceptibility to various types of cracking in the HAZ, especially cold cracking. This has been attributed to the formation of a very susceptible HAZ microstructure (25). Thus, it is important to study the microstructure of the HAZ in order to investigate the weldability of steels.

In this chapter, the thermal cycle experienced by the HAZ is discussed in Section 2.2. The solutions of the heat flow equations given by Rosenthal are discussed in this section, providing the correlation between welding heat input and cooling rate of the HAZ, since the cooling rate is one of the main factors determining the HAZ microstructure. The relationship between heat input and HAZ cooling rate also provides the theoretical basis for predicting HAZ microstructure resulting from a particular welding procedure.

The microstructure of the HAZ is discussed in Section 2.3 in terms of discussing the types of structure likely to occur in various sub-regions and the effect of multi-run welding on HAZ microstructure. Since coarse austenite grains significantly enhance the formation of undesirable microstructural constituents, the emphasis of this section is on the control and prediction of austenite grain size in the HAZ.
Fig. 2.1 A schematic diagram of the various sub-regions of the HAZ approximately corresponding to the alloy Co (0.15wt% C) indicated on the Fe-Fe3C phase diagram (ref. 17).
Finally, the literature review on HAZ microstructure simulation techniques and HAZ continuous cooling phase transformation diagrams is presented in Sections 2.4 and 2.5, respectively.

2.2 THE THERMAL CYCLE OF BASE METAL

Arc welding is a process in which a very intense, moving heat source is applied to the workpiece. The weld thermal cycle is applied for very short time and produces high temperatures. It causes very steep temperature gradients, which affect the solidification process in the weld metal and phase transformations in the heat-affected zone (HAZ) of the parent plate.

The weld thermal cycle experienced by a point in the HAZ has been measured (26,27), as well as being theoretically predicted (28). The prediction of the thermal cycle in the HAZ is based on the relationship between the cooling rate of the HAZ and the welding heat input and is derived below following the treatment by Easterling (17).

Rosenthal (29) assumed that the source energy input or heat input (HI)(measured in kJ/mm or MJ/m) moving at a constant speed S, is given by

$$HI = \eta \frac{VI}{S}$$

(2.1)

where V is welding voltage, I is welding current and \(\eta\) is the arc efficiency (\(\eta = 0.9-0.99\) for submerged arc welding).

When assuming the thermal conductivity (\(\lambda\))(Jm\(^{-1}\)S\(^{-1}\)K\(^{-1}\)) and the specific heat (\(\rho\)) x density product (c)(\(\rho c\) - the volume thermal capacity, Jm\(^{-3}\)K\(^{-1}\)) are constants, the
differential equation of heat flow for the co-ordinates in Fig. 2.2 is given by equation (2.2)

\[ \frac{\partial^2 T}{\partial x^2} + \frac{\partial^2 T}{\partial y^2} + \frac{\partial^2 T}{\partial z^2} = 2\alpha \frac{\partial T}{\partial t} \]  

(2.2)

where \( T \) is temperature (K), \( t \) is the time (s) and \( \lambda \) is the thermal conductivity.

For a moving coordinate system, \( x \) can be replaced by \( x' \), the distance between a fixed reference point and the point heat source moving at speed \( S \) after time \( t \) is given by

\[ x' = x - St \]  

(2.3)

Equation (2.2) becomes

\[ \frac{\partial^2 T}{\partial x^2} + \frac{\partial^2 T}{\partial y^2} + \frac{\partial^2 T}{\partial z^2} = 2\lambda S \frac{\partial T}{\partial x} + 2\lambda \frac{\partial T}{\partial t} \]  

(2.4)

In most welding situations, the temperature distribution around a heat source of uniform velocity will be a constant, i.e., if \( HI \) is a constant, \( \partial T / \partial t = 0 \). Therefore, equation (2.4) becomes,

\[ \frac{\partial^2 T}{\partial x^2} + \frac{\partial^2 T}{\partial y^2} + \frac{\partial^2 T}{\partial z^2} = -2\lambda S \frac{\partial T}{\partial x} \]  

(2.5)

Assuming the heat losses through the surface are negligible, for a fixed position in the HAZ at a radial distance \( r \) (\( r = (x^2 + y^2 + z^2)^{1/2} \)) from the heat source, the heat flow is three-dimensional for thick plate, for example, a bead-on-plate weld (Fig. 2.3a). Equation (2.5) becomes can be solved and simplified to produce

\[ T - T_0 = \frac{HI}{2\pi\lambda t} \exp\left(-\frac{r^2}{4at}\right) \]  

(2.6)
Fig. 2.2 Welding configuration in terms of a point heat source (q) and a constant velocity (S)(ref.17).

Fig. 2.3 Three- (a) and two-(b) dimensional heat flow in welding (ref.17).
and for thin plates (two-dimensional heat flow, Fig. 2.3b),

\[ T - T_0 = \frac{HI}{d(4\pi \lambda \rho c t)^{1/2}} \exp\left(-\frac{r^2}{4at}\right) \]  

(2.7)

where \( a (= \lambda/pc \text{ m}^2 \text{ s}^{-1}) \) is the thermal diffusivity and \( d \) is the plate thickness. \( T_0 \) (in K) refers to the initial temperature prior to welding. The above equations are obtained based on the assumption that the application of heat on the particular point (r) in HAZ is instantaneous.

The temperature range of 800-500°C approximately represents the A3-A1 transformation temperature range. It is claimed (17) that, the cooling time through this temperature range \((\Delta t_{8-5})\) is constant within the HAZ for a given welding process, weld geometry and material. \( \Delta t_{8-5} \) has been widely used to characterise the effect of a weld on the HAZ.

To calculate the cooling time \( \Delta t_{8-5} \), for thick plate, equation (2.6) is simplified to equation (2.8):

\[ T_p - T_0 = \left(\frac{2}{\pi e}\right)\frac{HI}{\rho c t^2} \]  

(2.8)

where \( T_p \) refers to the peak temperature of the thermal cycle and \( e \) is the base of natural logarithms (=2.718). The cooling time \( \Delta t_{8-5} \), therefore, is given

\[ \Delta t_{8-5} = \frac{HI}{2\pi \lambda \Theta_1} \]  

(2.9)

where \( \Theta_1 \) is defined by equation (2.10)

\[ \frac{1}{\Theta_1} = \left(\frac{1}{773-T_0} - \frac{1}{1073-T_0}\right) \]  

(2.10)
For the thin plate case,

\[ \Delta t_{8-5} = \frac{(HI)^2}{4\pi \lambda \rho_c \Theta_2^2 d^2} \]  

\[ \frac{1}{\Theta_2^2} = \frac{1}{(773-T_0)^2} - \frac{1}{(1073-T_0)^2} \]

The important correlations between cooling time \( \Delta t_{8-5} \) of the HAZ and the welding heat input (HI) are established by equations (2.9) and (2.11). That is, \( \Delta t_{8-5} \) is proportional to heat input for thick plate and to the square of heat input for thin plate. The higher the heat input applied during welding, the longer the cooling time, i.e., the slower the cooling rate of HAZ.

The choice of equation used in a particular case is governed not only by plate thickness, but also by the welding process and even the type of material. The critical thickness, \( d' \), which defines the cross-over or boundary condition between equations (2.9) and (2.11) for thick and thin plates, is obtained by equating both equations (17):

\[ d' = \left\{ \frac{HI}{2\rho_c} \times \left( \frac{1}{773-T_0} + \frac{1}{1073-T_0} \right) \right\}^{1/2} \]

It is seen that the critical thickness \( d' \) is dependent on welding process used (HI) and type of material (\( \rho_c \)).

The Rosenthal equations (2.9) and (2.11) indicate that, regardless of the values of voltage (V), current (I) and welding speed (S), welds deposited with the same value of heat input (HI) will have identical cooling times \( \Delta t_{8-5} \), and therefore, identical HAZ microstructure and properties.
However, Wingrove (30) demonstrated that the HAZ microstructure and hardness are not solely dependent on the final value of heat input, but also depend on the individual values of voltage (V), current (I) and welding speed (S). Since HAZ microstructure is determined by HAZ cooling rate, this implies that the cooling rate of the HAZ can be different at constant heat input, depending on the values of V, I and S which produce the heat input (HI). He claimed that the Rosenthal analysis, based on 'a moving point heat source', is an oversimplification of a complex phenomenon, since the 'pinching off' and travel of molten metal droplets from the tip of electrode are affected by arc temperature, droplet size distribution and rate of transfer of droplets. A similar result was reported by the CSIRO Division of Manufacturing Technology in Adalaide (192) based on cooling rate measurements in 4 wire submerged arc (SA) weld HAZs. It was found that the cooling rate was higher for a higher welding speed at constant heat input.

However, the Rosenthal approach appears to be adequate as a first approximation and if only one of the parameters: voltage, current and welding speed is varied to increase the heat input, it is commonly observed that the higher heat input does result in a slower cooling rate of the HAZ (31).

2.3 HAZ MICROSTRUCTURE

2.3.1 Grain Coarsened Region

The grain coarsened region (GCR or GCHAZ) of the HAZ is the area immediately adjacent to the fusion line (Fig. 2.1). In this region, the base material experiences a thermal cycle with peak temperature beyond the austenite grain coarsening temperature, in the range of 1100-1500°C for most steels. Although the final transformation product of this region is determined by many factors, it is generally a ferritic structure for lower
carbon equivalent (CE) steels and lower temperature transformation products, such as Widmanstatten ferrite sideplates, bainite and martensite for higher CE steels. A typical microstructure of a GCHAZ is shown in Fig.2.4a.

The GCHAZ is regarded as the most dangerous region of the HAZ, because cold cracks are most common in this area. The cold cracking results from the embrittlement by structures, such as Widmanstatten ferrite sideplates, upper bainite, martensite-austenite (MA) islands and twinned martensite. Increase in the austenite grain size effectively enhances the formation of these unfavorable microstructural constituents.

2.3.1.1 Precipitate retardation of austenite grain growth

Since fine austenite grain size in the GCHAZ is important to ensure the desired properties in the HAZ, significant attention has been focused on the refinement of grain size. The most practical way to achieve this goal is to utilize the pinning effect of particles, like carbides and nitrides, during austenite grain growth. To produce these particles, small amounts of strong carbide and nitride forming elements, notably niobium (Nb), titanium (Ti) and vanadium (V) can be added to steels. These microalloying additions play an important role in achieving adequate levels of strength and toughness in the steel as well as in the weld HAZ. However, in practice, the amount of microalloying elements does not always related to finer grain size. The final effect of microalloying elements on retardation of austenite grain growth is quiet complicated and depends on particle solubility, stability and formation temperature, as well as on the composition of the steel.

McCutcheon et al. (32) carried out an investigation on the effect of microalloy additions on the HAZ notch toughness for a C-Mn-Mo line pipe steel. They found that a small Ti addition (0.005 to 0.010 %) significantly restricted austenite grain coarsening in a
Fig. 2.4 Microstructures of various sub-regions of HAZ of a 4 wire Sub. Arc welded 45mm Australian Standard (AS 1204) 350 grade steel (10kJ/mm, 1000mm/min.) (ref.33). (a) grain coarsened region; (b) grain refined region; (c) partially transformed region; (d) unchanged base metal. (160x)
simulated HAZ structure, and, therefore, resulted in a improved HAZ toughness. Nb addition to the Ti steels resulted in further grain refinement in the HAZ.

The beneficial effect of Ti (34,35) on controlling austenite grain size in the weld HAZ and base steel has led to the development of Ti-modified steels. In manufacturing these steels, Nb and/or V are usually added to Ti in order to provide the required degree of grain refinement. Thus, Ti-bearing steels may contain a mixture of various carbonitrides, such as TiN, TiC, VN, AlN, and NbC. During a high heat input weld cycle, only the TiN is likely to survive the peak temperature of the GCHAZ without complete solution, but the coarsening of these precipitates is inevitable (17).

Ti-bearing steels usually have nitrogen concentrations above the stoichiometric level for TiN, to prevent formation of TiC, and are suitably heat treated to yield a high volume fraction of grain boundary pinning TiN precipitates. However, Edwards et al. (36) reported that for steels containing a high level of Al, Ti rather than N concentration greater than the stoichiometric level for TiN give better grain size control near the weld fusion boundaries. The reason suggested was that small TiN particles dissolve at the high temperature reached near the weld fusion line. The excess Ti in solution then appears to minimize austenite grain growth by a mechanism of solute drag.

A suitable size of TiN precipitates is critical in suppressing austenite grain coarsening in the weld HAZ. It is unlikely that large precipitates can play any useful role in refining the grain size. A study by Chen et al. (37) of concast Nb-V, Nb-V-Ti, and Nb-Ti steels indicated that addition of Ti to Nb containing HSLA steels resulted in undesirable precipitation in these steels, i.e. precipitation of large particles with various morphologies. These large particles were found to be stable up to 1150°C and some up to the melting point of the steels.
A detailed review on fundamental aspects of precipitation in microalloyed steel has been given by Honeycombe (38). He concluded that nitrides coarsen more slowly than carbides, so partial substitution of carbon by nitrogen in carbides is beneficial.

Concerning the austenite grain growth through a weld thermal cycle, Ikawa et al. (39) reported that the maximum increase in grain size occurs in the initial stages of grain growth. In fact, it was found that 80% of grain growth occurred during the heating part of thermal cycle for tungsten inert gas (TIG) and submerged arc (SA) welded microalloyed high strength steels. Easterling (17) explained that it is due to the higher driving force available at the initial stage of grain growth.

The rapid grain growth may also related to $\alpha \rightarrow \gamma$ transformation by diffusionless mechanism during heating, i.e., by a martensitic mechanism, as found by Sekino and Mori (40) and Albutt and Garber (41). The authors of both references proved that $\alpha \rightarrow \gamma$ transformation by a diffusionless mechanism could happen during heating rates of 200-300°C/s in the temperature range of 800-1100°C. This transformation produces some homogeneous plastic deformation in the austenite formed (42), which leads to a rapid recrystallization of austenite at high temperature. If the martensitic transformation mechanism operates and austenite recrystallization occurs with an enhanced driving force then, precipitates may not be effective in inhibiting austenite grain growth and reducing the grain size. It was suggested by Rasanen and Tenkula (43) that the only way to eliminate coarse grains would then be by reducing the heating rate in the temperature range 800-1100°C to such an extent that the phase change of $\alpha \rightarrow \gamma$ occurs by the normal diffusional mechanism.
2.3.1.2 Prediction of prior austenite grain size of HAZ

Since fine grain size is an important factor in avoiding HAZ cold cracking, it would clearly be useful if the prior austenite grain size in the HAZ produced by any specific heat input could be predicted. In this way, any potentially dangerous situations could be identified before fabrication proceeded. A number of papers (17,44) have been published on developing these models. The calculations of austenite grain growth in the HAZ were based on elementary kinetic models for grain growth and particle dissolution, integrated over the weld cycle. Generally, a reasonable correlation can be achieved between theoretical prediction of grain growth in a weld thermal cycle and that obtained experimentally using weld simulated specimens.

Although good agreement is expected between predicted grain size and that of weld simulated samples, in applying the equation to weld HAZs, Alberry et al. (28) found that the equation predicted a larger grain size than what is actually observed. This result suggests that an alternative growth-inhibiting mechanism is operating in the actual weld HAZ. He concluded that the steep thermal gradients in the actual HAZ produce 'thermal pinning' which inhibits austenite grain growth. To take this effect into account, he introduced a thermal pinning factor $N$ in the equation to predict the grain size in the weld HAZ.

2.3.2 Grain Refined Region

The temperature range of this region (GRHAZ) is about 900-1100°C. Compared to the grain coarsened region, the lower peak temperature does not allow the austenite grains to coarsen, and the grain size remains very small. Furthermore, the weld thermal cycle may not allow enough time for carbides to be dissolved completely. During cooling, the fine grained austenite and remaining carbides tend to produce a fine grained ferrite plus second phase structure due to the large austenite grain boundary area and carbides acting
as nucleation sites (Fig. 2.4b). The second constituents can be pearlite, bainite or martensite-austenite islands depending on the cooling rate and composition of the steel.

2.3.3 Partially Transformed Region

In the partially transformed region, the temperature range of 750-900°C resulted in partial austenisation. The pearlite regions of the base plate are austenitized due to their lower transformation temperature. The austenite regions formed are enriched in carbon and manganese contents which are higher than the average values of the base material. For this reason, the transformation product of this austenite during cooling can be pearlite, upper bainite, autotempered martensite or high-carbon martensite (Fig. 2.4c), depending on cooling rate.

2.3.4 Tempered Region

This region is also called the spheroidized carbide region. It corresponds approximately to the temperature range of 650-750°C. In this temperature range, a soft annealed structure is formed during welding. Little $\alpha\rightarrow\gamma$ transformation occurs during the rapid heating cycle, so that the most obvious change concerns degradation of the lamellar pearlite to spheroidal particles of Fe$_3$C.

2.3.5 Region of Unchanged Base Material

As this region corresponds to temperature below 650°C, no changes of microstructure can be observed optically in the base material (Fig. 2.4d). However, the welding stresses together with greater solubility of carbon and nitrogen can lead to dynamic strain aging.
The substructures, such as fine precipitates and dislocations, are changed as a result of the aging. This phenomenon can be intensified more by multi-run than by single-run welding (45).

2.3.6 HAZ Microstructure of Multi-pass Welds

In practical welding processes, particularly for welding of the thick plates, multi-pass welding is often required to fill up the joint gaps. Since the heat input is limited to avoid the coarse HAZ structure resulting from high heat input, the volume of deposited filler metal for each pass, which is proportional to the heat input, is also restricted. For the multi-pass weld HAZ, a large proportion of the previous pass HAZ is reheated to a certain extent. Therefore, study of the reheated HAZ microstructure is important in investigating the whole HAZ weld joint. Figure 2.5 shows a schematic diagram of single and multi-run welds.

Except for a small part of the HAZ being subjected to the highest peak temperature during the second weld run, most of the reheated region experiences a relatively low temperature in the second thermal cycle and is therefore subjected a relatively mild heat treatment. A major effect of reheating the HAZ is refining the structure (46). Depending on the distance between the beads, some original single pass weld HAZs may be subject to third or fourth reheating cycles. The final structure of the reheated HAZ is the result of the cumulative effect of each thermal cycle. It depends on a sequence of peak temperatures and cooling rates, and the precipitation behaviour during each weld cycle, as well as on the composition and initial microstructure of the material.

Since refinement of the HAZ structure leads to improvement of the mechanical properties of the HAZ, multi-layer welding is regarded as beneficial. However, as discussed in
Fig. 2.5  Schematic representation of structure distribution in HAZ of (a) single pass and (b) multi-pass weld deposits on flat plate (ref.47).
Chapter 3, Section 3.4.3, multi-pass welding with low heat input reduces the efficiency of the welding process. High heat input single pass welding is finding increasing application in the fabrication industry to increase welding productivity. Compared to multi-pass welding, single pass welding represents the most severe case with respect to the grain growth in the HAZ.

Alberry and Jones (47) first developed a computer model for calculating the microstructure in multi-pass weldments. They showed that with a knowledge of the phase-transformation behaviour (CCT diagram) and grain-growth kinetics, prediction of the multi-run weld HAZ structure is possible by utilizing Rosenthal's heat flow equation. In this work, reasonable agreement was obtained between computer-model predictions and measured weld HAZ structural distributions in multi-pass 0.5Cr-Mo-V weldments. With this model, it is possible to predict optimum weld parameters for a wide range of welding applications.

In 1985, based on the established relationship between dimensions of single manual metal arc weld bead HAZs and welding process parameters, Clark (48,49,50) was able to develop a model which allows the welding process parameters to be optimized with respect to dilution and refinement in the HAZ and weld metal.

2.4 MICROSTRUCTURE SIMULATION

The HAZ consists of a range of metal structures, as discussed in Section 2.3. Each type of structure is likely to possess different mechanical properties. The non-uniform microstructure of real weld HAZs causes difficulty in conducting mechanical property test on the HAZ. In order to obtain information about the microstructural and property gradient across HAZ, the weld simulation technique was developed by Nippes and Savage (51) to produce a synthetic HAZ microstructure in a small scale specimens.
2.4.1 Uses of Weld Thermal Simulators

The fracture toughness of the HAZ affects the overall weldability of a steel and is a key factor in determining the integrity of welded steel structures. Despite its importance, difficulties exist in carrying out toughness testing, such as Charpy V-notch impact and crack opening displacement (COD) testing, on real weld HAZs. These difficulties are:

(i) the problem of locating the notch tip in a volume containing a particular microstructure, due to the structure gradient in the HAZ;
(ii) specimens with standard dimensions for both tests are difficult to obtain from an actual weld, especially for the COD test.

A recent study by Ahmed and Yellup (52) proved that a large scatter exists among the values of Charpy impact energy and COD of weld HAZs. The sectioning and polishing of tested samples revealed that inaccuracy of notch tip location was the main reason for the scattering of results.

Although the influence of notch position on the critical COD value may be eliminated to a certain extent if the notch root of the COD specimen is prepared perpendicular to the surface (53), extensive toughness assessment of the HAZ has been carried out on the weld simulated HAZ microstructures (54,55,56).

Thermal simulators have also been used to study mechanical properties of the HAZ such as hardness (57,58), tensile strength (59,60,61), notch tensile strength (62) and stress rupture strength (63). Hot cracking problems associated with welding has also been investigated (64), where that the simulator was limited to hot ductility measurement.
equipment which could load the specimen rapidly to failure at any chosen point in the thermal cycle.

In addition, the weld simulation technique provides a very useful method for the continuous cooling transformation study of the HAZ (Section 2.5) and producing specimens for microstructural study of any required part of the HAZ. The effect of preheat and postweld heat treatments (PWHT), multi-run welds, etc., can also be easily incorporated in the programmed thermal cycle.

2.4.2 Weld Simulator

To achieve a very rapid heat rate similar to that experienced by the HAZ during welding, the specimen is usually heated by its own electrical resistance, or by a high frequency (HF) induction furnace. In resistance heating, specimens are clamped to a pair of water-cooled metal grips. The control of thermal cycle is via a thermocouple either spot welded to the surface or embedded inside the sample. The parameters of heating cycle, such as heating rate and peak temperature, are controlled by varying the current and voltage. The cooling rate during the cooling cycle is controlled by voltage and current inputs. In the case of air cooling, the cooling rate is controlled by adjusting the cooling water flow rate and the separation of the metal grips (65). To introduce fast cooling rates, helium cooling (66) and water spray cooling have also been used.

2.4.3 Comparison of Weld and Thermally Simulated HAZ Microstructures

There are two types of weld simulation.
(1) The weld simulation thermal cycle is programmed at different cooling rates, corresponding to different heat inputs according to Rosenthal's equation (Section 2.2)(67,68);

(2) the simulated thermal cycle is based on one measured from an actual weld HAZ (27,69).

In the first situation, thermal cycles with various peak temperatures and cooling times $\Delta t_{8-5}$ are used to simulate the microstructure corresponding to various positions in the HAZ and various heat inputs. The simulated microstructure does not directly correspond to the weld HAZ, although the approximate heat input can be calculated from the cooling time $\Delta t_{8-5}$ by employing the Rosenthal equations (2.9) or (2.11). However, since $\Delta t_{8-5}$ is related to heat input, the effect of heat input can be simulated by changing the HAZ cooling rate (or $\Delta t_{8-5}$). Similarly, various peak temperatures can be applied to simulate the microstructure developed in different positions in the actual HAZ.

In the second circumstance, the thermal cycle applied on simulated sample is kept as close as possible with that measured from the actual HAZ. In most cases, the results obtained from simulated samples and the actual HAZ were directly combined without any correction, as it was assumed that the microstructures were similar for both cases. However, this is clearly incorrect as it has been established (70) that even when the thermal cycles experienced by the actual HAZ and simulated sample are the same, difference in microstructure between both types of samples occurs.

Numerous attempts have been made to compare simulated and real welds. In general, there appears to be a fairly satisfactory correlation with respect to both microstructure and mechanical property measurements (71). These comparisons have been based mainly on qualitative examination of the microstructural constituents microstructures martensite or bainite. Little attention has been given to obtaining quantitative information on the proportions of the constituents and a comparison of the austenite grain sizes.
Dolby and Widgery (70) first reported that a discrepancy in austenite grain size was found for similar thermal cycles. Since austenite grain size is an important factor controlling the final transformation structure, a detailed investigation to determine the reasons for the discrepancy was carried out. The following possible reasons were examined:

1. difference in rate of heating to peak temperature between the weld HAZ and the simulator;
2. difference in rate of cooling from peak temperature between the weld HAZ and the simulator;
3. systematic errors in measurement of peak temperatures;
4. solution of precipitates, e.g. aluminium nitride, caused by their high local resistance to the heating currents in the simulated specimen, thus allowing easier grain growth; and
5. the narrow width of the weld HAZ restricting grain growth.

It was found that compared to the real weld HAZ, the larger austenite grain size of the thermally simulated specimen is caused by two major factors. They are the existence of a temperature gradient in the weld HAZ (reason 5) and errors in temperature measurement (reason 3). Any factor which controls the width of the HAZ, i.e., the temperature gradient, is very important in terms of the austenite grain size developed in the HAZ. These factors are heat input (31), welding speed (72) and plate thickness. Low heat input and high welding speed would be beneficial in keeping the austenite grain size to a minimum level, since a narrow HAZ results. The errors in temperature measurement during simulation were found to be a result of finite bead size of the thermocouples used, resulting in lower peak temperatures being recorded than the actual specimen temperature. The average error was about 29-31°C. The results suggested that care must be taken in interpreting the simulation results.
Due to the lower recorded temperature and larger austenite grain size in the simulated specimen, it was recommended that the recorded peak temperature of the simulated thermal cycle should be kept lower than that experienced in the actual HAZ. Good agreement was found between the microstructures of simulated specimens and the HAZ after cycling the simulated specimens to a lower peak temperature.

A difference in austenite grain size between the simulated specimen and the weld HAZ was also reported later by Berkhout (73). He found that the discrepancy in austenite grain size increased with decreasing of heat input and suggested that a correction to austenite grain size is necessary when comparing the austenite grain size in both HAZs, especially for welding with low heat input. This is consistent with Dolby and Widgery's conclusion that austenite grain growth is restricted by the temperature gradient in the actual HAZ. A smaller austenite grain size is expected for lower heat input welding because of the narrower HAZ.

More recently, a similar phenomenon was observed by Alberry et al. (28) when comparing the predicted austenite grain size with that measured from the actual HAZ. Supporting Dolby and Widgery's work, the authors concluded that the thermal gradient in weld HAZ causes the reduction of austenite grain size in the actual HAZ.

2.5 CONTINUOUS COOLING TRANSFORMATION DIAGRAMS

Systematic and detailed investigation of the $\gamma \rightarrow \alpha$ phase transformation reaction is necessary to elucidate the development of microstructure in the weld HAZ. The continuous cooling transformation (CCT) diagram relates the composition, cooling rate, and austenite grain size of the material to its $\gamma \rightarrow \alpha$ transformation temperature and the resultant microstructure.
Conventional CCT diagrams usually involve re-austenitization in the low temperature region of the austenite phase field (850-900°C), followed by continuous cooling by quenching, air cooling, or furnace annealing. A much higher austenite temperature, usually 1200-1400°C, is experienced by the GCHAZ during welding, and thus the conventional CCT diagrams cannot be applied directly to the HAZ for most welding situations. In addition, the thermal cycle of welding differs from that used to establish the conventional CCT diagram. Namely, the GCHAZ is very rapidly heated to a temperature just short of the solidus temperature, and then the cooling begins immediately.

For this reason, weld CCT diagrams have been developed (74) to predict HAZ microstructure and hardness.

The earliest publications describing CCT diagrams applicable to the weld HAZ were reported in the 1950s (75,76), following the first conventional CCT diagram produced by Christenson et al. (77) in 1945. In addition to many microstructural applications of CCT data which have allowed a fuller understanding of the γ→α transformation in the HAZ (78), the transformation product hardness values obtained have often proved to be a valuable source of data for predicting the actual HAZ hardness and as such have been incorporated into some of the schemes for avoiding HAZ hydrogen cracking (79)(Section 3.2) and predicting HAZ toughness (54).

A wider range of peak temperatures is experienced by the base metal adjacent to a weld, resulting in a wide range of microstructures in the HAZ (as discussed in Section 2.3). Thus, in principle, a large number of CCT diagrams is required to describe transformation behaviour in every region of the HAZ. Usually, the grain coarsened region (GCHAZ) represents the area in the HAZ most likely to have cracks as a result of an undesirable microstructure, such as twinned martensite and a local brittle zone. Thus the main effort in the study of weld CCT diagrams is concentrated on this region. A typical CCT diagram of the grain coarsened region (Tp=1400°C) is shown in Fig.2.6.
Fig. 2.6 Example of a weld HAZ CCT diagram for a medium strength C-Mn steel, (Tp=1300°C)(ref.74).
2.5.1 Methods for Determining CCT Diagrams

The principle of CCT diagram measurement is to heat samples quickly to certain peak temperatures, then cool them at various cooling rates. The $\gamma \rightarrow \alpha$ transformation temperatures are recorded during cooling. The weld simulator is usually used to produce appropriate thermal cycles similar to actual welding.

Various methods have been used to obtain HAZ CCT diagrams, but the main methods are dilatometry (54,80) and thermal analysis (81).

Dilatometry involves a dilatometer which is located on the center of the specimen which continuously monitors the diametral expansion and contraction throughout a thermal cycle. After plotting the curve of dilation recorded against temperature during cooling, transformation temperature is detected by the sudden change of dilatation of the specimen, as shown in Fig. 2.7.

The thermal analysis method has been reviewed extensively by Akselsen and Simonsen (82). They explained that during phase transformation of f.c.c. to b.c.c, the free energy is reduced. The enthalpy difference is liberated as heat, $\Delta H_{\text{trans}}$, resulting in a slower cooling rate through the transformation range which is used to monitor the onset of transformation. Akselsen et al. divided this method into three groups according to the processing of the signal: temperature-time ($T - t$), derived temperature-time ($dT/dt - T$ or $d^2T/dt^2 - T$), and differential and derived logarithmic derivation methods.

The temperature at which delayed cooling occurs due to the heat liberation of austenite transformation is considered as the start of transformation ($T_s$). The end of transformation ($T_f$) temperature is the temperature at which the temperature-time curve approaches its normal curvature, but with a displacement $\Delta t$ along the time axis, as shown
Fig. 2.7 Schematic illustration of a cooling curve (left) with the corresponding dilatation curve (right)(ref.82).
in Fig. 2.7. As can be seen, it is difficult to determine the precise location of both the start and end of transformation by using the temperature-time curve.

The derived temperature-time (65) analysis uses the curve of successive equal intervals \( \Delta T \) of temperature plotted against the temperature \((dT/dt - T)\). In this case, the start transformation temperature is shown at the temperature of marked deviation from linearity (Fig. 2.8). This method was considered to be more accurate than the temperature-time method since a more exact temperature for start of transformation can be defined by this method. To improve precision, the temperature-time cycle can also be analysed by plotting \( d^2T/dt^2 \) (Fig. 2.9), \( \ln(T-To) \) and \( d\ln(T-To)/dt \) versus time (Fig. 2.10) or temperature \((T)\).

Differential thermal analysis (81) involves a reference sample. It relies on the curve of temperature difference between tested and reference samples plotted against temperature (Fig. 2.11a). It is considered to be one of the most sensitive methods for recording the start of transformation. At the start of transformation, the temperature difference between tested and reference sample increases significantly. Applying the derived analysis \((dT/dt)\) on the differential curve can further improve the precision. After derived analysis, the curve is called the derived differential thermal analysis (Fig. 2.11b). The disadvantage of this method is that a reference sample is required.

Compared to the dilatometric technique, Phillips (81) reported that thermal analysis gave the most precise transformation temperature for fast cooling. At slow cooling, the dilatometric technique appeared to be the most precise method. Akselsen and Simonsen (82) later indicated that both methods give approximately the same temperatures for cooling time \( \Delta t8-5 \) in the range from 5-30 seconds. For slower cooling, corresponding to \( \Delta t8-5 \) longer than 30 seconds, the dilatometric analysis should be applied.
Fig. 2.8 Derived temperature-time analysis of temperature-time curves (ref.81).

Fig. 2.9 Thermal analysis of temperature-time cycle. Schematic illustration of the primary curve and the first and second derivatives. The start and finish of the transformation are marked (ref.83).
Fig. 2.10 Example of temperature recording (a) and signal processing (b and c)(ref.84).
Fig. 2.11  Derived differential thermal analysis curve. (a) Temperature difference between tested and reference sample against temperature; (b) Derived differential thermal analysis of curve (a)(ref.81).
2.5.2 CCT Characteristics and Hydrogen Cracking

In the past, CCT studies have been used to examine how HAZ transformation characteristics influence hydrogen cracking. In addition, data obtained from CCT diagrams are often used to help predict safe welding procedures. In most cases, a maximum HAZ hardness of 350HV is considered as the critical value with respect to cold cracking.

An early investigation by Cottrell (75) on the relationship between the HAZ CCT characteristics and HAZ hydrogen cracking indicated that the severity of HAZ hard zone cracking was linearly related to the 50% $\gamma \rightarrow \alpha$ transformation temperature, and was also loosely related to the martensite finish (Mf) temperature, as shown in Fig. 2.12.

Watkinson and Baker (85) used the data of transformation behaviour to predict welding conditions which, for a given composition, would result in an unhardened HAZ ($\leq$350HV). In early 1970, further improving the methods for predicting welding procedures to avoid HAZ cold cracking, Bailey (86,87) established linear relationships between carbon equivalent (CE) and the reciprocal of square root of critical cooling rate which gave a peak HAZ hardness of 350HV (Fig. 2.13). Based on this relationship, the maximum cooling rate for safe welding can be predicted for a given material.

More recently, it has become increasing apparent that safe welding procedure prediction based on 350 HV critical hardness for all steels is not reliable, because the critical hardness varies with carbon content (21,88). For low carbon steels, it was found (89) that the extent of HAZ embrittlement was more dependent on total fraction of martensite than the maximum hardness. Pavaskar and Kirkaldy (90) developed a method for assessing cold cracking susceptibility in low alloy steels, based on prediction of both % martensite and the hardness of the HAZ.
Fig. 2.12 Relationship between transformation temperature during continuous cooling and extent of cracking in tests weld (ref.75).

Fig. 2.13 Relationship between critical cooling rate giving 350HV (R350) and carbon equivalent (CE) for C-Mn steels containing 0.9-1.7%Mn and 0-0.6%Si (ref.87).
2.5.3 HAZ Toughness Prediction from the CCT Diagram

The weld thermal simulation technique has been used to assess HAZ toughness in the past. A number of these studies included the use of CCT data to help in the understanding of HAZ microstructure and toughness. These studies have provided useful information concerning a single weld HAZ.

For a quenched and tempered steel, Nippes et al. (76) used the thermal simulation technique to investigate the HAZ toughness and to obtain a HAZ CCT diagram. A fast cooling transformation product which is a mixture of low carbon martensite and bainite was found to have a higher toughness than a completely bainitic microstructure formed at a slower cooling rate. Therefore, it was concluded that best HAZ toughness was achieved when welding at low heat inputs without preheating, because of the presence of a high proportion of tough low carbon martensite.

However, contrary results were found for different steels for the effect of low carbon martensite on HAZ toughness. Inagaki and Sekiguchi (74) reported that an increase of toughness with increasing volume fraction of low carbon martensite is only true when the volume fraction of martensite is below 50%. For some steels (such as a as rolled and annealed steel), toughness declined significantly with increase of volume fraction of low carbon martensite in the vicinity of the critical cooling rate at 50% martensite. They were unable to explain the reason for this phenomenon. It was later suggested by Harrison and Farrar (91) that this may be the result of different degrees of autotempering of martensite during the final stage of cooling. Similarly, it was reported by Shorshorov et al. (92) that an increase in the proportion of martensite to ~100%, caused by rapid cooling, resulted in a sharp reduction in ductility. In addition, it was concluded that for the low carbon C-Mn-Nb steel investigated, the HAZ would be resistant to brittle fracture and cold cracking if the cooling times $\Delta t_{8-5}$ were maintained between 5-300 seconds.
A schematic diagram, shown in Fig. 2.14, was proposed by Sato and Yamato (93) to explain how low carbon, lower bainite type structures have high toughness, whereas high carbon upper bainitic structures exhibit low toughness. This conclusion was later supported by Inagaki and Hiroyuki (94). This research has led to the development of Ti-B and Ti-O type steels (95) in which ferrite nucleants are used to ensure that harmful upper bainite structures are replaced by acicular ferrite.
Fig. 2.14  Schematic diagram showing relationship between HAZ microstructures and toughness: M martensite, BL low bainite, BU upper bainite, F+P ferrite and pearlite, vTre transition temperature of absorbed energy (ref.93).
CHAPTER 3

PROPERTIES OF THE HEAT AFFECTED ZONE
3.1 INTRODUCTION

As pointed out in Chapter 1, the weldability of a steel is defined as the susceptibility of a steel to cracking problems associated with welds. Cracking behaviour, thus, is the key to investigate the weldability.

Of several types of cracking, cold cracking (or hydrogen induced cracking) is the most serious, least understood, and most widely encountered cracking problem associated with the HAZ. The occurrence of cold cracking is the result of three factors: a sufficient hydrogen concentration, an applied tensile stress and a susceptible microstructure. For a given level of hydrogen and a weld joint restriction which determines the stress of a weld, the risk of cold crack is dependent on the microstructure of HAZ.

HAZ hardness and fracture toughness are two important mechanical properties of weld joints. HAZ hardness reflects the type of microstructure and gives a direct indication of the strength, toughness and hardenability of the HAZ. Fracture toughness is an important property since it represents the crack resistance of a tested material, and HAZ toughness testing has been widely used to assure the quality of the HAZ as well as the whole weld joint.

In this Chapter, a brief review of HAZ cold cracking is given in Section 3.2, which covers the effect of microstructure (Section 3.2.1) and prediction of cold cracking by carbon equivalent value (Section 3.2.2). In Sections 3.3 and 3.4, the HAZ hardness and toughness are reviewed, respectively.
3.2 COLD CRACKING

Cold cracking (or hydrogen induced cracking) is the most serious of all weld cracking problems. An example of cold cracking is shown in Fig. 3.1. Cold cracking can occur in the weld metal as well as in the HAZ, but, except for very highly restrained welds or weld-metal strength exceeding 550 Mpa, this problem is encountered mainly in the HAZ due to the susceptible microstructure (96), as discussed later in detail.

3.2.1 Effect of Microstructure

Figure 3.2 shows that the solubility of hydrogen in ferrite is much less than in austenite. During the $\gamma \rightarrow \alpha$ transformation, a decrease in hydrogen solubility occurs which results in an increase in diffusivity. Hydrogen become more mobile to accommodate the change of solubility in $\alpha$ and $\gamma$ during transformation. The austenite is gradually enriched in hydrogen with the transformation. The lower temperature product is transformed in this region from more hydrogen-enriched austenite. Twinned martensite is considered very susceptible to cold cracking since it is formed at the lowest transformation, originating from the most hydrogen-enriched austenite.

Generally, martensite transformation is likely to be associated with a relatively high hydrogen enrichment at defects such as lath-packet boundaries or twinned plate boundaries. Overall, the lower temperature transformation products, such as martensite, bainite or Widmanstatten ferrite side plates, have higher susceptibility to cold cracking. Any factors that decrease the transformation temperature, like increasing austenite grain size and increasing carbon equivalent, will increase the susceptibility of the material to the cold cracking. The grain coarsened region of the HAZ is the region in which the greatest risk of hydrogen cracking generally exists for both C-Mn and low alloy steels (79), as a result of coarse grain structure.
Fig. 3.1 Examples of hydrogen or cold cracking in the HAZ of weld (ref.97).
Fig. 3.2 The solubility of hydrogen in iron (ref.98).
In general, the harder the microstructure, the greater is the risk of cracking (79,99). This is consistent with the trend that lower transformation temperature products have higher susceptibility to cold cracking because of their higher hardness.

The HAZ microstructure produced in any steel is dependent upon the composition, hardenability of the steel and the cooling rate experienced in the thermal cycle. The parameters related to HAZ microstructure derived from the above factors may be listed as follows: carbon equivalent (CE), time between 800-500°C, (Δt8-5) (proportional to heat input) and HAZ hardness.

3.2.2 Prediction of Cold Cracking by Carbon Equivalent

The CE value of a steel is assessed in terms of how the alloying elements present affect the transformation characteristics, i.e. the CCT diagram of the steel. Its value provides an indication of the type of microstructure to be expected in the HAZ as a function of cooling rate from the peak temperature. More specifically, it gives an indication of transformation temperature, the possibility of martensite formation in the steel and the hardness and hardenability of the steel. The established CE formulae are based on hardness and cracking testing data (21).

For many years, CE formulae have been widely and successfully used for assessing the risk of cold cracking in the HAZ of steels. The most widely used carbon equivalent formula was developed by the International Institute of Welding (IIW) (23) and was incorporated into a British Standard in 1974 (22):

$$CE = C + \frac{Mn}{6} + \frac{Ni + Cu}{15} + \frac{Cr + Mo + V}{5}$$

---------(3.1)
Conflicting factors must be considered for developing a good quality steel with proper chemical composition, i.e. carbon equivalent. Since high hardness and the presence of hard structure like martensite are known to increase the risk of cold cracking in HAZ, the CE level should be kept as lower as possible. However, at the same time, it should be recognized that the strength of a steel must be maintained at high level to meet the increasing demand for high strength steel. Increase in strength is likely to be achieved by increasing alloying element levels which result in an increase in CE.

Despite the contrary factors, the steelmaker has been able to lower the hardenability, and lower the CE level in pursuit of improved weldability. This situation has been achieved by improvements in rolling technology (12) and use of microalloy additions to obtain fine grained steel with an optimum combination of strength and toughness (Section 2.3.1.1).

The purpose of applying the CE equation is to control the chemistry of the steel to help attain good weldability. Since the hardened microstructure of a weld HAZ is susceptible to cold cracking, good weldability is ensured by placing a maximum allowable value on HAZ hardness. HAZ hardness is controlled by keeping the CE below a critical level. A value of 350HV is generally considered to be the critical value of HAZ hardness. Limiting values of CE have been proposed to ensure that HAZ hardness is kept below 350HV. A CE value of 0.4 (100) is considered to correspond to a HAZ hardness of 350HV and is claimed to be the maximum allowable value for avoiding cold cracking.

The above recommendation may only be considered as a generalization. It is obviously not true that a CE value of 0.4 is invariably associated with a hardness of 350HV because HAZ hardness is also largely determined by the cooling rate \( \Delta t_{8-5} \). This criticism has been made out by Brisson et al. (101). In addition, it has been advised (21) that a distinction between hardenability and cracking is important since the risk of cracking is related not only to avoiding hardened microstructures (hardenability), but also to the inherent susceptibility of any given hardened microstructure to cracking under the developed
contraction stress. Any effect of composition on hardenability may not be the same as on susceptibility, and the final effect on the risk of cracking will be the balance of the individual influences on these factors.

Nevertheless, as Winterton (102) pointed out, carbon equivalents can generally be used for rough comparison of steels limited to the range used in developing the carbon equivalent in question. To cover wide ranges of steel compositions, CE equations were developed based on the theory that the susceptibility of HAZ to cold cracking can be detected by one of the many weldability cracking tests available and that the results of the cracking test are associated to the carbon equivalent by the link to the HAZ microstructure.

In recent years, questions have arisen concerning the accuracy of IIW formula. It has been shown (103) that simple extrapolation of the IIW formula derived for high CE steel to low CE steel is not appropriate. In particular, the effect of the elements Mn and Mo are likely to be overestimated in these reduced C-steels with regard to their influence on hardening (17). Furthermore, the cooling rate (or Δt8-5) should be considered for its effect on HAZ hardness.

By investigating the relative effects of 0.07-0.17C, 1.0-2.0Mn, 0-0.8Ni, 0-0.5Mo and 0-0.14C in C-Mn steels on the risk of HAZ hydrogen cracking and on HAZ hardenability, Hart and Harrison (21) developed equations relating composition to critical hardness and critical cooling time Δt8-5 for cracking. Critical hardness was found dependent of composition, being higher with increasing C, Mn, Mo, and decreasing V contents. They also established the equations for cooling time to achieve different levels of HAZ hardness in the range 250-450HV as a function of composition. The decrease in critical hardness with decrease of CE value means that control of HAZ hardness to the conventional 350HV maximum may not prevent cracking in some low CE (<0.4) steels.
In his work on Australian produced low CE steels, Squires (24) also reported that the critical hardness appears to decrease with decreasing CE.

Recently, Yurioka and Suzuki (233) gave a detailed review of prediction of susceptibility to cold cracking as a function of HAZ hardness and carbon equivalent. They concluded that a critical hardness is not an adequate index to assess the susceptibility to cold cracking, since recent experience in the welding of modern steels has indicated that there is a trend towards cracking at a lower HAZ hardness level as the carbon content or CE decreases. However, they also concluded that whatever the trend in critical hardness with the CE may be, reduction of the CE of a steel definitely leads to an improvement in the resistance to cold cracking. Various CE formula were discussed. For low carbon, low alloy HSLA steels, the CE and cracking parameter (Pcm) formulae shown below are considered to assess the weldability more satisfactorily than the CE formula proposed by the IIW (equation (3.1)).

\[
\text{CEHSLA} = C + \frac{\text{Mn}}{16} - \frac{\text{Ni}}{50} + \frac{\text{Cr}}{23} + \frac{\text{Mo}}{7} + \frac{\text{Nb}}{5} + \frac{\text{V}}{9} \quad \text{---------(3.2)}
\]

\[
Pcm = C + \frac{\text{Si}}{30} + \frac{\text{Mn}}{20} + \frac{\text{Cu}}{20} + \frac{\text{Ni}}{60} + \frac{\text{Cr}}{20} + \frac{\text{Mo}}{15} + \frac{\text{V}}{10} + 5\text{B} \quad \text{---------(3.3)}
\]

Table 3.1 lists values of CEs calculated by the formulae CEIIw, CEHSLA and Pcm for CR HSLA 80 steel and the similar strength steel, HY 80.

<table>
<thead>
<tr>
<th>Equation</th>
<th>CR HSLA 80</th>
<th>HY 80</th>
</tr>
</thead>
<tbody>
<tr>
<td>CEIIw</td>
<td>0.41</td>
<td>0.59-0.97</td>
</tr>
<tr>
<td>CEHSLA</td>
<td>0.12</td>
<td>0.22-0.26</td>
</tr>
<tr>
<td>Pcm</td>
<td>0.20</td>
<td>0.30-0.41</td>
</tr>
</tbody>
</table>

Table 3.1  Comparison of Pcm and CEs for CR HSLA 80 and HY 80 steels
3.3 HAZ Hardness

HAZ hardness still gives a useful indication of the composition of a steel, microstructural constituents and the presence of martensite. HAZ hardness serves as a basis for characterizing the HAZ strength, toughness and hardenability (104) of a steel.

Despite the difference in critical HAZ hardness for steels with different CE, it has been shown by numerous researchers (105,106) that HAZ hardness is superior to carbon equivalent in predicting cold cracking tendency, especially at low carbon levels. This is due to the fact that the HAZ hardness is a parameter that takes into consideration both composition and cooling rate. Furthermore, it has been shown (99,107,108) that HAZ hardness correlates reasonably well with implant cracking test results.

An example of a hardness traverse across the HAZ is shown in Fig. 3.3. From the base metal towards the fusion line, hardness increases to a maximum at the fusion line, indicating a harder microstructure and the presence of martensite as a result of a large austenite grain size in the grain coarsened region (GCHAZ) near the fusion line. Lower heat input, or faster cooling rate of the HAZ, results in a harder HAZ. A relationship between peak HAZ hardness and cooling time $\Delta t_8$ is shown in Fig. 3.4.

For most steels, the whole range of the HAZ is harder than the base plate. However, numerous instances of HAZ softening have been revealed in different steels (96,109). Softening in the outer regions of the HAZ has been related to the overtempering effect on carbides and precipitates. In age hardened steel, softening observed for heating above the
Fig. 3.3 Hardness traverse in the HAZ of a 45mm AS 3678 (formerly AS 1204), 350 Grade structural steel welded by 4 wire Sub. Arc welding at heat input of 10kJ/mm and welding speed of 1000mm/min (ref. 33).

Fig. 3.4 Mean Vickers hardness measurements in the HAZ of a medium strength 20mm C-Mn steel based on several different welding processes (ref. 74).
At temperature involves the loss of precipitation hardening in the HAZ and the formation of high carbon austenite and ferrite in the intercritical HAZ region (31). A softened HAZ has occasionally been revealed adjacent to the fusion line in Cr-Mo steels (102). This result is thought to be a reflection of carbon migration during heat treatment (annealing or normalizing) or postweld heat treatment (PWHT).

3.3.1 Hardness Measurement

HAZ hardness measurement is usually fairly easy to carry out. However, owing to the narrow width and various weld bead shapes, special caution must be taken to choose the test location and hardness load in order to achieve consistent results.

Hardness testing methods for welds were reviewed extensively in a paper published by Commission IX (Behaviour of metals subjected to welding) of the International Institute of Welding (IIW) (110). Compared to Brinell and Rockwell testing methods, the Vickers test was considered to be preferable, since it covers the whole range of hardnesses, from the lowest to the highest.

The width of the HAZ could be very narrow (ranging from a millimeter to a few millimeters) depending on the welding process and the heat input. Hardness testing in the HAZ should be performed using loads which are necessarily low to reduce the size of the indentation. Concerning the errors of hardness measurement as a function of load for hardnesses ranging from 250 to 450HV, Brooks and Hart (111) found that the errors increase with decreasing of loads and hardness measurements are more accurate for lower hardness material. It was concluded (106) that a loads of 5 and 10 kg are more appropriate for assessing the hardness in the HAZ in the light of the HAZ dimensions generally resulting from current welding processes.
For the various standard specification (112,113,114), HAZ hardness measurement is usually recommended either along a line parallel to the surface of the plate (tangential method) or along the contour of the fusion line (contour method)(Fig. 3.5). The Australian Standard (AS 2205.6.1)(115) adopts the tangential method with the hardness survey parallel to the specimen surface, at 2mm below the surface. For a sharp-penetration weld profile, as in Fig. 3.6, it has been found (106,116) that zones marked A and B show higher hardness values than at the point of maximum penetration. Therefore, the location of hardness testing according to the Australian Standard is not necessarily at the position of maximum hardness.

Because determining maximum hardness value in the HAZ is not specified for a large number of specifications, it is difficult to assess the accuracy of the reported maximum hardness. It seems arbitrary using only one maximum value as the maximum HAZ hardness without confirmation by surrounding hardness points. For this reason, some specifications (114,116) have adopted the arithmetic mean method, specifying the mean of the 3 hardest values lying within a range of 25 Vickers as the peak value.

Cochrane and Amin (117) investigated the factors contributing to the scatter of HAZ hardness data on bead on plate (BOP) weld HAZs. A survey of the accuracy and reproducibility of the BOP hardness testing technique and assessment of the effects of microstructural variables were conducted. It was established that the scatter arising from the hardness testing technique and the variation in operator performance were small compared with the overall scatter of 50HV found in a typical production weld. A large variation in HAZ hardness along the length of a weld run was observed, typically within a range of 24HV, supporting earlier results by Brooks and Hart (111). The base plate characteristics, such as heat treatment, ferrite grain size and segregation (or pearlite banding) had minor effects on HAZ hardness. The most significant effect on hardness was found to be the austenite grain size of the HAZ which was largely influenced by the amount of Al present as AlN.
Fig. 3.5  Tangential (a) and Contour (b) HAZ hardness testing methods (ref. 110). WM: weld metal; BM: base metal; HAZ: heat affected zone.

Transverse section

Fig. 3.6  Cross-section of a weld (ref.110).
3.3.2 Prediction of Maximum Hardness

Direct determination of HAZ hardness requires destructive testing. Moreover, the accurate peak hardness is difficult to achieve due to the reason stated in Section 3.3.1. It would therefore be very helpful to be able to assess the risk of cold cracking prior to actual fabrication by the predicted HAZ hardness, based on the carbon equivalent value.

One approach of empirically establishing the maximum hardness formula has been the use of measured hardness values, cooling time $\Delta t_{8.5}$ and calculated cooling transformation temperature. Based on this approach, Yurioka et al. (118) developed equations to predict the hardness and martensite content of HAZs for most transformable steels with particular emphasis on those with tensile strength around 500 MPa.

Suzuki (119) established a formula (BL70) for estimating HAZ peak hardness based on a backward logistic curve relationship between maximum hardness and cooling time $\Delta t_{8.5}$ ($H_{\text{max}} = H_{\infty} + K/(1+ \exp(a(Y-Y_S)))$, $Y = \log \Delta t_{8.5}$, shown in Fig. 3.7). Stepwise multiple regression analyses were performed for the material constants, $K$, $a$ and $Y_S$, with 11 alloying elements as independent variables, to establish the BL70 formula. He claimed that this formula is more reliable than other formulae proposed previous by Yurioka et al. (120), Lorenz and Duren (121) and Terasaki (122).

Duren (123) later examined the validity of the BL70 formula by using hardness data from 110 different steel grades (119). The BL70 formula was compared with the formula ($HV_c$) proposed by Lorenz and Duren (121) and an older formula (SM) previously proposed by Suzuki (124). The result showed that the $HV_c$ formula produces the closest coincidence between measured and calculated hardness values. It was suggested that the inaccuracy of the BL70 formula is due to variations in measuring accuracy and the inclusions in steels with highly dissimilar chemical compositions (e.g. structural steels, pipeline steels).
Fig. 3.7 A backward logistic curve between Hmax and log Δt8-5 (ref.119).
3.4 Toughness of HAZ

Fracture toughness is one of the most important mechanical properties of the HAZ since it represents the crack resistance of HAZ in industrial applications for a given welding procedure. It remains the key to the structural integrity of major steel constructions.

Toughness is the capacity of a material to absorb energy by deforming plastically before fracture. It is determined by the combined strength and ductility of a material and usually is measured by the amount of work absorbed during the propagation of a crack through a structural member or a standard specimen. The consideration of toughness during design of a part will permit selection of materials with low probability of failure by fracture.

3.4.1 Testing Methods

When making HAZ toughness measurements, two experimental approaches are commonly used. They are the Charpy V notch impact test (125) and the crack opening displacement (COD) test (126).

As mentioned in Section 2.4, Charpy impact and COD tests on the actual HAZ are difficult to conduct due to the difficulty of correct placement of the notch tip in a suspected low toughness region. Thus, much toughness testing of the HAZ is carried out on the simulated HAZ microstructure (Section 2.4).

Charpy impact energy is thought to rank material toughness in the right order which is particularly useful in quality control testing. Compared to the COD test, the impact test can be easily carried out since the test bar (10x10mm--standard size) is cheap to produce, and the testing procedure and testing machine are relatively simple. These advantages are
the main reasons for the Charpy impact test being widely used to evaluate the toughness of the HAZ.

However, this test has its limitation. The most important limitation is that a good impact value is not always a guarantee against failure by brittle fracture. The impact test bar of 10mm thickness is selected independent of the thickness of the tested material. Therefore, the result achieved does not reflect the size effect of the material. More importantly, the strain rate of the impact test (ca. $10^5$-$10^6$ s$^{-1}$) is much too high compared to the strain rate of the construction generated by normal work loading conditions. The difference in strain rate results in a different toughness value between the test sample and the material in a structure. Furthermore, the notch of a standard impact bar (2mm depth, angle 45° and a 0.25mm radius at the bottom of the notch) may not be sited in a position where actual defects, such as crack, slag inclusions and undercutting exist, thus giving an optimistically high toughness value. On the other hand, the notch root could co-incide with these defects, giving a falsely low toughness value.

Charpy impact tests on simulated specimens have been reported (127) to give consistently higher transition temperatures than tests on actual weld HAZs. This difference can be attributed to inaccuracy in simulating the microstructure, such as coarser grains in the simulated sample due to absence of the thermal gradients that exist in the weld HAZ (Section 2.4.3). Another possible reason is that unlike the actual weld HAZ, the simulated Charpy specimen has a uniform microstructure across the ligament below the notch which is likely to give lower toughness values. Because of these disadvantages of the Charpy impact test, COD testing has found increasing application in assessing HAZ toughness.

The dimensions of the COD specimen are according to the proposals of the Standard (128). The thickness of the specimen varies, but it is often equal to the original plate thickness and this is one of the advantages over the impact test. In addition, it has been
reported (129) that the results of the COD test are much more sensitive to the microstructure at the notch tip than the impact test.

One of the disadvantages of COD testing is that the specimen must be pre-cracked at the notch tip. The process to produce a fatigue crack at the root of a machined notch is generally time consuming and expensive. It also involves more complicated and expensive instrument and requires a long time to complete the test. In addition, the COD test result usually gives an extremely wide scatter, which is difficult to interpret.

For this reason, the Charpy impact test is still widely used in the welding industry, as it provides a useful quality control test for ranking the toughness of materials.

In a review titled "Incentives for fracture testing", Denys (130) stated that a straightforward application of above-mentioned testing options in assessing the integrity of welded joints is difficult, due to:

a. the exact level of toughness required to avoid brittle fracture is not precisely known;
b. the COD specimen geometry does not always model the actual constraint;
c. the relationship between structural performance and COD for thick material sections is not documented with experimental data;
d. there is concern about the current practice of extending the results of COD measurement to structural performance;
e. the differing stress-strain characteristics of the weld and base metal are not sufficiently accounted for.

As a consequence, he urged that new ideas and engineering judgement are urgently needed to arrive at realistic toughness requirements. Immediate actions are needed to re-establish the credibility of fracture mechanics experts. He suggested that the tension
fatigue test could be of great values as a supplementary test to routine tests because the extrapolation to structural performance is quite easy.

3.4.2 Influence of Metallurgical Factors

Microstructure and HAZ toughness relationships have been briefly discussed in Section 2.5.3 with respect to HAZ toughness prediction from the CCT diagram. A more detailed review on HAZ toughness in terms of metallurgical aspects is reported herein.

The microstructural constituents which generally cause a toughness deterioration in the HAZ are as follows:

1. coarse grains;
2. upper bainite structure;
3. martensite-austenite (MA) constituent;
4. a high proportion of martensite, especially twinned martensite, with a relatively high hardness;
5. age hardening occurring in a slow cooling stage or postweld heat treatment (stress relief heat treatment).

The occurrence and degree of embrittlement due to the above structures are determined by the steel type and welding procedure. In the following sections, the effect of coarse grains and MA constituents in the HAZ are reviewed in detail.

3.4.2.1 Coarse grain in the HAZ

Maximum embrittlement of the HAZ normally occurs in the grain coarsened region. The embrittlement in this region is associated with a coarse microstructure, such as grain
boundary ferrite and ferrite side plates which are linked to a coarse prior austenite grain size resulting from a relatively high peak temperature. These constituents have low cleavage resistance despite the low HAZ hardness (131,132).

To reduce the prior austenite grain growth, precipitates such as TiN can be used as a means of pinning the grain boundaries. The beneficial effect of TiN on restricting the austenite grain growth and refinement of microstructure has been demonstrated by many researchers (27,133,134), despite some lack of understanding of the precise mechanism. The retardation of austenite growth by precipitate particles is reviewed in Chapter 2, Section 2.3.1.1.

Reducing the heat input of a welding process will result in a finer grain size as a result of the fast heating and cooling during the weld thermal cycle, as well as the thermal pinning effect of a steep gradient of grain size across the HAZ.

3.4.2.2 Martensite-austenite constituent

As well as the grain coarsened region of the HAZ, embrittlement can also occur in other sub-regions of single and multi-run weld HAZs. By using the weld simulation technique, Tomita et al. (55) examined local brittle zones (LBZ) in single and multiple weld HAZs for a series of offshore structural steels. It was found that the LBZ consisted of the partially transformed region and the intercritically reheated grain coarsened region of the HAZ. Similar results were observed by Uchino et al. (135).

The low toughness in these regions was attributed to the formation of high carbon martensite-austenite (MA) islands. In both regions, the MA islands were formed from partially austenized regions as a result of reheating to the intercritical range. These islands were found to be concentrated in original pearlite colonies (67,135,136) and were formed
from the austenite enriched in carbon and alloying elements. As a result, high carbon, twinned martensite islands were formed from the austenite during cooling. As well as the partially transformed and intercritically reheated grain coarsened regions, MA islands have also been found in the grain coarsened HAZ (53). In this case, the MA islands transform from retained austenite after preferential ferrite formation and they are located in regions between the ferrite laths and are interlocked by these laths.

The deterioration of toughness due to the presence of MA islands has been observed by many researchers (137,138). Increasing the volume fraction of MA islands was found to cause a significant deterioration of toughness (53,67). Toughness of the LBZ has also been strongly linked with the microhardness of the MA islands with higher hardness associated with lower toughness (68).

In spite of the numerous observations mentioned above, the mechanism of the deterioration of the toughness by the presence of the MA constituent has not been clarified. However, it has been found that preferential fracturing occurs at the MA constituent which lowers the toughness (139) and that brittle cracks run along the boundaries of the MA islands and ferrite matrix (140). Since the interfacial energy between MA islands/matrix ferrite decreases with the segregation of carbon (141), Nakanishi et al. (53) suggested that the MA constituent fractures preferentially at low stress and/or strain levels, establishing a Griffith crack (142), which then propagates along the MA constituent/matrix ferrite interfaces.

It was also suggested and demonstrated that cracking initiates from voids in the interface of martensite and ferrite (143, 144). The voids are developed by interfacial decohesion at the ferrite/martensite interfaces.

A rather different mechanism was proposed by Akselsen et al. (67) and Ramberg et al. (68). Instead of in martensite, they suggested that the initiation of cleavage fracture,
occurs in the ferrite as a result of an extensive build-up of dislocations at the ferrite/martensite interface. Stress concentration develops at the interface due to the significant difference in yield strength between the two phases (231), and is enhanced by the difference in flow strength between ferrite and martensite (232,145). In addition, strain partitioning between the MA islands and the ferrite may occur when only the ferrite can accumulate strain. Consequently, the embrittlement in the intercritical HAZ containing MA islands is considered to arise from the extensive build-up of dislocations at the boundaries of MA island and ferrite, and the brittle fracture initiates in the ferrite.

One possible solution to the embrittlement of the LBZ mentioned above is to facilitate the decomposition of the MA constituent by subsequent welding runs or by postweld heat treatment (PWHT)(stress relief heat treatment). However, PWHT, can cause temper brittleness or precipitation brittleness with some steels (69,146). This kind of embrittlement can occur in copper-bearing steels as discussed in Chapter 4, Sections 4.3.4 and 4.4.4.

3.4.2.3 Influence of microalloying elements

Microalloying elements can affect the HAZ toughness through changing the HAZ microstructure. The effect of Ti on retardation of austenite grain growth in the HAZ by forming TiN is reviewed in Section 2.3.1.1. The fine grain size and structure are likely to improve the toughness of the HAZ. A detailed study by Homma et al. (95) showed that Ti-containing oxides (Ti2O3) formed during a submerged arc welding process considerably enhanced the nucleation of acicular ferrite. Due to its high toughness (147), the acicular ferrite structure improved the HAZ toughness. Ti2O3 is found to be more stable than TiN even at very high peak temperatures near the weld fusion boundary and provides active nucleation sites for fine intragranular ferrite formation. The authors showed that the newly developed Ti2O3 steel exhibited less deterioration in notch
toughness in the grain coarsened region of the HAZ as compared with a conventional TiN steel.

The effect of Nb on HAZ toughness is found to be quite contradictory. Hulka and Heisterkamp (148) showed that Nb contents remarkably refine the austenite grain in the HAZ. It was reported that Nb contents up to 0.18% have a small influence on the $\gamma \rightarrow \alpha$ transformation, shifting it to a shorter time and preventing martensite formation, even at very high cooling rate. Therefore, Hulka and Heisterkamp (148) concluded that the toughness of the HAZ is enhanced rather than impaired. However, the effect of Nb has also been reported to depress the transformation temperature. Cane and Dolby (149) using COD testing demonstrated a detrimental effect of Nb on submerged arc weld HAZs at heat inputs of 5 and 7kJ/mm. The lower HAZ toughness of the Nb steels resulted from the effect of Nb on the $\gamma \rightarrow \alpha$ transformation temperature at slow cooling rates. By depressing the transformation temperature, the more brittle upper bainite structure developed, lowering the resistance to cleavage.

It is reported that in the presence of TiN, the reprecipitating elements do not always mix homogeneously with the TiN, but instead may sometimes form shells around the original TiN particles (150). The coarsened particles offer little resistance to grain growth in the higher temperature ranges, thus resulting in a coarse grain structure.

Wang et al. (25) indicated that the combination of Ti and Nb produced much poorer toughness properties than Ti alone. This was due to the much higher content of sideplate ferrite in (Ti,Nb) steel than in Ti steel as a result of Nb forming caps on the surface of TiN particles and reducing its potency for nucleating acicular ferrite.

Like Nb, V was also found to cause the as-weld HAZ notch toughness to decrease as the heat input increased (151,152). In the study of Hannerz and Holmquist (152) on
0.15%C, 1.3%Mn steels, notch toughness improved initially as the V content increase. At higher levels, however, toughness was reduced (Fig.3.8).

In the presence of TiN, Ca has a beneficial effect on HAZ toughness through the formation of calcium oxysulfide particles to complement the effect of TiN on pinning due to their higher stability in the high temperature range of the weld thermal cycle (134). Ti-Ca bearing steels exhibited a higher toughness than Ti-bearing steels at every simulated peak temperature up to 1450°C in simulated HAZ specimens for a series of high strength offshore steel plates (134).

Boron (B) can be successfully utilized to improve the toughness of a welded joint in Ti-containing steels. The absorbed energy was reported (134) to increase with an increase in boron content, but decreased when the content exceeded 0.0015% (Fig. 3.9). Since small TiN particles dissolve near the fusion line and free nitrogen is released, the improvement in toughness due to B was considered to result from reduction of free nitrogen by fixing it as BN at an early stage of cooling cycle of welding.

3.4.3 Influence of Welding Parameters and High Productivity Welding

The welding parameters such as heat input, welding speed, current and voltage affect the HAZ microstructure and properties. Although the individual effect of current, voltage and welding speed is relatively unknown, increase in heat input usually results in a coarse HAZ structure as a result of slow cooling, consistent with Rosenthal's equation (Section 2.2). The coarse austenite grains enhance the formation of undesirable microstructures, such as coarse grain size, coarse grain boundary ferrite, upper bainite and martensite-austenite (MA) constituents (Section 3.4.2). By reducing the heat input, a faster cooling rate brings about refinement of microstructure, eliminates or reduces the formation of these unfavorable structures and improves the toughness (153,154). However, the heat
Fig. 3.8 Influence of vanadium on the simulated HAZ transition temperature (ref. 152).

Fig. 3.9 Effect of boron on the toughness in simulated HAZ test (ref. 134).
input must be kept beyond a critical value to avoid the formation of martensite, especially twinned martensite, which also causes the degradation of HAZ toughness (155). Therefore, in general, there is an optimum cooling rate (heat input) which produces the highest toughness in the HAZ.

In recent years, increasing attention has been focused on increased welding productivity to achieve overall cost savings in the construction of large engineering structures (52,156). Improving welding productivity can be achieved through increasing the deposit rate (kg of filled wire/time) by

1. increasing the heat input;
2. increasing the welding speed;
3. using multiple wires.

Although it is generally believed that increasing the heat input may cause the loss of HAZ toughness, high heat input welding can be used in fabrication, provided the construction materials can still satisfy the mechanical property requirements, particularly the HAZ toughness. Moreover, developing a favorable microstructure and reducing the grain size by utilizing a microalloying elements have allowed that high heat input welding procedures to be used safely for some steels without considerable deterioration of weld mechanical properties.

Cuddly et al. (157) examined the addition of Ti on the HAZ toughness of high heat input welds for ship steels. Plates were subjected to simulated weld thermal cycles corresponding to 7.2 and 40kJ/mm heat input for submerged arc and electroslag welding, respectively. There was an optimum Ti content for both kinds of welds to achieve the best HAZ toughness. The improvement in toughness caused by small Ti additions (~0.01%) was due to reduction in hardenability by forming TiN which refines austenite grain size, promotes intragranular nucleation of ferrite and alters ferrite morphology from
Widmanstatten to equiaxed, thus improving the morphology and distribution of the pearlite grains. Higher Ti and N additions degraded toughness by embrittling the matrix.

In the study by Itoh et al. (158), weld simulation of large heat input welding was carried out on a newly developed Ti-B-lowN HY-50 steel. The cooling time $\Delta t_{8-5}$ was 160s. This steel showed good HAZ toughness under large heat input welding because complex precipitates of TiN-MnS-Fe$_2$3(CB)$_6$ acted as nuclei for intragranular ferrite formation producing a drastic change in the microstructure and improving toughness in the coarse grained HAZ.

A comparison of HAZ toughness of three high heat input welded Nb-Ti based HSLA steels welded was conducted by Ahmed and Yellup (52). The heat input ranged from 5.3-13.4 kJ/mm. The increase in heat input produced a decrease in toughness for the two normalized microalloyed steels. A third high-strength, low-alloy steel which was controlled rolled and contained Ti showed high HAZ toughness over the whole range of heat inputs. The good HAZ toughness of this steel was attributed to the effect of Ti on the retardation of austenite grain growth and the lower carbon equivalent of this steel.

Concerning the effect of the welding speed on the HAZ properties, Eichhorn and Pyrasch (136) examined the suitability of high speed electrogas welding with higher-alloy metal powder compositions on normalised and water-quenched and tempered fine grained structural steels. The Charpy impact energy of all regions of the weld joints met the requirement value of 27J at -51°C.

'Undercutting' and 'humping' were reported as problems occurring in high speed welding. 'Undercutting' is the depression at the edge of a weld and 'humping' is the formation of a raised section (hump) in the centre of the seam. Both phenomena were claimed to be more pronounced in high speed welding (159,160,161). However, with
close control of welding parameters and knowledge of weld bead formation mechanisms, good quality joints can be achieved (162,163).

The application of multiple wire submerged arc welding to increase productivity in welding was first reported three decades ago (164,165). During welding, several electrodes were positioned in series, parallel with the welding direction and were fed simultaneously to the joint, to produce a multiple pass weld in a single traverse of the joint (Fig. 3.10).

Ahmed and French (156) have used a double and 4 wire submerged arc welding arrangement to increase joint completion rates. The effect of welding speed on weld properties was also examined on the basis of constant heat input. It was shown that high speed multiple wire submerged arc welding significantly increases deposition rate and productivity of welding procedure. It was concluded that an optimum welding speed existed for a particular material to achieve the best weld quality. They suggested that thorough procedure development is required before use of such multiple wire techniques in a particular application since non-optimal procedures can cause surface and internal defects, as well as toughness degradation.
Fig. 3.10 4 wire Sub. Arc Welding equipment in CSIRO Division of Manufacturing Technology, Adelaide Laboratory.
CHAPTER 4

COPPER-BEARING STEELS AND THEIR WELDABILITY
4.1 INTRODUCTION

Compared with other copper alloys, iron-base alloys containing copper are of quite recent in origin. On the basis of the total tonnage of copper produced, the amount of copper used in ferrous products is still small. Since copper is cheap to produce and usually only relatively small amount are required to confer desirable properties, increasing amounts of copper are being used in ferrite alloys as an alloying element.

In the early stages, addition of copper in steel was mainly to increase the corrosion resistance. As long ago as 1928, driven by the need to develop slower-rusting steels than mild steel which was then widely used, the British Iron and Steel Institute set up a corrosion committee, to carry out completely unbiased corrosion tests on steel, cast irons and wrought irons (166). They discovered that copper-bearing steel, in the same application as mild steel, showed much slower-rusting properties. But it was not until some thirty years ago, that users of steel were benefiting from the evidence provided by this research.

More recent investigations have shown that Cu, in addition to providing improved corrosion resistance, can make a significant contribution to the mechanical properties of steel. The use of Cu as an alloying element has the following advantages; increasing the strength by Cu precipitation hardening (1,12); retaining good formability (4) and toughness; high corrosion resistance (5); high fatigue cracking resistance (6) and resistance to hydrogen induced cracking (167, 168).

When Cu is combined with other microalloying elements such as B or Mo, a tough acicular or bainitic structure can be produced (169). Copper also increases resistance to softening on tempering (170). Its effect, which is due both to solid-solution strengthening and to the effect of precipitated particles, is more complex than that of other common alloying elements.
Lower carbon content and carbon equivalent level are possible for Cu bearing steels since strengthening can be achieved by Cu precipitation hardening, rather than by increasing the hardenability through a high carbon content and a high carbon equivalent level. The lower carbon equivalent thus results in good weldability (171)(Section 3.2.2) and HAZ toughness.

Despite the abovementioned advantages of Cu in steel, it has been traditionally believed that addition of Cu to steel may cause problems in conventional steels. These problems include (5):

(1) difficulty of descaling, contributing to bad surface quality of the products, especially for steels with high sulphur content; this problem is usually attributed to a build-up of copper by preferential oxidation of iron, leaving a copper and sulphur-rich surface layer with a lower melting point;

(2) hot shortness caused by relatively low melting point of the ε-Cu phase (~1090 °C); the hot shortness is characterized by small cracks on the surface or edges during hot-rolling or continuous casting;

(3) deterioration of some specific mechanical properties, for example, embrittlement during stress relief.

These problems have prevented the widespread acceptance and use of copper as a major alloying element.

However, the severity of these problems is dependent on the amount of Cu and other alloying elements in the steel. With a careful control of Cu content and use of modern steel process technology, these problems are not insurmountable and the otherwise beneficial effects of copper may be more fully utilized.
By the addition of Ni, which increases the solubility of copper in austenite, during the production of Cu-containing steel products (172); and by control of the process parameters, including heating rate, soaking time and temperature, and furnace atmosphere (173), the hot shortness can be overcome.

The avoidance of hot shortness and improved steelmaking practice have led to the introduction in recent years of several commercial copper bearing steels, especially high strength low alloying (HSLA) steels, designed to exploit the strengthening effect of copper.

In this Chapter, a literature survey on the development of Cu-bearing HSLA steels is presented in Section 4.2. The weldability of Cu-bearing steels is reviewed in Section 4.3, which covers the effect of Cu on the susceptibility of weld to hot cracking (Section 4.3.1) and cold cracking (Section 4.3.2); the effect of Cu on HAZ toughness (Section 4.3.3) and the embrittlement after postweld heat treatment (PWHT) due to the Cu precipitation hardening (Section 4.3.4). The recent development of steels based on the composition ASTM A710 steel and the weldability of these steels are discussed in Section 4.4.

4.2 Development of Copper-Bearing HSLA Steels

Cu-bearing HSLA steels, with the addition of nickel (Ni) or chromium (Cr), have assumed considerable importance in recent years. Their corrosion resistance is considerably better than of unalloyed copper-bearing steels and they have the added advantage of possessing a higher strength.

The corrosion resistance of Cu-bearing HSLA steels (Cor-Ten Type) is compared in Fig. 4.1 with that of mild steel and normal copper-bearing steel. The better corrosion
Fig. 4.1 Comparative corrosion curves for three grades of steel in an industrial atmosphere (ref. 166).
resistance of these steels is dependent on a protective rust film on the surface, enhanced by an additional sealing effect on the pores of the rust by Cr and Ni (166).

Some mechanical properties of the steels in Fig. 4.1 are given in Table 4.1 which indicates that the copper-bearing HSLA 80 steel showed an improvement in yield strength of 50 per cent due to the precipitation hardening of Cu. This enables a considerable reduction in section thickness of the structural members, reducing the overall cost of a structure. With the addition of Ni and Cr to avoid the hot shortness, Cu-bearing HSLA steels have been routinely produced commercially without hot-shortness problems being significant.

Krishnadev and LeMay (174) reported that alloys with 2wt% Cu can be heat treated to yield strength levels beyond 860MPa (~125 ksi) while retaining reasonable ductility. The high strength is attributed to the martensitic matrix structure from a quenching heat treatment before aging. Based on this result, extra-low-carbon HSLA steels with yield strengths in the range 830-1030MPa have been developed by combining precipitation strengthening due to copper and niobium (Nb) and transformation substructure strengthening due to manganese (Mn) and molybdenum (Mo)(175).

In 1981, a new pipeline steel was developed utilizing the precipitation hardening of Cu by Nippon Kokan Steel of Japan (176). Being a low C and high Cu steel with Ni and Nb added, this steel has a yield strength of 456-491MPa and Charpy absorbed energy of more than 100J at -60°C after normalizing and tempering. The toughness of seam welds and the field weldability are also good.

A research project was undertaken in Laval University (Canada) to develop copper containing steels for structural applications in energy and resource, transportation and conversion systems (177). The experimental steels are based on higher copper and lower
<table>
<thead>
<tr>
<th></th>
<th>Approximate Chemical Composition</th>
<th>Mechanical Properties</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>C</td>
<td>Mn</td>
</tr>
<tr>
<td>Mild Structure Steel</td>
<td>0.3</td>
<td>0.5</td>
</tr>
<tr>
<td>Copper-bearing Steel</td>
<td>0.3</td>
<td>0.5</td>
</tr>
<tr>
<td>Cor-Ten Type</td>
<td>0.1</td>
<td>0.3</td>
</tr>
</tbody>
</table>
carbon levels than normal weathering steels and contain niobium and titanium. It was concluded that there is considerable scope for the development of a new class of higher copper content weathering steels for low temperature applications. By control of processing variables, including thermomechanical processing and aging heat treatment, a wide ranging combination of properties (536-1016 MPa yield strength and -15 to -130°C transition temperature) can be achieved in these alloys.

A new family of Cu-containing HSLA steels, namely Cu-B steels for pressure vessel applications and Cu-Si-Ti-Nb-B steels for hydrogen resistance application have been developed (178). The high yield strength, low transition temperatures, good formability characteristics, intrinsic weldability, corrosion and fatigue resistance were achieved on the basis of the use of copper as a major strengthening element, its synergistic influence with boron on $\gamma \rightarrow \alpha$ transformation and the ability of Ti to impart hydrogen cracking resistance.

Another family of steels based on the ASTM specification A710 (7) is finding increasing applications, and will be reviewed separately in Section 4.4.

### 4.3 Weldability of Copper-Bearing Steels

In the early days, there seems to be agreement among various researchers that copper is permissible in steel in amounts up to 0.8 to 1.0% without introducing difficulties in welding (179). There was a little more uncertainty about the welding of steels containing more than 1.0% copper, in spite of numerous reports indicating good weldability of steels containing up to 2% copper (180).
4.3.1 Hot Cracking

Hot cracking of welds has long been regarded as a major problem in copper-bearing steels. Such a cracking phenomenon is induced by the low melting point of copper, and hot shortness can occur in the weld metal as well as in the HAZ. This once-prevalent belief has led many people to believe that copper steels cannot be welded satisfactorily.

Quite conflicting results have been obtained for studies of hot cracking phenomenon in welds of Cu steels. Ni was reported (181) to reduce the hot cracking in weld with the steels containing 1%Ni-2% Cu showing the best result.

Miyoshi et al. (182) studied hot cracking of linepipe steels using the Varestraint Test. No hot cracks were observed for up to 1%Cu in a 0.11%C, 1.1%Mn steel, but in a 0.16%C, 1.45%Mn steel, hot cracks were observed for Cu contents greater than 0.5% with 4% or more bending strain.

Ostrovskaya (171) developed equations for determining the comparative and total effects of certain elements on resistance to hot cracking of weld metal deposits for mild and low alloying construction steels. Copper is among the group of alloying elements which was considered to have a bad effect on the resistance to hot cracking.

For a steel with a base composition of 0.18%C, 0.9%Mn, 0.7%Cr, 0.2%Mo, 0.07%Zr, Matsuda (183) studied the effects of other elements on hot cracking. The worst element was Cu among those examined (Fig. 4.2). Intense hot cracks were observed with a copper content more than 0.3%. The hot cracking was caused by a liquid film, containing segregated Cu along the solidification fronts and cell boundaries.

Watanabe and Matsuzaka (184) showed that, even in small quantities, Cu increased the tendency to hot cracking for a 0.25%C, 0.35%Si and 0.9%Mn steel.
Fig. 4.2 The effect of added elements on the hot cracking of a steel as measured by the Varestraint test (ref. 182).
More recently, investigations of hot cracking susceptibility using Varestraint testing of control rolled and aged Cu-bearing A710 modified steel (12) and quenched and tempered (class 3) A710 steel (1) indicated that these low carbon, Cu-Ni-Mn steels are less susceptible to hot cracking than commonly used steels (also see Section 4.4.4).

It was suggested (5) that steel plates containing copper must be descaled before welding because Cu in the scale layer sometimes causes severe hot cracking.

In general, although Cu enhances the tendency towards hot cracking, with the addition of Ni, hot cracking can be avoided in Cu-containing steels and eliminated by descaling the steel plate and careful control of Cu content below the critical value determined for that composition.

4.3.2 Cold Cracking

A small effect of Cu on the tendency towards cold cracking would be expected since copper has a minor effect on the hardenability of a steel (5). The relatively low propensity toward hardening of the HAZ indicates that copper-bearing steels may be welded without special precautions, such as preheating or the use of specially designed electrodes to give low hydrogen.

In Chapter 3, Section 3.2.2, the carbon equivalent (CE) equations aimed to predict the risk of cold cracking are discussed. Compared to other alloying elements, the low coefficient for Cu in carbon equivalent equations (equation (3.1)) reflects the low hardenability effect of Cu in steel (23).

Miyoshi et al. (182) summarised the published data dealing with the effect of Cu on cold cracking (Table 4.2). In most cases, copper had a smaller effect than nickel. Although
Table 4.2  Effect of copper on cold cracking susceptibility (ref.182)

<table>
<thead>
<tr>
<th>Formula</th>
<th>Proposer</th>
<th>Coefficient of Cu</th>
<th>Coefficient of Ni</th>
<th>Composition Range</th>
<th>Remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Edson</td>
<td>1/47</td>
<td>1/16.4</td>
<td>Cu: Not specified</td>
<td>From the hardness of HAZ</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C : Not specified</td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>Heuschkel</td>
<td>1/99</td>
<td>1/99</td>
<td>Cu: ~1.06%</td>
<td>From the bending angles of T-bend test at 21°C</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C : ~0.48%</td>
<td></td>
</tr>
<tr>
<td>3</td>
<td>&quot;</td>
<td>1/28</td>
<td>1/28</td>
<td>&quot;</td>
<td>Same as above, Temp. at -29°C</td>
</tr>
<tr>
<td>4</td>
<td>Dearden and O'Neill</td>
<td>1/13 (Cu&gt;0.5%)</td>
<td>1/15</td>
<td>Cu: 0-0.65%</td>
<td>From the hardness of HAZ of Reeve test</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0 (Cu&lt;0.5%)</td>
<td></td>
<td>C : 0.10-0.31%</td>
<td></td>
</tr>
<tr>
<td>5</td>
<td>Bradstreet</td>
<td>0</td>
<td>1/15</td>
<td>Cu: 0-1.13%</td>
<td>From the end of transformation temperature</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C : 0.12-10.20%</td>
<td></td>
</tr>
<tr>
<td>6</td>
<td>Winterton</td>
<td>1/40</td>
<td>1/20</td>
<td>Cu: ~1.5%</td>
<td>From the 90% transformation temperature</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C : ~1.2%</td>
<td></td>
</tr>
<tr>
<td>7</td>
<td>Suzuki and Tamura</td>
<td>0</td>
<td>1/40</td>
<td>Cu: 0-0.50%</td>
<td>From y-groove restraint cracking test</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C : 0.07-0.22%</td>
<td></td>
</tr>
<tr>
<td>8</td>
<td>Ito and Bessyo</td>
<td>1/20</td>
<td>1/40</td>
<td>Cu: 0-1.60%</td>
<td>From y-groove restraint cracking test</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C : 0.07-0.22%</td>
<td></td>
</tr>
<tr>
<td>9</td>
<td>Masumoto</td>
<td>1/30</td>
<td>1/50</td>
<td>Cu: 0-1.60%</td>
<td>From the hardness of HAZ</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C : 0.01-0.13%</td>
<td>(4 kJ/mm on 25 mm thick plate)</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1/20</td>
<td>1/18</td>
<td></td>
<td>(10 kJ/mm on 25 mm thick plate)</td>
</tr>
</tbody>
</table>
some coefficients were obtained by maximum hardnesses which are not directly related to cold cracking, the effect of Cu on HAZ hardenability can be easily obtained from this table. The difference in coefficients for HAZs produced by different heat inputs by Masumoto et al. (185) in Table 4.2 indicates that the lower cooling rate for high heat inputs allows Cu to precipitate more easily, thus increasing the hardness and the risk of cold cracking.

4.3.3 HAZ Toughness

The low hardenability contribution of Cu means that little detrimental effect of Cu on HAZ toughness would be expected. In fact, Cu as an added element, can improve the toughness at the fusion line and HAZ. The improvement of HAZ toughness by the addition of Cu is due to the refinement of the microstructure caused by a change from upper to lower bainite (186).

The effect of various elements on toughness near the fusion line was studied by Tanaka et al. (187) for manual arc welding (4.5kJ/mm heat input) and electro-gas welding (20kJ/mm heat input). As shown in Fig.4.3, Cu improved the toughness in the case of the 4.5 kJ/mm heat input, and had no effect in the case of the 20 kJ/mm heat input.

Based on the Charpy V-notch transition temperature (TT) data from thermal simulated HAZs in low alloy steels containing Cu in a range of 0-1.6%, Masumoto et al. (185) reported that the following formulae applied to submerged arc welds of the 25 mm thick plate, with heat inputs of 4 and 10kJ/mm, respectively.
Fig. 4.3 The effect of alloying elements on the shift of the Charpy impact transition temperature at the fusion line after two different welding procedures (ref. 185).
\[ TT (4 \text{kJ/mm}) = \text{Const.} - 48x(\%\text{Cu}) + 14x(\%\text{Cu})^2 \]  
\[ TT (10\text{kJ/mm}) = \text{Const.} - 5x(\%\text{Cu}) - 20x(\%\text{Cu})^2 \]

The combination of Ni with Cu has been reported (187) to improve HAZ toughness in Cu-bearing steels.

The Cu-bearing HSLA steels recently developed around the world, which obtain high strength from Cu precipitation hardening, have also shown good HAZ toughness and weldability. A newly developed 1.2%Cu-bearing fitting pipe steel, containing Ni and Nb, exhibited a Charpy impact energy and critical COD value comparable to those of the base plate (176).

Tomita et al. (188) demonstrated that HAZ toughness is insensitive to Cu content, consistent with little change in microstructure of the HAZ as the copper content increased from 0 to 1.4%.

High HAZ toughness values were also reported for ASTM A710 steels which are reviewed separately in Section 4.4.4.

4.3.4 Stress Relief Embrittlement

Stress relief embrittlement occurs in the HAZ during stress relief heat treatment (or postweld heat treatment (PWHT)) of welds, in the temperature range of 450° to 650°C. It can occur in precipitation hardened steels and is usually attributed to the precipitation of fine particles within the ferrite and the formation of a precipitate denuded zone along the grain boundary (189). Creep strains developed during PWHT or high temperature exposure are concentrated in the weakened grain boundary region and this, coupled with
grain boundary sliding, produces intergranular cracks. Elements contributing to such embrittlement have been reported to be chromium, copper, sulphur, phosphorus, arsenic, antimony and tin (189).

Figure 4.4 shows the effect of Cu and Mo on the cracking sensitivity for low alloy steels, according to Ito and Nakanishi (190). The increase in cracking susceptibility with increase in alloying elements was attributed to secondary hardening caused by precipitates.

Stress relief cracking in A710 steels has been observed by some researchers (189,191) and is discussed further in Section 4.4.4.

4.4 A710 Steels and Their Weldability

A710 steels are low-carbon, Cu-Ni-Cr-Mo-Cb, copper precipitation hardened steels which have been identified by a number of designations over the years. According to the ASTM Standard Specification for A710 steel (7), there are two different grades. Grade A provides minimum yield strength levels ranging from 55 to 85 ksi (380 to 585 MPa) and Grade B has the minimum strength range of 75-85 ksi (515 to 585 MPa). Most development has occurred for Grade A steels. The composition and three classes of A710 Grade A steel are given in Table 1.1. All three classes involve an aging heat treatment to achieve high strength by Cu precipitation hardening.

The A710 steels were known as IN-787 steels during their early development in the late 1960s and first commercial production in 1970. IN-787 steel was a modification by the International Nickel Company of an earlier alloy system called "Nicuage" (174). The IN-787 family of steels was subsequently developed to ASTM specifications for structural (A710) and pressure vessel (A736) applications. A military specification, MIL-S-24645
Fig. 4.4 The effect of Cu and Mo on stress relief cracking of low alloy steels (ref. 189).
(SH), also known initially as 'HSLA 80', was developed by U.S. Navy (192) from the A710 steels (Section 4.4.2). Increasing tonnages of A710 steels are being used in various applications, such as ship building (3), machinery and offshore platforms (8).

In these steels, the heat treatments of quenching and normalizing (classes 3 and 2 of A710 steel, respectively) are conventionally applied before the aging treatment to improve mechanical properties. To further improve properties, especially the weldability, thermomechanical processing can also be utilized for HSLA steels with a modified A710 composition (11, 178)(Section 4.4.3).

4.4.1 Age Hardening Heat Treatment

As a precipitation hardening steel, proper aging treatment is important to achieve adequate properties. Jesseman and Murphy (8) showed that aging at 540-705°C raises yield strengths in all three classes of steels by up to 175 MPa. Strength and toughness levels are mainly influenced by the age hardening temperature with changes being gradual and easily controlled. Time at temperature and cooling rate after age hardening have only relatively minor effects.

Hicho et al. (193) reported their study of the effect of thermal processing variations on the mechanical properties and microstructure of A710 Grade A, Class 3 (Q&A) steel. The main conclusion was that the size and amount of fine copper-rich precipitates are sensitive to the aging treatment and these, in turn, determine the mechanical behaviour of the alloy. Manufactured plate is normally in an overaged condition. Hicho et al. (194) later demonstrated that variations in properties caused by heat treatment can be attributed to the variation in the Cu precipitate distribution.
Abe et al. (195) revealed that accelerated cooling and direct quenching after controlled rolling enhanced the formation of a microstructure of low carbon bainite which produced improvements in strength and toughness of A710 and A710-modified steels. The improvement of mechanical properties was partly attributed to the retardation of recrystallization during hot working due to a Cu addition over 1% and the suppression of e-Cu precipitation due to rapid cooling.

4.4.2 Certification of HSLA 80 Steel in Naval Construction

The attractive properties of A710 steels have led to the development of these steels in naval structures.

As pointed out in the introduction of this thesis (Chapter 1), most of the conventional high strength steels used in Navy construction, particularly HY 80 and HY 100, develop their strength levels from a quench-and-temper (Q&T) heat treatment. The result is a martensitic steel structure which requires the use of stringent welding process controls and specially designed filler materials to retain adequate properties in the as-welded condition. Unfortunately, these requirements, as well as the Q&T heat treatment of the base plate, increase costs considerably (196).

Beginning in 1980, the U.S. Navy carried out a certification program for HSLA steels for use in ship construction (1). The prime goal was to 'reduce shipbuilding cost through improvement of welding processes, materials, technologies, procedures, and techniques, whilst simultaneously improving overall quality' (1).

Three classes of HSLA steels were chosen for certifying as 80 ksi HSLA steel for use in destroyers and other surface ships. They were (a) copper precipitation strengthened, (b) control rolled, and (c) Mn-Mo-Cb quenched and tempered steels. Extensive mechanical
property and weldability testing was conducted on these steels. Tests of sample steels from each class showed that only the Cu-strengthened steel based on the ASTM A710 Grade A, class 3 (Q&A) steel could immediately meet the property goals without requiring any alloy development or modifications.

The successful certification of the Q&A HSLA 80 steel based on the A710 Grade A, class 3 steel has led to wide acceptance of this steel around the world. Significant tonnages have been produced as a replacement for HY 80 in cruiser deck (3,169), bulkhead and hull applications (3), fittings, machinery, and offshore platforms (8).

4.4.3 A710-Modified Steels Produced by Thermomechanical Control Rolled Processing

Thermomechanical control rolled processing (TMCP) has been utilized to produce offshore structural steel based on the A710 chemistry (12,188). To produce A710 steel by the TMCP process route, modification of chemical composition is necessary because the alloy design of this steel is more suited to Q&A processing (class 3), rather than to the as-rolled and aged condition (class 1), due to the high quench hardenability provided by Ni, Cr and Mo additions (Table 1.1). Consequently, A710 class 1 steel plate with a conventional composition exhibited a much lower toughness than that of class 3 steel (1,8). A TMCP A710 type steel with a modified composition has been reported to show improved toughness of the base plate and HAZ relative to the conventional A710 class 1 steel (12,188).

As offshore structures of larg size are often installed in cold regions and in deep seas, steels of high tensile strength and low-temperature toughness are required to cope with these conditions.
To meet the increasingly stringent property demands for offshore applications, steel plates with a maximum thickness of 80mm, yield strength ≥450 MPa, tensile strength ≥570 MPa and high notch toughness of the weld joint have been developed in Japan by Tomita et al. (188). This new Cu-bearing steel with a modified A710-type chemical composition has been produced by TMCP. In this work, Mn less than 1.5% was found to increase the strength substantially and at the same time to increase HAZ toughness. However, Mn was found to decrease the HAZ toughness when present in an amount larger than 1.5%. Thus, Mn was added up to a limit of 1.5%. The main modifications of composition of this A710 modified steel were reduction of carbon and increase in manganese content. The effect of Mn in increasing the strength of A710 steel was also realized by Wilson (9).

Recently, BHP Steel Slab and Plate Products Division (SPPD) at Port Kembla, Australia, developed an even higher strength grade (550Mpa (80 ksi) Y.S) for offshore and naval use by using a TMCP (or control rolling (CR) process) (12). A modified leaner alloy design has been employed to facilitate TMCP for plates up to 50mm thick. Test plates with a thickness less than 25mm showed a yield stress in excess of 550MPa (80ksi), and can be classified as HSLA 80 steel or (CR HSLA 80). Plates in the thickness range 25-50mm had lower yield strengths, but were in excess of 500MPa (73ksi).

The chemical composition of this steel is shown in Table 1.2, together with that of conventional A710 steel. Compared to conventional A710 steel (Table 1.2), the main differences in chemical composition are:

(1) the elimination of Cr and Mo additions;
(2) an increase in Mn content;
(3) a reduction of C and Nb contents; and
(4) introduction of a small amount Ti.
Cr and Mo are eliminated because high hardenability is not required when TMCP is utilized. The increase in Mn was designed to offset the strength reduction arising from the elimination of Cr and Mo and the reduction of C, as well as to increase the HAZ toughness as mentioned previously (188). The reduced carbon equivalent level due to the reduction of C, Cr, Mo and Nb, together with the small addition of Ti, provide enhanced weldability of this steel. Furthermore, the reduced alloying element content means that this steel can be produced more cheaply than conventional A710 steels.

This steel has demonstrated excellent notch toughness. High HAZ toughness was also established by Charpy V-notch and COD tests on the actual HAZ of submerged arc welds at 5kJ/mm for 50mm plate and 3.2kJ/mm for 12mm plate. The improved mechanical properties were largely attributed to the grain size refinement brought about by TMCP and by the Ti addition (see Sections 2.3.1.1, 3.4.2.3).

4.4.4 Weldability of A710 Steels

The enhanced weldability and no requirement for preheat despite the high strength and toughness levels have been the major motivations for the use of A710 steel. The good weldability of this steel is attributed to its lower carbon content and carbon equivalent level (CEIWW of 0.4-0.57) compared to similar strength grades (e.g. HY 80 steel (CEIWW of 0.59-0.96)).

An early report on the weldability of A710 steel was by Jessman and Schmid (197). In this work, weldability was tested by determining transverse tensile and Charp V-notch (CVN) impact transition properties on both the weld deposit and HAZ for submerged arc welds. The microstructures of the deposit and HAZ were examined. It was concluded that A710 Grade A alloy steel plates can be readily submerged arc welded without
preheat. HAZ toughness in the coarse-grained region (GCHAZ) was less than that in the base metal, but high CVN impact toughness was still maintained at -50°F (-46°C) or lower test temperatures.

To test the weldability of the new Q&A HSLA 80 steel based on the A710 class 3 steel, extensive laboratory testing has been conducted by Ingalls Shipbuilding, Pascagoula, Miss. (3) to characterize weld properties. This work further proved that Q&A HSLA 80 is an outstanding ship steel that possesses high yield strength and toughness; can be joined with all of the standard welding processes used to fabricate high yield strength steels; and can be welded and fabricated with results of excellent quality.

Deb et al. (198) studied the microstructural characterization of two (preheated and non-preheated) shielded metal arc weldments across the weld HAZ for Q&A HSLA 80 steel. Their paper reported that preheated and non-preheated samples exhibited the same microstructure in similar locations with respect to the fusion line. It is also concluded that this steel can be welded without preheating as hydrogen-assisted cracking is not expected to be a problem in the HAZ.

Isothermal and continuous cooling transformation behaviour of HAZ for the Q&A HSLA 80 steel was measured by Vandermeer and Vold (199) and Lundin et al. (66), respectively.

It has been found that A710 Grade A, class 3 steel can be successfully laser beam welded (14,15). Through mechanical testing and microstructural studies of welds, Lukens (16) demonstrated that thick sections of this steel can also be successfully welded by the autogenous buried gas tungsten arc process.

Smith et al. (13) carried out a close examination of the relationship between microstructure and mechanical properties for weld metal and HAZ of submerged arc
welded Q&A HSLA 80 steel. In the as-welded condition, a decrease in notch toughness of the grain coarsened HAZ was found with increase in heat input, due to (a) an increasing proportion of the GCHAZ taking part in the fracture; and (b) a change in microstructure of the GCHAZ from low-carbon martensite/bainite to coarse upper bainite.

A710 steels show much less susceptibility to hot cracking compared to HY80 steel (3) despite a higher Cu content. The CR A710 modified steel showed less susceptibility to hot cracking than a low strength structural steel (AS1204-250) and 345 MPa YS grade TMCP low carbon microalloy steel (12).

Wilson (9) suggested that care must be taken when stress relieving (or postweld heat treatment) of the welds because a local hardened HAZ region can be created which could result in deterioration of toughness.

Supporting Wilson's suggestion, embrittlement of HAZ after postweld heat treatment (PWHT) was reported by Smith et al. (12) for Q&A HSLA 80 steel. PWHT at a temperature of 1050°F (566.1°C), 1150°F (621.1°C) and 1250°F (677.1°C) for 1 hour was performed on the HAZ by Lundin et al. (179). It was demonstrated that the GCHAZ of this steel was susceptible to stress relief cracking (SRC) at these three temperatures. Intragranular precipitation of Cu-rich precipitates leading to grain matrix strengthening and concomitant strain accumulation in the prior austenite grain boundaries during PWHT are likely to be the reasons for this cracking phenomenon. Cracking was not observed, however, in the grain-refined or partially-transformed regions of the HAZ. It was later pointed out that PWHT at 454°C for 1 hour also induced embrittlement in the GCHAZ (66).

Comparison of the susceptibility to SRC of HY 80 and HSLA 100 steels was carried out by Balaguer et al. (191). HSLA 100 is a copper-containing steel a modified A710 steel
(1.58%Cu) designed to meet the mechanical property specifications for HY 100. The stress-relief cracking susceptibility of the HSLA 100 steel was found to be greater than that of the higher-carbon, quench and tempered, high yield strength HY80 steel.

Most of the weldability studies listed above were concerned with Q&A HSLA 80 steel based on A710 Grade A, class 3 steel. Very little work has been reported on the weldability of the TMCP or CR HSLA 80 steel, except for the limited information reported in references 12 and 188. The need for a close examination of the weldability over a wide range of welding conditions for this type of steel was the driving force for the present investigation.
CHAPTER 5

EXPERIMENTAL METHODS
5.1 INTRODUCTION

The aim of the present research was to investigate the weldability, through a study of the structure and properties of the HAZ, of BHP's newly developed CR HSLA 80 steel. As pointed out in previous sections, the microstructure, toughness and hardness of the HAZ are its most important characteristics. The experimental work was thus directed mainly towards analysis of the microstructure, measurement and analysis of the toughness and hardness of actual and simulated HAZs. The types of welds used and the experimental work carried out on different types of HAZs are illustrated in Fig. 5.1.

Since the 'weld' continuous cooling transformation (CCT) diagram provides detailed information on the $\gamma \rightarrow \alpha$ phase transformation temperature range, a partial HAZ CCT diagram of the CR HSLA 80 steel was obtained by a weld thermal cycle simulation technique.

Investigating the properties of each sub-region of the HAZ is difficult because of the small volumes involved (Section 2.4.1), and thus the weld simulation technique was employed to provide a larger volume of uniform microstructure similar to that of a selected region of the actual HAZ. Charpy impact toughness and Vickers hardness measurements were made on the simulated specimens.

As discussed in Section 3.4.3, the productivity of the welding process can be improved by increasing the heat input and welding speed, but increases in productivity are only useful if the quality and integrity of the weld zone structure are not adversely affected by the welding process.

A joint (GIRD) research project on 'Development of high productivity welding processes and procedures for joining structural steels of Australian manufacture' is being currently conducted at the University of Wollongong. The main partners of this project are CSIRO
Fig. 5.1 Diagram showing the experimental procedures carried out for each type of weld HAZ.
Division of Manufacturing Technology (Adelaide Laboratory), BHP Steel SPPD and Bisalloy Steel Pty Ltd. Several locally produced steels are being investigated and the CR HSLA 80 steel produced by BHP Steel is one of the steels under investigation. Research work related to the steel is presented as part of this thesis.

This Chapter provides a detailed description of the materials used in the present research (Section 5.2); the welding procedures and electrodes (Section 5.3); and heat treatment by a second heating cycle for the bead-on-plate submerged arc (SA) weld HAZs (Section 5.4). The HAZ weld thermal simulation experiments are outlined in Section 5.5. Determination of the $\gamma \rightarrow \alpha$ transformation temperatures for construction of the CCT diagram is described in Section 5.6. The methods of mechanical properties testing are presented in Section 5.7 and the metallographic analysis of actual and simulated HAZ structure are given in Section 5.8.

5.2 MATERIALS

The steel used in the present study was CR HSLA 80 steel developed by BHP Steel, SPPD, Port Kembla, Australia. The chemical composition of this steel is given in Table 5.1. The carbon equivalent is 0.41 according to the formula proposed by the International Institute of Welding (IIW)(23). This steel was produced by a thermomechanical controlled processing (TMCP) route, involving three stages (12) which are illustrated in Fig. 5.2. The three stages are:

(1) recrystallisation controlled rolling in the $\gamma$ recrystallisation temperature range to achieve fine $\gamma$ grains;
Table 5.1 Chemical composition of CR HSLA 80 steel

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
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<td>0.05</td>
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<td>0.003</td>
<td>0.85</td>
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</tr>
<tr>
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<td>Cr</td>
<td>Mo</td>
<td>Nb</td>
<td>Ti</td>
<td>CErw</td>
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</tr>
<tr>
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<td>-</td>
<td>-</td>
<td>0.02</td>
<td>0.013</td>
<td>0.41</td>
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</tbody>
</table>
Fig. 5.2 TMCP schematic representation for CR HSLA 80 steel (B) and LCE 350 steel (A) (ref. 12).
(2) non-recrystallisation rolling (finishing phase) at the lowest possible temperature in the austenitic range, with the aim of minimizing the recovery of potential ferrite nucleation sites during the time interval before transformation begins; and

(3) controlled cooling to 500-550°C after rolling to enhance the plate cooling rate, thereby restricting the extent of 'auto' aging by Cu precipitation during the cooling of the plate. The last stage is designed to maximise the Cu age-hardening increment on subsequent aging (195). Finally, the steel was age hardened at 550°C for 1/2 hour in order to induce Cu-based precipitation hardening.

This steel has a banded ferrite and pearlite structure as shown in Fig. 5.3. The ferrite grains are fine and slightly elongated in the rolling direction and the pearlite band spacing of thicker plate was higher than thinner plate. The volume fraction of pearlite, average ferrite grain size and pearlite banding spacing of 20mm plate in transverse sections were reported as 2.3%, 7.2μm and 51.9μm, respectively (200).

Typical mechanical properties of plates with thicknesses from 12 to 50mm are summarized in Table 5.2 which shows that plates up to 25mm thickness satisfy the requirement of 550 MPa (80 ksi) yield strength (YS) and can be classified as HSLA 80 steel, according to the military specification (192)(Section 4.4.2). Plate thicknesses studied in the present work were 20, 36 and 50mm, and even though the steel is referred to CR HSLA 80 steel, it should be borne in mind that 'HSLA 80' does not apply strictly to the 36 and 50mm grades; which have a yield stress lower than the specified 550 MPa minimum yield stress.

A low carbon equivalent (LCE) 350 MPa YS grade steel was chosen as a reference steel to investigate the effect of Cu and Ni contents on the HAZ CCT diagram. This steel has a composition similar to the CR HSLA 80 steel except for a lower content of Cu and Ni contents as shown in Table 5.3. The IIW carbon equivalent of this steel is 0.34 (23). A
Fig. 5.3 Microstructure of base metal of the 20mm CR HSLA 80 steel (320x).
Table 5.2  Mechanical properties of CR HSLA 80 steel
(ref.12 and unpublished data from BHP SPPD)

<table>
<thead>
<tr>
<th>plate thick (mm)</th>
<th>orientation</th>
<th>Tensile properties</th>
<th>Charpy-V impact properties</th>
</tr>
</thead>
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<tr>
<td></td>
<td></td>
<td>LYS (MPa)</td>
<td>TS (MPa)</td>
</tr>
<tr>
<td>12</td>
<td>L-T</td>
<td>-</td>
<td>-</td>
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<tr>
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<td>T-L</td>
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<tr>
<td>25</td>
<td>L-T</td>
<td>583</td>
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Table 5.3  Comparison of chemical compositions of (A) CR HSLA 80 and (B) low CE 350 steels (ref.12)

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Ni</th>
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<tr>
<td>A</td>
<td>0.05</td>
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<td>0.012</td>
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<tr>
<td>B</td>
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<table>
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<th>Steel</th>
<th>Cu</th>
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<th>Mo</th>
<th>Nb</th>
<th>Ti</th>
<th>CEI\text{tw}</th>
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<td>0.41</td>
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<tr>
<td>B</td>
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<td>0.002</td>
<td>0.021</td>
<td>0.013</td>
<td>0.34</td>
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</table>
similar TMCP process to that for the CR HSLA 80 steel was utilized to produce the LCE 350 MPa grade steel, except that controlled cooling and aging heat treatment were not employed after rolling (Fig. 5.2). The microstructure of this steel also consisted of banded ferrite and pearlite (Fig. 5.4); however, ferrite grains were more equiaxed and larger than in 20mm CR HSLA 80 steel (Fig. 5.3).

5.3 WELDING PROCEDURES AND ELECTRODES

5.3.1 Bead-on-Plate SA and FCA Welds

Steel plate of 36mm thickness was single pass bead-on-plate (BOP) welded on the surface of the plate parallel to the rolling direction at BHP SPPD, Port Kembla by submerged arc (SA) and gas shielded flux-cored arc (FCA) welding processes. The welding parameters, types of electrodes used and shielding gas are listed in Table 5.4. Arc efficiencies ($\eta$) of 1 are used in the calculation of heat input for both types of welding processes.

The chemical composition of the electrode, Lincolnweld L50S, used for SA welds is given in Table 5.5 and meets the requirements of Australian Standard AS1858.1 for the EMH12K type electrode (also shown in the table)(201). The typical undiluted weld metal chemical composition of the Dual Shield 101TM flux-cored electrode is listed in Table 5.6.
Fig. 5.4 Microstructure of base metal of the 20mm LCE 350 steel (320x).
Table 5.4 Bead-on-plate welding parameters and types of electrodes

<table>
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<tr>
<th>Welding process</th>
<th>Electrode type</th>
<th>Shielding gas</th>
<th>Current (A)</th>
<th>Voltage (V)</th>
<th>Welding speed (mm/min)</th>
<th>Heat input (kJ/mm)</th>
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</thead>
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<td>L50S</td>
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<td>576</td>
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<td>L50S</td>
<td>-</td>
<td>800</td>
<td>30</td>
<td>360</td>
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<td>Dual Shield II</td>
<td>75%Ar/25%CO₂</td>
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<tr>
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<td>1.5</td>
</tr>
<tr>
<td></td>
<td>All Position</td>
<td></td>
<td>250</td>
<td>28</td>
<td>168</td>
<td>2.5</td>
</tr>
</tbody>
</table>

Table 5.5 Chemical composition (wt%) of electrode L50S and requirement of AS 1858.1 (ref.201) EMH12K type electrode

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Cu(total)</th>
</tr>
</thead>
<tbody>
<tr>
<td>AS 1858.1</td>
<td>0.07-0.15</td>
<td>1.25-1.75</td>
<td>0.15-0.35</td>
<td>≤0.03</td>
<td>≤0.03</td>
<td>≤0.15</td>
</tr>
<tr>
<td>L50S</td>
<td>0.09</td>
<td>1.53</td>
<td>0.32</td>
<td>0.009</td>
<td>0.013</td>
<td>&lt;0.1</td>
</tr>
</tbody>
</table>
Table 5.6  Typical undiluted weld metal chemical composition of electrode Dual Shield 101TM

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.058</td>
<td>1.13</td>
<td>0.37</td>
<td>0.006</td>
<td>0.011</td>
<td>1.78</td>
</tr>
</tbody>
</table>
5.3.2 Four Wire Submerged Arc Welds

Plates of 20 and 50mm were welded by CSIRO DMT using specially developed 4 wire SA welding equipment. Figure 5.5 shows the schematic diagram of the 4 wire SA welding arrangement. A photograph of this welding equipment is shown in Fig. 3.10. Steels were welded along the rolling direction at three different heat inputs and three welding speeds to analyse the effects of both parameters on quality of weld and response to high productivity welding procedures.

The joint preparations, types of electrodes and welding parameters used are summarised in Table 5.7. Two different types of electrodes were chosen for the 20 and 50mm plates due to the difference in yield strength level of the two plates. The chemical compositions of both electrodes are listed in Table 5.8.

5.4 REHEATED BOP SA WELD HAZS

Samples of dimensions 5x2x22mm were cut from the cross-sections of the BOP SA welds for each of the three heat inputs (Fig. 5.6). These samples were rapidly heated in a muffle furnace to a pre-determined temperature, then immediately air cooled. A thermocouple was embedded in each specimen to record the heating and cooling rates.

A second weld pass produces a thermal pulse at a particular point in the original HAZ and the peak temperature experienced will vary with distance from the heat source. In the present 'simulation', every point in the HAZ experienced the same thermal pulse, i.e., the same second peak temperature, Tp2. In this experiment, the intercritical and subcritical reheating of the HAZ was simulated with the peak temperatures selected at 800°C (A(c1) < T < A(c3)) and 600°C (< A(c1)), respectively.
Travel direction of electrodes

Fig. 5.5 Schematic diagram of 4 wire Sumberged Arc welding.
Table 5.7  Welding parameters, joint preparations and types of electrode of 4 wire Submerged Arc welds

<table>
<thead>
<tr>
<th>Plate thickness (mm)</th>
<th>Heat input (kJ/mm)</th>
<th>Welding speed (mm/min)</th>
<th>Edge preparation</th>
<th>Electrode</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
<td>2.5</td>
<td>1000</td>
<td>single V</td>
<td>LA100</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1500</td>
<td>M</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>2000</td>
<td>M</td>
<td></td>
</tr>
<tr>
<td></td>
<td>5</td>
<td>1000</td>
<td>single V</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>1500</td>
<td>S</td>
<td></td>
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<td></td>
<td></td>
<td>2000</td>
<td>S</td>
<td></td>
</tr>
<tr>
<td></td>
<td>10</td>
<td>600</td>
<td>nil</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>1000</td>
<td>S</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>1300</td>
<td>S</td>
<td></td>
</tr>
<tr>
<td>50</td>
<td>2.5</td>
<td>1000</td>
<td>double V</td>
<td>TIBOR 33</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1500</td>
<td>M</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>2000</td>
<td>M</td>
<td></td>
</tr>
<tr>
<td></td>
<td>5</td>
<td>1000</td>
<td>double V</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>1500</td>
<td>M</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>2000</td>
<td>M</td>
<td></td>
</tr>
<tr>
<td></td>
<td>10</td>
<td>600</td>
<td>double V</td>
<td>S</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1000</td>
<td>S</td>
<td>S</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1300</td>
<td>S</td>
<td>S</td>
</tr>
</tbody>
</table>

M: multi-pass; S: single-pass.

Table 5.8  Chemical composition of electrodes LA 100 (A) and TIBOR 33 (B)

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>0.057</td>
<td>1.74</td>
<td>0.46</td>
<td>0.004</td>
<td>0.015</td>
<td>1.9</td>
<td>0.09</td>
</tr>
<tr>
<td>B</td>
<td>0.09</td>
<td>1.46</td>
<td>0.02</td>
<td>0.007</td>
<td>0.012</td>
<td>0.02</td>
<td>0.04</td>
</tr>
<tr>
<td>Element</td>
<td>Mo</td>
<td>Al</td>
<td>Cu</td>
<td>Ti</td>
<td>B</td>
<td>N</td>
<td>O</td>
</tr>
<tr>
<td>---------</td>
<td>--------</td>
<td>-------</td>
<td>-------</td>
<td>--------</td>
<td>--------</td>
<td>------</td>
<td>----</td>
</tr>
<tr>
<td>A</td>
<td>0.39</td>
<td>0.011</td>
<td>0.05</td>
<td>0.04</td>
<td>49 ppm</td>
<td>85 ppm</td>
<td>160 ppm</td>
</tr>
<tr>
<td>B</td>
<td>0.44</td>
<td>0.04</td>
<td>0.01</td>
<td>0.15</td>
<td>0.013</td>
<td>0.006</td>
<td>-</td>
</tr>
</tbody>
</table>
Fig. 5.6 Selection of samples from actual HAZ of BOP SA welds for the second thermal cycle treatment.
The heating rate of 15°C/s is likely to be much lower than under actual welding conditions, but was the highest possible for furnace heating of the samples used. The cooling time between 800 to 500°C (Δt8-5) was 80 seconds (cooling rate of 3.8°C/s), which corresponds to high heat input welding with a low cooling rate. Although each sample had a different heat input for the first welding pass, the effective 'heat input' of the simulated second pass was the same for each sample heated to the same peak temperature because of the constant cooling rate.

5.5 HEAT AFFECTED ZONE WELD SIMULATION

5.5.1 Weld Simulator

HAZ thermal simulation in the present work was carried out on a recently developed weld simulator at the Department of Materials Engineering, at the University of Wollongong (Fig. 5.7). The simulator is based on resistance heating, relying on a computer to control the heating and cooling cycle.

A bar shaped test piece, 11x11mm in cross-section and with length ranging from 100 to 125mm was sampled from the mid-thickness of the 20mm plate. The test piece was clamped by grips at both ends and a thermocouple was spot welded onto the surface at the mid-length, to transfer the temperature signal from the specimen to the computer. The thermocouple wires consisted of Pt/Pt 13% Rh.

Sample cooling following a heating cycle was effected by conduction of heat through the water cooled cast iron grips.
Fig. 5.7 Weld thermal simulator.
5.5.2 Control of Thermal Cycle Parameters

In order to simulate the HAZ microstructure of a particular HAZ region for Charpy impact testing and to obtain the HAZ continuous cooling $\gamma \rightarrow \alpha$ transformation temperature, specimens were air cooled during the cooling cycle so that thermal changes during cooling were detected and analysed in terms of the $\gamma \rightarrow \alpha$ transformation temperature. An example of a simulated weld thermal profile is shown in Fig. 5.8.

Cooling rate and peak temperature were the two thermal cycle variables of most concern in this investigation. Cooling rate relates to the heat input of the welding procedure and peak temperature relates to the position in the HAZ in respect to the fusion line. To obtain the required peak temperatures and cooling rates, trial and error methods were used on dummy samples.

5.5.2.1 Cooling rate

Because specimens were air cooled, the cooling rate of the specimen was controlled by the separation of the grips at constant flow rate of cooling water. Figure 5.9 shows the relationship between the cooling rate (between 800-500°C) and the separation of the grips at constant flow rate of cooling water. The preset peak temperature was 1140°C for the data given in Fig. 5.9. As can be seen, the wider the grips are apart, the lower the cooling rate.

5.5.2.2 Peak temperature

The actual peak temperature reached by the specimen was usually higher than the preset one due to overshoot. Therefore, the preset peak temperature was set below the desired
Fig. 5.8  An example of the thermal profile experienced by a sample heated in the weld thermal simulator (actual $T_p=1300^\circ$C, preset $T_p=1265^\circ$C, $\Delta t=60s$).

Fig. 5.9  Relationship between cooling rate (between 800-500$^\circ$C) and separation of grips (preset $T_p=1140^\circ$C).
peak temperature level. As for cooling rate, the overshoot of the peak temperature is dependent on the separation of the grips. It also depends on the preset heating rate and the selected peak temperature. As Fig. 5.10 shows, the actual peak temperature varies with separation of the grips for a preset temperature of 1140°C. The wider the separation, the lower the temperature overshoot.

Since the separation of the grips was selected to suit the required cooling rate, as mentioned in the last section, and the preset heating rate was kept at a constant value for different simulation conditions, the desired actual peak temperature was obtained by setting the preset peak temperature at a suitable value. This suitable preset peak temperature was determined on a dummy sample before the actual simulation run.

5.5.3 Simulation Conditions

The relationship between cooling rate (800-500°C) and heat input is needed to calculate the equivalent heat input from the cooling rate for the simulated HAZ.

As pointed out in Chapter 2, Section 2.2, the time between 800-500°C (Δt8-5) during cooling of the HAZ is related to the heat input of the welding process by Rosenthal's equations ((2.9) and (2.11)). Δt8-5 is proportional to the heat input for thick plate and is proportional to the square of the heat input for thin plate.

To decide which equation to use to estimate equivalent heat input from cooling time data (Δt8-5), the critical thickness which defines the boundary condition between the two equations was calculated from equation (2.13)(17). Equation (2.13), which shows that d' is dependent on heat input, is as follows,
HAZ temperature measurements have been carried out using embedded thermocouples by CSIRO DMT (202,203). The temperature profile of a point in the HAZ at the weld root was recorded for 4 wire BOP SA welds on both 20 and 50 mm plates. The data for the 20 mm plate was used to estimate the heat inputs corresponding to the highest and lowest cooling rates possible in the simulator without forced cooling (32 and 5°C/s.). If the BOP 4 wire SA welds on 20 mm plate are considered to correspond to the thin plate condition, equation 2.11 should apply and a linear relation between cooling time Δt8-5 and the square of heat input (or between cooling rate and reciprocal of the square of the heat input) is to be expected. Figure 5.11 shows that a linear relationship does exist between cooling rate and the reciprocal of the square of heat input for measured cooling rate data for the BOP 4 wire SA welds at three heat inputs (2.5, 5 and 10kJ/mm) and a welding speed of 1000mm/min (202). The square of the correlation coefficient was equal to 1, suggesting that the thin plate equation (2.11) was appropriate for the welded 20mm plate. The equivalent weld heat inputs corresponding to the cooling rate for each simulated HAZ sample were thus calculated on the basis of the empirical equation established by the data plotted in Fig. 5.11. Equivalent heat inputs of 5 and 1.9 kJ/mm were estimated for the observed minimum (5°C/s) and maximum (30°C/s) cooling rates. As a consistency check, the calculated heat inputs were substituted in equation 2.11 to calculate the changeover d' values. The values were respectively 43.2 and 24.5mm, both in excess of the dimension of the plate from which the cooling rate data were obtained. Therefore, application of the thin plate equation is valid in this case.

5.5.3.1 HAZ of single and multi-pass welding

The peak temperatures were chosen to simulate the local HAZ peak temperature of single and multi-pass welds as closely as possible.
Fig. 5.10  Relationship between actual peak temperature and separation of grips (preset Tp=1140°C)

\[
y = 0.53333 + 107.29x \quad R^2 = 1.000
\]

Fig. 5.11  Relationship between cooling rate and reciprocal of square of heat input.
For simulating the distinct microstructural regions from the HAZ boundary to the fusion boundary of a single weld run or the last welding pass of multi-pass welding, a series of single thermal cycles was used, where the peak temperature was varied from 600 to 1300°C (refer to Section 2.3). The peak temperatures were chosen at 600°C, 700°C, 800°C, 900°C, 1020°C, 1300°C. The cooling time from 800°C to 500°C ($\Delta t_{8-5}$) was 22.6 seconds which converts to an average cooling rate of 13.3°C/s, corresponding to an equivalent heat input of 2.9kJ/mm (from Fig. 5.11).

Generally, during multi-pass welding, the grain coarsened region (GCHAZ) near the fusion boundary is reheated by the subsequent pass up to temperatures which depend on the distance from the fusion line of the subsequent pass (Section 2.3.5). The reheated peak temperature is lower than the previous one and can result in tempering, partial re-austenitising or full re-austenisation plus grain refinement.

To simulate the microstructural evolution of the original GCHAZ in multiple pass welding, a second thermal cycle was applied, where the peak temperature of the first cycle ($T_{p1}$) was 1300°C, and peak temperatures of the second cycle ($T_{p2}$) were 600°C, 800°C and 900°C, corresponding to tempered (subcritically reheated), partially transformed (intercritically reheated) and grain refined (fully re-austenitised) regions of the HAZ.

The reheated region near the primary fusion boundary is reheated further by the successive welding passes, but the thermal pulse is relatively weak and attenuates with increasing pass number. To simulate this case, an additional thermal cycle (the third) was applied. The peak temperature of the third cycle was 600°C and only the tempering effect was studied, because the distance between the primary GCHAZ and the third pass fusion boundary is generally large and a high reheating temperature cannot be expected in an
actual welding process. The peak temperatures of the first and second cycles in this case were 1300 and 800°C, respectively.

5.5.3.2 The effect of heat input

To investigate the effect of heat input on microstructure and toughness of the GCHAZ, three additional cooling rates were chosen to modify the microstructure of the GCHAZ produced at a peak temperature of 1300°C. In addition to the cooling rate of 13.3°C/s indicated in experiments discussed in the last section, the three average cooling rates between 800-500°C were 5.0, 23.8 and 30.2°C/s which correspond to 4.9, 2.2 and 1.9 kJ/mm, respectively.

5.5.3.3 Postweld heat treatment

The effect of postweld heat treatment (PWHT) or stress relief heat treatment on properties of the GCHAZ was examined out in the present work. After being subjected to a peak temperature of 1300°C at a cooling rate (800-500°C) of 12°C/s (equivalent heat input of 3.1kJ/mm), specimens were postweld heat treated in a muffle furnace at 450°C, 550°C and 650°C for 1 hour.

5.6 DETERMINATION OF $\gamma \rightarrow \alpha$ TRANSFORMATION TEMPERATURE

Cooling curves (temperature versus time) for each cooling condition of simulated GCHAZ specimens were analysed to determine the start, 50% and completion $\gamma \rightarrow \alpha$ transformation temperatures to obtain data on the CCT diagram for the GCHAZ.
As mentioned in Chapter 2 (Section 2.5.1), the $\gamma \rightarrow \alpha$ transformation temperature can be obtained by dilatometry and by thermal analysis of cooling curves. In the present study, thermal analysis was used. The derived temperature-time method based on the curve of $dT/dt$ versus temperature ($T$) was used to determine the transformation temperatures. This method was adopted because of its relative simplicity and accuracy. An attempt was also made to improve the accuracy of definition of the transformation temperature by plotting curves of $d^2T/dt^2$ against temperature ($T$). However, these $d^2T/dt^2$ curves failed to show a more clearly defined transformation temperature than $dT/dt$ curves (Fig. 5.12). Thus, the curve of $dT/dt$ against temperature was used to determine the transformation temperature.

For the same cooling rate, three identical Charpy samples were prepared for impact toughness testing. The transformation temperatures (start, 50% and completion transformation) were measured from each of the three cooling curves obtained and the average result was used in the final CCT diagram.

5.7 MECHANICAL PROPERTIES TESTING

5.7.1 Impact Toughness

5.7.1.1 Simulated HAZs

Charpy-V notch (CVN) impact energy testing was carried out on samples with simulated HAZ structures. After thermal simulation, specimens of 55mm length were selected at mid-length of the heated area and surface machined to a cross-section of 10x10mm to remove the oxide scale on the surface, as well as to obtain the standard Charpy dimensions (10x10x55mm)(125). The notch was cut to produce samples with a T-L
Fig. 5.12  (a) A temperature-time cooling curve at cooling rate of 13.1°C; (b) Derived temperature-time analysis \( (A, \frac{dT}{dt} \sim T) \) of curve in (a) and derived analysis of curve A \( (B, \frac{d^2T}{dt^2} \sim T) \).
orientation. Three parallel samples were tested for each condition to obtain an average value. Since the 20mm plate meets the military requirement of HSLA 80 steel (192), the testing temperature was selected at -51°C to comply with the requirements of the military standard for this grade of steel. The testing of Charpy specimens was carried out at the Bisalloy Steel Pty Ltd laboratory.

5.7.1.2 Four wire SA welds

CVN tests were also carried out on actual HAZs of 4 wire SA welded 20 and 50mm plates by CSIRO DMT. Charpy test bars were also notched with a T-L orientation. Due to the difference in yield strength of the base plate, HAZs of 20 and 50mm thickness plates were tested at -50°C and -20°C, respectively, to comply with the requirements for each plate thickness.

5.7.2 Hardness

Vickers diamond pyramid hardness testing was employed to assess the hardness of actual and simulated HAZs. Before testing, specimens were metallographically polished and lightly etched as required by the Australian Standard AS 2205.6.1 (115). Combined polishing and etching procedures were necessary for actual welds to reveal the position of the HAZ so as to ensure the correct position for hardness testing. In all cases, the interval between the two closest indentations was greater than 2.5 times the diagonal of the largest indentation as required by the Australian Standard AS 1817 (205).

Five measurements were carried out for testing the peak hardness of the actual HAZs and the hardness of simulated HAZ specimens. The final results are shown as average values, with 95% confidence limits (206) for each average.
5.7.2.1 Bead-on-plate welds

Hardness traverses across the HAZs were measured along two lines 1mm apart and perpendicular to the fusion line for BOP SA and FCA weld HAZs, and as well as for the second thermal cycle reheated BOP SA welds. The hardness testing position was chosen at the widest HAZ, where the highest hardness is expected, as reported by other researchers (106,116)(Fig. 3.6 of Section 3.3.1). Hardness traverses were determined by means of a Vickers hardness tester with 5kg load. The interval between the two closest impressions was 1mm.

For the unheated BOP SA welds, hardness traverses across the HAZs were also measured at three different positions (a, b and c), corresponding to maximum, intermediate and minimum HAZ width as seen in Fig. 5.13. The hardness traverses in this case were determined by means of a Leitz micro-hardness tester. The hardness traverses were tested along two parallel lines at 0.3mm interval with the indentations offset by 0.15mm.

The peak hardness of the HAZ was measured near the fusion line by using a smaller load (200g) in order to locate the indenter as close as possible to the fusion line. This measurement was carried out on a Leitz micro-hardness tester. The peak hardness was measured at intervals of 200μm along the contour of the fusion line at a similar position to the hardness traverse. The distance between the fusion line and the indentations was 50μm.

5.7.2.2 Four wire submerged arc welds

Hardness traverse surveys of 4 wire SA weld HAZs for both 20 and 50mm thick plates
Fig. 5.13 Macrographs of cross-section of BOP SA welds. (2.5% nital)

a, b and c on each macrograph indicate the hardness traverse sampling positions.
were carried out along a line 2mm from the surface of the plate, from the weld metal across the HAZ (of the sealing run in the case of multiple runs) to the base metal. The hardness traverse was measured by indentations at 635μm intervals by using a 5 kg load. This method was adopted as it is recommended by the Australian Standard (AS 2205.6.1)(115).

Peak hardnesses were measured on HAZs of 20mm plate using a 5kg load, in a similar way to that used for bead-on-plate welds. Peak hardness testing was also performed at positions of maximum HAZ width and along the contour of the fusion line. Five measurements were made at intervals of 635μm and 150μm away from the fusion line.

5.7.2.3 Simulated HAZ

After Charpy impact testing, the surfaces of Charpy samples of the simulated HAZ structures were metallographically polished and lightly etched in 2.5% nital. Vickers hardness testing with a 5kg load was then carried out, followed by metallographic analysis at similar location to that of hardness testing.

5.7.3 Tensile Properties of 4 Wire Submerged Arc Welds

Since hardness traverses across the HAZs indicated that the hardness was lower in the HAZ than in the weld metal and base plate, transverse tensile tests were carried out to investigate whether the softening of the HAZ caused any reduction of strength of the weld joint.

Cylindrical tensile pieces of 10mm diameter were taken from the mid-thickness of each 4 wire SA welded 20mm plate sample. The test piece was oriented in the transverse
direction and sectioned with the weld metal centred about the mid-length position. The
gauge length was 50mm which was 5 times the diameter, as recommended by the
Australian Standard AS 1391 (207). Three parallel tensile tests were carried out for each
welding condition using the facilities of Bisalloy Steel Pty Ltd.

5.8 METALLOGRAPHY AND FRACTOGRAPHY

5.8.1 Optical Metallography and Fractography

Both macro- and micro-scopic metallographic techniques were used to examine the actual
and simulated HAZ specimens.

Transverse cross-section of welds were cut from the plate and metallographically
polished. HAZs were macroscopically revealed by deep etching in 2.5% nital solution.
After macrographic examination of the weld profile and the HAZ, photomacrophgraphs
were taken for all the welds using vertical illumination. Welds were then repolished and
lightly etched in 2.5% nital for microscopic examination using a Leitz and/or Nikon
microscope.

After Charpy impact testing and hardness testing, the notch side surfaces of the simulated
HAZ Charpy specimens were microscopically examined at the region near the notch
where hardness measurements were taken.

The fracture surfaces of Charpy impact specimens of the simulated and actual 4 wire SA
weld HAZs were macroscopically examined by stereomicroscopy and SEM to elucidate
the fracture characteristics and the fracture mechanism associated with different
microstructures. Similar observations were carried out on the fracture surfaces of tensile test pieces of 4 wire SA welded 20mm plate.

5.8.2 HAZ Width Measurement

The width of the HAZ perpendicular to the fusion line was measured along the fusion line for BOP SA and FCA welds. The measurement was carried out by measuring the apparent width in macrophotographs of etched cross-sections of the welds. The HAZ width was recorded as the distance between the fusion line and the termination of the partially transformed region of HAZ, corresponding to the apparent boundary of the HAZ on the macrophotograph. This measurement was carried out on a graphics tablet attached to an Apple computer.

5.8.3 Quantitative Analysis of Martensite-Austenite Islands

A quantitative metallographic study of martensite-austenite (MA) islands was performed on BOP SA weld HAZs (36mm) and 4 wire SA weld HAZs of 20mm plate. To reveal the islands clearly, a color etching method was used (208). This method was based on the modification of a etching method originally used for revealing sulphides in carbon and alloy steels (209). Samples were pre-etched in 2.5% nital and then stain etched in a solution of aqueous 20% sodium thiosulfate, 2.3% citric acid and 2.3% cadmium chloride. The ferrite grains were colored blue, and the islands were colored brown. The black and white image of a color etched specimen is shown in Fig. 5.14. Quantitative image analysis was carried out on an MD-20 Image Analyser equipped with a video camera connected to a Nikon optical microscope.
Fig. 5.14  Color etched GCHAZ showing MA islands for 20mm CR HSLA 80 steel plate welded at 10kJ/mm heat input and 600mm/min speed (640x).
For BOP SA weld HAZs, the quantitative analysis was carried on contiguous fields extending from the fusion line towards the base metal over a distance of 500µm at the root of the weld where the width of the HAZ is a minimum. This traverse included the grain coarsened and part of the grain refined regions of the HAZs. In excess of 1000 particles were sampled in determination of each data point for volume fraction or areal density.

For 4 wire SA weld HAZs, quantitative metallographic analysis was conducted at a position of maximum HAZ width near the fusion line. Five fields were measured along the fusion line. Each field covered an area of 4469µm². The results quoted in Chapter 6 are the average values for the five fields.

5.8.4 Measurement of Prior Austenite Grain Size

By using delineation of the prior austenite grain boundaries by ferrite allotriomorphs, the linear intercept austenite grain size of the HAZ adjacent to the fusion line was determined along the fusion line for each of the BOP SA HAZs and 4 wire SA weld HAZs of 20mm plate. The measurement was carried out on Reichert and MGK Olympus optical microscopes for BOP and 4 wire SA weld HAZs, respectively, with the aid of a scaled eye piece.

The prior austenite boundaries of the HAZ in the area adjacent to the fusion line of BOP FCA weld HAZs were revealed by etching in 2.5% nital (Fig. 5.15). The prior austenite grain sizes of this area along the fusion line for the three BOP FCA weld HAZs were measured by outlining grain boundaries on a graphics tablet attached to a Apple computer.
Fig. 5.15  Optical microstructure showing the prior austenite grain boundaries of the HAZ near the fusion line of BOP FCA weld HAZ (1.5kJ/mm)(320x).
Two layers of prior austenite grains adjacent to the fusion line were measured. The measured area covered half of the total length of the fusion line, from the surface to the root of the weld bead. The final austenite grain size of each HAZ is the calculated average grain size of all the measured grains for each weld.

5.8.5 Electron Micrography and Fractography

To reveal the fine detail of microstructure and Cu precipitates in the HAZ, transmission electron microscopy (TEM) was employed. Thin foil samples selected from the actual HAZ were prepared by jet polishing in a Struers Tenupol using a 5% perchloric acid-95% acetic acid solution. Foils were examined in a Jeol 2000FX electron microscope, after which the discs were etched in 2.5% nital and examined by optical microscopy to determine the precise location in the HAZ of the region sampled by TEM.

Scanning electron microscopy (SEM) was used to study the fine fractographic features of tensile and Charpy fracture surfaces. The specimens were ultrasonically cleaned in acetone and examined in an Hitachi S450 scanning electron microscope.
CHAPTER 6

EXPERIMENTAL RESULTS
6.1 INTRODUCTION

This Chapter presents the results of experimental work described in Chapter 5. The experimental results are discussed in five sections. The first four sections (6.2-6.5) cover the results obtained from four types of HAZs. Section 6.2 describes the results of the bead-on-plate (BOP) submerged arc (SA) welds; Section 6.3, the BOP flux-cored arc (FCA) welds; and Section 6.4 the 4 wire SA welds. Section 6.5 presents the experimental data for simulated single and reheated HAZs.

The effect of postweld heat treatment (PWHT) on the properties of the simulated grain coarsened HAZ region is described in Section 6.6, separately from the results of simulated HAZs (Section 6.5). Finally, the partial continuous cooling transformation (CCT) diagrams of the grain coarsened HAZ regions for the CR HSLA 80 and LCE 350 steels are presented in Section 6.7. A comparison of both diagrams is made to enable assessment of the effect of copper and nickel contents on $\gamma \rightarrow \alpha$ transformation in the HAZ.

6.2 BEAD-ON-PLATE SUBMERGED ARC WELDS

6.2.1 Microstructure of Various Regions

The HAZ is usually divided into five sub-zones or regions according to the peak temperature experienced in different positions of the HAZ during welding. The characteristics of each region are reviewed in Chapter 2, Section 2.3.

Figure 6.1 shows the typical microstructures of various HAZ regions for the CR HSLA 80 steel at 2.5kJ/mm heat input. Figure 6.1f gives the microstructure of the GCHAZ at a
Fig. 6.1 Various regions of the HAZ (BOP SA weld HAZ, 2.5kJ/mm).
(a) grain coarsened region; (b) grain refined region;
(c) partially transformed region; (d) tempered region;
(e) unchanged base metal (320x)(root of weld bead);
(f) GCHAZ at shoulder of weld bead (200x).
lower magnification for the same sample. It can be seen that the HAZ microstructure of the CR HSLA 80 steel is basically ferritic in nature. In the grain coarsened region (GCHAZ)(Fig. 6.1a), the structure consists of four distinct microstructural features: packets of coarse parallel ferrite laths, quasi-polygonal ferrite, acicular ferrite and second phase islands.

Transmission electron microscopic (TEM) analysis confirmed that the second phase islands are martensite-austenite (MA) constituents and consist of martensite, retained austenite and cementite (31). Transformation twins (Fig. 6.2a) were observed in some martensitic regions, indicating their formation from a relatively high carbon austenite (>0.4%C)(210). Similar twinned MA islands have been reported by many researchers (67,135,136,211). Small precipitate particles of cementite indicated in Fig. 6.2b are likely to be the result of autotempering of high carbon martensite after its formation during the weld cooling cycle.

The mixture of various types of ferrite with second phases is sometimes referred as 'granular structure' in the case of quasi-polygonal ferrite and acicular ferrite, and 'granular bainite' in the case of bainitic ferrite (212). The coarse laths shaped ferrite with MA islands in the present case may therefore be classified as 'granular bainite' (213,214,215), and is different from the mixture of ferrite laths and interlath cementite associated with upper bainite. Other workers would classify it as B1 type bainite (78) or B3 (150). A variety of granular structures, consisting of different ferrite matrices with second and third constituents, are often observed in TMCP steels (212).

The grain refined region of the HAZ (Fig. 6.1b) contained fine equiaxed ferrite grains (or polygonal ferrite) and MA islands. Compared to the GCHAZ, the finer structure is the result of the formation of fine recrystallized γ grains at a lower peak temperatures ranging from about 900°C to 1100°C (Section 2.3).
Fig. 6.2  Transmission electron micrographs of MA islands in the grain coarsened region (BOP SA weld HAZ, 6kJ/mm).
(a) high magnification bright field showing transformation twins;
(b) dark field corresponding to g=(110)_Fe3C in same region as (a);
(c) selected area diffraction (SAD) pattern of twinned M in (a);
(d) indexing of (c).
In the partially transformed region (Fig. 6.1c), the structure consisted of fine polygonal ferrite bands and bands of islands of higher carbon constituents which include untransformed but degenerated pearlite aggregates and MA islands. The banded structure obviously arose from the banding in the base material. The small MA islands resulted from the partially austenitised regions and were concentrated in layers corresponding to the original pearlite bands in the base plate. The pearlite region of the base plate was enriched in carbon and alloying elements and was the first region to revert to austenite on heating (lower A(c3) temperature). In the absence of homogenization during the thermal cycle, this region also has the lowest A(r3) temperature on cooling. Undercooling of ferrite is therefore promoted, favouring transformation to martensite.

The tempered region (Fig. 6.1d) has a similar structure to the base plate, except for degeneration of the lamellar pearlite in the base plate to spheroidal particles of Fe3C formed as a result of experiencing a weld thermal cycle with a peak temperature in the range of 650-750°C (Section 2.3.4).

The 'unchanged' region (Fig. 6.1e) adjacent to the tempered region has the same structure as the base plate which consists of polygonal ferrite and pearlite bands. Although no change in microstructure was apparent at an optical level, weld thermal cycling to a peak temperature below 650°C influences Cu precipitates on a submicroscopic scale, altering hardness and toughness properties in this region, as discussed in following sections.

TEM revealed that ε-Cu precipitates were present in the HAZ (Fig. 6.3). Precipitates in the aged base plate are shown for comparison in Fig. 6.3a. The Kurdjumov-Sachs (K-S) orientation relationship was confirmed to exist between ε-Cu particles and the ferrite matrix (Fig. 6.3e), indicating that Cu precipitation had occurred in ferrite. Since complete austenization occurred in the grain coarsened and refined regions, the presence of fine ε-Cu precipitates in both regions (Figs 6.3b and 6.3c) indicates that some Cu re-precipitation took place during cooling.
Fig. 6.3 ε-Cu precipitates in (a) the aged base plate; (b) grain coarsened region; (c) grain refined region; (d) partially transformed region. (e) Diffraction pattern consistent with the K-S orientation relationship between copper particles and ferrite matrix in (d) (BOP SA weld HAZ, 4kJ/mm); (f) indexing of (d).
6.2.2 Quantitative Metallography of MA Islands across the HAZ

Since the MA islands are detrimental to toughness (Section 3.4.2.2), a quantitative metallographic study of these islands was carried out in the present work.

The volume fraction of MA islands as a function of distance from the fusion line in each HAZ of BOP SA welds is shown in Fig. 6.4. The volume fraction decreased from the fusion line towards the base metal. The higher volume fraction near the fusion line is consistent with the coarser austenite grain size due to the higher peak temperature experienced, slightly faster cooling rate and therefore, higher hardenability in this region of the HAZ, compared with the grain refined region. Overall, the variation of volume fraction with the distance from the fusion line was small. The volume fractions of MA islands across the HAZ for the three welds were in the range of 4.5-6.5%.

Assuming that C partitioning takes place between ferrite and austenite during cooling and that the C content of the ferrite is about 0.02%, the estimated C level of MA island is about 0.6% or higher which is 12 times the average C content of the steel (0.05%). This calculation provides further support for the conclusion that a high carbon content exists in the MA islands.

Changes in number density (number/area) of MA islands with distance from the fusion line for the three weld heat inputs are shown in Fig. 6.5. With increasing distance from fusion line, the number density decreased, indicating that because of smaller austenite grain size, slightly lower cooling rate and thus lower hardenability, fewer MA islands were formed in the grain refined region. This observation is also consistent with the decrease in the volume fraction of the islands with distance from the fusion line (Fig. 6.4).
Fig. 6.4 Volume percentage of MA islands as a function of distance from the fusion line of BOP SA weld HAZs.
Fig. 6.5 Number density ($\mu$m$^{-2}$) as a function of distance from the fusion line of BOP SA weld HAZs.
Although maximum lengths up to 15μm were measured for particles in the grain coarsened region, the majority of the MA islands had maximum dimensions less than a few microns.

6.2.3 Hardness Traverses

Hardness traverses across the HAZs for the three welds were measured at three different positions (Fig. 5.13). Figure 6.6 summarises the results and shows a matrix of hardness-distance graphs corresponding to the three sampling positions and the three heat inputs. The local width of the HAZ and the subzones are also marked on each graph. In this case, the width was determined microstructurally in terms of the boundary between modified and unmodified structure, i.e. the termination of the tempered region.

Figure 6.6 shows that the peak hardness of the HAZ occurred in the grain coarsened region near the fusion line. Unlike many other steels for which the whole range of the HAZ is harder than the base metal (Section 3.3), hardness gradients in the Cu-bearing CR HSLA 80 steel show a minimum value which is well below the hardness of the base plate. The minimum hardness occurred in the partially transformed region. The low hardness zone also extended beyond the apparent microstructural boundary of the HAZ (termination of the tempered region) to unchanged base material regions, indicating that the hardness of the HAZ is altered submicroscopically by overaging of the copper-rich precipitates.

The widths of HAZ soft zones where the hardness is lower than that of base plate were measured at the three locations of HAZ for different heat inputs. As can be seen in Fig. 6.7, from a→b→c, the width of the low hardness region increased, similar to the width of HAZ. Higher heat input resulted in a wider HAZ soft zone.
Fig. 6.6 Hardness traverses at positions a, b and c for the three BOP SA weld HAZs. First row: 2.5kJ/mm; second row: 4kJ/mm and third row: 6kJ/mm.

1 - grain coarsened region; 2 - grain refined region; 3 - partially transformed region; 4 - tempered region.
Fig. 6.7 Width of HAZ softened zone as a function of heat input at three positions (a, b and c) for BOP SA welds.
6.2.4 The Effect of Heat Input

6.2.4.1 Macro- and micro- structures of HAZ

Figure 6.8 shows the macroetched weld cross-sections for three different heat inputs. As can be seen the three welds showed similar weld profiles. The higher heat input was associated with a larger volume of weld deposit, a longer fusion line length and a wider HAZ.

The widths of the HAZs perpendicular to the fusion line are recorded in Fig. 6.9 as a function of distance along the fusion line of the three welds. Significant variation in HAZ width can be seen along the fusion line (Fig. 6.9). The ratio of the maximum and minimum widths of the HAZ is about 2.5:1 for a given heat input, with the minimum width occurring at the root of the weld bead and the maximum width occurring at the shoulder of the bead (Fig. 6.8). Figure 6.9 also indicates that a higher heat input is associated with a wider average HAZ which is consistent with the macrographs shown in Fig. 6.8.

The microstructure of the HAZ changed with heat input. Of the various regions of the HAZ, the GCHAZ showed the most marked change with heat input. The GCHAZ is usually regarded as the most dangerous region of the HAZ, because embrittlement can occur in this region as a result of the formation of unfavorable brittle constituents (Chapter 2, Section 2.3.1). Therefore, the present study of the effect of welding parameters on HAZ microstructure and properties has been concentrated on the GCHAZ region.
Fig. 6.8 Macrographs of weld cross-sections for BOP SA welds.

Fig. 6.9 HAZ width along the fusion line of BOP SA welds.
Figure 6.10 shows the optical and SEM microstructure of the GCHAZ for the BOP SA weld HAZs at different heat inputs. It can be seen that a lower heat input resulted in a finer structure due to faster cooling rate of the HAZ during welding. At the highest heat input (6kJ/mm), grain boundary ferrite with aligned MA islands is observed.

From Fig. 6.4, it can be seen that similar volume fractions of MA islands were present in regions near the fusion line (~6%) for the three different heat inputs. This result indicates that heat input in this case had little effect on the volume fraction of MA islands in the HAZ.

In contrast, the number density of MA islands increased with decrease in heat input from 4 to 2.5kJ/mm (Fig. 6.5). The mean maximum dimension of the MA islands averaged over the grain coarsened and grain refined regions are shown in Fig. 6.11 as a function of heat input. The size of the islands increased from 2.5 to 4 kJ/mm heat input.

As there was little change with heat input of the volume fraction of MA in the grain coarsened region, the higher value of number density for the lowest heat input (Fig. 6.5) clearly shows that the major change in microstructure with decreasing heat input is the increasing level of structural refinement, consistent with the microstructures shown in Fig. 6.10. This conclusion is also consistent with the evidence of smaller mean maximum dimension of MA particles for the lowest heat input (Fig. 6.11). For low heat input, the higher cooling rate and lower "residence time" at high temperatures, where rapid grain growth can occur, collectively result in a smaller austenite grain size. The hardenability is increased by faster cooling, but is reduced by decreasing austenite grain size, so that no major increase in the volume fraction of MA island was obtained for the lowest heat input (2.5kJ/mm). However, both fast cooling and fine austenite grain size enhance the nucleation of diffusional products, resulting in transformation structures which are finer than those present in the grain coarsened region of the higher heat input welds.
Fig. 6.10  Low magnification optical microstructures (a) of GCHAZ at different heat inputs for BOP SA welds (at the shoulder of the weld bead) (200x). High magnification SEM micrographs (b) showing the MA islands of GCHAZ at three heat inputs (at the root of weld bead) (770x).
Fig. 6.11 Relationship between mean maximum dimension of MA islands and heat input for BOP SA weld HAZ.
The data for volume fraction (Fig. 6.4), particle density (Fig. 6.5) and mean maximum particle length (Fig. 6.11) indicate that there was little difference in the characteristics of the MA islands for 4 and 6 kJ/mm heat input, but that significant refinement occurred for the 2.5kJ/mm weld.

Linear intercept prior austenite grain sizes were measured along the fusion line for each BOP SA weld HAZ. Figure 6.12 shows the distribution of grain size along the fusion line at three heat inputs. As can be seen, the grain sizes measured at the shoulder of the weld bead (maximum HAZ width) were scattered, but the average value at this position was larger (134.8µm) than at the root of the bead (81.3µm). Together with the marked variation of HAZ width along the fusion line, this result indicates that profoundly different local cooling rates exist along the fusion line in the HAZ.

The average prior austenite grain sizes along the fusion line for three welds are plotted against heat input in Fig. 6.13. It is evident that the coarser grain size in the HAZ resulted from higher heat input welding.

6.2.4.2 Peak hardness

Peak hardnesses of the HAZs were measured in the GCHAZ near the fusion line at three HAZ positions marked in Fig. 5.13 for all three welds. Each value is the average of 5 measurements and 95% confidence limits have been calculated for each average value.

For the BOP SA welds, peak hardness of the HAZ decreased with increase in the heat input for the corresponding positions (a, b and c). Figure 6.14 shows the change of peak hardness with heat input for position c. For all the weld HAZs, the maximum hardnesses were well below the limiting hardness value (350HV) for prevention of cold cracking (Section 3.2.2).
Fig. 6.12  Distribution of prior austenite linear intercept grain sizes along the fusion line of BOP SA weld HAZs. (a) shoulder of weld bead (maximum HAZ width); (b) root of weld bead (minimum HAZ width).
Fig. 6.13  Average austenite grain size along the fusion line as a function of heat input for BOP SA weld HAZs.

Fig. 6.14  Peak hardness of HAZ as a function of heat input at position c for the BOP SA welds. Bars represent 95% confidence limits of the means.
6.2.5 Second Thermal Cycle Reheating of HAZ

The second thermal pulse heat treatment of BOP SA welds was described in Chapter 5, Section 5.4. The cooling time $\Delta t_{8-5}$ of the second thermal pulse was 80 seconds, which is typical of high heat input and an associated slow cooling rate.

Figure 6.15 shows a comparison of the microstructures of the grain-coarsened region of the HAZ before and after reheating to a peak temperature ($T_{p2}$) of 800°C (intercritical reheating). After the reheating, the grain-coarsened and grain-refined regions of the three welds showed many small MA islands aligned in bands which corresponded to the positions of the original pearlite bands in the base plate. Higher magnification photomicrographs (Fig. 6.16) showed that these second phase particles were mainly elongated in cross-sectional shape. Transmission electron microscopy confirmed the tendency towards banding of these particles (Fig. 6.17) and the presence of twinned high carbon martensite in the particles (Fig. 6.18). The decreasing prominence of the bands of MA islands with increasing heat input is evident from Fig. 6.15.

There was little change in the HAZ structure at an optical microscopic level after subcritical reheating ($T_{p2}=600^\circ C$). However, TEM analysis revealed (Fig. 6.19) that the original MA islands in this region were tempered by the reheating.

Figure 6.20 shows the hardness gradients across the HAZ for the original weld and for the superimposed heat treatments to peak temperatures of 800°C and 600°C. It should be noted that the last two hardness gradients do not simulate those expected for a second welding pass, as the whole range of the HAZ was subjected to the thermal cycle with the same peak temperature ($T_{p2}$ of 800°C or 600°C). The graphs indicate, however, the change in hardness to be expected if a particular point in the original HAZ was subjected to the imposed thermal cycle. For example, at a point 3mm into the HAZ from the fusion
Fig. 6.15  Microstructures of grain coarsened region of HAZ (a) initial BOP SA weld and (b) after reheating to a peak temperature of 800°C.
Fig. 6.16  High magnification micrograph showing the shape and distribution of MA islands in intercritically reheated GCHAZ (BOP SA weld HAZ at 4kJ/mm).

Fig. 6.17  TEM micrograph showing the tendency towards banding of MA islands in the GCHAZ after intercritical reheating (BOP SA weld at 4kJ/mm).
Fig. 6.18  Twinned martensite in intercritically reheated GCHAZ (BOP SA weld at 4kJ/mm).

Fig. 6.19  TEM micrograph showing the tempered MA islands in subcritically reheated HAZ (BOP SA weld at 4kJ/mm).
Fig. 6.20 Hardness traverses of HAZ for single pass BOP SA welds before and after reheating to peak temperatures of 600°C (subcritical) and 800°C (intercritical).

1-grain coarsened region; 2-grain refined region; 3-partially transformed region; 4-tempered region.
line for the 2.5 kJ/mm weld, both the Tp2 of 800°C and 600°C treatments raised the hardness by at least 10 points.

The effect of the second thermal cycle on the base plate hardnesses is also indicated by Fig. 6.20. For Tp2 of 600°C, the hardness level was raised about 10 points above the original average hardness. This result indicates that the 550°C aging treatment for 1/2 hour following control rolling does not produce maximum hardening and that the superimposed thermal spike to a 600°C peak temperature gives an additional hardening increment. In contrast, intercritical treatment to 800°C resulted in softening in the base plate to levels similar to the minimum produced in the HAZ by the single pass weld.

For the 4.0 and 6.0 kJ/mm samples, there was no significant change in the hardness gradient in the GCHAZ region after reheating to 600°C. However, for the 2.5 kJ/mm heat input, the second thermal cycle with peak temperature of 600°C resulted in an increase in hardness of the grain-coarsened region of this sample (Fig. 6.20a). Reheating to 800°C did not have any significant effect on GCHAZ hardness.

6.3 BEAD-ON-PLATE FLUX-CORED ARC WELDS

Figure 6.21 shows the macroetched weld cross-sections for the BOP FCA welds for three heat inputs. As can be seen the welds showed similar weld profiles at different heat inputs. However, the weld profiles of BOP FCA welds were greatly different compared to BOP SA welds (Fig. 6.8). A higher heat input was associated with a larger volume of weld deposit, a longer fusion line length and a wider HAZ, similar to the BOP SA welds.

The widths of the HAZs perpendicular to the fusion line are also recorded in Fig. 6.22 as a function of distance along the fusion line. For the FCA welds, the variations of HAZ width along the fusion line are quite different to the SA welds, reflecting the difference in
Fig. 6.21  Macrographs of weld cross-sections of BOP FCA welds.

Fig. 6.22  HAZ width along the fusion line for BOP FCA welds.
weld profiles observed in Fig. 6.21. The ratios of maximum to minimum HAZ width are different at different heat inputs and are 2.1:1, 1.3:1 and 1.4:1 for 1.0, 1.5 and 2.5kJ/mm, respectively. The measurement of HAZ width (Fig. 6.22) indicates that higher heat input resulted in a wider HAZ, consistent with the macrographs in Fig. 6.21 and with the results obtained for BOP SA welds.

Figure 6.23 shows the optical microstructure of the GCHAZ for BOP FCA weld HAZs at three different heat inputs. The structures of the GCHAZ for the BOP FCA weld HAZs are similar to those for BOP SA welds (Fig. 6.10). It is evident that a lower heat input resulted in a finer structure. Compared to the BOP SA welds, the BOP FCA welds showed some dark etching structural constituents in the GCHAZ (marked in Fig. 6.23) which contained a very fine lath structure. Lower heat input resulted in a higher percentage of this dark etching structure. Transmission electron microscopy (TEM) revealed that the structure consisted of finely distributed sheaf-like ferrite (Fig. 6.24), consistent with low carbon lath martensite (212). In general, the structure of the GCHAZ, for low heat input and the associated fast cooling rate, consists of a mixture of granular bainite and lath martensite (212).

The average prior austenite grain sizes in the GCHAZ along the fusion line for BOP FCA welds are plotted against heat input in Fig. 6.25. Much lower prior austenite grain sizes were obtained from FCA welds over those of SA welds due to the faster cooling rates associated with the lower heat inputs. Consistent with the results for the BOP SA welds, high heat input resulted in a coarser grain size in the HAZ.

Peak hardnesses of the HAZs were measured in the GCHAZ near the fusion line at maximum HAZ width for the welds. As for the BOP SA welds, the peak hardness of the HAZ decreased with increase in the heat input (Fig. 6.26). The BOP FCA welds showed a much more significant change in peak hardness with heat input than the BOP SA welds.
Fig. 6.23  Microstructures of GCHAZ region for different heat inputs in BOP FCA welds (256x). The arrow indicates the dark etching structure.
Fig. 6.24  TEM micrographs showing low carbon martensite in HAZs of BOP FCA welds.
Fig. 6.25  Average austenite grain size along the fusion line as a function of heat input for BOP FCA weld HAZs.

Fig. 6.26  Peak hardness of HAZ as a function of heat input for the BOP FCA welds. (Bars show the 95% confidence limits)
6.4 FOUR WIRE SUBMERGED ARC WELDS

6.4.1 The Effect of Heat Input

Figure 6.27 shows the macroetched weld cross-sections for 4 wire SA welds. The weld profiles of the 4 wire SA welds were similar to those of the BOP SA welds. Higher heat input was also associated with a larger volume of weld deposit, a longer fusion line length and a wider HAZ.

Optical microstructures of the GCHAZ at different heat inputs in Figure 6.28 indicate that at the same welding speed, a lower heat input resulted in a finer structure in the HAZ. This observation is consistent with the results obtained from BOP SA and FCA welds.

Quantitative measurement of MA islands was also carried out on 4 wire SA welds. Similar results were obtained to those for BOP SA welds, in that variation in welding conditions (heat input and welding speed) appeared to have little effect on the volume fraction of MA islands in the GCHAZ (Fig. 6.29a). The volume fractions of MA islands in the GCHAZ for all the welding conditions (three heat inputs and three welding speeds) were within the range of 3-6%. However, as heat input decreased, the average maximum dimension of the MA islands decreased (Fig. 6.29b), consistent with the measurements for BOP SA welds (Fig. 6.11). These results confirm that a finer HAZ structure is obtained at lower heat input.

The average linear intercept prior austenite grain sizes along the fusion line for the 4 wire SA welds (averaged over the three speeds) are plotted against heat input in Fig. 6.30. It is evident that a coarser grain size in the GCHAZ results from higher heat input welding. The two types of SA welding procedures (BOP and 4 wire SA welds) showed similar prior austenite grain sizes in the HAZ, whereas the FCA welds showed much smaller austenite grain sizes (Fig. 6.31).
Fig. 6.27 Macrographs of weld cross-sections of 4 wire SA welds in 20mm plate.
Fig 6.28  Comparison of GCHAZ microstructures for different heat inputs in 4 wire SA welds at 1000mm/min welding speed on 20mm plate (256x). (a) 2.5kJ/mm; (b) 10kJ/mm.
Fig. 6.29 Volume fraction (a) and maximum dimension (b) of MA islands in HAZ near the fusion line at the shoulder of weld bead of 4 wire SA welded 20mm CR HSLA 80 steel plate.
Fig. 6.30  Average austenite grain size along the fusion line as a function of heat input (averaged over the three welding speeds) for 4 wire SA weld HAZs.

Fig. 6.31  Average austenite grain size along the fusion line as a function of heat input for the three types welds (BOP SA, BOP FCA and 4 wire SA).
Peak hardness of the HAZ of 4 wire SA welding decreased with increase in the heat input (Fig. 6.32). The result at each heat input is the average value over three welding speeds. Table 6.1 and Fig. 6.33 summarize the peak hardmesses for the three types of welding processes. Compared to the two types of SA welding processes, the BOP FCA welds showed a more significant change in peak hardness with heat input than the SA welds. The 4 wire SA welds exhibited lower peak hardmesses than the BOP SA welds, probably as a result of lower cooling rates in the multiple wire welds of thinner plate (20mm versus 36mm).

Although lower heat input results in a harder HAZ, relative to the hardness of base plate, the hardening of the HAZ is rather mild for the 4 wire SA welds. The 2.5kJ/mm HAZ in the 4 wire SA weld showed a peak hardness slightly higher than the hardness of the base plate (Table 6.1), but the peak hardness for the 5 and 10kJ/mm heat input welds were lower than that of the base plate. For all the weld HAZs, the maximum hardmesses were well below the limiting hardness value (350HV) for prevention of cold cracking (Section 3.2.2).

Charpy impact tests were conducted on 4 wire SA weld HAZs of 20 and 50mm plates by CSIRO DMT at -51°C and -20°C, respectively (216,217). Individual CVN values for all weld HAZs are listed in Table 6.2, together with the calculated average values for each HAZ. The average CVN values are also plotted in Fig. 6.34 and 6.35 for 20 and 50mm plates, respectively. It should be noted that in this case, the CVN values obtained for the HAZs were obtained using a notch running across the structural gradient of the HAZ, as difficulty exists in locating the notch tip within a narrow region of the weld HAZ (Section 3.4.1). The CVN results are average values across the HAZ. Both Figs. 6.34 and 6.35 show that toughness of the HAZ did not vary markedly with heat input in the range of 2.5 to 10kJ/mm.
Fig. 6.32  Peak hardness of HAZ (averaged over three welding speeds) as a function of heat input for the 4 wire SA welds. Bars represent 95% confidence limits of the means.
Table 6.1 HAZ peak hardnesses of three types of welds

<table>
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<th>Types of welds</th>
<th>Heat input (kJ/mm)</th>
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<th>95% confidence limit</th>
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<tbody>
<tr>
<td>BOP, FCA</td>
<td></td>
<td></td>
<td></td>
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<tr>
<td>36mm</td>
<td>1.0</td>
<td>309.4</td>
<td>289.6-329.2</td>
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<tr>
<td></td>
<td>1.5</td>
<td>298.2</td>
<td>291.7-302.7</td>
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<td>2.5</td>
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<td>246.6-281.6</td>
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<td>248.2</td>
<td>242.5-253.9</td>
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<td>4.0</td>
<td>240.7</td>
<td>231.9-249.5</td>
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<td>6.0</td>
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<td>20mm</td>
<td>2.5</td>
<td>226.2</td>
<td>220.6-231.8</td>
</tr>
<tr>
<td></td>
<td>5.0</td>
<td>218.6</td>
<td>213.5-223.8</td>
</tr>
<tr>
<td></td>
<td>10.0</td>
<td>213.2</td>
<td>205.6-220.8</td>
</tr>
<tr>
<td>base plate</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>36mm</td>
<td></td>
<td>218.0</td>
<td>205.6-230.4</td>
</tr>
<tr>
<td>20mm</td>
<td></td>
<td>221.7</td>
<td>218.6-224.8</td>
</tr>
</tbody>
</table>
Fig. 6.33  Peak hardness of HAZ as a function of heat input for the three types of welds (BOP SA, BOP FCA and 4 wire SA).
Table 6.2 Charpy-V notch impact energy of 4 wire SA welds (ref.216,217)

<table>
<thead>
<tr>
<th>Plate thickness (mm)</th>
<th>Heat input (kJ/mm)</th>
<th>Welding speed (mm/min)</th>
<th>CVN values (J)</th>
<th>Mean CVN values (J)</th>
</tr>
</thead>
<tbody>
<tr>
<td>20mm (-51°C)</td>
<td>2.5</td>
<td>1000</td>
<td>146, 89</td>
<td>118</td>
</tr>
<tr>
<td></td>
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<td>1500</td>
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<td>180</td>
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<td>2000</td>
<td>210, 216, 214</td>
<td>210</td>
</tr>
<tr>
<td></td>
<td>5</td>
<td>1000</td>
<td>178, 155</td>
<td>167</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1500</td>
<td>90, 126, 130, 87</td>
<td>108</td>
</tr>
<tr>
<td></td>
<td></td>
<td>2000</td>
<td>172, 118, 147, 164</td>
<td>150</td>
</tr>
<tr>
<td></td>
<td>10</td>
<td>600</td>
<td>47, 95, 102, 72</td>
<td>79</td>
</tr>
<tr>
<td></td>
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<td>170, 170, 148</td>
<td>163</td>
</tr>
<tr>
<td></td>
<td></td>
<td>1300</td>
<td>118, 114, 60</td>
<td>97</td>
</tr>
<tr>
<td>50mm (-20°C)</td>
<td>2.5</td>
<td>1000</td>
<td>251, 255, 227, 208</td>
<td>235</td>
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<td></td>
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<td>1500</td>
<td>240, 261, 228, 206</td>
<td>234</td>
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<td></td>
<td></td>
<td>2000</td>
<td>226, 223, 234, 220</td>
<td>226</td>
</tr>
<tr>
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<td>5</td>
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<td>210, 189, 207, 194</td>
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<td></td>
<td></td>
<td>1500</td>
<td>174, 183</td>
<td>179</td>
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<td></td>
<td>2000</td>
<td>207, 246</td>
<td>227</td>
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<tr>
<td></td>
<td>10</td>
<td>600</td>
<td>194, 204, 210, 196</td>
<td>201</td>
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<td></td>
<td></td>
<td>1000</td>
<td>166, 198, 164, 176</td>
<td>176</td>
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<td></td>
<td></td>
<td>1300</td>
<td>186, 240, 200, 186</td>
<td>203</td>
</tr>
</tbody>
</table>
Fig. 6.34 CVN values at -51°C for 4 wire SA weld HAZs as a function of welding speed at three heat inputs for 20mm CR HSLA 80 plate (ref.216, 217).

Fig. 6.35 CVN values at -20°C for 4 wire SA weld HAZs as a function of welding speed for three heat inputs for 50mm CR HSLA 80 plate (ref.217).
Tensile properties were measured for the 4 wire SA welds of 20mm plate. Of all the welding conditions, four contained defects. These defects were either lack of fusion or slag entrapped in the weld metal. Tensile test specimens of these welds failed at the defects, showing relatively low strength values. These results are not considered here. Table 6.3 shows the tensile results for the remaining welds.

For heat inputs of 2.5 and 5kJ/mm, the welds showed a similar yield strength to the base plate, but a slightly lower tensile strength. However, the welds at 10kJ/mm exhibited lower yield strength values than the base plate. The values were also below the minimum requirement (80ksi or 550MPa) for the HSLA 80 steel. Figure 6.36 shows that the width of the HAZ, which has a hardness lower than that of the base plate, increased with heat input. The wider softened HAZ associated with higher heat input is the likely reason for the lower strength.

Examination of the failure location of each weld indicated that all samples failed at the HAZ where softening occurred. Each sample exhibited a cup and cone fracture appearance, consisting of a central cup region and a shear lip outer region, as shown in Fig. 6.37. Both regions showed transgranular fracture by microvoid coalescence (dimpled appearance)(Fig. 6.38a and 6.38b), but few inclusions were found inside the dimples. The rough central area also indicated that the fracture was very ductile, which correlates with the large values of reduction of area in Table 6.3. The dimple size varied significantly in the central region (Fig. 6.38a), but the outer shear lip region showed dimples of similar size.
Table 6.3 Tensile properties of 4 wire SA welded HSLA 80 plate (20mm)

<table>
<thead>
<tr>
<th>H.I (kJ/mm)</th>
<th>W.S (mm/min)</th>
<th>UYS (MPa)</th>
<th>UT (MPa)</th>
<th>El. (%)</th>
<th>ROA (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>10</td>
<td>1000</td>
<td>534, 524</td>
<td>576, 598, 601</td>
<td>20, 20, 21</td>
<td>76, 76, 76</td>
</tr>
<tr>
<td></td>
<td>1300</td>
<td>510, 517</td>
<td>590, 597, 601</td>
<td>20, 21, 21</td>
<td>74, 72, 70</td>
</tr>
<tr>
<td>5</td>
<td>1500</td>
<td>609, 589, 607</td>
<td>638, 635, 635</td>
<td>18, 18</td>
<td>78, 76, 76</td>
</tr>
<tr>
<td></td>
<td>2000</td>
<td>608</td>
<td>639, 640, 644</td>
<td>17, 17</td>
<td>76, 72, 70</td>
</tr>
<tr>
<td>2.5</td>
<td>2000</td>
<td>608, 574</td>
<td>657, 660, 654</td>
<td>19, 20</td>
<td>73, 70</td>
</tr>
<tr>
<td>base</td>
<td></td>
<td>602, 598, 604</td>
<td>672, 668, 675</td>
<td>28, 24, 30</td>
<td>71, 71, 70</td>
</tr>
</tbody>
</table>

Fig. 6.36  Hardness traverses across the HAZs of 4 wire SA welds in 20mm plate. (a) 2.5kJ/mm, 1500mm/min; (b) 5kJ/mm, 1500mm/min; (c) 10kJ/mm, 1300mm/min.
Fig. 6.37  Cup and cone fracture appearance of weld transverse tensile specimen (HSW 20, UYS = 534 MPa).

Fig. 6.38  SEM fractographs showing the dimpled appearance at the (a) central region (440x) and (b) side shear lip (330x) in the tensile fracture surface (HSW20, UYS=534MPa).
6.4.2 Effect of Welding Speed on the Structure and Properties of 4 Wire SA Welds

The effect of welding speed on HAZ microstructure and mechanical properties was studied by comparing the results of welds produced at different welding speeds at constant heat input for 4 wire SA welding.

As shown in Fig. 6.27, there was no significant change in weld profile with welding speed. However, a slight reduction in HAZ width with increasing speed is evident in Fig. 6.27. The weld cross-section for the lowest welding speed at each heat input showed an internal boundary within the weld bead (Fig. 6.39), indicating that instead of a single weld bead, some solidification occurred on passage of the previous electrodes before passage of and remelting by the following electrodes.

Linear intercept HAZ prior austenite grain sizes along the fusion line of 4 wire SA welds of 20mm plate are plotted as a function of welding speed in Fig. 6.40. Except for low values at the lowest welding speeds for 2.5 and 5kJ/mm heat inputs, the general trend with increasing welding speed is towards smaller austenite grain size.

The unexpectedly small austenite grain size at the slowest welding speed for 2.5 or 5kJ/mm heat inputs may be related to the additional fusion lines within the weld bead as shown in Fig. 6.39. The internal fusion lines are associated with the complexity of the 4 wire welding procedure and will be explained in the next Chapter.

Both optical microstructural observations and quantitative metallography of the MA islands (Fig. 6.29) showed that no significant HAZ microstructural change occurred with change of welding speed at constant heat input, despite the reported slight cooling rate increase with increasing welding speed (203).
Fig. 6.39 Optical micrograph showing the additional fusion lines in the weld bead of last pass of 4 wire SA weld at 2.5kJ/mm heat input and 1000mm/min speed (2.5% nital etched, 16x).

Fig. 6.40 Average prior austenite linear intercept grain size of the HAZ along the fusion line as a function of welding speed for 4 wire SA welded 20mm plate.
Table 6.4 lists peak hardnnesses of 4 wire SA weld HAZs of 20mm plate for different heat inputs and welding speeds. These are averages based on 5 measurements and the 95% confidence limits for these means are also shown in Table 6.4. It can be seen that peak hardness of the HAZ is not very sensitive to changing welding speed at constant heat input.

As Figs. 6.34 and 6.35 show, HAZ toughness is similar to peak hardness in that it is not sensitive to change in welding speed for both 20 and 50mm thick plates.

There is insufficient data in Table 6.3 to indicate how the welding speed affects the tensile properties of the weld. However, no significant difference in transverse tensile properties existed for the two different welding speeds at 10 and 5kJ/mm, which suggests that only a minor change in tensile properties is to be expected with changing welding speed.

6.5 SIMULATED HEAT AFFECTED ZONE

The Charpy impact energy and hardness results obtained from simulated HAZ samples are presented in the following sections. They are firstly summarized in Table 6.5-6.8 which covers the results: for simulated single weld HAZs heated to different peak temperatures (Table 6.5); for second and third cycle reheated GCHAZ (Table 6.6); for simulated GCHAZ at different heat inputs (Table 6.7); and for the PWHT treated GCHAZs (Table 6.8). These Tables provide the detailed data source for all the figures related to the simulated HAZ. Fibrosity and lateral expansion of some simulated HAZ Charpy samples are also listed in these Tables.
Table 6.4 HAZ peak hardness of 4 wire SA welded 20mm CR HSLA 80 steel

<table>
<thead>
<tr>
<th>Heat input (kJ/mm)</th>
<th>Welding speed (mm/min)</th>
<th>HVmax (5kg load)</th>
<th>95% confidence limit</th>
</tr>
</thead>
<tbody>
<tr>
<td>2.5</td>
<td>1000</td>
<td>228.6</td>
<td>222.7-234.5</td>
</tr>
<tr>
<td></td>
<td>1500</td>
<td>229.9</td>
<td>223.4-236.4</td>
</tr>
<tr>
<td></td>
<td>2000</td>
<td>220.2</td>
<td>215.9-224.5</td>
</tr>
<tr>
<td>5</td>
<td>1000</td>
<td>221.8</td>
<td>214.2-229.4</td>
</tr>
<tr>
<td></td>
<td>1500</td>
<td>218.5</td>
<td>213.3-223.7</td>
</tr>
<tr>
<td></td>
<td>2000</td>
<td>215.6</td>
<td>213.0-218.2</td>
</tr>
<tr>
<td>10</td>
<td>600</td>
<td>215.7</td>
<td>207.3-224.2</td>
</tr>
<tr>
<td></td>
<td>1000</td>
<td>212.6</td>
<td>206.4-218.8</td>
</tr>
<tr>
<td></td>
<td>1300</td>
<td>211.3</td>
<td>203.2-219.4</td>
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</table>

Table 6.5 Charpy impact properties and hardness of simulated single weld HAZ at equivalent heat input of 2.9kJ/mm (13.3°C/s)

<table>
<thead>
<tr>
<th>Tp1 (°C)</th>
<th>CVN (J) (-51°C)</th>
<th>Mean CVN</th>
<th>Fibrocity (%)</th>
<th>Lat. Exp. (mm)</th>
<th>HV5</th>
</tr>
</thead>
<tbody>
<tr>
<td>600</td>
<td>113, 185, 116</td>
<td>138</td>
<td>98, 98, 100</td>
<td>1.32, 1.84</td>
<td>207</td>
</tr>
<tr>
<td>700</td>
<td>210, 210, 194</td>
<td>205</td>
<td>-</td>
<td>-</td>
<td>197</td>
</tr>
<tr>
<td>800</td>
<td>227, 217, 236</td>
<td>227</td>
<td>100, 100, 100</td>
<td>2.14, 1.9, 2.0</td>
<td>192</td>
</tr>
<tr>
<td>900</td>
<td>235, 223, 225</td>
<td>228</td>
<td>-</td>
<td>-</td>
<td>198.6</td>
</tr>
<tr>
<td>1020</td>
<td>178, 162, 123</td>
<td>154</td>
<td>100, 100, 90</td>
<td>1.66, 1.7, 1.26</td>
<td>215.2</td>
</tr>
<tr>
<td>1300</td>
<td>62, 124, 51</td>
<td>79</td>
<td>30, 60, 30</td>
<td>0.69, 1.32, 0.58</td>
<td>222.4</td>
</tr>
</tbody>
</table>
Table 6.6  Charpy impact properties and hardresses of simulated multi-pass HAZ at equivalent heat input of 2.9kJ/mm (13.3°C/s)

<table>
<thead>
<tr>
<th>Temp. (°C)</th>
<th>CVN (J) (-51°C)</th>
<th>Mean CVN</th>
<th>Fibrosity (%)</th>
<th>Lat. Exp. (mm)</th>
<th>HV5 (av.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tp1 = 1300°C</td>
<td>600 169, 90, 156</td>
<td>138</td>
<td>-</td>
<td>212.6</td>
<td></td>
</tr>
<tr>
<td></td>
<td>800 90, 140, 200</td>
<td>143</td>
<td>45, 55, 70</td>
<td>220.2</td>
<td></td>
</tr>
<tr>
<td></td>
<td>900 175, 153, 160</td>
<td>163</td>
<td>-</td>
<td>208.4</td>
<td></td>
</tr>
<tr>
<td>Tp2 (Tp2 = 800°C)</td>
<td>600 174, 146, 184</td>
<td>168</td>
<td>60, 55</td>
<td>218.4</td>
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Table 6.7  Charpy impact energies and hardresses of simulated GC HAZ (Tp1=1300°C) at various cooling rates (equivalent heat inputs)

<table>
<thead>
<tr>
<th>Cooling rate (°C/s)</th>
<th>Heat input (kJ/mm)</th>
<th>CVN (J) (-51°C)</th>
<th>Mean CVN</th>
<th>Fibrosity (%)</th>
<th>Lat. Exp. (mm)</th>
<th>HV5</th>
</tr>
</thead>
<tbody>
<tr>
<td>30.2 1.9</td>
<td>150, 59, 160</td>
<td>123</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>236.3</td>
</tr>
<tr>
<td>23.8 2.2</td>
<td>125, 70, 151</td>
<td>115</td>
<td>60, 45, 55</td>
<td>1.3, 0.71, 1.51</td>
<td>-</td>
<td>230.9</td>
</tr>
<tr>
<td>13.3 2.9</td>
<td>62, 124, 51</td>
<td>79</td>
<td>30, 60, 30</td>
<td>0.69, 1.32, 0.58</td>
<td>-</td>
<td>222.4</td>
</tr>
<tr>
<td>5.0 4.9</td>
<td>70, 52, 141</td>
<td>88</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>203.2</td>
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Table 6.8  Charpy impact energies and hardesses of simulated GC HAZ after PWHT at various temperatures for 1 hour (equivalent heat input of 3.1kJ/mm)

<table>
<thead>
<tr>
<th>PWHT Temp. (°C)</th>
<th>CVN (J) (-51°C)</th>
<th>Mean CVN</th>
<th>HV5</th>
</tr>
</thead>
<tbody>
<tr>
<td>450</td>
<td>147, 144, 70</td>
<td>120</td>
<td>216.8</td>
</tr>
<tr>
<td>550</td>
<td>49, 19, 19</td>
<td>29</td>
<td>242.7</td>
</tr>
<tr>
<td>650</td>
<td>160, 172, 160</td>
<td>164</td>
<td>209.6</td>
</tr>
</tbody>
</table>
6.5.1 Single Pass HAZ

The microstructures of the various HAZ regions were simulated by applying single thermal cycles with different peak temperatures (600-1300°C) to base material as described in Chapter 5, Section 5.4.3.1. Figure 6.41 shows the simulated HAZ microstructures for various peak temperatures. The cooling rate between 800-500°C was 13.3°C/s which corresponds to a heat input of 2.9kJ/mm. Comparison of the simulated HAZ microstructure (Fig. 6.41) with that of the actual weld HAZ for a similar heat input (Fig. 6.1) suggests that the microstructures of the actual HAZ have been successfully simulated.

The hardness of the single thermal cycle simulated HAZ as a function of peak temperature is shown in Fig. 6.42. The value near zero peak temperature in Fig. 6.41 is the hardness value of the base metal. The bars in the figure represent the 95% confidence limits for the mean values. The hardness trend in Fig. 6.42 is similar to the hardness traverses across the actual HAZs as shown in Fig.6.6. As for Fig.6.6, Fig. 6.42 indicates that the peak hardness of the HAZ occurred in the grain coarsened region near the fusion line (Tp=1300°C). The minimum hardness, which is well below that of the base plate, occurred in the partially transformed region (Tp=800°C). The low hardness zone also extended beyond the apparent microstructural boundary of the HAZ (termination of partially transformed region) to the tempered region (Tp=700°C) and into the unchanged base material (Tp=600°C) due to the overaging of the copper-rich precipitates.

Charpy impact energies at -51°C for simulated HAZ structures of 20mm CR HSLA 80 steel are shown in Fig. 6.43 as a function of peak temperature. The values near zero peak temperature represent the CVN energy of the CR HSLA 80 base plate. CVN values for various peak temperatures exceeded the Military Standard concerned with the qualification of submerged arc weld consumables for HY80/100 steels (47J at -51°C)(218). As peak temperature increased, CVN energy increased to maximum values in the range of 800°C -
Fig. 6.41 Optical microstructures of the simulated HAZ as a function of peak temperature (Tp1) (2.9kJ/mm)(320x).
(a) Tp1=1300°C; (b) Tp1=900°C; (c) Tp1=800°C;
(d) Tp1=700°C; (e) Tp1=600°C.
Fig. 6.42  Hardness as a function of peak temperature (T_p1) for single cycle simulated HAZ of 20mm CR HSLA 80 steel. The value near 0°C represents the hardness of the base plate. (Bars represent 95% confidence limits)

Fig. 6.43  CVN values for simulated HAZ at -51°C as a function of peak temperature (T_p1). The values near 0°C show the impact values of the base plate.
900°C (partially transformed and grain refined regions), and decreased with further increase in peak temperature. The grain coarsened HAZ region near the fusion line showed the lowest CVN value. Overall, compared to the base plate, little deterioration of toughness occurred for the HAZ sub-region structures, although the GCHAZ region showed slightly lower values than the base plate.

Samples with higher CVN values showed higher amounts of fibrosity and lateral expansion (Table 6.5). Except for a peak temperature of 1300°C, Charpy samples cycled to peak temperatures from 600°C to 1020°C showed almost totally fibrous fracture surfaces, consistent with the high impact toughness values (≥113J at -51°C).

Figure 6.44 shows the typical fracture surfaces for lower (Fig. 6.44a) and higher peak temperatures (Fig. 6.44b). Figure 6.45 is a schematic diagram which illustrates various fracture zones for a typical fracture surface (219). It consists of shear lips at both sides of the fracture surface; a tearing shear lip at the final fracture region; a ductile initiation zone (Diz) adjacent to the notch which acts as a stress concentrator and a central area of brittle and/or ductile fracture.

Shear lips at either side of the fracture surface are typically ductile as indicated by the dimpled fracture appearance (Fig. 6.46). The tearing shear lip at the final fracture region showed a similar fracture morphology except that the dimples were elongated and pointed in the direction opposite to that of crack propagation (Fig. 6.47).

The Diz showed rough steps or bands under stereo-microscopy (Fig. 6.44) and in the SEM image (Fig. 6.48). Elongated dimples were also found in this area (Fig. 6.49).

At lower peak temperatures (600°C and 800°C), the central part of the fracture surface showed internal open lips parallel to the rolling direction as seen in Fig. 6.44a. SEM examination of the matching sides of the lips indicated that some area exhibited a totally
Fig. 6.44 Fracture surface appearances of Charpy impact specimens (-50°C) of simulated HAZ (2.9kJ/mm) for (a) Tp₁=600°C, 115J; (b) Tp₁=1020°C, 162J; (c) Tp₁=1300°C, 62J.
Fig. 6.45 Schematic diagram showing the various zones of the fracture surface of a Charpy impact specimen (ref.219).
Fig. 6.46  SEM micrograph showing the dimpled appearance of the shear lips at the side of the Charpy impact fracture surface (440x) (T_p=1020°C, 162J at -50°C).

Fig. 6.47  Elongated dimples in tearing shear lip pointing towards the opposite direction to that of fracture in the final fracture region (440x) (T_p=1020°C, 162J at -50°C).
Fig. 6.48  SEM micrograph at low magnification showing the rough steps and bands in the ductile initiation zone near the notch of the impact specimen (66x). (Tp1=1020°C, 162J at -50°C)

Fig. 6.49  Elongated dimples in ductile initiation zone (440x)(Tp1=1020°C, 162J at -50°C).
dimpled appearance, corresponding to shear lips, other areas showed cleavage fracture mode, indicating that these areas can be regarded as 'splits' (220). For a high peak temperature where there were no internal splits or shear lips (Fig. 6.44b), and the central part was predominantly transgranular cleavage type of fracture (Fig. 6.50).

The size of the ductile area (Diz and shear lips) varied with the peak temperature as did the Charpy impact energy. Specimens with higher CVN values showed larger ductile areas. A marked increase in thickness of the Diz accompanied the increase of CVN value with decrease in peak temperature from 1300°C to 800°C.

6.5.2 Reheated Grain Coarsened HAZ Region

Overlapping of HAZs occurs in actual multi-pass welds. The microstructure and properties of this reheated HAZ region is usually different to the single pass HAZ. The properties of the reheated region are very important in terms of overall quality of the weld joint. Since the grain coarsened region exhibits the lowest toughness of the various HAZ regions, the effect of reheating of the GCHAZ has been investigated in the present work.

Changes in optical microstructure of the GCHAZ due to reheating to three different temperatures is shown in Fig. 6.51. The three reheating temperatures (Tp2=600°C, 800°C and 900°C) were selected to correspond to the tempered, the partially transformed and the grain refined regions, respectively.

No obvious microstructural changes occurred after reheating to 600°C, as expected, since only tempering occurs at this temperature. However, compared to the structure before reheating, the original MA islands (white etching) were darkened after reheating due to the tempering of MA islands.
Fig. 6.50  Transgranular cleavage type of fracture in central region of fracture surface (440x)(Tp1=1020°C, 162J at -50°C).
Fig. 6.51 Optical microstructures of simulated grain coarsened HAZ region (2.9kJ/mm, $T_{p1}=1300^\circ$C)(320x). (a) before reheating; (b) $T_{p2}=600^\circ$C; (c) $T_{p2}=800^\circ$C; (d) $T_{p2}=900^\circ$C.
Intercritical reheating to 800°C modified the granular bainitic structure of the original single weld GCHAZ, by refining the structure and forming bands of MA islands. The final structure was fine equiaxed ferrite grain with MA islands in discontinuous bands. Further reheating to 600°C (Tp3) in a third cycle resulted in tempering of these MA islands.

The microstructure of the GCHAZ after reheating to 900°C (temperature of grain refinement) was similar to that reheated to 800°C, except that the MA islands were more randomly distributed, and banding was less obvious.

The effect of the second thermal cycle on the hardness of the original GCHAZ (Tp1=1300°C) is shown in Fig. 6.52 and Table 6.6. The hardness value before reheating is plotted near zero temperature for comparison. It is obvious that the subsequent reheating to all three peak temperatures reduced the hardness relative to that of the single weld GCHAZ. The intercritically reheated GCHAZ (Tp2=800°C) showed the smallest hardness reduction.

A third thermal cycle with peak temperature (Tp3) of 600°C (Table 6.6) showed little effect on the hardness of the intercritically reheated GCHAZ (Tp2 of 800°C).

Figure 6.53 and Table 6.6 shows Charpy impact energies (CVN) at -51°C for double thermal cycled original GCHAZ specimens. The impact values near zero temperature represent the results before the second thermal treatment, which corresponds to a single cycle to Tp1=1300°C.

Little difference was found amongst the CVN values for the three second peak temperatures of 600°C, 800°C and 900°C. The impact values were lower than those resulting from a single cycle to the same peak temperature (Fig. 6.43), but the toughness was slightly improved relative to the simulated GCHAZ before reheating (Fig. 6.53).
Fig. 6.52  Hardness (HV5) of reheated GCHAZ as a function of second peak temperature (Tp2). The value plotted near 0°C is for the unheated GCHAZ (Tp1=1300°C). (Bars represent 95% confidence limits of the means).

Fig. 6.53  CVN values at -51°C for the reheated GCHAZ as a function of second peak temperature (Tp2). The values plotted near 0°C are for the GCHAZ (Tp1=1300°C) with no reheating.
CVN values for samples subjected to three heating cycles are given in Table 6.6. The peak temperatures in this case were 1300°C (Tp1), 800°C (Tp2) and 600°C (Tp3). Following the third thermal cycle with peak temperature of 600°C, the specimens did not show any significant change in Charpy value compared to the double cycled specimens (Table 6.6).

Examination of the fracture surfaces of doubly and triply cycled samples showed little difference in morphology compared to samples subjected to a single cycle to 1300°C, except that the size of the Diz and shear lips increased, consistent with the improved CVN values.

6.5.3 The Effect of Heat Input

From the previous discussion, it is clear that the unreheated GCHAZ shows the lowest impact toughness among single and multi-run weld HAZs. It is therefore regarded as having the potential to govern the toughness of welded joints. On this basis, the effect of welding variables, and particularly heat input, on the properties of this region will be significant in terms of weldability of this material.

Figure 6.54 shows the optical microstructure of the simulated GCHAZ (Tp1=1300°C) at different heat inputs. The most significant microstructural change in GCHAZ with increasing heat input is the coarsening of the structure. This result is consistent with the actual BOP SA, FCA and 4 wire SA weld HAZs, as discussed previously. Increasing amounts of dark etching structure, identified as lath martensite in BOP FCA weld HAZs were found at lower heat inputs, as a result of faster cooling.

The effect of heat input on hardness of the simulated GCHAZ structure is shown in Fig. 6.55. Bars represent 95% confidence limits on the mean values of 5 measurements.
Fig. 6.54 Microstructures of simulated GCHAZs for a peak temperature of 1300°C and various equivalent heat inputs (320x). (a) 1.9kJ/mm; (b) 2.2kJ/mm; (c) 2.9kJ/mm; (d) 4.9kJ/mm.
Fig. 6.55  Hardness of simulated GCHAZ as a function of equivalent heat input.
The value plotted near 0kJ/mm is for the base plate. (Bars represent 95% confidence limits of the means)
Similar to the results from actual weld HAZs, Fig. 6.55 indicates that lower HAZ hardness resulted from higher heat input welding. The hardness of the simulated GCHAZ at the lowest equivalent heat input (1.9kJ/mm) was only about 13 points higher than the hardness of the base plate, and the GCHAZ at the highest heat input (4.9kJ/mm) was about 20 points softer than the base plate.

Charpy impact energies for the simulated GCHAZ for different equivalent heat inputs is summarized in Table 6.7. These CVN values are plotted as a function of heat input in Fig. 6.56, together with the CVN values of the base plate for comparison. Similar to the results obtained on actual 4 wire SA weld HAZs, Fig. 6.56 indicates that the impact toughness of the GCHAZ is apparently insensitive to changing heat input, but a large scatter exists in the CVN values.

6.5.4 The Effect of Postweld Heat Treatment

Postweld heat treatment (PWHT), or stress relief heat treatment, of both the base metal and simulated GCHAZ (Tp1=1300°C) for 3.1kJ/mm equivalent heat input was carried out at three temperatures (450, 550 and 650°C) for 1 hour. The effects of PWHT temperature on the hardness of the simulated GCHAZ and the base plate are shown in Table 6.8, Fig. 6.57 and Fig. 6.58.

For the simulated GCHAZ, the hardesses before PWHT at slightly lower equivalent heat input (2.9kJ/mm) are shown near the zero point of the temperature in Fig. 6.57. It can be seen in Fig. 6.57 and Table 6.8 that the GCHAZ showed the maximum hardness after PWHT at 550°C. The hardness value in this case was higher than that before PWHT and higher than that of the base metal (~222HV5). PWHT at 450 and 650°C resulted in a reduced hardness in the GCHAZ.
Fig. 6.56 CVN values at -51°C of simulated GCHAZ as a function of equivalent heat input. The values near 0°C are for the base plate.

Fig. 6.57 Hardness of simulated GCHAZ at 3.1kJ/mm equivalent heat input as a function of PWHT temperature for 1 hour. The value near 0°C is for the GCHAZ (Tp1=1300°C) before PWHT. (Bars represent 95% confidence limits of the means)
Fig. 6.58  Hardness of the base plate as a function of PWHT temperature for 1 hour. The value plotted near 0°C is for the base plate prior to PWHT. (Bars represent 95% confidence limits of the means)
PWHT reduced the hardness of the base material (Fig. 6.58), with the higher the PWHT temperature, the lower the hardness of the base metal.

Figure 6.59 and Table 6.8 show the CVN values at \(-51^\circ C\) as a function of PWHT temperature for simulated GCHAZ samples (Tp1=1300°C). The Charpy energy of the GCHAZ before PWHT is shown near the zero point of the temperature axis for comparison.

It can be seen that impact toughness showed the opposite trend to that of hardness with PWHT temperature. The lowest CVN value occurred after PWHT at 550°C. The CVN value in this case is much lower than that before PWHT and lower than that of the base plate. The toughness of the GCHAZ was improved after PWHT at 450°C and 650°C, especially at the higher temperature.

The fracture surfaces of Charpy samples for the simulated GCHAZ after PWHT were examined by stereo-microscopy. As for the samples with single thermal cycle HAZs (Section 6.5.1), the fracture surface of the simulated GCHAZ specimens after PWHT also consisted of four typical zones as shown in Fig. 6.45. Samples with lower CVN values showed a narrower Diz and a smaller area of shear lips at both sides and tearing shear lips at the final fracture region. Almost no Diz and shear lips were observed in samples after PWHT at 550°C for 1 hour. The central area of each sample after PWHT exhibited a similar fracture mode, consisting of transgranular cleavage type fracture.

### 6.6 CONTINUOUS COOLING TRANSFORMATION DIAGRAMS

For the 20mm CR HSLA 80 steel plate, a partial continuous cooling transformation (CCT) diagram of the GCHAZ was obtained by measuring the start, 50% and finish transformation temperatures on cooling curves at different cooling rates (Fig. 6.60). The
Fig. 6.59  CVN values at -50°C of simulated GCHAZ at 3.1kJ/mm equivalent heat input as a function of PWHT temperature for 1 hour. The values near 0°C are for the GCHAZ (Tp=1300°C) before PWHT.

Fig. 6.60  Partial CCT diagram of GCHAZ for 20mm CR HSLA 80 steel. Peak temperature Tp=1300°C.
peak re-austenitising temperature was 1300°C and the equivalent heat input ranged between 1.9 to 4.9kJ/mm corresponding to cooling rates (800-500°C) of 30.2-5.0°C/s. Microstructure, hardness and toughness for various heat inputs are reported in Section 6.5.3.

To study the influence of Cu and Ni alloying elements on HAZ transformation behaviour, a similar partial CCT diagram was prepared for a reference steel: a low carbon equivalent (LCE) 350 MPa grade steel (Fig. 6.61), which has a similar chemical composition except for much lower Cu and Ni contents (Table 5.3). The same peak temperature (1300°C) was used as for the CR HSLA 80 steel. The heat input range in this case was 1.8-4.8kJ/mm, corresponding to cooling rates of 32.6-5.1°C/s. Compared to Fig. 6.60, the LCE 350 MPa grade steel showed a lower $\gamma \rightarrow \alpha$ transformation temperatures for heating and cooling conditions simulating those of the GCHAZ.

The hardness of the simulated GCHAZ of the LCE 350 MPa steel for different equivalent heat inputs is shown in Fig. 6.62. Compared to Fig. 6.55, the LCE 350 MPa steel showed a lower GCHAZ hardness than that of the CR HSLA 80 steel for a similar heat input.

Optical microstructures of the GCHAZ at different heat inputs for LCE 350 steel (Fig. 6.63) are similar to those of the CR HSLA 80 steel (Fig. 6.54). However, much of the high carbon island constituent is diffusional transformation product (pearlite or bainite) as shown by the dark etching particles, whereas for the CR HSLA 80 steel, most of the islands were white etching (Fig. 6.54), indicating the non-diffusional MA product. The formation of MA islands in the CR HSLA 80 steel is associated with the lower $\gamma \rightarrow \alpha$ transformation temperature and the higher carbon equivalent (CE). The MA islands contribute to the higher hardness obtained in the GCHAZ of this steel.
Fig. 6.61 Partial CCT diagram of GCHAZ for 20mm LCE 350 steel. Peak temperature $T_p=1300^\circ$C.

Fig. 6.62 Hardness of the simulated GCHAZ of LCE 350 steel as a function of equivalent heat input. The value near 0kJ/mm is for the base plate. (Bars represent 95% confidence limits of the means)
Fig. 6.63  Microstructures of simulated GCHAZ at different heat inputs for LCE 350 steel (320x). (a) 1.8kJ/mm; (b) 2.2kJ/mm; (c) 2.8kJ/mm; (d) 4.8kJ/mm.
7.1 INTRODUCTION

In understanding and optimizing the weldability of a steel in terms of HAZ properties, it is very important to analyse the inter-relationships between HAZ structure, properties and welding conditions. In this Chapter, the experimental results presented in Chapter 6 are discussed in four sections. In each section, an attempt is made to characterise the relationship between structure and properties of the HAZ. A summary is given at the end of each section.

Because a structure gradient exists across the HAZ, the properties at different positions of the HAZ vary. The structures and properties of the various HAZ regions are discussed in Section 7.2, with the aim of establishing the region or regions which exert the greatest influence on the overall HAZ properties. The hardness and HAZ width variation along the fusion line and with heat input for BOP SA welds is discussed in Section 7.3.

In Section 7.4, the effect of various factors associated with the welding process on the HAZ structure and properties are considered. These factors include multi-pass welding (Section 7.4.1), heat input (Section 7.4.2), welding speed (Section 7.4.3) and postweld heat treatment (Section 7.4.4).

Finally, the $\gamma\rightarrow\alpha$ phase transformation behaviour of the grain coarsened HAZ region is discussed in Section 7.5. The partial HAZ CCT diagram for the CR HSLA 80 is presented in Section 7.5.1 and is compared with the diagram for low carbon equivalent (LCE) 350 steel (Section 7.5.3) to assess the effect of copper and nickel on transformation of austenite in the HAZ. The properties of the GCHAZ are summarised in Section 7.5.2.
7.2 STRUCTURE AND PROPERTIES OF THE HAZ OF SINGLE PASS ACTUAL AND SIMULATED WELD

Hardness traverses across the actual and simulated HAZ indicated that peak hardness occurred in the grain coarsened region (GCHAZ), and the partially transformed region showed the minimum hardness which was well below that of the base material (Figs. 6.6 and 6.42). Most of the HAZ was softer than the base material. To explain the softening of the HAZ, it is necessary to discuss the Cu precipitation hardening behaviour in the HAZ, since the softer HAZ is attributed to the different state of Cu precipitation in the HAZ and base plate.

Since the solubility of Cu in austenite is higher than in ferrite (Fig. 7.1), relatively fast cooling from high temperature retains much of the Cu in supersaturated solid solution in ferrite. Subsequent aging to produce a fine dispersion of Cu clusters in the ferrite matrix produces a considerable increase in hardness and strength. It has been reported that coherent metastable b.c.c precipitates (Cu clusters or G.P. zones) contribute strongly to the precipitation hardening (175,221). These Cu clusters are not resolved in the electron microscope because they are either too small to be seen by TEM (222,223) or the contrast between Cu clusters and the ferrite matrix is low due to the similarity of the scattering factors of iron and copper and negligible strains are produced by the precipitate (224). Particles visible by TEM are overaged Cu clusters which are f.c.c ε-Cu particles.

In the GCHAZ and grain refined (GR) HAZ regions, complete austenitisation occurs. As a result, Cu precipitates in the ferrite of the base metal re-dissolve during α→γ transformation because of the higher solubility of Cu in austenite. Whether or not re-precipitation of Cu occurs in the HAZ during cooling after welding is dependent upon the cooling rate of the HAZ. The observation of ε-Cu precipitates in the GCHAZ and GRHAZ (Fig. 6.3b and 6.3c) in the present research suggests that reprecipitation of Cu
Fig. 7.1 Partial Fe-Cu phase diagram illustrating the solubility of Cu in Fe (ref.225).
occurred in the HAZ during cooling for the welding conditions investigated. Cu particles were uniformly distributed in the ferrite matrix rather than preferentially on dislocations (Fig. 6.3). This observation is consistent with reports by several investigators (174,175). Hornbogen (226) concluded that the probability of dislocation and matrix nucleation of the f.c.c phase at temperatures below 700°C is almost equal due to a low strain energy term for nucleation. This factor leads to the uniform distribution of particles. The Kurdjumov-Sachs (K-S) orientation relationship between ε-Cu particles and the ferrite matrix (Fig. 6.3e and 6.3f) also indicated that Cu-rich coherent b.c.c. metastable precipitates (clusters) occur initially, and then transform *in situ* into f.c.c. ε-Cu particles (226).

Although Cu reprecipitation occurred in the HAZ during cooling, the resulting rehardening was very minor compared to the precipitation hardening generated by aging heat treatment of the base plate (at 550°C for 1/2 hour). This difference arises because of different cooling rates in TMCP and after welding, and the absence of an aging heat treatment in the case of the welded samples. The higher hardness in the GCHAZ than the GRHAZ is attributed to the coarse prior austenite grain size which increases the hardenability, resulting in a coarse ferrite lath structure (comparing Fig. 6.1a to 6.1b) with a higher volume fraction of MA islands than other regions of the HAZ (Fig. 6.4).

The partially transformed region had the lowest hardness in the HAZ. The hardness value in this region was well below that of the base material. A similar result was reported for gas tungsten arc (16) and submerged arc welded (9) Q&A A710 steel. Compared to the base metal, the significant loss of hardness is attributable to the overaging of Cu clusters in untransformed ferrite, and re-solution of Cu precipitates in austenitised regions (as in the GCHAZ and GRHAZ). Heating by the weld thermal cycle to a peak temperature between A(r1) - A(r3) ensured rapid overaging of Cu clusters in the untransformed ferrite grains, as shown by the coarse ε-Cu particles in Fig. 6.3d. Re-austenitised regions transformed on cooling to ferrite and MA islands which are supersaturated with Cu.
Although austenite transformation does not occur in the tempered and apparently unchanged base metal regions, the reduced hardness compared with the base material is due to overaging of Cu clusters. The softening is demonstrated in Fig. 6.6 for the BOP SA welds and in Fig. 6.42 for Tpt of 600 and 700°C in simulated weld HAZ samples. Furthermore, there is a contribution to softening by degradation of lamellar pearlite to spheroidal particles of carbide in the tempered HAZ region.

CVN values for various HAZ regions exhibited the opposite trend to hardness, i.e. a higher hardness was associated with a lower impact toughness value (Figs. 6.42 and 6.43). This result confirms a commonly accepted conclusion drawn by many investigators that harder structures generally show a lower toughness (99, 131).

The GCHAZ showed the lowest CVN value and the highest hardness of the various sub-regions. The relatively low toughness in this region results from a structure with the highest volume fraction of MA islands (Fig. 6.4) which contribute to the higher hardness of this region. The deterioration in toughness associated with the MA islands has been reported previously (55, 67), as discussed in Chapter 3, Section 3.4.2.2. The fact that an increasing volume fraction of MA islands causes significant reduction of toughness is also widely accepted (53, 67). In addition, the relatively low toughness of the GCHAZ is also associated with the presence of the high carbon, twinned martensite of the MA islands (Fig. 6.2). Twinned martensite has been reported to be one of the microstructural constituents which is most detrimental to toughness (155).

However, compared to the base plate, little or no deterioration in toughness occurred in the simulated HAZ structures (Fig. 6.43) for the present steel. The high HAZ toughness of this steel is consistent with work reported on other steels based on the ASTM A710 steel composition (see Chapter 4, Section 4.4.4). The toughness is associated with the
much lower carbon content and carbon equivalent of the A710 steels than conventional steels of same grade (e.g. HY 80). As a result, a tough ferritic structure is formed in the HAZ. In the present steel, the high HAZ toughness may also attributed to the grain refining influence of TiN which restricts grain growth in the GCHAZ.

Significant degradation of toughness in the partially transformed region and the intercritically reheated GCHAZ has been demonstrated by several authors (55,67) in offshore structural and low carbon microalloyed steels. The deterioration of toughness in both regions was reported to be related to the formation of MA islands.

In contrast, simulated HAZ structures for the present steel showed the highest toughness in the partially transformed region and improved toughness in the intercritically reheated GCHAZ (Fig. 6.43 and 6.53), despite the presence of MA islands. Charpy impact energies in both regions were higher than in the base plate. This result is associated with the low hardness in both regions as the result of loss of the Cu precipitation hardening. The better toughness in both regions may also associated with the low carbon content of the steel and the generally low volume fraction of MA islands in the HAZ.

The hardness and structure of the MA islands are influenced strongly by the carbon content. The carbon content in the present steel (0.055%C) is low compared to the steels mentioned above which showed low CVN values in both the partially transformed and intercritically reheated GCHAZ regions (≥0.08%C). Therefore, the volume fraction and/or the carbon content of the MA islands must be lower in the present steel. Thus the hardness is expected to be lower, and provided that the MA islands are not morphologically unfavourable, a higher toughness should be exhibited. Coarse interfacial plates of brittle constituents are unfavourable morphologies with respect to toughness. Quantitative metallography indicated that the MA islands in the GCHAZ of BOP SA welds had an average maximum dimension of about 1 μm (Fig. 6.11), although particles
up to 15µm were measured. The fine MA islands are consistent with the relatively high measured toughness in the partially transformed region.

The CVN impact properties of the partially transformed HAZ region has been reported to be closely related to the volume fraction of MA islands (67), with high volume fractions resulting in lower toughness in this region. It has been found from investigations of several types of steels that the toughness in this region starts to deteriorate when the volume fraction of MA islands exceeds a critical value of about 6% (67). In the present case, both bead-on-plate SA welds (Fig. 6.4) and 4 wire SA welds HAZs (Fig. 6.29a) showed that the volume fraction of MA islands in the GCHAZ is about 6% or lower and is relatively insensitive to changes in heat input and welding process. Since the volume fraction of MA islands decreased from the GCHAZ to the partially transformed region, the volume fraction in the partially transformed region for the various welding conditions examined was much lower than the critical value of 6%.

In summary, high HAZ toughness was obtained for all major regions of the HAZ of simulated weld structures in the CR HSLA 80 steel, consistent with relatively high CVN values in the HAZ of actual welds. The overall high HAZ toughness of this steel is associated with the low carbon content and carbon equivalent and the grain refining effect of Ti, all of which result in a low volume fraction of MA island in the HAZ. In addition, the high toughness of the HAZ relative to the base plate is associated with the softening caused by the loss of Cu precipitation hardening.

The critical region governing the toughness of the HAZ is considered to be the GCHAZ. The relatively low toughness in this region is the result of its higher hardenability which results in a microstructure of coarse ferrite sideplates and a relatively high volume fraction of MA islands. The partially transformed and grain refined regions exhibited highest toughness and lowest hardness values because of the grain refinement and maximum loss of hardness in these regions.
7.3 HARDNESS AND HAZ WIDTH VARIATION ALONG THE FUSION LINE OF BOP SA WELDS

The variation of HAZ width along the fusion line is significant as shown in Fig. 6.9 for the three different positions (a, b and c). Variation in width was also marked with changing heat input with the average HAZ width increasing with increasing heat input. Since high heat input is associated with slow cooling rate (equations (2.9) and (2.11)), it is inferred that a wide HAZ corresponds to a slow HAZ cooling rate.

For welds at different heat inputs, the relationship between cooling rate \( \propto 1/\Delta t^{8.5} \) and width of HAZ can be deduced from Rosenthal's equations.

For thick and thin plate, equations (7.1) and (7.2) apply to a certain fixed position from the heat source as defined by a radial distance \( r \)

\[
T_p - T_0 = \left( \frac{2}{\pi e} \right) \frac{HI}{\rho c \pi r^2} \quad \text{(7.1)}
\]

\[
T_p - T_0 = \left( \frac{2}{\pi e} \right)^{1/2} \frac{HI}{d \pi c r} \quad \text{(7.2)}
\]

where \( T_p \) refers to the peak temperature (in K) of the thermal cycle and \( e \) is the base of natural logarithms (=2.718); \( HI \) (J/m) is heat input; \( \rho c \) is the specific heat per unit volume in Jm\(^{-3}\)K\(^{-1}\) and \( d \) is the thickness of the plate.

The HAZ width is roughly equal to the difference in \( r \) (\( \Delta r \)) between \( T_{1500\degree C} \) and \( T_{700\degree C} \). From equations (7.1) and (7.2), the following relationship exists between the width of the HAZ (\( X \)) and the heat input (\( HI \)) for thick and thin plates, respectively.
The above equations, (7.3) and (7.4), suggest that at a constant heat input, the HAZ width is constant. This is obviously not true since significant variation of HAZ width is observed around the periphery of BOP SA and FCA welds (Fig. 6.9 and 6.22). This observation indicates that the Rosenthal approach, based on 'a moving point heat source', is an oversimplification of the complex phenomenon of arc welding in which the heat source is extended and variable in shape. Therefore, the above equations can only strictly be applied at similar HAZ position for different heat inputs.

The following equations have been developed to relate $\Delta t_{8.5}$ and heat input (Chapter 2).

For thick plate,

$$\Delta t_{8.5} = \frac{HI}{2\pi\lambda\Theta_1} \quad (2.9)$$

where $\lambda$ is the thermal conductivity, and $\Theta_1$ is defined by equation (2.10)

$$\frac{1}{\Theta_1} = \left(\frac{1}{773-T_0} - \frac{1}{1073-T_0}\right) \quad (2.10)$$

For the thin plate case,

$$\Delta t_{8.5} = \frac{(HI)^2}{4\pi\lambda\rho c\Theta_2^2 d^2} \quad (2.11)$$

$$\frac{1}{\Theta_2^2} = \frac{1}{(773-T_0)^2} - \frac{1}{(1073-T_0)^2} \quad (2.12)$$
The relationship between HAZ width and cooling time ($\Delta t_{8-5}$), for both the thick and thin plate cases, is

$$X \propto (\Delta t_{8-5})^{1/2} \quad (7.5)$$

Therefore, it is concluded that for welds of different heat inputs, the HAZ width at a given position is proportional to the square root of cooling time $\Delta t_{8-5}$, and is therefore inversely proportional to the square root of cooling rate, with a wider HAZ corresponding to a slower cooling rate.

The wider HAZ, the coarser structure in the GCHAZ and the lower HAZ peak hardness (Figs. 6.9, 6.10 and 6.14) at higher heat input for the BOP SA welds are consistent with Rosenthal's equations (2.9) and (2.11) for $\Delta t_{8-5}$ and the derived equations (7.3), (7.4) and (7.5), for corresponding positions in welds of different heat inputs.

However, the significant variation of HAZ width from position a to c for the BOP SA welds (a ratio of 2.5:1) indicates that the local cooling rate in the HAZ varies at different positions along the fusion line.

Since the HAZ maximum hardness correlates approximately with cooling rate (17), which is inversely proportional to the square of HAZ width ($X^2$) (equation (7.5)), the maximum hardnesses at different positions along the fusion line and for the three heat inputs can therefore be replotted as a function of the inverse square of HAZ width (Fig. 7.2). The dashed curves in Fig. 7.2 show the variation of hardness with heat input for corresponding positions in the HAZ, and therefore give similar information to Fig. 6.14. As can be seen in Fig. 7.2, the change in cooling rate ($\propto 1/X^2$) with heat input at a certain position is relatively small compared to that with changing positions (a→b→c) at a certain heat input. Similar results are obtained from the change in prior austenite grain size (Fig. 7.3). For position a, for example, the grain size changed from 80 to 100 $\mu$m as the heat
Fig. 7.2  Mean peak hardness as a function of inverse of square of HAZ width ($X$). Heat inputs and sampling positions are indicated.

Fig. 7.3  Mean austenite grain size at positions a and c in the grain coarsened region of the HAZ, as a function of heat input.
input increased from 2.5 to 6kJ/mm. It is inferred that since the prior austenite grain size is similar, the increase in cooling rate associated with decreasing heat input dominates in determining the hardness trend for corresponding positions in the GCHAZ. The higher cooling rate at lower heat input does not substantially change hardenability, since the volume fraction of MA islands is largely independent of heat input (Fig. 6.4), but it stimulates a higher nucleation rate, refinement of the ferrite laths and corresponding refinement of the interlath regions of MA constituent. (Fig. 6.10).

The full curves in Fig. 7.2 give the variation in hardness with cooling rate at the three sampling positions within the one sample (i.e. at constant heat input). It should be emphasized again that the square of HAZ width ($X^2$) variation is much greater within a sample than for varying heat input over the range studied (Fig. 7.2). It follows that the cooling rate variation with position within a sample (at constant heat input) is greater than that effected by changing the heat input. In general, the full curves trend slightly downwards with increased cooling rate from position c to a. Two counteractive effects on hardenability should be considered to explain the hardness trend. They are the cooling rate and austenite grain size. From position c to a, cooling rate increased strongly (Fig. 7.2) and austenite grain size decreased substantially (Fig. 7.3). These two effects are counteractive in terms of hardenability, and it is inferred that grain size is over-riding in this case, resulting in a higher hardness at position c, at least for the two lower heat inputs. Another possible influence is the greater extent of solution of Cu, Nb(CN) and TiN at position c because of a longer dwell time at elevated temperature. The subsequent lower cooling rate then allows re-precipitation of these species and consequent re-hardening. Thus, the local hardness in the HAZ will be determined by the complex interactions amongst the austenite grain surface area, the extent of solution of hardening species, the cooling rate, the amount and type of re-precipitation and the nature of the austenite transformation product.
The width of the softened zone (defined as the region with hardness less than the base plate level of 215 HV1) is plotted as a function of heat input in Fig. 6.7. This width increased with heat input and with change in sampling position from a→b→c, as a result of the decreasing cooling rate. Slow cooling increases the volume of material which is softened by overaging and/or re-solution of copper. For the 4.0 and 6.0kJ/mm welds, the softened zone extended for 4-6.5 mm in positions b and c (Fig. 6.6) and the transverse tensile properties could therefore be adversely affected, as was confirmed for the 4 wire SA welds (see Section 7.4).

7.4 FACTORS AFFECTING THE HAZ STRUCTURE AND PROPERTIES

7.4.1 Effect of Multi-pass Welding

7.4.1.1 Intercritical reheating of BOP SA welds

The microstructure of the intercritically reheated actual and simulated GCHAZ consisted of bands of small MA islands in a matrix of refined and more equiaxed ferrite grains than in the GCHAZ (Fig. 6.15). The tendency of the MA islands to be distributed in bands is shown in Figs. 6.16 and 6.17, suggesting the preferential nucleation of austenite in original pearlite band areas of the base plate where there the carbon and alloying element contents are higher (227).

The preferential nucleation of austenite in the original pearlite band sites in the intercritically reheated GCHAZ clearly indicates that even in the GCHAZ, complete redistribution of carbon and alloying elements did not occur. The lower A(c3) temperature due to the higher carbon and alloy contents allows the original pearlite band regions to
revert to austenite first on reheating. The more uniformly distributed MA islands in the intercritically reheated GCHAZ produced at higher heat input (Fig. 6.15) suggests that the low cooling rate (at higher heat input) allowed longer time at high temperature for the redistribution of carbon and alloying elements, thus resulting in a less segregated structure.

In analysing the effect of intercritical reheating on the hardness of the GCHAZ, two factors need to be considered. They are hardening due to Cu-rich precipitate hardening of ferrite during cooling and softening caused by the microstructural change from a coarse granular structure to a mixture of fine equiaxed ferrite and a lower volume fraction of MA islands (Fig. 6.15). The relatively unchanged hardness of the GCHAZ after reheating to 800°C (Fig. 6.20) for the BOP SA weld HAZs indicates balancing of these two counteractive factors.

7.4.1.2 Subcritical reheating of BOP SA welds

For reheated BOP SA welds, the hardness of the GCHAZ was raised for the lowest heat input (2.5kJ/mm) after reheating to 600°C (subcritical reheating), whereas for the higher heat inputs (4 and 6kJ/mm), there was relatively little change in hardness of the GCHAZ after subcritical reheating (Fig. 6.20).

In the case of subcritical reheating of the 4 and 6 kJ/mm welds, any Cu precipitation hardening in the GCHAZ was counteracted by the softening due to the tempering of MA islands, as shown by TEM (Fig. 6.19). For the 2.5kJ/mm heat input, a higher level of supersaturation of Cu resulted after the first weld thermal cycle because of the faster cooling rate at lower heat input (compared to 4 and 6kJ/mm). Thus, during subcritical reheating, more significant precipitation hardening was induced, resulting in an increase in hardness compared to that of the original GCHAZ. This result implies that some
deterioration of toughness may occur in the subcritically reheated GCHAZ after applying a second weld pass to an original HAZ produced by a low heat input.

7.4.1.3 Intercritically and subcritically reheated simulated GCHAZ

Both intercritical ($T_{p2} = 800^\circ C$) and subcritical reheating ($T_{p2} = 600^\circ C$) of the simulated GCHAZ resulted in a reduction of hardness (Fig. 6.52). This result is different to that of the reheated BOP SA welds (Section 7.3.1.1 and 7.3.1.2). In the case of BOP SA welds, after subcritical reheating, the HAZ associated with the lowest heat input (2.5kJ/mm) showed an increase in hardness, while relatively little change in hardness occurred for HAZs produced by higher heat inputs (4 and 6kJ/mm). A relatively unchanged hardness of the GCHAZ was also found in all three BOP SA weld HAZs after intercritical reheating. The different hardness changes after reheating between BOP SA weld GCHAZs and simulated GCHAZ may be due to the different heating and cooling rates associated with the second thermal cycles in the two cases.

For the simulated GCHAZ, the cooling rate between 800-500°C for the second thermal cycle was 13.3°C/s, which is much higher than that in the case of BOP SA welds (3.8°C/s). The faster cooling rate of the second thermal cycle induced less re-precipitation hardening by Cu in the ferrite of the GCHAZ after reheating to 600°C or 800°C. As a result, the overall effect of reheating to 600°C and 800°C on the simulated GCHAZ was a reduction of hardness in this region.

The softening of the simulated GCHAZ after reheating to 900°C is due to the change of the structure from a mixture of coarse ferrite laths and MA islands to fine equiaxed ferrite grains and a lower volume fraction of fine MA islands, aligned in bands (Fig. 6.51).
Impact toughness was improved after reheating the simulated GCHAZ to the three temperatures (600, 800, and 900°C) (Fig. 6.53). The improved toughness is the result of the softening after reheating (Fig. 6.52), and also to grain refinement in the case of the reheating to 800 and 900°C.

In work reported on significant loss of toughness in the intercritically reheated GCHAZ (55, 67), the MA islands were reported to be distributed along the prior austenite grain boundaries. Since the grain boundaries are weaker than the grain interiors, the intergranular MA islands weaken the grain boundaries, causing embrittlement. For the present CR HSLA 80 steel, the MA islands were present in bands after intercritical reheating. The difference in the location of the MA islands may explain the relatively high toughness in the present steel. Furthermore, the resistance to significant deterioration of toughness in this region is due mainly to the low carbon content of this steel and the low volume fraction of MA islands in the HAZ.

7.4.1.4 Summary

The reheating of the simulated GCHAZ to tempering (600°C), partial transformation (800°C) and grain refining (900°C) temperatures improved the toughness and reduced the hardness of the original GCHAZ at a cooling rate (between 800-500°C) of 13.3°C/s (equivalent heat input of 2.9kJ/mm). However, significant rehardening could occur if a low heat input weld were followed by a second weld thermal cycle with a low cooling rate (associated with either first pass preheating or a higher second pass heat input).
7.4.2 Effect of Heat Input

7.4.2.1 Weld profile and HAZ width

The effect of heat input on weld profile and HAZ width is shown in Figs. 6.8, 6.9, 6.21, 6.22, and 6.27. The weld profiles were similar for the bead-on-plate and 4 wire SA welds, but the FCA welds showed a markedly different weld profile, which is associated with the different type of welding process. An increase in the size of the weld bead and the width of the HAZ with increasing heat input is evident in the FCA welds. The larger weld bead for a higher heat input in both types of welds is consistent with the higher energy input per unit length which results in a larger volume of weld metal. The wide HAZ associated with high heat input is consistent with the equations (7.3) and (7.4), deduced from Rosenthal's equations (Section 7.3). The wider HAZ for higher heat input indicates slower HAZ cooling rate (equation (7.5)).

Despite low heat input, the BOP FCA welds had a wider HAZ than the BOP SA welds (Fig. 6.9 and 6.22). However, the optical microstructures of the GCHAZ for the two types of welds (Fig. 6.10 and Fig. 6.23) showed that a higher cooling rate appeared to be associated with the FCA weld HAZs, as indicated by finer structure and higher percentage of the dark etching lath martensite. The faster cooling rates of the FCA weld HAZs are also reflected by the much higher HAZ peak hardness than for the SA welds (Fig. 6.33).

Compared to the BOP SA welds (Fig. 6.8), the BOP FCA welds (Fig. 6.21) showed much shallower weld penetration and a wider weld bead. The heat flow direction from the molten weld pool for the FCA welds is, therefore, towards the bottom of the plate, whereas for the SA welds, the deep weld penetration allows the heat flow to be three dimensional. The heat from the molten weld pool in the case of SA welds was largely
extracted by the 'infinite' volume of metal in the direction parallel to the surface, whereas
heat flow in the FCA welds was directed towards the 'finite' volume of metal bounded by
the bottom surface of the plate. However, compared to the SA weld HAZs, the initial
cooling rate of FCA weld HAZs was high because of the low heat input and the
distributed heat source. Although heat was extracted quickly in the initial stages, the
cooling rate of the HAZ at a later stage of the cooling cycle was reduced due to warming
up of the plate. The slow down in the cooling rate thus resulted in a wide HAZ for the
BOP FCA welds.

7.4.2.2 HAZ peak hardness

The changes of HAZ peak hardness with heat input are discussed in Section 7.3 for the
BOP SA welds. Figure 6.33 shows the hardness change with heat input for the three
types of actual welds and in Fig. 6.55 for the simulated GCHAZ. In all cases, a lower
heat input resulted in a higher peak hardness of the HAZ. This observation is consistent
with a faster cooling rate of the HAZ at lower heat input, which results in a harder HAZ
structure due to structural refinement and increased amount of lath martensite (Figs. 6.10,
6.23, 6.28 and 6.54). The higher GCHAZ hardness for lower heat input is associated
with a lower $\gamma \rightarrow \alpha$ transformation temperature, as shown by the partial CCT diagram for
the GCHAZ (Fig. 6.60).

7.4.2.3 HAZ impact toughness

In spite of the increase in hardness with decreasing heat input, toughness did not change
substantially with heat input for simulated GCHAZ structures produced at equivalent heat
inputs between 1.9-4.9kJ/mm (Fig. 6.55). A similar trend was observed for 4 wire SA
weld HAZs with heat inputs of 2.5-10kJ/mm for 20 and 50mm plates (Table 6.2, Figs 6.34 and 6.35). In general, this result can be explained on the basis of the effects on toughness of hardening and structural refinement at lower heat inputs; and softening and structural coarsening at higher heat inputs (Section 3.4.3).

It is evident that, at high heat input, the slow cooling rates produce relatively coarse microstructural constituents which limit the cleavage resistance (Section 3.4.2), despite the low HAZ hardness. The coarse structure at higher heat input is indicated by larger average prior austenite grain size in the GCHAZ along the fusion line of the actual welds (Figs. 6.13, 6.25 and 6.30); and the low number density (Fig. 6.5) and relatively high maximum dimension of the MA islands (Figs. 6.11 and 6.29b). The optical microstructures shown in Figs. 6.10, 6.23, 6.28 and 6.54 qualitatively support this conclusion. On the other hand, at low heat input, toughness is limited because the fast cooling rates produce a relatively hard microstructure as shown by higher peak hardness values (Figs. 6.33 and 6.55).

The relatively small change of HAZ toughness with heat input is also consistent with the relatively unchanged volume fraction of MA islands for various heat inputs as shown in Fig. 6.4 and 6.29a.
7.4.2.4 Transverse tensile properties of 4 wire SA welds

Tensile properties were measured for 4 wire SA welds of 20mm plate. Examination of the failure location indicated that the welds failed in the HAZ as a result of softening in the HAZ. Nevertheless, yield strengths consistent with that of the base plate were obtained for the two lowest heat inputs (2.5 and 5kJ/mm) (Table 6.3). A significant reduction of yield and tensile strengths occurred, however, at the highest heat input (10kJ/mm). The upper yield strengths at two different welding speeds for the 10kJ/mm welds were much lower than that of the base plate and lower than the minimum requirement (80 ksi or 550 MPa) for HSLA 80 steel. The reduced strength at the highest heat input (10kJ/mm) is clearly associated with the wide HAZ which is mostly softer than the base plate (Fig. 6.36). This result suggests that any factors which reduce the width of the HAZ will improve the strength of the HAZ. Another possible way to increase HAZ strength is by applying postweld heat treatment to induce Cu precipitation hardening in the HAZ (as discussed in Section 7.3.4).

Each fracture surface exhibited a typical cup-and-cone tension-overload fracture, consisting of a fibrous central region and an outer annular shear lip zone. Both regions showed typical transgranular fracture by microvoid coalescence (Fig. 6.38). The fibrous fracture surfaces, the large values of reduction of area and the high elongation values indicate that the samples were very ductile (Table 6.3). Ductile tension fracture implies a high toughness in the fracture area, i.e. in the HAZ, consistent with the impact toughness results for the actual and simulated HAZs obtained in the present work.

In summary, high HAZ toughness was a feature of the HAZ produced by 4 wire SA welding and by simulation at heat inputs between 2.5-10kJ/mm and 1.9-4.9kJ/mm, respectively. HAZ toughness for all welding conditions for the 20mm plate met or exceeded the military standard for HY 80/100 steels. However, to be used as HY 80
steel, the 20mm CR HSLA 80 steel should be SA welded at a heat input less than or equal to 5kJ/mm to avoid a wide softened HAZ which can cause a significant reduction in the transverse tensile properties of the weld joint.

Increasing heat input had little effect on HAZ toughness, despite a significant decrease in hardness, because of coarsening of the ferrite. The small change in HAZ toughness with heat input is also consistent with a minor change in the volume fraction of MA islands.

7.4.3 Effect of Welding Speed on the HAZ of 4 Wire SA Welds

The effect of welding speed on HAZ microstructure and properties was investigated by examining 4 wire SA weld HAZs produced by different welding speeds at constant heat input.

Macrography of the weld cross-sections indicated a slight decrease in HAZ width with increasing speed. The reduction of HAZ width is associated with a higher HAZ cooling rate at high welding speed, as reported by CSIRO, DMT (203).

Comparing the weld profile of a single pass (or the last pass in the case of multi-pass welding), the weld cross-sections in Fig. 6.27 showed that there is no major change in weld profile with welding speed, as expected for the same welding procedure. However, instead of showing a uniform weld bead, the weld cross-section for the lowest welding speed for each of the three heat inputs showed multiple fusion boundaries (or solidification lines) within the weld bead. For the sample of 2.5kJ/mm at 1000mm/min speed, Fig. 6.39 shows four fusion boundaries within a weld bead, corresponding to the passage of the four wires used in the welding process.
At low welding speed, the time delay between the passage of the previous and the following wire is significant, and there is sufficient time for the weld pool resulting from the first wire to partially solidify before the passage of the following wire, which causes only partial remelting. A continuous single weld pool was not formed because of the relatively slow speed and the separation of the wires. Since the four wires were arranged essentially as two pair (Fig. 5.5), a point in the HAZ experiences two major thermal pulses producing the double peak nature of the thermal profile in the HAZ, as reported by CSIRO, DMT (202). The double fusion lines in the 10kJ/mm weld at lowest welding speed (600mm/min) further reflects the double tandem nature of the welding process.

The general trend with increasing welding speed was a smaller prior austenite grain size except for the unexpectedly low values for the lowest welding speed at 2.5 and 5kJ/mm heat inputs. The general trend is due to the faster cooling rate at higher speed leading to a short dwell time above the austenite grain coarsening temperature. The unexpectedly small austenite grain size for the two low heat inputs is also associated with the the nature of the 4 wire welding process. The separation of thermal pulses experienced by the HAZ evidently reduces the dwell time above the grain growth temperature. The effect of separate weld pools is greater at lower heat input, since the faster cooling rate enables the weld pool associated with the previous wire to solidify quickly and only a small volume of this solidified weld pool is remelted by the second pair due to the lower energy input. In contrast, the highest heat input (10kJ/mm) did not show a low austenite grain size at the lowest speed, because the two energy pulses are still very high (~5kJ/mm each).

Although the cooling rate of the HAZ increases with increasing welding speed as demonstrated by CSIRO, DMT (203), the effect of welding speed on HAZ microstructure is minor. This observation is consistent with the small change in volume fraction and average maximum dimension of the MA islands with welding speed (Fig. 6.29). The small change of HAZ microstructure with welding speed, apart from a slight change in austenite grain size, is associated with a relatively small change in peak HAZ hardness.
(Table 6.1), in HAZ toughness (Figs. 6.34 and 6.35) and in tensile properties (Table 6.3) with welding speed.

In summary, the welding speed at constant heat input did not have a substantial effect on HAZ hardness, toughness and tensile properties, consistent with the limited variation of HAZ microstructure. This result suggests that improvement in welding productivity of CR HSLA 80 can be effected by increasing speed, without significant deterioration of mechanical properties.

Due to the nature of the 4 wire SA welding procedure used, a single weld pool may not form. The tendency for separate weld pools is high for the low heat input at low welding speed. Under these circumstances, the 4 wire process is similar to quick succession, multiple pass welding.

7.4.4 Effect of Postweld Heat Treatment

Postweld heat treatment (PWHT) at 450, 550 and 650°C for 1 hour was carried out on the simulated GCHAZ at an equivalent heat input of 3.1kJ/mm for 20mm plate. Peak hardening due to Cu precipitation occurred in the GCHAZ after PWHT at 550°C (Fig. 6.59), accompanied by significant deterioration in toughness (Fig. 6.59). The impact toughness of the GCHAZ after PWHT at 550°C was lower than the minimum requirement of HSLA 80 steel (47J at -51°C). However, PWHT at both 450 and 650°C resulted in a softer structure and improved toughness.

The present result is different to that reported by Lundin et al. (66) for PWHT behaviour of 37.5mm thick Q&A A710 steel welded with 2kJ/mm heat input (Section 4.4.4). The authors demonstrated that the GCHAZ of Q&A A710 steel is susceptible to stress relief cracking (SRC) after 1 hour at 454, 566 and 677°C. The different PWHT behaviour of
the GCHAZ in the present case is likely to be due to the difference in chemical composition and thermomechanical processing for these two types of A710 steel.

It is interesting to note that the 25mm control rolled plate also showed that peak hardening occurred at 500-550°C and that the strength level was not greatly affected by small changes in the aging temperature and time for temperatures up to about 600°C. Although the strengthening gained from aging was accompanied by a degradation in toughness (12), the degradation in toughness was found not to be simply related to the strength increment. In the overaged condition, the strength increment is achieved without a serious reduction in toughness (12). Aging treatment at 550°C for 1/2 hour was therefore selected by BHP to obtain an appropriate balance between strength and toughness for the CR HSLA 80 steel.

A similar precipitation hardening response has also been reported recently for Q&A (class 3) ASTM A710 steels (8,193,194,195). Despite the difference in plate thickness and thermal process prior to the aging treatment, it has been commonly reported that aging at 550°C induces peak hardening of hot rolled plates. The effect of aging time is considered to be minor compared to that of aging temperature (8,193).

The similar peak hardening temperatures of the unaged control rolled (CR) base plate and the GCHAZ suggests that the Cu precipitation hardening behaviour for both conditions is similar. It also implies that the majority of the Cu-rich clusters in the aged base plate (aged at 550°C for 1/2 hour) were redissolved in the HAZ regions during the weld heating cycle. Thus, Cu precipitation hardening in the base plate due to the aging treatment at 550°C for 1/2 hour is almost totally lost, contributing to the softening of the HAZ. As a result, the Cu content in solution in the HAZ reaches a similar level to that in the CR plate. The precipitation behaviour in the HAZ during further aging is therefore similar to that of the unaged hot rolled base plate.
Although the hardness of the base plate increased after subcritical reheating to 600°C for the BOP SA welds as indicated in Fig. 6.20, it should be noted that in this case, only a short duration thermal pulse was applied to the sample. This result indicates that, although the commercially treated plate is considered to show the maximum yield stress after an aging treatment at 550°C for 1/2 hour (12), this treatment does not, in fact, induce the maximum Cu precipitation hardening response.

In contrast, tempering by PWHT at 450, 550 and 650°C for 1 hour led to a reduction in hardness of the base plate, as a result of overaging of the Cu-rich precipitates (Fig. 6.58). As shown in Fig. 6.58, the higher the PWHT temperature, the higher degree of overaging and the lower the hardness value.

PWHT of the GCHAZ can result in Cu precipitation hardening as well as softening caused by tempering of the MA islands. Figure 7.2 shows a comparison of the GCHAZ microstructure before and after PWHT. The dark islands after PWHT (Fig. 7.4b) indicate the decomposition of the MA islands (white etching in Fig. 7.4a) due to tempering.

In the case of PWHT at 550°C for 1 hour, the rehardening by Cu precipitation must outweigh the softening caused by tempering of the MA islands, because of the observed overall hardening effect. In the case of PWHT of 450°C, softening due to tempering of the MA islands is the main reason for the slight decrease in hardness since the temperature is too low to induce any significant Cu precipitation hardening (12). For PWHT at 650°C for 1 hour, the softening effect of tempering of MA islands is added to the softening due to overaging of the Cu clusters, resulting in an overall reduction in hardness and improved toughness.

PWHT did not change the fracture mode of Charpy impact specimens. As for the fracture surfaces for the samples simulating the various HAZ regions, all samples after PWHT showed fracture surfaces consisting of the ductile initiation zone (Diz), and shear lips at
Fig. 7.4 Comparison of simulated grain coarsened HAZ region microstructures before (a) and after (b) PWHT at 550°C for 1 hour (3.1kJ/mm, Tp1=1300°C)(320x).
the sides of the fracture surface. Transgranular cleavage type of fracture (Fig. 7.5) occurred in the central part of fracture surface, similar to the GCHAZ before PWHT.

Specimens showing higher CVN values exhibited larger areas of shear lip and ductile initiation zone, again similar to observations for the simulated weld HAZ structures produced by a single thermal cycle to different peak temperatures.

To increase the HAZ strength without reducing HAZ toughness below the minimum toughness requirement, the temperature of PWHT should be set outside the age peak hardening range (500-600°C)(12). Since Cu precipitation hardening is approximately fully induced in the base plate, any additional aging treatment will result in overaging of the Cu clusters and reduction of strength in the base plate, especially at a temperature higher than 600°C (Fig. 6.58). To avoid significant reduction of base plate strength, the PWHT temperature should be kept below the peak hardening temperature range (<500°C). A detailed investigation is needed to determine the appropriate PWHT temperature and time to achieve optimum properties of the HAZ and base plate.

In summary, care must be taken when applying PWHT to the HAZ. Embrittlement of the GCHAZ was induced after PWHT at 550°C for 1 hour, as indicated by significant reduction of GCHAZ toughness and increase in hardness due to Cu precipitation rehardening. The Cu precipitation response of the GCHAZ was found to be similar to that of the CR base plate. As a result of the hardening, the toughness value for the GCHAZ was lower than the minimum requirement for HSLA 80 steel. PWHT of the GCHAZ at 450 and 650°C for 1 hour resulted in reduced hardness and improved toughness. PWHT at all three temperatures induced various levels of softening in the base plate because of overaging of the Cu clusters. Therefore, the PWHT temperature should be kept below the Cu precipitation peak hardening temperature of 500°C to avoid substantial loss of base plate strength.
Fig. 7.5 Quasi-cleavage appearance of central area of fracture surface for the simulated GCHAZ (Tp1=1300°C) after PWHT at 550°C for 1 hour Charpy impact sample (19J at -51°C).
7.5 TRANSFORMATION IN THE GCHAZ AND RESULTANT STRUCTURE AND PROPERTIES

7.5.1 Partial CCT Diagram of CR HSLA 80 Steel

Although transformation in the HAZ under isothermal (199) and continuous cooling conditions (66) has been reported for Q&A (class 3) A710 steel, the HAZ transformation behaviour of the present CR HSLA 80 steel, based on a modified A710 composition, is less certain.

The partial CCT diagram for the GCHAZ of 20mm plate is shown in Fig. 6.60. The peak temperature used was 1300°C. The equivalent heat input range corresponding to the four cooling conditions was 1.9-4.9kJ/mm.

The transformation start temperature was between 600-650°C, being lower at lower heat input. A much lower $\gamma \rightarrow \alpha$ transformation temperature (~150°C lower) was determined by Lundin et al. (66) for 12.5mm Q&A (class 3) A710 steel for a similar equivalent heat input range. The higher transformation temperature for the present steel can be attributed to the lower carbon and alloying element content (CE of 0.41) compared to the conventional A710 steel (average CE of 0.49). In addition, the TMCP procedure for the present steel resulted in a smaller grain size in the base plate, providing a larger grain boundary area for nucleation sites for austenite on reheating and resulting in smaller austenite grains. During cooling, the $\gamma \rightarrow \alpha$ transformation would be stimulated by a larger austenite grain boundary area acting as nucleation sites for transformation product, contributing to the higher transformation temperature.

Study of phase transformation in the GCHAZ indicated that the microstructure of the GCHAZ resulting from equivalent heat inputs between 1.9-4.9kJ/mm is a mixture of
various ferrite structures and MA islands (granular structure and/or granular bainite). As heat input decreased, the microstructure of the GCHAZ was refined, accompanied by an increase in hardness of this region. For the lowest heat input (1.9kJ/mm), the fast cooling rate enabled the formation of a small volume fraction of lath martensite.

7.5.2 Properties of the Simulated GCHAZ

Toughness, hardness and microstructure for various heat inputs have been discussed previously in Section 7.3.2, with reference to the effect of heat input on HAZ structure and properties. The microstructure for each heat input consists predominantly of various ferrite structures with MA islands. A structural coarsening effect is obvious as the heat input increases (Fig. 6.54). At lower heat input, a small amount of a lath martensite is observed in locations corresponding to the original pearlite bands in the base plate. The hardnesses of the simulated GCHAZs were higher for a faster cooling rate (lower heat input) (Fig. 6.55).

Results for the simulated GCHAZ can be summarised as follows:

(1) the hardnesses of the simulated GCHAZs for the equivalent heat input range of 1.9-4.9kJ/mm were between 203-236HV5, much lower than the conventional critical hardness (350HV) for susceptibility of cold cracking (Section 3.2);
(2) the CVN values for the simulated GCHAZ were similar to those of the base plate (Fig. 6.56), indicating a relatively high HAZ toughness;
(3) the $\gamma \rightarrow \alpha$ start temperature for the GCHAZ was in the range 600-650°C, being lower at lower heat input.
7.5.3 Effect of Copper and Nickel Contents on HAZ Microstructure and Hardness

To study the combined effect of Cu and Ni contents on HAZ transformation behaviour, a similar partial CCT diagram (Fig. 6.61) was obtained for a low carbon equivalent 350 MPa grade steel (LCE 350) with similar composition to the CR HSLA 80, except for a slightly higher carbon and significantly lower Cu and Ni contents (Table 5.3). The equivalent heat input range in this case was 1.8-4.8kJ/mm, very close to the range for the CR HSLA 80 steel.

Compared to the CR HSLA 80 steel, a higher HAZ transformation temperature was obtained for the LCE 350 steel. However, the transformation temperature increment was relatively small (~25°C). The lower $\gamma \rightarrow \alpha$ transformation temperature for the CR HSLA 80 steel is consistent with the higher contents of Cu and Ni which suppress the transformation of austenite. The reduction of transformation temperature due to the higher Cu and Ni contents in CR HSLA 80 steel must be greater than the increase in transformation temperature because of the slightly lower carbon content. As a result, the CR HSLA 80 steel showed an overall lower transformation temperature. The mild effect of Cu in suppressing austenite transformation has been reported by Culter et al. (177). The effect of Ni on the stability of the austenitic phase in iron is also well documented (228,229,230).

The simulated GCHAZ microstructures of the LCE 350 steel at different heat inputs were also a mixture of ferrite and second phase islands as shown in Fig. 6.63. The trend in microstructural change with heat input was similar to the CR HSLA 80 steel, i.e. a finer microstructure was obtained at a lower heat input. However, compared to the CR HSLA 80 steel, the carbon enriched second phase constituents were mainly diffusional transformation products (pearlite or bainite - the dark etching constituents in Fig. 6.63), instead of MA islands (Fig. 6.54). The structure of the GCHAZ for various heat inputs for the LCE 350 steel is softer than that of the CR HSLA 80 steel (compare Figs. 6.62
and Fig. 6.55), as a result of the higher transformation temperature of the LCE 350 steel and the lower volume fraction of MA islands.

In summary, comparison of the partial CCT diagrams for the GCHAZs of the CR HSLA 80 and LCE 350 steels revealed that for the HSLA 80 steel, the higher contents of Cu and Ni suppressed HAZ transformation to a lower temperature, increased HAZ hardness and changed the second phase islands in the GCHAZ structure from mainly a diffusional product (pearlite and/or bainite) to a nondiffusional product (martensite and retained austenite).
CHAPTER 8

CONCLUSIONS
The weldability of the newly developed control rolled HSLA 80 steel has been investigated in relation to heat-affected zone microstructure and mechanical properties. The present research was concentrated on six areas:

(i) the microstructure and properties of a single weld HAZ;
(ii) the effect of type of welding process on HAZ structure and properties;
(iii) the effects of heat input and welding speed on HAZ properties;
(iv) simulation of the HAZ structure in actual welds;
(v) the effect of multi-pass welding and postweld heat treatment on HAZ properties; and
(vi) the effect of copper and nickel on HAZ transformation.

The conclusions of this work are as follows.

(1) In general, the HAZ of welded CR HSLA 80 steel showed a relatively high Charpy impact toughness. This high HAZ toughness is related to the low carbon content and the low carbon equivalent of the steel, which generate a predominantly ferritic microstructure under normal welding conditions.

(2) Hardness surveys of the HAZ of welds produced by bead-on-plate submerged arc (HI=2.5-6kJ/mm); bead-on-plate flux cored arc (HI=1.0-2.5kJ/mm); and 4 wire SA (HI=2.5-10kJ/mm) welding; as well as simulated weld HAZs (HI=1.9-4.9kJ/mm), indicated that in all cases the hardness was well below 310 HV. For similar heat inputs, the bead-on-plate single-run weld HAZs showed higher peak hardnesses than the 4 wire SA welds. These results are directly related to the welding process through the arc force, weld bead shape and the subsequent cooling rate in the HAZ.

(3) The importance of weld bead profile on HAZ properties is reflected in the significant changes in HAZ structure and hardness with changes in the welding process, despite a nominally constant heat input.
(4) The commonly accepted proposition arising from the Rosenthal analysis that heat input determines the cooling rate in the HAZ and hence its structure and properties, is inconsistent with (3) and with the observation that marked variations occurred in the width, structure and hardness of the HAZ within a single BOP SA weld. The shape of the weld bead and changing heat transfer conditions around the fusion line dictate local variations in cooling rate and hence HAZ characteristics. However, the Rosenthal analysis was found to be more satisfactory when comparing the same HAZ position for welds made with the same process, but different welding parameters.

(5) The microstructure of the GCHAZ generally consisted of ferrite laths and interlath MA islands, but some lath martensite was detected in low heat input welds, particularly those produced by FCA welding.

(6) Among the various HAZ sub-regions, the grain-coarsened region (GCHAZ) exhibited the highest hardness and lowest impact toughness value because of the coarse structure, which contained a relatively high volume fraction of MA islands (~6%). In contrast, the partially transformed region showed the highest toughness and a hardness well below that of the base plate. The softening of the HAZ was related to the loss of Cu precipitation hardening.

(7) Reheating of the simulated grain-coarsened region to a subcritical tempering temperature of 600°C, an intercritical temperature of 800°C; and a grain refining temperature of 900°C improved the toughness of the GCHAZ region. This result suggests that overlapping and multi-pass welding can be safely conducted for this steel without deterioration in the toughness of the GCHAZ.

(8) Although the peak hardness of the HAZ increased with decreasing heat input for the four types of welds, the effect of decreasing heat input on HAZ toughness was relatively small. This effect is due to the opposing influences on toughness of increasing microstructural refinement and increasing hardness with decreasing heat input. Another
contributing factor is the relatively small change with decreasing heat input of the volume fraction of MA islands which are considered to be detrimental to toughness.

(9) Changing the welding speed by a factor of about 2 in 4 wire SA welds also had only a slight effect on HAZ microstructure and therefore, a minor effect on the toughness, hardness and strength properties of the HAZ.

(10) Strength compatible with that of the base plate was measured in transverse tensile tests of the weld joint produced by 4 wire SA welding at heat inputs of 2.5-5kJ/mm. However, for a higher heat input, significant deterioration of weld strength occurred due to the greater width of the HAZ which was softer than the base plate. The investigation showed however, that HAZ strength can be increased by postweld heat treatment at a temperature of or below 500°C, by inducing minor copper precipitation hardening, without a substantial reduction in the base plate strength.

(11) Postweld heat treatment of the simulated GCHAZ structure resulted in a significant decrease in toughness after treatment at 550°C for 1 hour. This embrittlement is associated with peak hardening by copper precipitation in the GCHAZ. Postweld heat treatment at both 450 and 600°C resulted in improved toughness and a softer HAZ, due to the tempering of the MA islands at both temperatures and overaging of Cu precipitates at 600°C.

(12) The partial continuous cooling transformation diagram determined for the grain coarsened HAZ region showed that the $\gamma \rightarrow \alpha$ start transformation temperature was between 600-650°C for a simulated heat input range of 1.9-4.9kJ/mm. A lower transformation temperature was obtained at a lower heat input (faster cooling rate). The microstructure of the simulated grain coarsened HAZ region was close to that of actual weld HAZs and consisted of a mixture of ferrite laths and MA islands. The major change in microstructure of the simulated GCHAZ with increasing cooling rate (decreasing heat input) was structural refinement and the appearance of low carbon lath martensite at low
(13) By comparison of the CR HSLA 80 steel with a similar reference steel with minor levels of Cu and Ni, the effect of these two elements on $\gamma \rightarrow \alpha$ transformation was established. The Cu (1.1%) and Ni (0.85%) in the CR HSLA 80 steel suppressed $\gamma \rightarrow \alpha$ transformation in the simulated GCHAZ. The $\gamma \rightarrow \alpha$ transformation start temperature was reduced by about 25°C. The major microstructural change brought about by the presence of copper and nickel was a change in the second constituent from a diffusional decomposition product (pearlite or/and bainite) to a nondiffusional product (MA islands). A higher HAZ hardness was also obtained in the HSLA 80 due to the formation of the harder second constituent (MA islands rather than pearlite or bainite) as a result of the lower $\gamma \rightarrow \alpha$ transformation temperature and the higher hardenability of the HSLA 80 steel.
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