2.1 INTRODUCTION

This chapter deals with the present status related to the welding metallurgy of high strength low alloy steels in general, and that of the quenched and tempered (Q&T) steel in particular. Since occurrence of hydrogen induced cracking (HIC) is the main problem in welding of the Q&T steel, factors that cause HIC, methods to control HIC and weldability tests to determine HIC resistance of the steel are discussed. Methods used to characterise the different factors that cause HIC are also discussed.

In the present work, one important type of welding electrode that is used for welding of the Q&T steel is austenitic stainless steel. There are certain metallurgical difficulties like solidification cracking and embrittlement due to the occurrence of sigma phase in austenitic stainless steel weld metal. The present status related to these problems are also discussed in this chapter.

2.2 HYDROGEN INDUCED CRACKING (HIC)

Hydrogen Induced Cracking (HIC) is also known as cold cracking, delayed cracking or underbead cracking. HIC is a defect that forms as the result of contamination of the weld microstructure by hydrogen. It is usually a delayed phenomenon. It requires both a requisite microstructure and a threshold level of stress. It also needs a critical level of hydrogen which is dependent on alloy and microstructure [1-6].

2.2.1 MECHANISM OF HIC FORMATION

The mechanism of HIC formation is still being studied by many investigators. Currently, the most widely accepted hypothesis involves the presence of pre-existing defect sites in...
the material - small cracks or discontinuities caused by minor phase particles or inclusions. In the presence of existing stress, these sites may develop local areas of very high biaxial or triaxial tensile stress. Hydrogen diffuses preferentially to these sites of dilated lattice structure. It has been postulated that both the pre-existing defects and an applied tensile stress will increase the actual amount of hydrogen at the crack tip. The increased solubility of hydrogen under an applied axial tensile stress has been estimated to be 5 times higher than the nominal solubility [7-9]. Also the hydrogen in front of the crack tip is about 1.2 to 1.4 times the nominal bulk hydrogen concentration [7]. As the local hydrogen concentration increases, the cohesive energy and strength of the lattice decrease. When the cohesive strength falls below the local intensified stress level, fracture occurs spontaneously. Hydrogen then evolves in the crack volume, and the process is repeated. This model of HIC is consistent with the relatively slow and discontinuous nature of the process [10].

During welding, hydrogen is absorbed by the weld pool from the atmosphere of the arc. Hydrogen readily decomposes in iron to its atomic state as per the reaction given below [11]:

\[ \text{H}_2 \rightarrow 2 \text{[H]} \]

During cooling, much of this atomic hydrogen escapes to the atmosphere, from the solidified bead, by diffusion; but some atomic hydrogen also diffuses into the heat affected zone (HAZ) and the parent metal. The quantity which does so depends on several factors such as the original amount absorbed, size of the weld, decreasing solubility (Figure 2.1) and the time-temperature conditions of cooling [9].

HIC has certain well established characteristics [11]. They include:

- Crack growth is a time dependent process.
- Crack growth occurs extremely non-uniformly, with periods of rapid growth followed by periods of arrested growth.
- Crack propagation is often of a semi-brittle, quasi-cleavage in nature, although in coarse grained material in which the grain boundaries are decorated with particles, it can often be intergranular.
- Hydrogen accumulation is only weakly dependent upon stress.
Figure 2.1
Effect of temperature on the solubility of hydrogen in iron
(Source: Ref. 9)

Studies on weldability of a quenched and tempered high strength low alloy steel
2.2.2 FACTORS RESPONSIBLE FOR HIC AND THEIR CONTROL

HIC occurs when the four factors (1) diffusible hydrogen, (2) stress level, (3) susceptible microstructure and (4) low temperature, are present simultaneously. Sources of these factors and methods to control them are discussed in the following sections [1-5,10].

(1) Diffusible hydrogen:

During welding, hydrogen is absorbed by the weld pool from the arc atmosphere. During cooling, much of this hydrogen escapes from the solidified bead to the atmosphere by diffusion but some diffuses into the HAZ and the parent metal. The principal sources of hydrogen in welding consumable include moisture and other hydrogenous compounds in the electrode coating or flux, oil, dirt, rust and grease on the surface of welding electrodes or wires. The principal sources of hydrogen from the base material include oil, grease, dirt, paint, rust etc. on the surface and adjacent to the weld preparation. In addition some hydrogen pick-up may arise from moisture in ambient atmosphere.

Hydrogen from welding consumable is controlled by proper storage like storing in dry rooms and drying or baking of coated electrodes and fluxes. Many of the earlier works have shown that reduction of weld metal hydrogen is one of the most effective means of reducing HAZ and weld metal cracking [12].

(2) Stress level:

Stresses are developed by thermal contraction of the cooling weld. The stress acting upon a weld is a function of weld size, joint geometry, fit-up, external restraint and the yield strength of the parent steel and weld metal. *HIC does not occur below a minimum stress level known as critical stress.* In one type of Q&T low alloy steel, called HY-80, W.F. Savage has reported that a stress of the order of the yield strength must be present to produce HIC [13]. Stresses may be reduced by preheating or by adjusting the welding procedure or by redesigning the weldment or by fabrication sequence to reduce restraint on the joint.
(3) **Type of microstructure:**

HIC occurs in regions containing certain microstructures which are hard and coarse grained. The HAZ that is closest to the fusion boundary is raised to a sufficiently high temperature to produce a coarse grain size. This high temperature region, because of its coarse grain size, is not only more hardenable but also less ductile than regions farther from the fusion boundary. It is thus this region in which the greatest risk of cracking exists.

As a general rule, the harder the microstructure the greater is the risk of cracking. Soft microstructure can tolerate more hydrogen than the hard one. Soft microstructures can be obtained by using steel with low contents of carbon and alloying elements to reduce the hardenability of the HAZ. Additionally the use of high heat input, thin plate and preheat will reduce the quenching rate thus leading to soft microstructure in HAZ.

(4) **Temperature:**

HIC occurs normally at ambient temperatures. Cracking is unlikely to occur above about 150°C and below about –100°C. It is therefore possible to avoid cracking in a susceptible microstructure by maintaining it at a sufficiently high temperature by post-heating until sufficient hydrogen has diffused away.

2.2.3 **CARBON EQUIVALENT (CE)**

As discussed earlier, the type of microstructure that is susceptible for HIC mainly depends on the hardenability of the steel. The hardenability of the steel is governed by its composition, and a useful way of describing hardenability is to assess the total contribution of all the elements to the hardenability. This is done by an empirical formula which defines a carbon equivalent (CE) that takes into account the important elements which are known to affect hardenability [10]. One important formula that is widely used is known as IIW (International Institute of Welding) formula. This is given as:

\[
CE_{IIW} = C + \frac{Mn}{6} + \frac{Cr+Mo+V}{5} + \frac{Ni+Cu}{15}
\]
This formula was first proposed by Dearden and O'Neil in 1940 [14]. This formula may be used for steels with carbon contents more than 0.18% or in the case of welding conditions requiring slow cooling, which means $t_{8/5}$ (time taken for cooling from 880 to 500°C) is longer than about 12 sec. [15,16].

Several other CE formulae have been proposed from time to time. One formula developed in Japan for steels with low carbon content, whose behaviour with regard to HIC is not well described by IIW formula, is known as composition parameter ($P_{cm}$) formula. This is given as:

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

This formula was first proposed by Ito and Bessyo in 1968 [17]. This formula may be preferred for steels with carbon contents less than 0.22% and in the case of rapid cooling, which means, $t_{8/5}$ is shorter than about 6 sec [16]. $P_{cm}$ gives an increased importance to carbon and includes the effect of the micro-alloying element boron. It has been shown that the $P_{cm}$ formula accurately predicts the cracking tendencies of mechanically processed low-carbon steels, especially those that were being utilised in the oil and gas industry for pipe line and structural plate. Since the maximum hardness developed at any location depends on the absolute carbon content and not on the carbon equivalent, B.A.Graville has constructed a diagram to classify the relative weldability of all the ferritic steels according to their relative ease of weldability [18]. This diagram, called Graville diagram, incorporates both the CE and the absolute carbon content (figure 2.2). This diagram has three zones. Zone-I steels have low carbon and low hardenability and are not very susceptible to cracking. Zone III steels have both high carbon and high hardenability, and all welding conditions will produce crack-sensitive microstructures. Therefore, to avoid HIC in Zone III steels, the user must apply low hydrogen procedures, including preheat and post weld heat treatment. Zone II steels have higher carbon levels with lower hardenability.
Susceptibility of steels to HIC is given relative to carbon content and CE, where CE = %C + (%Mn + %Si)/6 + (%Ni + %Cu)/15 + (%Cr + %Mo + %V)/5. Susceptibility to HIC progressively increases as steels migrate from zones I to II, to III.

Figure 2.2
Graville diagram showing susceptibility of steels to HIC (source: Ref.18)

Figure 2.3
Wong diagram showing susceptibility of steels to HIC as a function of diffusible hydrogen content and $P_{cto}$. (Source: Ref.19).
Thus it is possible to avoid crack sensitive microstructures in this zone by restricting the HAZ cooling rates. This can be accomplished through control of heat input and to a minor extent, with preheat. It may be noted that the quenched and tempered steel which is studied in the present work with C=0.3% and CE=0.7% falls in Zone III and hence this steel will require special considerations for welding.

From the carbon equivalent concept, we can conclude that low CE steels can tolerate high diffusible hydrogen while high CE steels cannot do so. Hydrogen cracking susceptibility has been related to diffusible hydrogen content and $P_{cm}$, in the form of diagrams that map cracking versus non-cracking behaviour, as shown in Figure 2.3 [19].

CE is sometimes used as a basis to decide whether preheat and post heat treatments for steels are required, as below:

$$
\begin{align*}
\text{CE} &< 0.45\% & \text{No specified treatment is required.} \\
\text{CE} &0.45 - 0.60\% & \text{Preheat alone is required.} \\
\text{CE} &> 0.6\% & \text{Preheat and post heat are required.}
\end{align*}
$$

### 2.2.4 USE OF AUSTENITIC STEEL AND NICKEL ALLOY WELD METAL TO CONTROL HIC

For steels which are very highly prone for HIC, methods like preheating and post heating are not sufficient. The only alternative is then to use a proper combination of welding process and consumable which virtually prevent the introduction of hydrogen into the HAZ so as to produce a weld metal insensitive to hydrogen. This is can be achieved by the use of austenitic stainless steel or nickel alloy electrodes [10, 20-23]. Weld metals made by both the austenitic steels and the nickel base alloys have face centred cubic structure, and hence have much higher solubilities for hydrogen in comparison with ferritic weld metals. This restricts the hydrogen that is diffused into the HAZ to a very low value.
Moreover the hydrogen which has diffused from weld metal into the HAZ (at high temperature when the HAZ was austenitic), will diffuse back into the weld metal during solidification. This is because during solidification, the HAZ transforms from austenitic structure to ferritic/martensitic structure, whereas the weld metal remains austenitic throughout.

### 2.2.4.1 Justification of undermatching of weld metal

Weld metals made of austenite stainless steels and nickel alloys have low strength levels, compared to the Q & T parent steel. Some applications of the Q & T steel do not require that the hardness and strength of the weld metal match those of the base metal [4]. The location of weld joints and the orientation of applied stresses can be such that the weld properties need not match base metal properties after a quench and temper treatment. In such cases, the austenitic stainless steel and nickel alloy electrodes can be used.

### 2.2.4.2 Austenitic steel electrode compared to nickel alloy electrode

There are some important differences between these two types of electrodes [10]. Nickel alloy electrodes are less prone to give cracked martensitic regions along the fusion boundary, which may occur with austenitic stainless steel electrodes. Nickel alloy electrodes, unless selected to minimise the problem, may suffer solidification cracking whereas austenitic electrodes would be immune for the same. Suitable nickel alloy electrodes are not available to give the high strength levels possible with some of the austenitic stainless steel types. Nickel alloy fillers have the advantage of lower coefficient of thermal expansion than stainless steels, and this can reduce shrinkage strains and thus reducing the risk of cracking in highly restrained joints. If post weld heat treatment (PWHT) is required, a nickel alloy filler may well become the preferred choice to avoid formation intermetallic during heat treatment loading to embrittlement.
2.2.4.3 Choice of nickel alloy electrode types

Taking into account the tolerance to base metal dilution and the strength requirement, inconel types of electrodes like ENiCrFe-3 and ENiCrFe-2 can be used for welding of Q&T Steels [4].

2.2.4.4 Choice of austenitic steel electrode types

When selecting austenitic stainless fillers, it is necessary to ensure that dilution from the base steel can be satisfactorily accommodated. The normal choice of austenitic consumable for shielded metal arc welding is from types 23Cr-12Ni (type 309), 29Cr-9Ni (type 312) or 20Cr-9Ni-3Mo (type 308Mo) [10, 18, 20-24]. The first named is most commonly used and is suitable for giving deposits containing sufficient ferrite to suppress solidification cracking, with little or no martensite in the bulk deposit. The 29Cr-9Ni (type 312) may be preferred for high dilution runs to avoid a fully austenitic deposit. This electrode will yield a weld microstructure of predominantly austenitic plus 5 to 10% ferrite even at dilution levels of around 45%. However in a low dilution situation, the weld metal will contain a high ferrite percentage, perhaps as high as 35%. Although this is of possible benefit in tolerating the pick-up of sulphur from the parent steel, and also in giving a high strength, this type of high ferrite weld deposit should not be subjected to post weld heat treatment, since it will show marked embrittlement as a result of the formation of the sigma phase during heat treatment.

2.3 EFFECT OF MICROSTRUCTURES ON HIC IN Q&T STEELS

2.3.1 EFFECT OF MICROSTRUCTURE IN LOW CARBON Q&T STEELS:

Low carbon Q&T steels have carbon content not exceeding 0.25%. Steels covered by ASTM specifications like A514, A517, A533, A543, A553 and A678 and the steels covered by military specifications like HY-80, HY-100, HY-130 are examples of low
carbon Q&T steels. Welding of these steels involves choosing the welding conditions which promote the formation of martensite (low carbon martensite).

In these steels, preheat temperatures are limited to the lowest possible values, because preheat reduces the cooling rate of the HAZ. If the cooling rate is too slow, the reaustenitised zone adjacent to the weld metal can transform either into coarse bainite or into ferrite with regions of high-carbon martensite. Both these microstructures lack high strength and good toughness. On the other hand, if the cooling rate is rapid, austenite will transform to low carbon martensite and soft bainite, resulting in good toughness. The toughness of martensite structures has been shown to vary inversely with carbon content [25, 26]. The low-carbon martensite is a constituent with good notch toughness [27-30].

2.3.2 EFFECT OF MICROSTRUCTURE IN MEDIUM CARBON Q&T STEELS:

Medium carbon Q&T steels have carbon content of 0.25 to 0.45% with sufficient chromium, nickel, molybdenum and other alloying elements to allow the plate to be quenched to martensite throughout their cross section regardless of their thickness. These steels are sometimes referred to as Heat Treatable Low Alloy (HTLA) steels [3,4]. These steels seem to have had their beginning with the armour plates used extensively in ships and tanks during world war II [1]. Welding of these steels involves choosing the welding conditions which suppress the formation of martensite (high carbon martensite) and promote formation of soft bainite [4, 15].

The combined carbon and alloy contents of these steels are sufficiently high to promote the formation of high carbon martensite, which may make the material more susceptible to HIC. The high carbon martensite is a brittle phase with low notch toughness [27]. The best approach to welding these steels is to preheat the joint area, thereby slowing the cooling rate, enough to permit formation of softer bainite instead of hard martensite.

For welding of these medium carbon Q&T steels, the minimum preheat and interpass temperatures required to prevent cracking depends on carbon content and alloy content, heat treatment condition, section thickness or amount of joint restraint and available hydrogen during welding. The ideal preheat temperature is about 30°C above the
temperature at which martensite starts to form on cooling (Ms). Holding at this
temperature for a time after welding will produce a bainite structure in the weld HAZs. It
will also permit dissolved hydrogen to escape. However the Ms temperature of many of
these steels is above 250°C. A preheat temperature of 280°C or higher will contribute to
welder discomfort and promote formation of a thin oxide layer on the joint faces.
Welding at a preheat temperature below 200°C will require that the hydrogen content be
kept extremely low to prevent cracking.

Thus the successful welding of medium carbon Q&T steels requires the following :

♦ Proper preheat and interpass temperature requirements.
♦ Achievement of a low hydrogen content in the weld metal especially when the
  preheat temperature is less than 200°C.
♦ Need to maintain preheat after welding is completed (post heat).

2.3.3 EFFECT OF GRAIN SIZE

Previous studies have found that coarsening of the prior austenite grain facilitates HIC
[30-32]. Accordingly HAZ that is the closest to the fusion line (coarse grained zone) is the
most vulnerable for HIC. High heat inputs will facilitate more duration for grain growth.

2.4 EFFECT OF MARTENSITE START TEMPERATURE ON HIC

Traditionally, welding experts have used carbon equivalent type expressions, which
include chemical composition and cooling rates, to determine the susceptibility of a steel
to HIC. More recently, however, there is a growing concern that the hydrogen
distribution across the weldment is non-uniform, due to different regions of weldment
undergoing austenitic decomposition at different times.

2.4.1 AUSTENITE DECOMPOSITION AT DIFFERENT TIMES

Depending on the hardenability of the HAZ (base metal) and the weld metal, these
regions may undergo austenite decomposition (more specifically, martensite
transformation) at different times [19,33,34]. As a result of the earlier or later transformation of the weld metal (compared with the HAZ), hydrogen may accumulate in the weld metal or in the HAZ. Selective distribution of hydrogen will lead to decreased or increased hydrogen damage susceptibility of the welded joint.

Austenite (FCC structure) has higher solubility but much lower diffusion rate of hydrogen compared to martensite (BCT structure) and ferrite (BCC structure) [35]. As a result, different amounts of hydrogen will be retained in the HAZ, depending on the austenite decomposition behaviour of the alloys.

When the weld metal and the HAZ have different transformation temperatures, the hydrogen is redistributed across the interface between ferrite and austenite over a period of time. At high temperature, both sides of the fusion boundary are austenite and hydrogen can diffuse from the weld metal to create an equal concentration on either side of the fusion boundary. On cooling, depletion of hydrogen in the ferrite, and accumulation of hydrogen in the austenite takes place. This continues to develop as long as the phases are different. When the austenite finally transforms, the enriched spike and the depleted zones of hydrogen will redistribute themselves.

Graville has shown an example of the calculated hydrogen distribution across the interface, as shown in Figure 2.4, where instantaneous transformation at 100°C on one side and 500°C on the other side is assumed [34]. The location of the peak at room temperature (in Figure 2.4) is about 1 mm from the interface, which is typically the location of cracks observed in high carbon martensitic HAZs. If one side (e.g., the weld metal) remains austenite at room temperature, the other side is depleted to about 40% of the original hydrogen level. For most HSLA steels the transformation temperatures are sufficiently high so that the hydrogen has time to completely redistribute, and the enriched spike and depleted zones are removed before cooling to room temperature. It is however not known whether this local concentration phenomenon plays a role in the susceptibility of martensite / austenite microphases when transformation occurs at low temperatures.
Figure 2.4
Distribution of hydrogen across the interface between two phases at various temperatures
(Source: Ref.34)
2.4.2 MARTENSITE START TEMPERATURE EXPRESSIONS

From the earlier discussion, it can be seen that the martensite start temperature is both a measure of the microstructure evolution and the ability to have available a phase for rapid hydrogen transport. The effect of specific alloying additions on the martensite start temperature is commonly represented by the formulation of Andrews and shown in the equation given below [19]:

\[
Ms = 539 - 423C - 30.4Mn - 17.7Ni - 12.1Cr - 7.5Mo
\]

This equation has been greatly used since 1965, and the predictions are quite accurate for alloy steels with C = 0.6% max., Mn = 4.9% max., Cr = 5% max., Ni = 5% max. and Nb = 5.4% max.

J.A. Self et al. developed more comprehensive expressions for both the HAZ and weld metal which are able to predict the martensite start temperature over a large compositional range. These expressions are given below [36,38]:

For plate: \[Ms = 521 - 350C - 14.3Cr - 17.5Ni - 28.9Mn - 37.6Si - 29.5Mo - 1.19CrNi - 23.1(Cr+Mo)C\]

For weld metal: \[Ms = 521 - 350C - 13.6Cr - 16.6Ni - 25.1Mn - 30.1Si - 20.4Mo - 40Al - 1.07CrNi + 21.9(Cr+0.73Mo)C\]

2.5 HYDROGEN AVAILABLE FOR CRACKING

In this section, it is intended to highlight that portion of the total hydrogen that diffuses towards the crack-susceptible regions and relate it with time for the occurrence of the cracking process using diffusion phenomenon.
2.5.1 POTENTIAL HYDROGEN VERSUS WELD HYDROGEN

Direct laboratory measurement of the moisture or hydrogen level of any consumable produces a result which is termed the potential hydrogen level of the process [10]. This term has been chosen because it is known that not all the measured hydrogen is absorbed by the weld pool. It is potentially available and, in general, the higher the potential hydrogen level, the higher will be the actual weld hydrogen. Measurement of potential hydrogen is intended to reveal the amount of hydrogen that is potentially available for absorption by the weld pool during cooling. It therefore provides a means of characterising the quality of a consumable with respect to hydrogen before that consumable is put to use. In case of electrode coatings (shielded metal arc welding) and submerged arc fluxes, measurement of the moisture content is assumed to represent the potential hydrogen level. Typical values of moisture content for different types of coated electrodes are as follows:

<table>
<thead>
<tr>
<th>Coating Type</th>
<th>Moisture Content</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cellulosic coating</td>
<td>6 - 10 %</td>
</tr>
<tr>
<td>Rutile coating</td>
<td>0.5 - 2.5 %</td>
</tr>
<tr>
<td>Basic coating (low hydrogen)</td>
<td>Less than 1.0 %</td>
</tr>
<tr>
<td>Basic coating (very low hydrogen)</td>
<td>Less than 0.3 %</td>
</tr>
</tbody>
</table>

Weld hydrogen level represents the amount of hydrogen absorbed by the weld metal during welding. Weld hydrogen depends on potential hydrogen level, as well as, on other factors like welding conditions and cooling conditions. Weld hydrogen is the summation of residual hydrogen and diffusible hydrogen.

Residual hydrogen is that portion of the weld hydrogen which will not be released by diffusion at ambient temperature, but may be extracted at higher temperatures, e.g. at 650°C. It is believed that this hydrogen is trapped in molecular form, in the lattice, in voids (particularly in association with non-metallic inclusions) or in chemical combination with other elements, such as carbon (when methane is formed). Diffusible hydrogen is that portion of the weld hydrogen which will be released by diffusion when the steel is held for several days at ambient temperature.
Of all the hydrogen parameters defined above, diffusible hydrogen alone gets practical significance, since the same is responsible for creating conditions for cracking in HAZ. Typical values of diffusible hydrogen for different types of coated electrodes are as below:

- Cellulosic coating: 70 - 100 ml / 100 g of deposited metal
- Rutile coating: 20 - 33 ml / 100 g of deposited metal
- Basic coating (low hydrogen): 5 - 10 ml / 100 g of deposited metal
- Basic coating (very low hydrogen): Less than 5 ml / 100 g of deposited metal

### 2.5.2 HYDROGEN AVAILABLE AT THE CRACK TIP

In welding, hydrogen available for the microcracking process is determined by the competing factors of hydrogen diffusion to the crack tip under a stress gradient and the escape of hydrogen from the weld metal to the atmosphere [39]. This situation is depicted for both high and low hydrogen concentration in the weld produced by E7010 and E7018 (properly dried) electrodes, respectively in Figure 2.5 [40].

It is assumed that, initially, the weld bead and the HAZ next to the fusion line in the vicinity of the notch (where crack initiates) have a uniform distribution of hydrogen concentration of about 30 ppm in the case of the E7010 electrode and 5 ppm for the E7018 electrode. With the passage of time after welding, the average hydrogen concentration in the matrix (weld metal and coarse grained HAZ) will decrease due to the outgasing of the charged weld metal to the atmosphere and to the surrounding base metal. The regions of high tri-axial stress (near voids, just ahead of the notch or near internal cracks as dictated by the microstructure) have a high affinity for hydrogen, and the local concentration in this region becomes greater than that of the surrounding matrix. Thus, while the hydrogen concentration tends to increase in the regions of high triaxial stress, there is a decrease in the average hydrogen concentration in the matrix. This situation is depicted in Figure 2.5(b).
Figure 2.5
Schematic showing diffusion tendencies of hydrogen in weldment
(Source: Ref.40)

(a) Escape paths for hydrogen
(b) Variation of hydrogen concentration in the matrix and at the crack tip with time

Figure 2.6
Loss of hydrogen from an infinite plate as a function of time
(Source: Ref.41).
As the crack propagates, part of the atomic hydrogen that had diffused into the triaxially stressed region is lost as molecular hydrogen behind the growing crack. This, together with the continual escape of hydrogen from the weld metal to the atmosphere with the passage of time, provides a situation wherein lesser and lesser amounts of hydrogen are available in the matrix for diffusion to the triaxially stressed regions.

Ultimately, the local hydrogen concentration at the crack tip will also decrease with time as shown in Fig. 2.5(b). Finally when the hydrogen concentration at the crack tip falls below the critical level, further crack growth should cease. This is the situation for implant test specimens loaded below the lower critical stress.

2.5.3 THEORETICAL ESTIMATION OF HYDROGEN CONTENT IN THE CRACKING ZONE AFTER THE DELAY PERIOD

The diffusible hydrogen test can be used to determine the amount of hydrogen initially solidified into the weld pool. However since fracture occurs sometime after the weld has cooled down, and some hydrogen would have been lost by diffusion, these results must be analysed to determine the amount of hydrogen remaining in the cracking zone at the instant of fracture.

The amount and distribution of hydrogen remaining in an implant specimen as a function of time after welding can be estimated from hydrogen removal curves [7,10]. These curves have been generated from the diffusion formulae that relate the hydrogen concentration with time. The curve, shown in Figure 2.6 shows the percentage of the original hydrogen concentration left at the centre of an infinite sized plate as time(t) progresses [41]. Time appears in the dimensionless expression, $D t / L^2$, which forms the abscissa, where $D$ is the appropriate overall diffusivity coefficient for the temperature and material under consideration, and $L$ is the critical dimension of the solid, represented by the half-thickness of plate. Similar curves can be generated for geometric shapes other than plates. This type of hydrogen-removal calculations can be particularly used to calculate the duration of post heating required to avoid HIC. Bailey et al., while formulating the above procedure, have assumed that hydrogen diffuses at approximately
the same rate either in weld metal, or in the HAZ or the parent metal [10]. This assumption may not be applicable for the diffusion of hydrogen from welds deposited using austenitic or nickel alloy fillers.

2.6 EXPERIMENTAL METHODS TO DETERMINE DIFFUSIBLE HYDROGEN

Experimental determination of diffusible hydrogen by any method involves depositing a single weld bead, quenching it rapidly, maintaining the quenched specimens at room temperature for a sufficient time to release its content of diffusible hydrogen, measuring the diffusible hydrogen by appropriate volumetric methods and reporting it as on unit of mass of fused metal. Relevant national standards from ANSI/AWS, BS, ISO and BIS serve as useful guides to determine the diffusible hydrogen [42-45]. Depending on the type of analytical apparatus used for collection and measurement of diffusible hydrogen, 3 methods are commonly available: (1) Mercury displacement method, (2) Gas chromatography method and (3) Glycerine method. The first two methods are known to measure atleast 90% of the total diffusible hydrogen in the test specimen. The Glycerine method has an important disadvantage of a certain quantity of hydrogen solubility in glycerine fluid which leads to error in the measured quantity of diffusible hydrogen.

2.7 EXPERIMENTAL METHODS OF VARIATION OF DIFFUSIBLE HYDROGEN

Many times it is required to have methods to vary the diffusible hydrogen in order to relate the varying amounts of diffusible hydrogen with the weldment characteristics. Researchers have followed different methods to vary the diffusible hydrogen.

W.F. Savage varied the diffusible hydrogen in gas tungsten arc welding (GTAW) by bubbling the argon shielding gas through a certain fixed column of water [46]. This had caused sufficient moisture pick-up to provide diffusible hydrogen upto 9ml/100g of weld
metal. Another method of varying hydrogen in GTAW is through addition of hydrogen in varying amounts from 1 to 5%, to the argon gas shielding [47].

In shielded metal arc welding (SMAW), iron powder basic coated electrodes can be baked at 400°C for 1 hour to yield a very low diffusible content of upto 2.0ml/100g [48,49]. In order to increase the hydrogen level, W.F.Savage had recommended soaking these electrodes in water (estimated H content 30-35 ml/100g) or to expose to air for a long time (estimated H content 8-10 ml/100g) [49].

2.8 IDENTIFICATION OF MICROSTRUCTURE IN OPTICAL METALLOGRAPHY

In welding of Q&T steels, one of the factors that causes HIC in HAZ is susceptible microstructure. In the earlier sections, it has been shown seen that for the medium carbon Q&T steel (the material used for the present study), formation of high carbon martensite makes the steel prone for cracking. Formation of softer bainitic structure is desirable. Many times, one comes across mixture of martensite and bainite at HAZs of these steels. Hence it is appropriate to have a proper etchant to differentiate martensite and bainite in optical metallography.

LePara has suggested an etchant, especially for identification of HAZ microstructures in the weldments of high strength dual phase steels as detailed below [50,51]:

**Etchant Preparation & Etching procedure:**

Solution-a: 1% Sodium Metabisulphate in distilled water

Solution-b: 4% Picric acid in Ethanol mixed in 1:1 volume ratio.

The specimen should be etch-polished using the solution-'b' (4% picral) to remove all disturbed metal (approximately three etch-polished sequences are sufficient). The etchant
is then prepared by mixing solution-'a' and solution-'b' at 1:1 ratio and the specimen is immediately immersed in it for 5-10 seconds.

Effect on microstructure: Martensite appears white; Bainite appears black; Ferrite appears tan (or grey). Also shows coring of solidification structure in weld metal.

2.9 CALCULATION OF COOLING RATE

The type of microstructure, and its susceptibility to cracking for a given steel, is strongly dependent on the peak temperature attained during welding and the cooling rate through the austenite transformation temperature region. Available methods to theoretically calculate the cooling rates, through equations and nomograms, are briefly reviewed in the following sections.

2.9.1 COOLING RATE EQUATIONS

Theoretical equations have been developed for calculation of peak temperatures and cooling rates as a function of material properties and welding parameters (25,33,52,53). These equations are give below:

(a) For three dimensional heat flow (thick plates): \[ R = \frac{2\pi k(T_c - T_0)^2}{H_{net}} \]

(b) For two dimensional heat flow (thin plates): \[ R = 2\pi k \rho C \left( \frac{h}{H_{net}} \right)^2 (T_c - T_0)^3 \]

Where
- \( R \) = cooling rate at weld centreline, °C/s
- \( k \) = thermal conductivity of the metal, J/mm.s. °C
- \( T_0 \) = Initial / Preheat temperature of plate, °C
- \( T_c \) = Temperature at which the cooling rate is calculated, °C
- \( h \) = thickness of base metal, mm
- \( \rho \) = density of base metal, g/mm³
\( C \) = specific heat of base metal, J/g, °C
\( \rho C \) = volumetric specific heat, J/mm³, °C (0.0044 for steels)
\( H_{\text{net}} = \text{Net heat input} = f \cdot \frac{VI}{S} \)
\( V \) = Arc voltage, volts;
\( I \) = Welding current, Amps
\( S \) = Travel speed of heat source, mm/s
\( f \) = Heat transfer efficiency (which varies from 0.8 to 1.0 for most consumable electrode processes).

The thick plate equation is used when heat flow is three dimensional (heat is conducted downward as well as lateral from the weld). The thin plate equation would apply to any weld in which heat flow is essentially lateral, i.e., a base metal thin enough that the difference in temperature between the bottom and the top structures is small in comparison to the melting temperature.

The critical temperature for most steels is assumed to be 550°C, and hence the cooling rates, using the above equations are calculated at \( T_c = 550°C \). A dimensionless quantity called "relative plate thickness", \( \tau \), is used to define the terms, "thick plate" and "thin plate", as given below:

\[
\tau = h \sqrt{\frac{\rho C (T_c - T_o)}{H_{\text{net}}}}
\]

The thick plate equation applies when \( \tau \) is greater than 0.9 and the thin plate equation applies when \( \tau \) is less than 0.6. When \( \tau \) falls between 0.6 and 0.9, the upper bound of the cooling rate is given by the thick plate equation, and the lower bound by the thin plate equation. The relative plate thickness between 0.6 and 0.9 are sometimes referred as "2.5D behaviour" [54].

2.9.2 NOMOGRAMS FOR COOLING TIME BETWEEN 800 AND 500°C

To obtain an accurate and reproducible prediction of the maximum HAZ hardness, the cooling rate or cooling time between a certain temperature range should either be measured or calculated by available nomograms and prediction equations. The HAZ thermal cycle has been variously characterised by cooling rates at 300°C, 540°C, 700°C,
but recently the cooling time between 800 to 500°C, denoted as $t_{g/5}$, has found universal acceptance because of ease and accuracy in its determination [76].

Several authors have published nomograms for the determination of cooling time, $t_{g/5}$ [54,55]. Inagaki has suggested the nomograms as shown in Figures 2.7 to 2.9, for shielded metal arc welding (SMAW), gas metal arc welding (GMAW) and submerged arc welding (SAW) respectively [56-58]. Using these, the cooling time, $t_{g/5}$ can be simply measured off, based on gross heat input, plate thickness and preheat temperature. Inagaki et al. have constructed these nomograms based on experimental data from actual welding. These nomograms are applicable for a maximum plate thickness of 34 mm and heat inputs limited to those shown in the figures. Within the applicable heat input range, $t_{g/5}$ for thicknesses greater than 34 mm will be similar to that for 34 mm.

There are nomograms, which are purely based on heat flow calculations, some of which can be found in the work of Lundin et al.[54]. Similarly, a number of equations are available in literature to calculate $t_{g/5}$ from the welding parameters. Some of them can be found in the works of Lundin et al.[54] and Dieter Radaj et al.[55].

2.10 DISTRIBUTION PATTERN OF RESIDUAL STRESSES IN WELDMENTS

One of the factors to cause HIC in welded structures is the presence of tensile residual stresses at the regions containing crack susceptible microstructures. It is important to understand the stress distribution pattern because only those regions which are subjected to tensile stresses are prone for cracking and regions subjected to compressive stresses are not affected by hydrogen embrittlement. In this section, a brief account of the different components that will constitute the residual stress and the effects of these individual stress components on the stress distribution pattern are discussed.
Figure 2.7
Cooling time, $t_{g5}$, for shielded metal arc welding dependent on gross heat input ($q_w$), plate thickness ($h$) and preheat temperature ($T_0$)
(Source: Ref. 56-58)

<table>
<thead>
<tr>
<th>$Q_w$ [kJ/mm]</th>
<th>$t_{g5}$ [S]</th>
<th>$T_0$ [°C]</th>
<th>$h$ [mm]</th>
</tr>
</thead>
<tbody>
<tr>
<td>S</td>
<td>F</td>
<td>O</td>
<td>W</td>
</tr>
<tr>
<td>4</td>
<td>100</td>
<td>1000</td>
<td>400</td>
</tr>
<tr>
<td>3</td>
<td>60</td>
<td>400</td>
<td>8</td>
</tr>
<tr>
<td>2</td>
<td>20</td>
<td>300</td>
<td>10</td>
</tr>
<tr>
<td>1</td>
<td>10</td>
<td>200</td>
<td>12</td>
</tr>
</tbody>
</table>

Covered electrode Welding

Example

S: Surfacing Weld
F: Fillet Weld
W: With Preheating
O: Without Preheating
Figure 2.8

Cooling time, $t_{\text{g/3}}$, for gas metal arc welding dependent on gross heat input ($q_w$), plate thickness ($h$) and preheat temperature ($T_0$)

(Source: Ref.56-58)
Figure 2.9

Cooling time, $t_{85}$, for submerged arc welding dependent on gross heat input ($q_w$), plate thickness ($h$) and preheat temperature ($T_0$)

(Source: Ref. 56-58)


2.10.1 THREE COMPONENTS OF RESIDUAL STRESS

In welding, residual stresses are built up during weld pool cooling. Residual stresses built up in any welded construction include three components, namely shrinkage stress, quenching stress and phase transformation stress[59].

2.10.2 RESIDUAL STRESS DUE TO SHRINKAGE

These stresses are produced because the shrinkage in the weld zone and in the adjacent HAZ is hindered by the restraint offered by the adjacent colder regions. As a consequence, tensile residual stresses develop at the weld centre and compressive stresses elsewhere.

The tensile stresses in the weld grow with progressive cooling. Even if the tensile stresses are diminished by relaxation, a stress distribution as shown in Figure 2.10-a, is observed for the residual stresses in the weld direction. The equilibrium conditions ensure that the stress amplitude zero points occur at the ends of the weldment. This distribution is typical for butt-welds in unalloyed steels.

The shrinkage hindered in the lengthwise direction also introduces residual stresses perpendicular to the weld (Figure 2.10-b). Due to shrinkage in the direction perpendicular to the weld, tensile stresses occur at the centre of the weld, which are in equilibrium with the compressive stresses near the end of the weld. For long plates, the middle of the plate will be free of residual stresses in this direction. Generally the maximum amplitude of the stress in the transverse direction is smaller than that in the longitudinal direction.

The amplitude of shrinkage residual stresses grows with the ratio of plate to weld thickness. The tensile residual stress is larger for smaller widths of the weld, for higher thermal coefficient of expansion, and for higher Young's modulus. From the knowledge of shrinkage in welds, a 'golden rule' is that tensile stress occurs in the regions which are the last to cool.
Figure 2.10
Stress distribution, parallel ($\sigma_x$) and perpendicular ($\sigma_y$) to a butt weld
(Source: Ref.59)

(a) W-form,  (b) M-form

Figure 2.11
Two forms of stress distribution, perpendicular ($\sigma_y$) to a weld
(Source: Ref.59)
2.10.3 RESIDUAL STRESS DUE TO QUENCHING

The cooling rate has no influence on the residual shrinkage stresses if homogeneous cooling conditions over the whole plate thickness are assumed. For thick plates this condition is not fulfilled and additional quenching residual stresses are observed. Zones near to the surface of the weld metal and HAZs cool much faster than the interior zones.

If the thermal stresses become higher than the yield strength, after cooling to room temperature, compressive residual stress occurs in the near surface zone of the weld and the HAZ, with tensile residual stress in the inner region of the weld.

Quenching residual stresses increase with both decreasing tensile yield strength and increasing temperature gradients. They also increase with plate thickness and the rate of cooling.

2.10.4 RESIDUAL STRESS DUE TO PHASE TRANSFORMATION

These stresses are caused by the transformation of austenite to ferrite or martensite. This transformation does not occur at the same time in near surface zones and in the inner regions, nor does it occur at the same time in the weld and in HAZ. As a consequence, tensile residual stress is produced in regions where the transformation occurs first and compressive stress in regions where the transformation occurs later.

2.10.5 RESIDUAL STRESS DISTRIBUTION

In practice, welds show a superposition of all the three effects described above and the form of the residual stress distribution is therefore determined by the dominant process. The residual stress distribution in a weld in general has a W-form or an M-form, as in Figure 2.11. The most conventional pattern of residual stresses takes the W-form, with the tensile stresses along the weld centre line. This occurs when the shrinkage residual stress is dominant compared to phase transformation residual stress.
The M-form of residual stress pattern occurs when the phase transformation stresses are dominant compared to shrinkage stresses. This situation arises when the transformation takes place at relatively low temperatures where yield strength of the steel will be quite high. In particular, when austenite (FCC) transforms to martensite (BCT), increase in volume occurs. This sets up compressive stresses large enough to neutralise and overcome the tensile stresses developed due to shrinkage. When the transformation occurs at relatively low temperatures, these compressive stresses will be quite dominant. Any significant stress relief may not occur because of low temperatures. A good example where M-form of residual stress occurs is the welding of maraging steel [60].

2.11 EXPERIMENTAL METHODS FOR MEASUREMENT OF RESIDUAL STRESSES IN WELDMENTS

Many techniques are used for measuring residual stresses in metals. Two important techniques are stress-relaxation technique and X-ray diffraction technique [52,61].

In the stress-relaxation techniques, residual stresses are determined by measuring elastic strain release. This takes place when residual stresses are relaxed by cutting the specimen into pieces or by removing a piece from the specimen. A variety of techniques exist depending upon the methods of sectioning the specimens to determine the residual stresses. Electric or mechanical strain gages are used for measuring the strain release. Strain release can also be determined using a grid system, brittle coatings, or photoelastic coatings. An inherent disadvantage of the stress-relaxation techniques is that they are destructive. However still they provide reliable and quantitative data.

X-ray diffraction (XRD) method is a widely recognised non-destructive method applied to the measurement of residual stress. The x-ray diffraction measurement is applicable to any material that is crystalline, relatively fine grained and produces diffraction for any orientation of the sample surface. However, XRD method is only capable of measuring stresses within about 0.025 mm of the surface due to the rather soft characteristic x-ray
radiation that is used. XRD method provides the quantity and directions of surface stresses.

In XRD, the change in the interplanar spacing brought about by the applied stress is used for the stress measurement. It exploits the fact that when a metal is under stress, applied or residual, the resultant elastic strains cause the atomic planes in the metallic crystal structure to change their spacing. The macroscopic stresses that extend through a large area relative to the grain size of the material are of general interest in design and failure analysis. Such macro-stresses for a given location and direction is determined by measuring the strain in that direction at a single point.

### 2.12 SOLIDIFICATION CRACKING

Solidification cracking is a type of hot cracking that occurs in weld metal when the solidifying weld metal cannot support the thermally or mechanically induced strain during welding [4]. These cracks are interdentritic or intergranular, and they are caused primarily by the segregation of low-melting elements such as sulphur, phosphorous, carbon, nickel, niobium and boron. High sulphur content is one of the most commonly reported causes of solidification cracking in steels. Manganese and silicon are elements that tend to reduce the solidification cracking of steel.

Solidification cracking is often longitudinal along the centreline of the weld bead. Deep, narrow weld beads are more prone to solidification. Large weld pools, created by high heat-input welding procedures are more susceptible to solidification cracking than small pools of low heat-input procedures [62,63].

### 2.13 SOLIDIFICATION CRACKING IN AUSTENITIC STEEL WELD METAL

Solidification cracking in austenitic stainless steel welds results from the formation of certain low-melting liquid films along grain boundaries during the last stages of
solidification [64]. Description of solidification cracking in austenitic stainless steel is given in the following sections, since in the present work, an austenitic stainless steel welding electrode has been used to weld the Q&T steel.

2.13.1 EFFECT OF DELTA FERRITE ON SOLIDIFICATION CRACKING

Weld solidification cracking in austenitic stainless steel weld metal is reduced by introducing controlled amount of delta ferrite in it [3,4,62-67]. Therefore, many of the commonly used austenitic stainless steel filler metals are formulated such that the as-deposited weld metal contains ferrite in the range of ferrite number (FN) 3 to 18. These filler metals are avoided only in cases where the presence of ferrite poses problems with respect to service performance, e.g., mechanical degradation at cryogenic temperatures and problems due to sigma-phase embrittlement at elevated temperatures.

Several theories have been proposed to explain the beneficial effects of delta ferrite in austenitic stainless steel weld metals. They include the following:

♦ Impurity elements, such as sulphur and phosphorous, have a higher solubility in delta ferrite relative to austenite. Thus delta ferrite acts as a preferential sink for impurities that promote cracking.

♦ Weld metals that contain delta ferrite have a smaller solidification temperature range.

♦ The difference in interfacial energies between austenite and delta ferrite in the solidification range prevents the formation of continuous liquid films resulting from segregation of alloy and impurity elements. In fully austenitic weld metal, continuous liquid films are more easily formed.

♦ The two-phase weld microstructure inhibits crack propagation relative to a single phase austenitic microstructure. Thus, ferrite-containing weld metals present a more "tortuous" cracking path than those that are fully austenitic.
2.13.2 EFFECT OF SOLIDIFICATION MODE

In austenitic stainless steels, the primary phase of solidification may be either austenitic or ferrite. This is dictated predominantly by composition. Welds that solidify as primary austenite are inherently more susceptible to weld solidification cracking than welds that solidify as primary ferrite. Thus control of weld solidification behaviour is critical to ensuring crack-free welds [3,4,63].

When solidification occurs as primary austenite (modes A and AF), the as-deposited microstructure can either be fully austenitic (mode A) or contain a small amount of ferrite, generally less than FN3 (mode AF). When solidification occurs as primary ferrite (mode FA), the as-deposited weld metal can exhibit levels of ferrite ranging from FN 3 to FN 45.

When delta ferrite is the primary phase of solidification, considerably higher impurity elements (such as sulphur and phosphorous) can be tolerated and the weld metal is extremely resistant to solidification cracking. When solidification occurs as primary austenite, the cracking susceptibility increases unless impurity levels can be maintained at very low levels.

The relationship between weld solidification cracking susceptibility, composition and solidification behaviour is illustrated in Figure 2.12. In this figure, composition is given in terms of the ratio of chromium equivalence (Cr-eq) and nickel equivalence (Ni-eq) [4]. As described previously, cracking susceptibility is high when the weld metal is fully austenitic and drops rapidly as solidification behaviour shifts to primary ferrite.
Figure 2.12
Cracking susceptibility Versus ferrite number
(Source: Ref. 4)

![Graph showing cracking susceptibility versus ferrite number.](source)

Figure 2.13
Schaeffler constitution diagram for stainless steel weld metal
(Source: Ref.68)

![Schaeffler diagram showing constitutional phases.](source)
It is also significant from this figure that the susceptibility to weld solidification cracking may increase with increasing ferrite contents above approximately FN 30. This is attributed to the transition to completely ferritic solidification (from FA to F mode). It has been proposed that the presence of ferrite-ferrite solidification boundaries (F mode) during the final stages of solidification results in an increase in cracking susceptibility relative to the case when some austenite forms (type FA).

2.13.3 CONSTITUTION DIAGRAMS

Because chemical composition has the greatest influence on weld microstructure, a number of empirical relationships and constitution diagrams have been developed to predict microstructure based on actual or approximated composition. In all the cases, the concept of chromium equivalence (Cr-eq) and nickel equivalence (Ni-eq) has been used to normalise the effect of various alloying additions on the ferrite-forming and austenite-forming potency, respectively.

The Schaeffler diagram shown in Figure 2.13 was developed in the late 1940s in an attempt to predict weld microstructure for a wide range of stainless steels [68]. This diagram was refined as the Delong diagram (Figure 2.14) which pertained to the range of compositions associated with 300 series stainless steels and filler metals [69]. In 1988, the Welding Research Council introduced a further improvement, which was called the WRC-1988 diagram (Figure 2.15) [70]. This diagram was further modified in 1992 by adding copper to the Ni-eq formula (Figure 2.16) [71].

The Schaeffler diagram is especially useful for predicting microstructures between dissimilar alloys using a tie line between the two alloy compositions and the appropriate alloy dilution. The diagram was developed using the shielded metal arc welding (SMAW) process. This diagram described the behaviour of most of the alloys in use at that time, and is relatively accurate for most 300 series alloys when using conventional welding processes.
Figure 2.14
Delong constitution diagram for stainless steel weld metal
(Source: Ref. 69)

The Schaeffler austenite - martensite boundary is included for reference

Figure 2.15
WRC-1988 constitution diagram for stainless steel weld metal,
including solidification - mode boundaries
(Source: Ref. 70)

Studies on weldability of a quenched and tempered high strength low alloy steel
Figure 2.16
WRC-1992 constitution diagram for stainless steel weld metal, including solidification-mode boundaries. (Source: Ref. 71)
In the WRC-1992 diagram, the alloying elements are limited to nickel, carbon, nitrogen and copper in the Ni-eq and to chromium, molybdenum and niobium in the Cr-eq. However this diagram is applicable for manganese upto 10%, molybdenum upto 3%, nitrogen upto 0.2% and silicon upto 1%. One important feature of this diagram is the inclusion of boundaries between different solidification modes. The most important boundary is the one separating primary ferrite from primary austenite solidification.

2.13.4 SIGMA PHASE EMBRITTLEMENT

The delta ferrite introduced in austenitic stainless steel weld metal to avoid solidification cracking may sometimes lead to another metallurgical difficulty, namely formation of the brittle sigma phase, under certain conditions [3,4,64]. Formation of sigma phase is always undesirable, since it decreases the ductility and toughness.

The sigma phase is a hard, brittle phase which forms in stainless steels when they are held in the temperature range 600 - 800°C. Though sigma can form over a wide range of compositions in stainless steels, the tendency increases with increasing amounts of ferrite-forming elements like chromium, molybdenum and silicon and decreases with the austenite-forming elements like nitrogen, nickel and carbon. As a consequence of alloy partitioning during both solidification and the solid state transformation of ferrite to austenite, ferrite is enriched in those alloying elements which promote the sigma phase and depleted in those elements that retard sigma formation. Thus, ferrite in austenitic stainless steel welds is especially susceptible to sigma-phase formation. The sigma-phase has been observed to nucleate preferentially at the ferrite-austenite interface in weld metal. It is well known that the notorious effect of sigma phase is a decrease of ductility and toughness.

2.14 WELDABILITY TESTS

Weldability tests are used to evaluate the effects of welding on such properties as base-metal cracking and weld-metal cracking, weld penetration, weld pool shape and
fluid flow [3]. Weldability testing can provide valuable data on new alloys, welding procedures, and welding processes. These tests are generally qualitative and cannot quantitatively predict fabrication experience [52,62]. The majority of weldability tests have been developed to evaluate the susceptibility of the base metal and weld metal to cracking.

Weldability tests evaluate various responses of the base metal to different welding conditions or simulated welding conditions. Most tests are used to evaluate susceptibility to one of the following types of cracking: Hydrogen induced cracking (HIC), hot cracking, lamellar tearing, and reheat cracking. Some tests are useful for both hot cracks and HIC. The weldability tests discussed in this section are classified as self-restraint tests or externally loaded tests [6].

### 2.15 SELF-RESTRAINT WELDABILITY TESTS

Self-restraint tests utilise the restraint, or stress, within the specimen to cause weld metal or base metal cracking. No external loading of the specimen occurs. The specimen is designed to produce variable restraint on the weld joint, thus causing cracking. Some of the self-restraint tests are described below:

#### 2.15.1 LEHIGH RESTRAINT TESTS

Lehigh restraint test (for HIC and hot cracking) is illustrated in Figure 2.17 [72]. This test uses a steel plate with slots machined into the sides and ends. A groove for weld joint is machined along the centreline of the plate, and a single-pass weld is produced along this groove. The restraint from the plate and slots produces a weld with various levels of cracking. The level of restraint is altered by changing the length of the slots. The restraint is quantified as $2x$, where $x$ is the distance from the centreline of the weld groove to the bottom of the slots. Numerous samples with different slot lengths are welded, and a threshold level of restraint is determined as the width that is just sufficient to cause cracking. Cracking can be evaluated by metallographic examination of cross sections of the weld. This test has also been used to evaluate reheat cracking in steels.
**Figure 2.17**
Lehigh Restraint Test - Schematic showing slotted plate specifications  
(Source: Ref. 72)

![Lehigh Restraint Test Schematic](image)

**Figure 2.18**
NRL Keyhole restraint test - Plate specifications  
(Source: Ref. 72)

![NRL Keyhole restraint test Schematic](image)
2.15.2 NRL KEYHOLE RESTRAINT CRACKING TEST

The Naval Research Laboratory (NRL) keyhole restraint cracking test (for HIC and hot cracking) is a simplified version of the Lehigh test (Figure 2.18) [72]. The specimen is welded along the groove beginning at the open end and progressing towards a hole deep inside the plate. This imposes a varying degree of restraint along the weld, with a maximum at the hole and a minimum at the edge where welding commenced. Cracks form at the hole and extend outward to a point where the restraint is low enough to arrest the crack growth. The crack length is the measure of crack susceptibility.

2.15.3 TEKKEN TEST

The Tekken test (for HIC), developed in Japan, uses several specimen dimensions and joint designs (Figure 2.19) [73]. The Y-joint provides more restraint than the U or double-U joint designs. Most of the Japanese studies relating to safe or crack-free welding procedures and their dependence on plate composition, hydrogen and restraint were conducted by using the Tekken test [74-76]. In these tests, preheat and welding parameters are varied to alter the stress state, rather than changing the specimen design.

2.15.4 CIRCULAR PATCH TEST

The circular patch test (for HIC and hot cracking), utilises a plate with circular hole cut out and replaced by a patch (Figure 2.20) [72]. Both the plate and the patch have a weld joint machined into them. One or more weld passes are made in this circular groove, and cracking is detected by visual, radiographic or liquid penetrant inspection.

2.15.5 WIC RESTRAINT TEST

This test, developed (for HIC) at the Welding Institute of Canada (WIC), has the test pieces fillet welded along their sides to a steel plate (Figure 2.21) [77]. The base contains a stiffener to minimise rotational distortion. It is possible to vary the amount of restraint by changing the length or position of the fillet welds.
Figure 2.19
Tekken test - Sample specifications
(Source: Ref. 73)

Modified Lehigh restraint specimen

Figure 2.20
Circular patch test - plate specifications
(Source: Ref. 72)
Figure 2.21
WIC Restraint cracking test
(Source: Ref. 77)

Figure 2.22
Controlled Thermal Severity (CTS) test - Plate dimensions
(Source: Ref. 72)
2.15.6 CONTROLLED THERMAL SEVERITY (CTS) TEST

The CTS test specimen (for HIC) consists of a square plate bolted and anchor welded to a larger rectangular plate (Figure 2.22) [72]. After the anchor welds have cooled to room temperature, two test welds are made on the specimen. The fillet weld along the plate edges is made first. This weld, called bi-thermal weld, has only two-dimensional heat flow. After this weld has cooled, the lap weld is made on the other edge of the top plate, near the centre of the bottom plate. Because this weld has nearly three dimensional heat flow, it is called a tri-thermal weld. The specimen is held for 72 hours at room temperature, then the degree of cracking is determined by measuring crack length on three metallographic specimens cut from cross sections of the two welds.

Each specimen is assigned a thermal severity number (TSN) according to plate thickness using the following equation:

For bi-thermal welds, \( TSN = 4(t+b) \)

For tri-thermal welds, \( TSN = 4(t+2b) \)

Where \( t \) and \( b \) are thicknesses of the top and bottom plates respectively.

A series of plate thicknesses, which provide varying cooling rate, are tested. The crack susceptibility of the base metal and filler material combination is determined by the minimum TSN that produces cracking.

2.15.7 CRUCIFORM CRACKING TEST

The cruciform cracking test (Figure 2.23) (for HIC) uses three plates that are surface ground along their interfaces to ensure consistent and reproducible fit-up [72]. The ends are tack welded together to form a double T-joint. Four test fillet welds are deposited in the sequence shown in fig. 2.23, with complete cooling between passes. The specimen is held at room temperature for at least two days before stress relieving at about 620°C. Cracking is determined by evaluation of metallographic cross sections of the four welds.
Figure 2.23
Cruciform cracking test - specimen specification
(Source: Ref. 72)

Figure 2.24
Varestraint hot cracking test
(Source: Ref. 78)
The tests can be erratic and are very susceptible to sample preparation. Five to ten samples are typically required to provide reliable results for cracking sensitivity level.

2.16 EXTERNALLY LOADED WELDABILITY TESTS

Externally loaded tests involve the application of an external load to the specimen during weld testing. This external load is varied to alter the stress state and thus the severity of cracking. This allows a more quantitative measure of the stress, than can be obtained from the self-restraint tests.

2.16.1 IMPLANT TEST

The implant test is used to evaluate HIC. This is considered a very important test for making a quantitative measurement of the HAZ cracking due to hydrogen. Since implant testing of the Q & T steel forms a major part of the present research work, this test is described in detail in section 2.17.

2.16.2 VARESTRAINT TEST

The varestraint (variable restraint) test (for hot cracking) was developed at the Rensselaer Polytechnic Institute [78]. It is the most common test used to evaluate hot cracking sensitivity. In this test, one end of a rectangular bar (typically 50 X 205 X 6.3 mm) is firmly mounted in a fixed position while the opposite end is attached to a hydraulic or pneumatic plunger (Figure 2.24). The weld is produced on the top side of the plate, along its longitudinal centreline, beginning at the unsupported end and moving toward the fixed end. When the Welding arc reaches a predetermined location over a die block, the plate is bent to confirm to the radius of the die block. This induces an augmented longitudinal strain on the welded surface of the specimen.

The approximate value of the strain, \( \varepsilon \), is given as \( \varepsilon = t / (2R) \), where 't' is the specimen thickness and R is the die block radius. Various die blocks with different radii can be used on samples of the same material to characterise the strain level at which cracking occurs.
2.16.3 TRANS-VARESTRAINT TEST

A variation of the varestraint test is called trans-varestraint test, in which the sample is bent in a direction normal to the welding direction instead of the longitudinal direction. The loading in this test is tensile and normal to the welding direction. This usually causes centreline cracks to form. The length of centreline cracks is used as a measure of hot crack susceptibility.

2.16.4 MUREX TEST

In this test (Figure 2.25) transverse plastic strain is imposed on a fillet weld between two plates by rotation of one of the plates during welding [79]. The length of crack is an indication of hot cracking susceptibility. It is a very severe test and is usually used for relative ranking of electrodes.

2.17 IMPLANT CRACKING TEST

2.17.1 GENERAL PRINCIPLE

The implant test for testing of high strength steels for susceptibility to HIC was first proposed by Henri Granjon [80]. By "implant" is meant a cylindrical sample prepared from the steel to be investigated for its weldability. This sample, with a groove on its one end, is inserted into a hole of the same diameter drilled into a base plate also made of the same steel or a thermally equivalent steel. A weld bead is deposited on the base plate and over the end of the test specimen. The weld bead penetration shall be such that the notch is located in the coarse grained area in the HAZ. After welding and before complete cooling, the implant shall be submitted to tensile static loading, which may cause HIC to occur in the HAZ. The base plate may be the same grade as the steel of the implant or it may be different. The essential thing is to achieve identical thermal conductivity between implant and the parent plate.
Figure 2.25
Murex hot cracking test
(Source: Ref. 79)
The basic principle which underlies the implant method lies in the fact that at the same distance "d" from the fusion line, the heat cycles at a point B of the implant and at a point A of the base plate are the same (Figure 2.26). Henri Granjon has shown that the rate of cooling in the implant is governed by the nature of the base plate [80]. Thus an implant of austenitic steel, inserted in a low carbon steel base plate, cools just as if it were low carbon steel and an implant of low carbon steel inserted in stainless steel base plate cools as if it were stainless steel.

As an implant (inserted into a base plate), undergoes the same thermal cycle as the base plate itself, all that is needed in order to study HIC is to exert a tensile force on the implant immediately after welding. In this way all the three features causing cracking, viz. transformation, hydrogen and stresses are studied. The nature of the transformation is governed by the welding method used. The hydrogen content is either a function of the type of electrode (when welding with covered electrodes) or of the welding process used. The stress exerted on the implant may be any value lower than the yield point of the parent metal. By varying the test conditions it is possible to determine the stress values for cracking. Variations introduced by the various laboratories in implant testing have concentrated on the shape of the implant specimen (particularly the shape of the groove in it), geometry of the base plate, variation of welding parameters, loading methods and the method of detection of cracks [25,39,40,81-101]. To a large extent, a document released by International Institute of Welding - IIS/IW-IX-802-84 incorporates all the implant variables and hence serves as a comprehensive standard [98].

2.17.2 ADVANTAGE OF IMPLANT TEST

Among the various testing methods for assessing HIC, the implant test has become one of the most popular for scientific investigations of the cracking phenomenon in welds. This is due to the fact that the stress, hydrogen level and microstructure can be independently varied and controlled in this test [7].
Figure 2.26
Equivalence of heat cycles in plate and implant
(Source: Ref.80)

Figure 2.27
Implant specimen and base plate (20mm thick) for implant testing
(Source: Ref.98)
2.17.3 SAWHILL'S MODIFICATION OF GRANJON'S TEST

With Granjon's test the uncertainty of the exact location by the circular notch during specimen preparation is a serious disadvantage. Ideally the notch should be located in the most susceptible region of the HAZ to maximise the sensitivity of the test procedure. This is particularly difficult since it is not always possible to predict the depth of weld penetration at all points along a weld due to minor fluctuations that are always present within the welding arc. For this reason, a modification of the Granjon test was proposed by Sawhill [87]. He replaced the single circumferential notch in the test specimen with a continuous helical notch. Such a notch then provides a stress raiser throughout the entire HAZ from the fusion line well into the base material, thus eliminating the uncertainty of placement of stress riser placement associated with the Granjon method.

2.17.4 BASE PLATE AND IMPLANT SPECIMEN DIMENSIONS

IIW Document No. IIS/IIW-IX-802-84 has standardised the dimensions for the base plate and the implant specimen [98]. They are as given in Figure 2.27.

The implant specimen is usually made of an 8 mm diameter cylinder and a part to be held by the fastening device at one end. The cylinder length shall be greater than the thickness of the base plate and a fillet shall be provided between the gauge part and the fastening part. At the opposite end, the implant specimen has a helical notch with shape and size described below:

\[
\begin{align*}
\text{Shape: V-notch;} & \quad \text{Depth: 0.5 } \pm 0.05 \text{ mm;} & \quad \text{Notch angle: 40 } \pm 2^\circ; \\
\text{Pitch: 1 or 2 mm;} & \quad \text{Root radius: 0.1 } \pm 0.02 \text{ mm.}
\end{align*}
\]

Higher diameter has the advantage of easy formation of cracks. Lower diameter has the advantage of easy fracture at lower stress levels. Though 8 mm diameter has been widely recommended in earlier works, 6 mm diameter has also been used, as done by Christensen [102]. Apart from the requirement of lower tensile loads, 6 mm diameter has the advantage of enabling the experiments to be conducted on small weld beads which do not cover 8 mm implants. In selecting the implant diameter, it must be remembered...
among other things that an adequate covering of the specimen by the weld bead is essential. The implant clearance in the plate hole shall range from 0.05 to 0.15 mm (sliding fit).

Regarding the base plate preparation, almost all experiments have adopted a plate that is flat without any special preparation. However, it has been reported that several Japanese laboratories have introduced a variation which consists of a groove in the base plate with a profile as shown in Figure 2.28 [97,103]. It has been pointed out that if the test is used as a complimentary information test, the original arrangement of a flat plate is simpler, and better. Where research work is undertaken, the Japanese variation is preferred.

2.17.5 CHARACTERISATION OF THE THERMAL CYCLE

The thermal cycle during weld deposition in implant testing is characterised by the cooling time between 800 and 500°C (tₖ/₅). Temperature measurements are conveniently made by means of thermocouples placed either in the weld metal or in the HAZ.

2.17.6 IMPLANT LOADING

Regarding the application of tensile static loading, the IIW document No. IIS/IIW-IX-802-84 recommends that it should be done as follows [98]:

(a) If there is no post heating:
The loading should be done at temperature Tₐ, after the end of welding. The loading temperature Tₐ is chosen depending on the preheating temperature Tᵥ, as shown below:

\[
\begin{align*}
Tₐ &= 150°C, & \text{if } Tᵥ \leq 100°C \\
Tₐ &= Tᵥ + 50°C, & \text{if } 100°C < Tᵥ < 200°C \\
Tₐ &= 250°C, & \text{if } Tᵥ \geq 200°C
\end{align*}
\]

The specified load should be reached within a time frame of 20 to 60s and before the temperature reaches 100°C.
Figure 2.28
Base plate for implant test with groove
(Source: Ref. 103)

Figure 2.29
Typical plot of rupture stress Vs failure time in Implant test
(Source: Ref. 86)
(b) **If there is post heating:**

The loading should be done as soon as the temperature of post heating is reached.

The specimen is released after holding the specimen for a certain time, after which HIC may not occur. In most of the earlier work, this holding time has been taken as 16 hours or 1000 min. In some of the work it has been taken as 24 hours [87,92].

R. Vasudevan et al. have suggested that the tensile static loading should be applied two minutes after completion of the weld deposit as it is the time taken for the weld to cool down to 125°C [40]. In some other studies, two minutes after welding, the bead was quenched for four minutes [82]. The implant was then set up in a modified creep testing machine and appropriate load was applied to the specimen, two minutes after quenching.

In several studies, loading was done when the temperature of the weld bead surface reached 150°C [86]. This is considered acceptable on the basis that it was below the $M_f$ of the material being tested, and above the temperature at which cracking generally begins in C-Mn steels and low alloy steels.

### 2.17.7 INSPECTION OF SPECIMENS AFTER LOADING FOR CRACK SUSCEPTIBILITY

If the implant specimen fractures due to the static loading, the load and the time of fracture are recorded. If no fracture occurs up to 16 hours, the cracks that might have formed at the notch under the applied load can be detected using appropriate methods. One such method is metallographic examination at magnification of 400 to 600X on several longitudinal sections located in the weld bead direction. An acoustic emission technique can also be used to detect the crack growth during static tensile loading [46, 87].

If fracture occurs, then for a given set of welding conditions, plots of rupture stress as a function of time are constructed similar to the one shown in Figure 2.29 [83]. The lower critical stress (LCS), below which rupture does not occur up to a specified time (16 hrs or
1000 min. in most of the cases) is taken as an index of the ability of the material to resist HIC [87]. Lower critical stress is also referred to as static fatigue limit [95]. In addition, an "Embrittlement Index", EI, is often calculated from the following relationship [87]:

$$EI = \frac{NTS - LCS}{NTS}$$

where $NTS = \text{Short-time notched tensile strength}$; $LCS = \text{Lower critical stress}$. It may be noted that as the susceptibility to HIC increases, the Embrittlement Index also increases. Evans et al., have related the LCS with the hydrogen concentration as per the relationship [95], $LCS = A - B \log[H]$, where the coefficients A and B depend on microstructure of the HAZ (which in turn may be related to the steel composition and cooling rate).

### 2.18 FRACTURE MORPHOLOGIES OF FRACTURED SPECIMENS

The fracture features of tested samples provide considerable information as to how the fracture occurs. Conclusions regarding stress state, strain rate, environment and microstructure may be deduced from fracture appearance. A major requirement in materials usage is avoidance of brittle fracture. If a material (or weldment) should fail, it is preferable that the failure occurs in a ductile manner. A brief review of literature about different fracture morphologies is given in the following section [104]. This will, in particular, help in identifying the fracture characteristics in implant testing of specimens in the present work.

Differentiating brittle and ductile microscopic fracture modes is often troublesome. From a practical standpoint, a metal is considered brittle if it breaks with little or no plastic deformation, during slow to moderate strain rate loading, or with little energy absorbed in impact loading. Macroscopically, dull or fibrous fracture surface appearance is usually associated with ductile failures, and shiny or crystalline fracture surface appearance is usually associated with brittle failures.
The range of fracture morphologies encountered on fracture surface of welded specimens range from ductile microvoid coalescence (MVC) fracture observed on tension, bending, impact and fatigue specimens, to brittle cleavage fracture observed on impact test specimens. Semi-brittle fractures are also observed. Mixed-mode fractures consisting of mixtures of distinct cleavage and microvoid coalescence are also observed in specimens fractured by both bending and impact. A quantitative fracture map showing regions of different fracture types is shown in Fig. 2.3.0 [1]

2.18.1 DUCTILE FRACTURES : MICROVOID COALESCEENCE (MVC)

MVC fractures are usually associated with the presence of inclusions or precipitates which initiate voids. Creation of microvoids occurs in places where the compatibility between macroscopic and localised deformation is no longer possible.

2.18.2 BRITTLE FRACTURES : CLEAVAGE

Cleavage is generally considered to be brittle fracture in steels. Brittle rupture by cleavage occurs when the stress rises beyond the local cohesion of a metal, resulting in rupture of interatomic bonds. Cleavage fractures occur generally after a very limited macroscopic deformation has occurred.

Cleavage fracture propagates along the [100] planes in BCC metals and sometimes along interfaces separating the matrix and mechanical twins. River patterns and "tongues" are characteristic features seen on cleavage fractures. River patterns originate because the crack front does not restrict itself to a unique plane. Instead, it deviates from one plane to a neighbouring one or separates into several components because it encounters structural discontinuities. The end result is a system of parallel cracks propagating simultaneously. Rivers tend to merge because the formation of steps takes energy which slows down the propagating crack front. Cleavage 'tongues' are small local decohesions along planes other than the primary cleavage plane.
2.18.3 "SEMI-BRITTLE" FRACTURES
(MIXED MODE AND QUASI-CLEAVAGE)

Mixed-mode fractures occur when the fracture propagates discontinuously by both ductile, and brittle forms. "Quasi-cleavage" is a fracture mode that resembles cleavage with its large, flat facets but occurs on unspecified planes and contains tear edges. The fracture initiates within the facets and often exhibits short, angular river pattern markings. Quasi-cleavage is not a specific fracture mode but rather a unique fracture appearance which has some aspect of cleavage as well as plastic deformation. The occurrence of quasi-cleavage fractures in the weld metal is intimately associated with the presence of weld discontinuities.

The transition region, between ductile and fracture morphologies, has areas of intergranular faceting (hydrogen embrittlement) below or next to areas of microvoid coalescence (ductile fracture). The resulting quantitative type of map developed for this type of specimen will be as depicted in Figure 2.30 [7].

2.18.4 FISH-EYE INCLUSIONS

A number of low hydrogen samples may have areas of "fish-eyes". A fish-eye is an inclusion that is locally surrounded by an area of hydrogen embrittlement. The local area of hydrogen embrittlement surrounding a fish-eye is presumed to be due to hydrogen trapping at the inclusion. Numerous investigations have found that hydrogen can be trapped at inclusions [9].
Figure 2.30
Quantitative fracture map showing regions of fracture types
(Source: Ref. 7)