A METHODOLOGY FOR QUANTIFYING HEAT-AFFECTED ZONE LIQUATION CRACKING SUSCEPTIBILITY

DISsertation

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By

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* * * * *

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To My Parents
ACKNOWLEDGMENTS

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ABSTRACT

The hot-ductility, spot- and longitudinal-Varestraint (mini-Varestraint) tests have long been recognized as useful weldability tests for both quantifying heat-affected zone (HAZ) liquation cracking susceptibility and developing a fundamental understanding of liquation cracking phenomena. However, test results from the spot-Varestraint and hot-ductility tests, interpreted using "conventional" techniques, were inconsistent in predicting the HAZ liquation cracking susceptibility of Incolloys 903 and 909. This discrepancy was attributed to the lack of a physical relationship between weldability test results and the metallurgical response of a material during testing. In the present study, a test methodology characterizing a thermal crack susceptible region (CSR) has been developed based on the properties of a material during welding as obtained from the hot-ductility test and the criteria assumed in the development of liquation cracking theories. This CSR was experimentally verified using the longitudinal- and spot-Varestraint tests performed on A-286 and Type 310 stainless steels. The thermal CSR is material-specific and represents a true quantification of weldability. Metallographic and fractographic examination of test samples revealed that CSR defined a region in the HAZ in which the bulk material exhibits partial melting during welding. The boundary of the CSR is the location at which grain boundaries exhibit partial melting or liquation. The development of this model has (1) provided a quantitative methodology for predicting HAZ liquation cracking susceptibility, (2) elucidated the physical relationship among weldability test results, liquation cracking theories and material properties, (3) clarified the interpretation of test data from the hot-ductility, the spot-and longitudinal-Varestraint tests and (4) confirmed the criteria assumed in the development of liquation cracking theories which state that cracking results from the localized loss of grain boundary ductility due to liquation. This dissertation addresses the theoretical background and experimental procedure in the development of the new methodology. The methods of data interpretation of the three tests and the metallurgical behavior of a material in the CSR are discussed. The utility of conventional methods of data interpretation of the three tests are also critically reviewed.
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<tr>
<td>BTR</td>
<td>Brittle Temperature Range</td>
</tr>
<tr>
<td>CHL</td>
<td>Cracked Heat-Affected Zone Length in the longitudinal-Varestraint test</td>
</tr>
<tr>
<td>CHL_{OC}</td>
<td>On-cooling portion of the cracked HAZ Length</td>
</tr>
<tr>
<td>CHL_{OC,NST}</td>
<td>On-cooling portion of the cracked HAZ Length at a point in the HAZ which experiences a peak temperature equal to the NST</td>
</tr>
<tr>
<td>CHL_{OC,TL}</td>
<td>On-cooling portion of the cracked HAZ length at a point in the HAZ which experiences a peak temperature equal to the $T_L$</td>
</tr>
<tr>
<td>CHL_{OH}</td>
<td>On-heating portion of the cracked HAZ Length</td>
</tr>
<tr>
<td>CR_{T}</td>
<td>Cooling Rate during weldability testing</td>
</tr>
<tr>
<td>CR_{W}</td>
<td>Cooling Rate for actual welding conditions</td>
</tr>
<tr>
<td>CSR</td>
<td>Crack Susceptible Region</td>
</tr>
<tr>
<td>DRT</td>
<td>Ductility Recovery Temperature</td>
</tr>
<tr>
<td>DRT_{NST}</td>
<td>Ductility Recovery Temperature corresponding to a peak temperature equal to the NST</td>
</tr>
<tr>
<td>DRT_{TL}</td>
<td>Ductility Recovery Temperature corresponding to a peak temperature equal to the $T_L$</td>
</tr>
<tr>
<td>G_{T}</td>
<td>Temperature Gradient during weldability testing</td>
</tr>
<tr>
<td>G_{W}</td>
<td>Temperature Gradient for actual welding conditions</td>
</tr>
<tr>
<td>HAZ</td>
<td>Heat-Affected Zone</td>
</tr>
<tr>
<td>MCL</td>
<td>Maximum Crack Length</td>
</tr>
<tr>
<td>NDR</td>
<td>Nil-Ductility Region</td>
</tr>
<tr>
<td>NDT</td>
<td>Nil-Ductility Temperature</td>
</tr>
<tr>
<td>NST</td>
<td>Nil-Strength Temperature</td>
</tr>
<tr>
<td>PMZ</td>
<td>Partially-Melted Zone</td>
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<td>$t_c$</td>
<td>Cooling Time over which cracking occurs in the spot-Varestraint test</td>
</tr>
<tr>
<td>$t_{c,NST}$</td>
<td>Cooling Time over which cracking occurs at a point in the HAZ which experiences a peak temperature equal to the NST</td>
</tr>
<tr>
<td>$t_{c,TL}$</td>
<td>Cooling Time over which cracking occurs at a point in the HAZ which experiences a peak temperature equal to the $T_L$</td>
</tr>
<tr>
<td>TCL</td>
<td>Total Crack Length</td>
</tr>
<tr>
<td>TCN</td>
<td>Total Crack Number</td>
</tr>
<tr>
<td>$T_{FL}$</td>
<td>Temperature at the fusion line after arc extinction</td>
</tr>
<tr>
<td>$T_L$</td>
<td>Liquidus Temperature</td>
</tr>
<tr>
<td>$T_p$</td>
<td>Peak Temperature of a thermal cycle</td>
</tr>
<tr>
<td>$T_{tip}$</td>
<td>Temperature at the crack tip during weldability testing</td>
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<td>$V_T$</td>
<td>Travel Speed during longitudinal-Varestraint testing</td>
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<td>$V_W$</td>
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INTRODUCTION

HAZ liqation cracking is a type of high temperature weld cracking which occurs in the HAZ adjacent to the fusion boundary. This type of cracking is often encountered during the welding of a variety of engineering alloys. It is particularly prevalent in nickel- and aluminum-based alloys, and in fully austenitic stainless steels. The metallurgical basis for cracking involves the presence and persistence of liquid films at grain boundaries and the inability of these films to accommodate the thermally and/or mechanically induced strain experienced during weld cooling. Although the precise mechanisms for liqation cracking are not fully understood, it is recognized that the simultaneous presence of a susceptible microstructure and a critical level of restraint is necessary to promote cracking. Since the control of weld restraint is often difficult, reduction in liqation cracking susceptibility is normally achieved by adjusting the weld and base metal composition, and microstructure.

The effects of composition and microstructure on liqation cracking susceptibility have been studied extensively using weldability testing techniques since the 1950’s. To date, over 150 separate and distinct techniques have been developed to quantify weld solidification and HAZ liqation cracking susceptibility (Ref. 1). These tests vary widely in their approach and utility, but can generally be classified as representative (self-restraint) or simulative (external-restraint) test techniques (Refs. 2, 3).
Representative test techniques, such as the circular patch test, seek to reproduce the actual welding condition as closely as possible in an effort to accurately "represent" the situation of interest. These tests depend on self-restraint induced by the specimen design and/or fixturing. In most cases, these test techniques only provide a simple "go or no-go" solution for a specific material/process/restraint combination. They are ineffective in quantifying weldability among different materials because of the difficulties in isolating the material factor from the test results.

Simulative test techniques, such as the Varestraint test, attempt to simulate some aspects of the thermo-mechanical response of the material to the welding process. These tests normally involve the application of an external augmented strain or stress whose magnitude can be easily quantified. The thermally-induced restraint of the specimen or/and fixturing is usually negligible compared with the relatively large amount of externally applied strain or stress. This approach allows the metallurgical and compositional factors associated with cracking to be isolated from the mechanical factors and permits their effects to be studied and quantified. As a result, simulative tests have been successful in providing a comprehensive order ranking of families of alloys or heats of a given alloy. However, the test conditions, especially the combination of thermal and mechanical history, are inherently different from actual welding situations. This difference may result in a disparate cracking behavior. The approach used to discern the effects of test conditions on the cracking behavior relies on developing a physical link between weldability test results and the metallurgical response during testing. The lack of this link in essentially all the weldability test techniques, however, has provoked difficulties in standardizing the test procedures and
resulted in poor reproducibility and correlation. Discrepancies between test results and field experience are not uncommon.

Despite these drawbacks, weldability testing is generally perceived as an efficient and economic method for predicting susceptibility to liquation-related cracking during welding. Among the weldability testing techniques, the longitudinal-Varestraint (mini-Varestraint), spot-Varestraint and hot-ductility tests are three of a few methods which can be utilized in quantifying the HAZ liquation cracking susceptibility or developing a fundamental understanding of the cracking phenomena. Despite their widespread use, there has been little effort to correlate the data generated by these tests or to rigorously apply this data to "real-world" situations.

In the present study, a preliminary investigation aimed at comparing the HAZ liquation cracking susceptibility of Incoloy 903 and 909 was performed using both the hot-ductility and spot-Varestraint tests. Results showed that these two tests were inconsistent in predicting the HAZ liquation cracking susceptibility, when "conventional" techniques were utilized to interpret test results. These results stimulated an evaluation program to systematically and fundamentally study the hot-ductility, spot- and longitudinal-Varestraint tests.

The evaluation program was devised to define and quantify a thermal "crack susceptible region (CSR)" in the HAZ, based on the ductility of the material during welding as obtained from hot-ductility test and the criteria assumed in the development of liquation cracking theories. The thermal CSR was experimentally demonstrated to be inherently material-specific by the spot- and longitudinal-Varestraint tests performed on two alloys with well recognized susceptibility to HAZ liquation cracking, namely A-286 and Type 310 stainless steels. Subsequent metallurgical characterization on the
samples from the three tests showed that the CSR is equivalent to the steady-state partially melted zone that surrounds the weld pool during welding. This approach has (1) elucidated the physical relationship among weldability test results, cracking theories and material properties, (2) provided a more precise interpretation of hot-ductility, spot- and longitudinal-Varestraint test results, (3) added important insight as to the cracking mechanisms, (4) defined a method for determining a material-specific measure of HAZ liquation cracking susceptibility, and (5) developed an approach for extending this technique for predicting HAZ liquation cracking in actual weldments.

In this dissertation, the liquation cracking mechanisms are discussed in chapter 1 with an emphasis in the concepts for designing a weldability test to quantify liquation cracking susceptibility. The test procedures and data interpretation of the spot-, longitudinal-Varestraint and hot-ductility tests are also reviewed. The conventional hot-ductility and spot-Varestraint test results of Incoloy's 903 and 909 are presented in chapter 2. Chapter 3 outlines the objectives of this study. The experimental design and procedures for the development of the new methodology are described in chapters 4 and 5. The correlation of test results and methodologies of data interpretation of the three aforementioned tests are discussed in chapter 5. The application of the new methodology in predicting HAZ liquation cracking susceptibility in actual welding conditions and the conventional methodologies of data interpretation of the three tests are also discussed in chapter 5. The results of metallographic and fractographic analyses on the hot-ductility, longitudinal- and spot-Varestraint test specimens of the four alloys investigated in this study: Incoloy 903, Incoloy 909, A-286 and Type 310 are presented in chapter 6. A summary of this dissertation and some future work related to this study are provided in chapter 7 and 8, respectively.
CHAPTER I
LITERATURE REVIEW

HAZ liquation cracking, by definition, is a type of high temperature weld cracking, which occurs in the HAZ adjacent to the fusion line and is associated with the formation of liquid films at grain boundaries. (Ref. 4). This form of cracking has been commonly observed in the welding of most engineering alloys including nickel-base alloys (Refs. 5-21), stainless steels (Refs. 22-45), ferritic steels (Refs. 46-51), aluminum alloys (Refs. 52-55) and iron-base superalloys (Refs. 56-63). Metallurgically, this type of cracking is associated with the occurrence of liquation, and as a result, most effort in studying HAZ liquation cracking mechanisms has been directed to characterize the evolution of the liquid. Currently, there is not an accepted mechanistic approach for describing the solidification of the liquid films and resultant cracking in the HAZ. Neither do any theories exist that quantify the potential for HAZ liquation cracking based only on the amount and/or characteristics of the liquid evolved.

The mere presence of liquid films at grain boundaries is not sufficient to induce a crack. In order for cracking to occur, it is essential that a crack susceptible microstructure be subjected to a sufficient tensile strain (or stress). During welding, the tensile strain or stress does not generally develop until the weld begins to cool. As a result, liquation cracking occurs during the solidification of the liquid films (Ref. 64).

The behavior of the liquid films in the HAZ during weld cooling is analogous to the final
stages of weld solidification, although the origin of the liquid films and the microstructural boundaries in which the liquid resides may be different (Ref. 65). Consequently, to a first approximation, the criteria that govern weld solidification cracking can be adopted to explain the solidification of liquid films and resultant cracking in the HAZ.

In this chapter, the theories regarding the evolution of liquid in the HAZ are reviewed, and the mechanisms describing the solidification of the liquid films and resultant cracking are addressed. Based on these theories, the concepts for designing a weldability test or test technique to quantify liquation cracking susceptibility are discussed. The test procedures and data interpretation of the longitudinal-Varestraint (mini-Varestraint), spot-Varestraint and hot-ductility tests are described.

1.1 The Evolution of Liquid in the HAZ

The evolution of liquid in the HAZ is indigenous to the fusion welding processes. According to Savage et al (Refs. 66-67), a weldment can generally be divided into four distinct regions, as shown in Figure 1, namely, the composite zone, the unmixed zone, the partially melted zone and the true heat-affected zone. The composite zone combined with the unmixed zone forms the fusion zone. By definition, the fusion zone refers to the region where the peak temperatures of the weld thermal cycles are above the liquidus of the parent material. Adjacent to the fusion zone is a region where the degree of melting ranges from 0 to 100%. This region is defined as a partially melted zone (PMZ). The PMZ in association with the true heat-affected zone forms the entire heat-affected zone (HAZ). Based on this definition, the evolution of liquid in the HAZ is confined to the PMZ where the peak temperatures experienced
fall between the liquidus \( T_L \) and effective solidus of the parent material. Because segregation is invariably present in a commercial alloy, the effective solidus is always below the equilibrium solidus for the nominal composition of the alloy. In addition, metallurgical reactions, which occur during dynamic weld thermal cycling, may cause further depression of the effective solidus.

There are several models which have been proposed to describe the evolution of liquid at the grain boundaries and resultant formation of the PMZ. In this section, the partial melting hypothesis, the penetration mechanism and the segregation mechanism are reviewed. The infiltration of grain boundaries by liquid weld metal resulting in liquid metal embrittlement is also discussed.

![Diagram of a heterogeneous weld](Ref. 67)
1.1.1 Partial Melting Hypothesis

The partial melting hypothesis was proposed by Savage et al (Ref. 68) based on their studies of single phase alloys. According to this hypothesis, the effective melting temperature of an alloy varies from point to point because of the localized variations in solute content, as shown schematically in the upper part of Figure 2. Superimposed on this plot are two hypothetical temperature gradients. Note that the locations near the fusion zone where the instantaneous temperature lies above the effective melting temperature should experience localized melting. The location and size of these molten regions, assuming no movement of the grain boundaries had yet occurred, are shown schematically (for the shallow gradient) by the cross-hatched areas in the central portion of Figure 2. At the high temperatures experienced near the fusion line, rapid grain boundary migration would occur as a result of normal grain growth. When the migrating grain boundary intersects the melted regions, the liquid may penetrate into the boundary to form a liquated grain boundary.

This model explains the localized "melting" phenomenon of an alloy. However, fusion welding is a non-equilibrium thermo-mechanical process. Several metallurgical reactions such as solid-state diffusion may occur in the HAZ during welding. As a result, the material in the HAZ may undergo "liquation", which is a liquid formation process due to transient solute and/or impurity redistribution which occurs during weld thermal cycling, before the localized melting temperature is reached. The detailed mechanisms for liquation have not been well understood. To date, the penetration mechanism and the segregation mechanism have been successful in providing general explanation for the onset of liquation in the HAZ.
1.1.2 Penetration Mechanism

The penetration mechanism involves the interaction of a migrating HAZ grain boundary with liquating matrix particles. The metallurgical basis for this mechanism is the constitutional liquation theory, which was first hypothesized by Savage from a
strictly theoretical standpoint in 1959. Subsequent experiments by his students (Refs. 69-72) demonstrated that the proposed theory was valid.

Constitutional liquation is a non-equilibrium process, which occurs in alloys containing thermally stable constituent particles and is associated with the rapid heating rates experienced in the HAZ adjacent to the fusion line (Refs. 71-72). Figure 3 schematically illustrates a portion of the phase diagram of an alloy consisting of a matrix A and particles A\textsubscript{x}B\textsubscript{y}. As this alloy is rapidly heated to a temperature above the eutectic temperature (T\textsubscript{e}) of the system, liquation occurs along the particle/matrix interface. This temperature is below the solidus temperature of the bulk material. The liquation is a consequence of the transient solute redistribution at the particle/matrix interface at elevated temperatures. As the composition at the particle/matrix interface reaches the eutectic composition of the system (C\textsubscript{e}), liquation occurs at a temperature T\textsubscript{e} to form a liquid "skin" along the particle matrix interface. If the temperature is higher, for instance T\textsubscript{e} in Figure 3, the compositional range capable of undergoing liquation is larger (C\textsubscript{a} - C\textsubscript{b}) resulting in a larger amount of liquid at the interface. The occurrence of constitutional liquation depresses the effective solidus temperature dramatically. As shown in Figure 3, the equilibrium solidus temperature for an alloy C\textsubscript{0} is T\textsubscript{o}. The effective solidus decreases to T\textsubscript{2} due to compositional variation from C\textsubscript{1} to C\textsubscript{2} (partial melting hypothesis). The constitutional liquation further depresses the effective solidus temperature to T\textsubscript{e}.

The constitutional liquation theory is well-substantiated from a metallurgical standpoint, as evidenced by the number of studies published in the literature supporting this phenomena (Refs. 6-12, 31-32, 34-36, 56-63). Study of the kinetic aspects of this theory, however, is limited. The first kinetic study was conducted by
Pepe et al (Ref. 71) using the Gleeble™ thermal simulator. Results from an isothermal treatment on 18-Ni maraging steels showed that the amount of liquid formed increased initially and then decreased as time increased. Evidence of liquation disappeared after a sufficient period of exposure. Recent studies by Radhakrishnan et al (Ref. 73) also confirmed the same behavior in Inconel 718.

In addition to the temperature experienced, the amount of liquid which forms along the interface also depends on the heating rate, the dissolution kinetics of the constituent particle and the diffusivity of solute atoms in the matrix. During welding, the heating rate to a given temperature above $T_o$ is the factor which controls the...

Figure 3. Schematic diagram of a portion of a hypothetical phase diagram for an $A_A\cdot B_y$ alloy system (Ref. 71).
amount of liquid formed. In the case of extremely rapid heating rates such as in
electron beam or laser beam welding, solid state solute diffusion may be negligible due
to limited processing time and constitutional liquation would be minimized. Conversely,
if the heating rates are very low, as in an electroslag welding, the particles may
dissolve before the temperature reaches $T_o$ and the onset of liquation is thus inhibited.
In practice, constitutional liquation is commonly observed when susceptible materials are
welded using the conventional GTAW, PAW, GMAW and SMAW welding processes.

Constitutional liquation alone is not sufficient to support HAZ liquation cracking.
Since HAZ liquation cracking is a grain boundary phenomena, the presence of liquid at
grain boundaries requires the migration of grain boundaries and the interaction of these
boundaries with the liquated constituent particles, unless the constituent particles are
already at grain boundary sites. As HAZ grain boundaries migrate due to normal grain
growth, the liquated particles intercept grain boundaries and penetrate into the
boundaries. The degree of liquid penetration is dependent on the wetting
characteristics of the liquid in contact with the grain boundary substrate and the
amount of liquid. If wetting is effective and the amount of liquid is sufficient, the HAZ
grain boundaries (over the temperature range where liqutation is possible) will be
coated by a liquid film. This results in a brittle microstructure consisting of liquid films
at the grain boundaries and a solid phase inside the grain, and can be defined as a
 crack susceptible microstructure.

The constitutional liquation theory has been successful in explaining the onset
of liquation and resultant cracking in the HAZ in a variety of alloys such as Inconel 718
(Refs. 11-12), cast 718 (Refs. 6-7, 17), Waspaloy (Ref. 8), Hastelloy X (Refs. 9-10),
Udimet 700 (Ref. 8), Incoloy 800 (Refs. 56, 63), Incoloy 903 (Refs. 58, 60), Incoloy 905
(Refs. 58, 61), Incoloy 907 and 909 (Ref. 58), Type 321 (Refs. 32, 34), Type 347 (Refs. 31-32, 35-36) and A-286 (Refs. 37-38) stainless steels. The particles which have been observed to undergo constitutional liqution include (Refs. 4, 58):

1. sulphide inclusions such as MnS;
2. oxide-type inclusions, possibly some low melting point silicates and spinels;
3. carbides: MC, M₆C, M₇₃C₆;
4. boron-carbides: M₂₃(C,B)₆;
5. borides: M₃B₂, Ni₃B₃;
6. carbonitrides: Ti(C,N), Zr(C,N);
7. other constituents such as Laves phase or Ni₁₀Nb₆Si G-phase.

1.1.3 Segregation Mechanism

The constitutional liqution theory is one of the most important and well documented mechanisms for describing the metallurgical phenomena occurring in the HAZ during welding. Despite its utility, HAZ liqution cracking is often encountered in alloys which do not contain particles capable of constitutionally liquating during welding. As a result, the segregation mechanism, which rationalizes the onset of grain boundary liqution in the absence of constituent particles, was proposed (Ref. 34).

The segregation mechanism provides a model for solute and/or impurity elements to segregate to the grain boundaries, thereby reducing the liqution temperature of the boundaries relative to the surrounding matrix. Above a critical temperature during the HAZ thermal cycle, preferential melting occurs along these boundaries. This model differs from the partial melting hypothesis in that elemental partitioning in the
segregation mechanism occurs during weld thermal cycling, while the segregation (compositional inhomogeneity) in the partially melting hypothesis is remnant from material production.

The metallurgical basis for the segregation mechanism is the migration of solute or impurity elements to interfaces such as grain boundary. The tendency for segregation to grain boundaries is related to the solubility of the elements in a given matrix phase. The degree of segregation can be quantified using a grain boundary enrichment factor ($\beta_b$), which is defined as (Ref. 74):

$$\beta_b = \frac{C_b}{C_o}$$  \hspace{1cm} (1.1)

where, \( C_b \) = the equilibrium concentration of the element at the boundary;
\( C_o \) = the equilibrium concentration in the matrix.

Higher $\beta_b$ indicates a greater tendency for the solute or impurity to segregate to the boundaries. $\beta_b$ has been measured in a number of alloys using Auger spectroscopy. A summary of these results, reported by Hondros et al (Ref. 74), is given in Figure 4. These results indicate that in iron-base systems, S, C, B, P, N, Sb and Sn, in decreasing order of effectiveness, were prone to adsorption. These same elements have been shown to be the major contributors to the occurrence of liquation cracking (Ref. 75).

The actual mechanisms by which solute and/or impurity atoms segregate to the HAZ grain boundaries are complex. There has been considerable attention paid to an understanding of the grain boundary sweeping mechanism. The pipeline diffusion mechanism has been cited although there has not been experimental evidence supporting it. The Gibbsian segregation mechanism while important in static systems, may be ineffective during the rapid thermal excursions associated with welding. These three mechanisms are reviewed below.
Figure 4. The effect of the solid solubility of a number of alloy systems on grain boundary enrichment (Ref. 74).

1.1.3.1 Grain Boundary Sweeping Mechanism

This mechanism has been studied extensively in a variety of materials including iron base superalloys (Ref. 56), stainless steels (Ref. 34) and ferritic steels (Refs. 46, 48). A schematic illustration of this model is shown in Figure 5. A detailed description
of this mechanism was provided by Lippold et al (Ref. 22). According to this model, a solute and/or impurity enriched area may exist in the matrix before the grain boundaries migrate, as shown in part A of Figure 5. This solute and/or impurity enriched area may be a segregation band or rolling band (Refs. 34, 46, 48), which forms during material production, or results from the dissolution of constituent particles on weld heating (Ref. 34). As the HAZ reaches the threshold grain growth temperature, the grain boundaries migrate and intercept these solute and/or impurity enriched areas, as shown in part B of Figure 5. The solutes and/or impurities are then swept up, assimilated into and dragged along the migrating grain boundaries, as shown in part C. Those solutes and/or impurities depress the melting temperature at the grain boundaries and cause localized liquation, as shown in part D. The elements which are surface active and/or exhibit a low solubility in the matrix (higher $b_p$) would have a highest probability of being swept into and assimilated with the boundary. At the temperature in the HAZ above the localized grain boundary melting temperature, liquation in the grain boundaries occurs and the region of the HAZ within a critical embrittlement temperature range becomes susceptible to liquation cracking.

Duvall et al (Ref. 76) have argued that grain boundary migration actually results in boundary “breakaway” from solute atmospheres and that the instantaneous composition of the mobile boundary reflects the local composition of the matrix through which it is moving. They provided metallographic evidence for this argument by showing “ghost” grain boundaries in the HAZ at the sites where breakaway occurred during initial boundary migration. Lippold et al (Refs. 22, 56) pointed out, however, that while some solute elements are not mobile enough to migrate with the boundary, others are extremely mobile and may either move with the original boundary or be
Figure 5. Schematic illustration of the grain boundary sweeping mechanism (Ref. 22).
swept from the matrix as the boundary migrates. Thus, larger, slow-diffusing atoms remain behind to form "ghost" boundaries while surface-active, fast-diffusing elements will move with the boundary. Tamura et al (Refs. 34, 46, 48) experimentally showed the evidence of enrichment of solute elements such as Ni and Cr in mobile grain boundaries in stainless steels and Ni-Cr-Mo high strength steel during rapid heating while rolling bands were present.

More recently, Romig et al (Ref. 77) have investigated the potential for grain boundary sweeping in simulated HAZ microstructures in Alloy 800. Using analytical electron microscopy (AEM) techniques, no clear evidence for grain boundary enrichment via a sweeping mechanism could be detected in samples heated into the liquation temperature range and then quenched in-situ. Instead, a "micro-constitutional liquation" mechanism was detected whereby submicron Ti(C,N) particles react with the surrounding matrix and liquate.

1.1.3.2 Pipeline Diffusion Mechanism

Another proposed mechanism for HAZ grain boundary segregation is called the pipeline diffusion theory (Refs. 22, 78). A schematic illustration of this mechanism is shown in Figure 6. Since weld solidification occurs via an epitaxial growth process, grain boundaries are continuous from the fusion zone across the fusion boundary into the HAZ. By the nature of the solidification process, partitioning of solute and impurity elements along the solidification grain boundary results in solute/impurity enrichment relative to the adjacent, epitaxial HAZ boundary. This boundary provides a high diffusivity "pipeline" for elemental segregation and subsequent enrichment of solute and/or impurity on HAZ grain boundaries. Since the rate of diffusion along grain
Figure 6. Schematic illustration of the pipe-line diffusion mechanism (Ref. 22).

boundaries is greatly accelerated relative to bulk diffusion, this mechanism provides a plausible explanation for grain boundary liquation adjacent to the fusion boundary. However, experimental evidence of the existence of this mechanism has never been reported, although this mechanism has occasionally been cited (Refs. 22, 58).

The distance that an atom can travel from the grain boundary in the fusion zone into the HAZ during welding, however, is very limited. Assume that the diffusion
coefficient (D) is $10^{-5}$ cm$^2$/sec (a nominal diffusion coefficient of an element in liquid metal) and the time available (t) for diffusion to occur is 5 second. Using a basic diffusion equation to calculate the diffusion distance (X), $X = 2(Dt)^{0.5}$, it can be found that the distance is only 0.14 mm. Considering that the actual diffusion coefficient and time are normally less than the magnitude assumed in this calculation, the distance in the HAZ in which the grain boundary compositions are affected by the fusion zone is quite small. Results from this simple calculation indicate that pipeline diffusion mechanism may not contribute to expand the size of a region in the HAZ in which liquation cracking may occur. However, it is possible that HAZ grain boundary segregation over a limited regime adjacent to the fusion line is enhanced via this mechanism.

1.1.3.3 **Gibbsian Segregation Mechanism**

Another possible mechanism for grain boundary segregation in the HAZ is the Gibbsian segregation mechanism (Ref. 79). This model invokes equilibrium diffusion, where the segregated region is localized within a few atom diameters on either side of the boundary. This type of grain boundary segregation is driven by the free energy difference between a solute or impurity atom in a matrix site versus a grain boundary site. Since the defect structure in a grain boundary is greater than that in the adjacent matrix, a reduction in the local free energy can be achieved by diffusion of the atom to the boundary site. This segregation is intensified by the high temperature experienced in the HAZ and further enhanced when the solute/impurity element is highly surface active and/or exhibits a low solubility in the parent material. This model rationalizes
segregation to a static HAZ grain boundary. However, it is more difficult to apply to mobile grain boundaries.

Again, using the same procedure described in the previous section to calculate the distance that an atom can travel during welding would be less than 1 μm from the grain boundary, because the lattice diffusion coefficient is normally smaller than $10^{10}$ cm$^2$/sec. As a result of this very short distance, this mechanism is generally discounted in the contribution of the evolution of liquid in the HAZ during welding.

1.1.4 Liquid Metal Embrittlement Mechanism

The other possible mechanism for the formation of liquid films at HAZ grain boundaries is the liquid metal embrittlement mechanism (Ref. 80). The theoretical basis for this mechanism is the decrease in the cohesive strength of a grain boundary in the presence of liquid. According to the Griffith theory of cracking (Ref. 81), the applied stress required to make a thin crack grow in an elastic material is proportional to the square root of the surface energy of the new surface. From Figure 7, the surface energy requirements for the possible types of cracking per unit plan area of each, are as follows:

1. transgranular cracking: $2\gamma_s$
2. grain boundary cracking: $2\gamma_s - \gamma_b$
3. grain boundary cracking with liquid: $2\gamma_{si} - \gamma_b$

where, $\gamma_s$ = surface energy for a free surface;
$\gamma_b$ = surface energy for a grain boundary;
$\gamma_{si}$ = surface energy for a solid/liquid interface.
Because $\gamma_{ai}$ is normally much lower than $\gamma_{a}$, the development of a crack along a grain boundary due to liquid embrittlement should be much easier, from a viewpoint of thermodynamics, than that of transgranular cracking or solid-state cracking.

As mentioned previously, the nature of epitaxial grain growth in the weldment has provided a path for the last-to-solidify liquid in the fusion zone to penetrate into the grain boundaries of the HAZ and promote cracking. This mechanism, therefore, can only explain the HAZ cracks in contact with the fusion line. However, this does account for HAZ cracking which develops in a segregation-free material. A study performed by Matthews et al. (Ref. 82) on the infiltration of the grain boundaries in HY-80 steel by liquid copper-nickel weld metal showed that the grain boundary cohesive strength was reduced by the adsorption of liquid copper-nickel weld metal and resulted in an increase in the HAZ liquation cracking susceptibility. This mechanism is also supported by a phenomenon called copper contamination cracking (Refs. 83-84).

Figure 7. Development of a crack along a grain boundary (Ref. 80).
The formation of liquid films along HAZ grain boundaries is complex and it is likely that more than a single mechanism may operate simultaneously in an alloy. An example of interactive effects of the constitutional liqation and grain boundary sweeping mechanisms was given by Tamura et al (Refs. 34, 46, 48), where they studied the HAZ liqation cracking mechanisms of alloys containing Ni and Cr, and concluded that Cr and Ni were swept up by the mobile grain boundaries from a rolling band existing in the matrix. The melting temperature at the grain boundaries was depressed due to the eutectic reaction involving Cr and Ni.

1.2 The Solidification of Liquid Films and Resultant Cracking in the HAZ

As the molten weld pool passes, both the fusion zone and the PMZ solidify. The solidification of the fusion zone starts from complete liquid. Yet in the PMZ, solidification only occurs at the grain boundaries, because 0-100% of solid already exists. The initial stages of solidification in the PMZ depend on the distance relative to the fusion boundary. However, the last stage of solidification, which is the critical period in terms of cracking, is very similar to that in the fusion zone, although the origin and composition of the liquid films and the microstructural boundaries along which they reside may be different from that during weld solidification. Consequently, to a first approximation, weld solidification cracking theories can be adopted to explain the solidification of liquid and resultant cracking in the HAZ.

Many mechanisms have been proposed to describe weld solidification cracking. Among these, the shrinkage-brittleness theory (Refs. 85-88), the strain theory (Refs. 89-91), the Generalized theory (Refs. 75, 92-96) and the technological strength theory (Refs. 97-98) are best known. Aspects of both the shrinkage-brittleness theory and the
strain theory have been incorporated into the Generalized theory. For the purpose of conciseness, only the Generalized theory and the technological strength theory are reviewed. In particular, the utilization of these theories in the design of a weldability test to quantify liquation cracking susceptibility is discussed.

1.2.1 The Generalized Theory

The Generalized theory deals with weld solidification cracking from a metallurgical standpoint (Ref. 93). The solidification of an alloy is divided into four stages, as illustrated in Figure 8. In stage 1, the solid phase is dispersed while the liquid is continuous. Both liquid and solid phases are capable of relative movement to accommodate any applied and/or thermally induced strain. Cracking is impossible. In stage 2, dendrite interlocking begins but the liquid is capable of relative movement. Cracking may occur in this stage, but sufficient liquid is still present to heal any cracks which have developed. In stage 3, the further development of solid restricts liquid to interdendritic regions. The limited amount of interdendritic liquid may not be able to accommodate the large amount of strain resulting from solidification. As a result, cracking is possible and the healing of cracks is unlikely to occur in this stage. In stage 4, the solidification process is completed and the material recovers enough ductility to resist cracking.

Borland also addressed the importance of liquid wetting characteristics in his theory. The wettability of the boundary is dependent on the relative interfacial energies of the liquid and solid in contact and is described by the following relationship:

\[
\gamma_{sl}/\gamma_{ss} = \frac{1}{2 \cos(\theta/2)}
\]  

(1.2)

where,
\( \gamma_{SL} \) = the interfacial solid/liquid energy;
\( \gamma_{SS} \) = the interfacial solid/solid energy;
\( \theta \) = the dihedral angle.

For cracking to occur, the liquid should be present over a relatively wide temperature range and must cover almost all the grain faces. It implies that if \( \gamma_{SS} \) is higher (as in the case of boundary between two solid grains with the same phases), wetting would be more efficient. This criterion has been utilized to rationalize the higher cracking resistance during dual phase solidification of stainless steels (Refs. 99-100) as well as in low alloy steels (Ref. 101).

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Figure 8. Schematic illustration of the Generalized theory (Ref. 93).
Based on this theory, stage 3 (Figure 8) is the critical stage for the initiation of cracking. The temperature interval in this stage is called the critical solidification range. This is analogous to the brittle temperature range (BTR), defined by the shrinkage-brittleness theory. According to Borland (Ref. 95), the upper temperature limit of the BTR is the coherent temperature at which strength recovery begins, and the lower limit is the temperature at which the ductility of the grain boundaries is sufficient to accommodate the strain imposed upon them. A larger BTR allows more strain to accumulate and, thus, it is more likely that cracking will occur.

This theory was further modified by Matsuda et al (Ref. 96) based on experiments permitting in-situ observation of weld solidification cracking. They observed that stage 1 occurs over a much shorter temperature range than originally proposed by Borland and that significant solid networks form very rapidly upon cooling below the liquidus temperature. This is schematically illustrated in Figure 9. Stage 3, as defined by Borland, is further divided into a "film stage" (3h) and a "droplet stage" (3l). Cracking occurs in the liquid film stage and propagation proceeds in either the film stage or the droplet stage. Crack initiation is not possible in the droplet stage of solidification due to extensive solid contacts.

The Generalized theory describes several metallurgical factors relating to liquation cracking susceptibility. These factors are:

1. the magnitude of the BTR over which cracking can occur, with a larger BTR indicating a greater susceptibility to cracking;

2. the quantity of liquid present in the final stage of solidification, with a greater quantity of liquid increased cracking resistance due in part to cracking healing;
3. the dihedral angle exhibited by the liquid and the grain boundaries, with a smaller dihedral angle (greater grain boundary wetting) promoting cracking.

The concept of the BTR has been utilized in the design of the Transvarestraint test (Ref. 102). The quantity and dihedral angle of the last-to-solidify liquid have also been employed to rationalize the cracking susceptibility in different alloys (Refs. 95, 103). These three factors, in combination, are probably most important in determining cracking susceptibility. Quantifying weldability by only one of the three factors may not reflect the actual cracking behavior of a material during welding. However, there is not a cracking index available to date which takes all these factors into account.

![Diagram](Figure 9. Schematic illustration of the modified Generalized theory (Ref. 96).)
1.2.2 The Technological Strength Theory

Complementary to the Generalized theory, the technological strength theory deals with solidification cracking from a mechanical standpoint. This theory provides a ductility curve for a material during solidification, as shown in Figure 10. The material exhibits a ductility dip in the BTR. Outside the BTR, the ductility of the material is assumed to be large enough to accommodate the thermally and/or mechanically induced strain during welding. In addition to the ductility curve, the welding-induced strain is also shown in the same diagram by assuming that the strain is linearly proportional to the decrease in temperature upon weld cooling. The occurrence of cracking is a result of competition between the accumulation of strain and the recovery of ductility during weld cooling. Cracking would occur if the ductility of the material is exhausted by the thermally and/or mechanically induced strain.

In contrast to the Generalized theory, this theory ignores the metallurgical reactions during solidification. Rather, it provides identifiable elements for quantifying cracking susceptibility. According to this theory, the major factors which influence cracking susceptibility are:

1. the magnitude of the BTR over which the material exhibits a ductility dip, with a larger BTR indicating a greater cracking susceptibility;
2. the ductility within the BTR, with a lower minimum ductility indicating a greater cracking susceptibility;
3. the deformation conditions experienced during welding or weldability testing, including both the total strain experienced and the rate at which this strain is applied.
To a first approximation, these three factors can be combined to form a liquation cracking index by applying an external strain at a constant rate relative to the decrease in temperature to exhaust all the ductility remaining in the material within the BTR. A critical deformation rate above which cracking occurs can be determined. This index (critical deformation rate) was termed as a technological strength index by Prokhorov, as shown in Figure 10. The critical deformation rate criterion has been utilized to design strain rate controlled weldability testing techniques such as the LTP1-6M machine (Ref. 104) and the program-controlled deformation cracking test (PVR) (Ref. 105).

\[
\varepsilon_w = \alpha \varepsilon_{w \text{ BTR}} \quad \text{free contraction deformation}
\]

\[
\varepsilon_f = \alpha \varepsilon_{f \text{ BTR}} \quad \text{deformation caused by the change in shape}
\]

\[
\varepsilon_m = \alpha \varepsilon_{m \text{ BTR}} \quad \text{amount of ductility available}
\]

\[
\varepsilon_{\min} = \alpha \varepsilon_{\min \text{ BTR}} \quad \text{ductility of the material}
\]

Technological Strength Index \[ \alpha \varepsilon_f = \alpha \varepsilon_{\min} - \alpha \varepsilon_{w \text{ BTR}} \]

Figure 10. Schematic illustration of the technological strength theory (Ref. 97).
The shape of the ductility curve in this theory was criticized by some investigators such as Yakushin (Ref. 106). They claimed that the ductility curve within the BTR may depend on the material and that the lowest ductility may not always occur at the lower bound of the BTR. Direct observation by Matsuda et al (Ref. 96) indicated that the minimum ductility occurs within the BTR close to the high temperature bound of the BTR. With sufficient strain, cracks initiated within the BTR at the temperature where the material exhibited the minimum ductility and propagated toward both the high temperature and low temperature boundary. This argument does not affect the validity of using the critical deformation rate as a cracking index, because the critical deformation rate is also dependent on the shape of the ductility curve. As shown in Figure 10, the critical deformation rate is a line tangent to the ductility curve. If the temperature at which the material exhibits a minimum ductility shifts, the critical deformation rate also changes.

1.3 Weldability Testing Techniques

According to liquation cracking theories, two basic elements must be present simultaneously to induce cracking. These are a crack susceptible microstructure and a threshold strain. To date, all the weldability testing techniques except the hot-ductility test were designed with the same concept by developing a crack susceptible microstructure and simultaneously applying a sufficient amount of strain (or stress) to produce a liquation crack. The susceptibility to cracking is quantified by the threshold strain or stress to induce a crack, or the degree of cracking under a specific level of strain or stress.
There have been many review papers classifying the types of weldability testing techniques (Refs. 2, 107-110). According to Baeslack and Lippold (Ref. 2), weldability testing techniques can be categorized into representative tests and simulative tests depending on the manner in which the strain or stress is applied. Based on this definition, the classification of weldability testing techniques can be summarized in Table 1 with typical examples. Representative tests attempt to reproduce the actual welding condition in a small test sample. These tests depend on self-restraint induced by the specimen design and/or fixturing. The circular patch test (Ref. 111) and Howaldt test (Ref. 112) are the typical examples in this group. In most cases, results from these tests represent a specific material/process/restraint combination. Any variation in process (welding parameters) or restraint factor would affect test results. It is generally very difficult to isolate the material factor from the test results. Thus, these tests are ineffective in quantifying the difference in liquation cracking susceptibility among different materials.

Simulative tests seek to simulate some aspects of the actual welding conditions. These tests involve the application of an external augmented strain or stress. The external restraint can be applied as a constant stress, such as in the Sigmajig test (Ref. 113) or as a tensile strain by controlling the absolute amount of strain, such as the Gleeble hot cracking test (Ref. 114), or the application rate, such as the variable tensile strain hot cracking test (Ref. 115), the Russian LTP1-6M machine (Ref. 104) or the program controlled deformation cracking test (PVR) (Ref. 105). Alternatively, the external restraint can be applied as a bending strain by controlling the absolute amount of strain, such as in the Varestraint test (Refs. 116-117), or the application rate, such as the slow bending type Transvarestraint (Ref. 118). The
Table 1. Classification of weldability testing techniques.

- **Representative Tests**
  - E.g.: Circular patch test, Houldcroft test

- **Weldability Testing Techniques**
  - **Simulative Tests**
    - **Tension Type**
      - E.g.: Gleeble hot cracking test
    - **Bending Type**
      - E.g.: Varestraint test, Spot-Varestraint test, Transvarestraint test
  - **Strain Rate Control**
    - **Tension Type**
      - E.g.: LTP1-6M, PVR
        - Variable tensile strain hot cracking test
    - **Bending Type**
      - E.g.: Slow bending type, Transvarestraint test
  - **Stress Control**
    - E.g.: Sigmajig test

- **High Temperature Mechanical Test**
  - E.g.: Hot-ductility test, Hot-strength test
thermally-induced restraint of the specimen and/or fixturing is usually negligible compared with the relatively large externally applied strain or stress. This approach allows the restraint factor to be measured. However, according to the technological strength theory (Refs. 97-98), the ductility of a material varies in a weld thermal cycle. The temperature at which a sufficient strain is applied on a material during weldability testing may be different from that under actual welding conditions. This difference may result in a disparate cracking behavior. Without comprehension of metallurgical response during weldability testing, it would not be appropriate to directly apply weldability test results to predict cracking susceptibility of materials in actual welding conditions.

Among the weldability testing techniques, only a few methods have been utilized in quantifying the susceptibility of a material to HAZ liquation cracking. Based on the published literature, the hot-ductility test, the longitudinal-Varestraint (mini-Varestraint) test and the spot-Varestraint test appear to be the most widely utilized methods (Ref. 1). In the present study, these three methods were rigorously evaluated. The test procedures and methodologies of data interpretation, in particular, are reviewed and critically discussed in the following sections.

1.3.1 Longitudinal-Varestraint test

The original Varestraint test was developed by Savage et al (Refs. 116-117) at Rensselaer Polytechnic Institute in the mid-60’s, and soon became one of the most widely utilized weldability testing techniques for quantifying the susceptibility of a material to weld solidification cracking. During the past several decades, three different modified versions: the mini-Varestraint test, the spot-Varestraint test (originally
called Tiga-ma-jig) and the Transvarestraint test, have been developed based on the original Varestraint concept. The mini-Varestraint test uses a smaller test sample, nominally 163 X 25 X 6.4 mm (6.5 X 1 X 0.25 in.) as compared with the original Varestraint test, nominally 300 X 50 X 13 mm (12 X 2 X 0.5 in.). In order to avoid confusion with the spot-Varestraint and Transvarestraint tests, the mini-Varestraint test is called the longitudinal-Varestraint test in this report. Although this method is mainly utilized in characterizing the weld solidification cracking, it has also be applied to determine HAZ liquation cracking susceptibility (Refs. 39-43, 54-55).

A schematic of the longitudinal-Varestraint test is shown in Figure 11. A specimen is supported as a cantilever beam at one end and a gas tungsten-arc weld (GTAW) is produces in the center section of the specimen. When the arc approaches the center of a radiused die block (marked "A" in Figure 11), a pneumatically operated ram is triggered and forces the specimen to conform to the surface of the die block. Meanwhile, the arc travels onward and is subsequently interrupted in the run off area "C". Two auxiliary bending bars are added in order to ensure that the specimen conforms to the contour of the die block. The applied augmented strain (¢) of the top surface of the specimen can be varied by adjusting the radius of the die block (R) following the equation, $\epsilon = t / (2R + t)$, where $t$ is the specimen thickness. In this manner, both weld solidification cracks and HAZ liquation cracks can be produced. The HAZ liquation cracks are normally located adjacent to the fusion boundary and perpendicular to the welding direction.

Cracking susceptibility is assessed by measuring each crack length on the as-tested specimen surface. The threshold strain (Table 2) to cause cracking and the degree of cracking at a specific strain level have been generally utilized as cracking
indices (Refs. 39-43, 54-55). The degree of cracking was quantified by the maximum crack length (MCL) (Table 2), the total crack length (TCL) (Table 2) or the cracked HAZ length (CHL) (Table 2).

Figure 11. Schematic illustration of the longitudinal-Varestraint test.
1.3.2 Spot-Varestraint Test

The spot-Varestraint test (Refs. 14, 116-117) evaluates the cracking susceptibility of a material by applying an augmented strain when the material is experiencing a crack susceptible microstructure, which is created by producing a spot weld. A schematic of the spot-Varestraint test is shown in Figure 12. During the spot-Varestraint testing, a gas tungsten-arc (GTA) spot weld is produced in the center section of a small rectangular specimen, nominally 150 X 25 X 6.4 mm (6 X 1 X 0.25 in.). After a predetermined weld time, the arc is extinguished and the specimen is forced to conform to the surface of a radiused die block. In this manner, HAZ liquation cracks can be generated on the surface of the specimen adjacent to the GTA spot weld. The applied augmented strain ($\varepsilon$) of the top surface of the specimen is approximated by the equation, $\varepsilon = t / (2R + t)$, where $t$ is the specimen thickness and $R$ the radius of the die block.

Cracking susceptibility is determined by measuring the length of each crack on the as-tested specimen surface. The threshold level of strain to cause cracking or the degree of cracking (quantified by MCL or TCL) at a specific strain level have been generally accepted as cracking indices since the introduction of this test (Refs. 6-7, 14, 18-23, 52, 56, 58-63).

1.3.3 Hot-Ductility Test

The concept in the design of the hot-ductility test is different from most of other weldability testing techniques. Instead of quantifying the cracking susceptibility by the degree of cracking, it characterizes the ductility of the material in the HAZ during welding and relates these hot ductility data to cracking susceptibility. A schematic
Figure 12. Schematic illustration of the spot-Varestraint test.
illustration of this method is shown in Figure 13. Basically, small tensile samples are fractured rapidly at some specific temperatures during either the on-heating or on-cooling portion of a duplicated weld thermal cycle in a thermo-mechanical simulator called a "Gleeble™". The reduction-in-area of the fractured sample is subsequently determined providing a measure of ductility of the material. Both the on-heating and on-cooling ductility curve in the vicinity of the solidus temperature can be obtained, as shown in Figure 13, and the nil-ductility temperature (NDT) (Table 2), nil-strength temperature (NST) (Table 2) and ductility recovery temperature (DRT) (Table 2) determined. In addition to the hot ductility, the hot strength of the material can also be measured.

In the past, experimental methodologies for hot-ductility testing have been the subject of considerable investigation. The methodologies involve the determination of test parameters and resultant data interpretation. These two subjects are interrelated. Since the ductility curves are the principal results, most investigators focused their attention on the effects of parameters on the hot ductility. Variation of the specific temperatures such as the NDT, NST and DRT due to different test conditions was overlooked. In testing methodology, the most important factors to be considered are: heating rate, cooling rate, peak temperature for the on-cooling test, hold time at the test temperature and peak temperature, and the stroke rate (crosshead speed). These factors have been studied by Lundin et al (Ref. 119) and Yeniscavich (Refs. 15, 120). Their results indicated that heating rate (62 - 248°C/sec) and stroke rate (3.8 - 127 mm/sec) did not significantly affect the hot ductility. The effects of cooling rate and hold time are controversial. For 316L, 304L and 347 stainless steels, Lundin et al (Ref. 119) claimed that the ductility decreased at lower cooling rates or longer hold.
Figure 13. Schematic illustration of the hot-ductility test.
times. However, Yeniscavich (Refs. 15, 120) observed an opposite trend with Inconel 600. Among the testing parameters, the peak temperature is the most critical factor influencing the recovery of ductility for the on-cooling test (Refs. 8-9, 15, 120). The on-cooling ductility decreases as the peak temperature increases. Different temperatures such as the NDT, the NST, the temperature between NDT and NST, and an arbitrary temperature have been used as a peak temperature for the on-cooling tests in the published reports. The use of different peak temperatures has made the comparison of hot ductility test results very difficult. The methods for interpreting hot ductility test results for HAZ cracking assessments vary significantly from investigator to investigator. The several criteria that have been proposed to date include:

1. classification of hot ductility curves;
2. the extent of the nil-ductility region and the amount and rate of ductility recovery on-cooling;
3. nil-ductility temperature range between the NST and DRT;
4. incorporation of the extent of ductility and strength recovery;
5. the zero ductility range and mid-temperature ductility dip range;
6. the temperature range between NDT and DRT;
7. ratio of ductility recovery (RDR), ductility recovery rate (DRR) and nil ductility temperature range (NDR).

1.3.3.1 Classification of Hot Ductility Curves

The criterion relating both on-heating and on-cooling ductility at elevated temperatures to HAZ liquation cracking susceptibility was first proposed by Nippes et al (Ref. 121) soon after the development of the Gleeble. After an extensive study using
this test and comparing the test results with field experience, Nippes et al (Ref. 122) concluded that the hot ductility behavior of stainless steels can be grouped into five classifications, as shown in Figure 14. According to their criterion, when a material has an H2 on-heating behavior, it should be rejected because of high HAZ cracking susceptibility, and when the material is in the H1 on-heating class, on-cooling tests have to be conducted to further evaluate the HAZ cracking propensity. The C1, C2 and C3 on-cooling classifications are the most commonly encountered after an on-heating behavior equivalent to H1. The material with behavior described as C1 is readily weldable without any tendency toward HAZ liquation cracking. The C2 and C3 behaviors predict less desirable liquation cracking tendencies with C3 indicative of the highest susceptibility to HAZ cracking.

1.3.3.2 The Extent of Nil-Ductility Region and the Amount and Rate of Ductility Recovery On-Cooling

Nippes's criterion provides a qualitative measurement of cracking susceptibility. However, the classification of hot ductility curves is sometimes difficult. Also, the on-heating hot ductility curve was considered to be unimportant because cracking is presumed occurred on weld cooling (Ref. 64). As a result, the amount and rate of ductility recovery on cooling was considered to be the critical factor relating to HAZ liquation cracking susceptibility (Refs. 8-9, 13, 36, 39-40, 42-44, 103, 123-124).

As mentioned previously, the on-cooling hot ductility behavior depends on the peak temperature for the on-cooling test. In general, the on-cooling hot ductility at a given temperature decreases as the peak temperature increases. Different peak temperatures from NDT - 28°C (50°F) (Ref. 125) to NST have been used and it is
Figure 14. Classification of hot ductility curves (Ref. 122).
almost impossible to compare results from different peak temperatures. Although a few investigators still use NDT as a peak temperature, most researchers agree that higher peak temperatures are more appropriate in terms of discriminating differences in cracking susceptibility among materials. Based on extensive testing of Ni-base superalloys, Duvall and Owczarski (Ref. 8, 9) proposed that the extent of the nil-ductility region (NDR) (Table 2) and the rate and amount of ductility recovery are the factors that best correlate with the degree of crack sensitivity. A smaller nil-ductility region and a faster recovery of ductility indicate a greater resistance to HAZ liquation cracking. The nil-ductility region is a region in the HAZ within which the material exhibits negligible ductility. This region is composed of two components: the on-heating nil-ductility region, \( T_L \cdot NDT \), and the on-cooling nil-ductility region, the temperature range between a peak temperature and its corresponding DRT. Based on a rough calculation of the weld pool size, Duvall et al (Ref. 9) stated that cracking occurred outside the nil-ductility region. Because defining the nil-ductility region is a time consuming task, they further suggested that the on-cooling hot ductility results from a peak temperature equal to the NST normalized relative to the NST provided a direct comparison of cracking susceptibility. This criterion has been widely utilized in the interpretation of the hot ductility data over the past 20 years (Refs. 8-9, 13, 39, 42-43).

1.3.3.3 Nil-Ductility Temperature Range Between the NST and DRT

In order to provide a quantitative measure of cracking susceptibility, the magnitude of the temperature range between the NST and DRT was proposed as a quantitative cracking index in the early 60's (Refs. 36, 39, 44, 123). An increase in this
temperature range represents a higher sensitivity to HAZ liquation cracking. This cracking index is the temperature range on a weld cooling cycle over which the material exhibits negligible ductility. Physically, it is the BTR of a point in the HAZ which experiences a peak temperature equal to the NST. Muesch (Ref. 36) further derived a cracking factor, \((\text{NST} - \text{DRT}) / \text{NST} \times 100\), as a cracking index based on this temperature range. This temperature range has received much attention in the interpretation of hot ductility test data.

1.3.3.4 **Incorporation of the Extent of Ductility and Strength Recovery.**

Because the amount and rate of recovery of ductility may not always correlate the cracking susceptibility with field experience, the importance of hot strength was proposed. This criteria was addressed by Kreischer (Ref. 13), Williams (Ref. 123), Weiss et al (Ref. 124), and also supported by Nippes et al (Ref. 121). They claimed that with similar on-cooling hot ductility behavior, the material with a higher hot strength recovery rate was less susceptible to cracking.

1.3.3.5 **The Zero Ductility Range and Mid-Temperature Ductility Dip Range.**

These criteria were proposed by Yeniscavich (Refs. 15, 120, 126) based on the hot ductility test results of IN 600 and Hastelloy X nickel base alloys. The zero ductility range (ZDR) is the temperature range between the \(T_L\) and DRT. A larger ZDR represents a greater sensitivity to HAZ liquation cracking. This cracking index was also supported by Arata et al (Ref. 127) and Holsberg (Ref. 128). Arata et al (Ref. 127) claimed that this index correlated well with the critical heat input to avoid microcracking during electron beam welding of heat resistant superalloys.
Yeniscavich (Refs. 15, 120, 126) also proposed another mid temperature ductility dip range criterion. He stated that some alloys exhibited a ductility dip at intermediate temperatures during weld cooling. If the ductility dips to zero in the mid-temperature range, the alloy is classified as crack-sensitive. If the ductility in the mid-temperature range remains high, the alloy is crack-resistant.

1.3.3.6 The Temperature Range between NDT and DRT

This cracking index was evaluated by Arata et al (Ref. 127). They claimed that this temperature range correlates well with the critical heat input to avoid microcracking during electron beam welding of heat resistant superalloys. With a smaller temperature range, the material is more resistant to cracking. Again, the DRT is dependent on the peak temperature. They employed different peak temperatures for different alloys with the peak temperatures determined based on the criterion that the microstructure achieved in hot-ductility tests were similar to those adjacent to the fusion line under actual welding conditions.

1.3.3.7 Ratio of Ductility Recovery, Ductility Recovery Rate and Nil Ductility Temperature Range

Recently, Lundin et al (Refs. 40, 129) measured the fracture strain in the far-HAZ and extrapolated the fracture strain into the PMZ. They concluded that the fracture strain in the PMZ was about 5%. The temperature on cooling from the NDT at which the material recovered 5% ductility was called a critical strain temperature (CST). Based on these results, they proposed three new indices, as illustrated in Figure 15, for relating hot-ductility data to HAZ liquation cracking. These indices are:
(1) ratio of ductility recovery (RDR) which is the ratio of the area under the on-cooling curve to the area under the on-heating curve from the peak temperature (NDT) to the maximum on-heating hot ductility temperature; (2) ductility recovery rate (DRR) which is the ratio of the on-cooling ductility to the on-heating ductility measured at the maximum on-heating hot ductility temperature; and (3) nil ductility temperature range (NDR) which is the temperature range between the peak temperature (NDT) and the CST. They claimed that these three indices were successfully utilized to evaluate HAZ liquation cracking susceptibility of nuclear grade stainless steels.

![Diagram](image)

\[
\text{DRR (\%)} = \frac{\text{area (BCF)}}{\text{area (ACF)}} \times 100
\]

\[
\text{RDR (\%)} = \frac{\text{area (BCF)}}{\text{area (ACF)}} \times 100
\]

\[
\text{NDR (C)} = \frac{\text{AC}}{\text{DF}}
\]

Figure 15. Schematic illustration of RDR, DRR and NDR criteria (Ref. 129).
Table 2. Definition of terms.

**Brittle Temperature Range (BTR):** The temperature range in a thermal cycle on weld cooling within which the material is susceptible to liquation cracking due to the localized loss of grain boundary ductility.

**Cooling Time (t<sub>c</sub>) or Delay Time:** The time period between arc extinction and specimen bending for the spot-Varestraint test.

**Cracked HAZ Length (CHL):** The length in the HAZ measured along the fusion boundary over which cracking is observed in the longitudinal-Varestraint test.

**Crack Susceptible Region (CSR):** A region in the HAZ adjacent to the fusion line within which the material is susceptible to liquation cracking due to the localized loss of ductility.

**Ductility Recovery Temperature (DRT):** Temperature on-cooling from a peak temperature above the NDT at which perceptible ductility (>5%) of the material is apparent.

**Maximum Crack Length (MCL):** The maximum length of cracks on the as-tested specimen surface in longitudinal- or spot-Varestraint tests. Traditionally, this crack length is measured from tip to tip of a crack. In this study, the MCL represents the distance between the isotherm at the crack tip and the isotherm at the fusion line.

**Nil-Ductility Region (NDR):** A region in the HAZ adjacent to the fusion line within which the ductility of the material is essentially zero.

**Nil-Ductility Temperature (NDT):** Temperature on-heating at which the ductility of the material drops to zero.

**Nil-Strength Temperature (NST):** Temperature on-heating at which the strength of the material drops to essentially zero.

**Saturated Strain:** The applied augmented strain level above which the maximum crack length remains constant, or saturates, for the spot- and longitudinal-Varestraint tests.

**Threshold Strain:** The applied augmented strain above which cracking occurs in the longitudinal- or spot-Varestraint tests.

**Total Crack Length (TCL):** Cumulative length of all cracks on a as-tested specimen surface in spot- or longitudinal-Varestraint tests.
CHAPTER II
PRELIMINARY STUDY

2.1 Introduction

As reviewed in chapter 1, the hot-ductility and spot-Varestraint tests have been widely utilized in quantifying HAZ liquation cracking susceptibility. However, there has been little effort to correlate the results obtained from these two tests. As a consequence, a preliminary study was devised to compare the HAZ liquation cracking susceptibility of Incolloys 903 and 909 using both the hot-ductility and spot-Varestraint tests. It was found that these test results were inconsistent in predicting HAZ liquation cracking susceptibility when "conventional" techniques were utilized to interpret the results. A greater MCL in Incoloy 909 was obtained from the spot-Varestraint test and a less rate and amount of ductility recovery in Incoloy 903 was obtained from the hot-ductility test. This inconsistency was rationalized by using a newly-developed methodology. The results from this preliminary study stimulated a research program to fundamentally evaluate the two techniques and resulted in the development of a new methodology, which will be discussed in the next three chapters. This chapter summarizes the quantitative spot-Varestraint and hot-ductility test results of Incoloy 903 and 909. Both the "conventional" and the newly-developed methods for interpretation of results from the two tests are discussed.

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2.2 Materials

The 900-series Incoloy alloys are precipitation-strengthened Fe-Ni-Co superalloys designed for elevated temperature applications requiring high strength, a low coefficient of thermal expansion and superior resistance to thermal fatigue (Refs. 130-131). The excellent room and elevated temperature strengths exhibited by these alloys result from the presence of Al, Ti and Nb (about 7 total wt%), which promote formation of the γ' Ni₃(Al, Ti) strengthening phase. The highly desirable combination of properties offered by these alloys has promoted their increased use in aerospace applications requiring close dimensional tolerance over a wide range of temperatures.

Recent work on Incoloy 903, 907 and 909 by Baeslack et al (Refs. 58-61) showed these alloys to be susceptible to weld HAZ liquation cracking. Metallurgical characterization of the weld fusion boundary region has shown that liquation originates, in part, from the constitutional liquation of Nb-rich carbides in Incoloy 903 and Nb and Si-rich Laves phase or G-phase in Incoloy 909 (Refs. 58-60). To date, the HAZ hot ductility characteristics of these alloys and their effects on liquation cracking susceptibility have not been reported.

Both Incolloys 903 and 909 were evaluated in this preliminary study. The chemical compositions of the materials tested are listed in Table 3. Both materials were received in the form of 6.35 mm (0.25 in.) thick plate. Before testing, Incoloy 903 was solutionized at 982°C (1800°F)/1 hr, water quenched, and Incoloy 909 at 1038°C (1900°F)/1 hr, water quenched, in order to produce nearly equivalent grain size. As shown in Figure 16, the two alloys exhibited comparable recrystallized austenite grain sizes in the range from ASTM 4 to 6. Incoloy 903 contained a bi-modal distribution of carbides. Coarse, MC-type carbides, rich in Nb and Ti, were randomly dispersed in
the matrix, while fine Nb-rich carbides were located primarily at grain boundaries (Ref. 132). Incoloy 909 showed a distribution of fine particles along prior-austenite grain boundaries and only a few coarse carbide particles. These fine particles were shown to be enriched in Nb and Si, suggesting Laves phase or Ni<sub>6</sub>Nb<sub>5</sub>Si<sub>3</sub>, G-phase (Ref. 133).

Table 3. Chemical compositions of Incoloy 903 and 909 investigated in the preliminary study (wt%).

<table>
<thead>
<tr>
<th>Element</th>
<th>Incoloy 903</th>
<th>Incoloy 909</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni</td>
<td>38.24</td>
<td>38.61</td>
</tr>
<tr>
<td>Co</td>
<td>14.96</td>
<td>12.79</td>
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<tr>
<td>Nb</td>
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<td>5.03</td>
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<tr>
<td>Ti</td>
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<tr>
<td>Si</td>
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<td>Cu</td>
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<tr>
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<tr>
<td>B</td>
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<td>0.002</td>
</tr>
<tr>
<td>Fe</td>
<td>bal.</td>
<td>bal.</td>
</tr>
</tbody>
</table>
Figure 16. Microstructure of the solution heat treated Incoloy 903 and 909 base materials.
2.3 Experimental Procedures

2.3.1 Hot-Ductility Testing

Hot-ductility tests were performed under an argon atmosphere using a Gleeble™ 1500 with specimen dimensions of 100 X 9.53 X 5.08 mm (4 X 0.375 X 0.2 in) with a freespan (jaw spacing) of 19 mm (0.75 in). The heating cycle used in this investigation was similar to that published by Nippes et al (Ref. 122) for stainless steels. Using this thermal cycle, specimens were rapidly heated to selected peak temperatures in 9.5 seconds and fractured immediately at a crosshead speed of 20 cm/sec (8 in/sec). The NST was determined under similar heating conditions, but with a reduction in heating rate to 10°C/sec above 1300°C. Samples were fractured under a static load of about 10 kg. On-cooling tests were performed from two peak temperatures at a constant cooling rate of 20°C/sec for both alloys. One peak temperature was the NST (1340°C for Incoloy 903 and 1268°C for Incoloy 909) and the other peak temperature was midway between the NST and NDT (1290°C for Incoloy 903 and 1219°C for Incoloy 909). The test conditions are summarized in Table 4. The reduction-in-area of each specimen was subsequently measured using a caliper to provide a measure of ductility at the corresponding test temperatures.

<table>
<thead>
<tr>
<th>Table 4. Conditions for hot-ductility testing of Incolloys 903 and 909.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Heating Time: 9.5 seconds</td>
</tr>
<tr>
<td>Holding Time at peak temperature: 0.03 second</td>
</tr>
<tr>
<td>Cooling Rate: 20°C/sec (36°F/sec)</td>
</tr>
<tr>
<td>Holding Time at test temperature: 0.03 second</td>
</tr>
<tr>
<td>Stroke Rate: 200 mm/sec (8 in/sec)</td>
</tr>
<tr>
<td>Freespan: 19 mm (0.75 in)</td>
</tr>
<tr>
<td>Atmosphere: Argon</td>
</tr>
</tbody>
</table>
2.3.2 Spot-Varestraint Testing

Spot-Varestraint testing was performed using an apparatus as illustrated in Figure 12. Dimensions of test specimens were 15.2 X 2.5 X 0.64 cm (6.0 X 1.0 X 0.25 in). The test conditions are listed in Table 5.

Quantitative cracking data were obtained by measuring the length of each crack projected in a direction perpendicular to the fusion line on the as-tested surface using a binocular microscope under 60X magnification. Thus, the MCL in this preliminary study represents the distance of a crack tip from the fusion line. The number and length of both the open and back-filled cracks were counted.

Table 5. Conditions for the spot-Varestraint testing of Incolloys 903 and 909.

<table>
<thead>
<tr>
<th>Condition</th>
<th>Specification</th>
</tr>
</thead>
<tbody>
<tr>
<td>Current</td>
<td>90 amps</td>
</tr>
<tr>
<td>Voltage</td>
<td>12 volts</td>
</tr>
<tr>
<td>Weld Time</td>
<td>40 seconds</td>
</tr>
<tr>
<td>Cooling Time</td>
<td>0.05 seconds</td>
</tr>
<tr>
<td>Augmented strain</td>
<td>2%</td>
</tr>
<tr>
<td>Electrode</td>
<td>W-2%ThO₂, 2.4 mm (3/32 in.) dia., 60° included angle</td>
</tr>
<tr>
<td>Shielding Gas</td>
<td>Argon, 15 CFH</td>
</tr>
<tr>
<td>Air Pressure</td>
<td>0.55 MPa (80 psi)</td>
</tr>
<tr>
<td>Strain Rate</td>
<td>approximately 13%</td>
</tr>
</tbody>
</table>

2.3.3 Metallurgical Characterization

Representative test specimens were metallographically prepared by polishing through 0.05 micron alumina and etched with a mixed acid solution comprised of equal parts of concentrated nitric, hydrochloric and lactic acids. The microstructures were examined using an optical microscope at magnifications up to 400X. Fractographic examination of hot-ductility test specimen surfaces was performed using a scanning
electron microscope (SEM) at 25 kV accelerating voltage and magnifications up to 2000X.

2.4 Results

2.4.1 Hot-Ductility Test

The hot-ductility test results are presented in Figure 17. The on-heating ductility of Incoloy 903 decreased rapidly above 1200°C and approached zero at 1240°C (NDT). On-cooling from 1290°C, the ductility of this alloy did not recover significantly until it had been cooled to below 1075°C. For a higher peak temperature equal to the NST (1340°C), the rate and amount of ductility recovery decreased with a DRT of about 1050°C (DRTNST). Incoloy 909 exhibited a similar on-heating behavior with a lower NDT of 1150°C. The on-cooling behavior of Incoloy 909, however, was quite different from that of Incoloy 903, with ductility recovering rapidly at 1140°C from the peak temperature of 1219°C. For a higher peak temperature equal to the NST (1268°C), this alloy also exhibited a slower rate of ductility recovery with a DRT of about 1100°C (DRTNST). Notice that the NDT-DRTNST temperature range was only 50°C for Incoloy 909 compared to 190°C for Incoloy 903. This large difference in temperature range was attributed to the difference in the liquating constituents. In Incoloy 903, Nb-rich MC-type carbides liquated during heating and resolidified to a Laves phase/austenite eutectic. In Incoloy 909, Nb- and Si-rich Laves phases liquated and resolidified to an identical Laves phase (Ref. 58).

Figure 18 shows microstructures of Incoloy 903 and 909 tested at the NST. Considerable intra- and intergranular liquation is apparent. At lower peak temperatures of 1290°C and 1219°C for Incoloy 903 and 909, respectively, liquid films at grain
Figure 17. Hot-ductility test results of incoloy 903 and 909.
Figure 18. Microstructures of hot-ductility specimens tested at the NST.
boundaries were not evident under an optical microscope, despite the observable liquation of Nb-rich constituent phases in both alloys (Figure 19). The observed restriction of grain growth above the NDT, however, confirmed that grain boundary liquation occurred and that it was effective in pinning the grain boundaries.

Fractographic analysis of hot ductility specimen fracture surfaces, as shown in Figure 20, confirmed an intergranular, liquation-related fracture (Ref. 124). This also suggested that relatively thin liquid films were present at grain boundaries prior to fracture.

2.4.2 Spot-Varestraint Test

Spot-Varestraint test results are shown in Figure 21. As indicated, Incoloy 909 exhibited a longer maximum, open + backfilled crack length and a greater open + backfilled total crack length versus Incoloy 903. However, Incoloy 909 did contain a fewer total number of cracks. It is of interest to note that Incoloy 909 also exhibited an appreciably lower maximum open crack length and total open crack length than Incoloy 903. Examination of the as-tested and metallographically prepared specimens showed this difference to result from the more extensive backfilling of HAZ liquation cracks with liquid from the fusion zone and a region in the PMZ with extensive localized melting (and/or liquation) in Incoloy 909. Frequently in Incoloy 909, the entire crack length from the weld fusion boundary to the tip of the crack was backfilled. Although such crack backfilling can heal liquation cracks in actual fusion weldments, this effect is significantly enhanced during spot-Varestraint testing. In addition, the backfilling is not taken into account in the hot-ductility test. Consequently, in order to provide a basis for the comparison of the results from both tests, the entire open + backfilled crack
Figure 19. Microstructures of hot-ductility specimens of Incoloy 903 and 909 tested at a peak temperature midway between the NST and NDT.
Figure 20. Fracture surface of hot-ductility specimens of Incoloy 903 and 909 tested at a peak temperature midway between the NST and NDT.
lengths were utilized as quantitative indices of cracking susceptibility in this preliminary study.

Typical weld HAZ microstructures of spot-Varestraint specimens in each alloy are shown in Figure 22. Regions directly adjacent to the fusion line (A-B in Figure 22) experienced appreciable intergranular liquation in both alloys, with this region being more extensive in the Incoloy 909. No cracks were observed within this region. Further from the fusion line, open and backfilled intergranular cracks were observed.

![Graphs showing crack length and total number for Incoloy 903 and 909](image)

Figure 21. Spot-Varestraint test results of Incoloy 903 and 909.
Figure 22. HAZ microstructure of the spot-Varestraint test specimens of Incolloys 903 and 909.
2.5 Discussion

Spot-Varestraint and hot ductility weldability tests are normally considered complementary in predicting HAZ liquation cracking susceptibility and typically yield quantitatively consistent results. A study performed by Brooks (Ref. 62) indicated that these two test results were consistent in accessing the susceptibility to HAZ liquation cracking of different heats of modified A-286 stainless steels. However, the hot ductility results interpreted either by the amount and rate of ductility criterion (Ref. 9), as shown in Figure 23, or by the quantitative NST-DRT$_{\text{NST}}$ temperature range (Figure 24) are contradictory to the TCL or MCL determined with the spot-Varestraint test in determining the cracking sensitivity of Incolloys 903 and 909. This discrepancy can be rationalized by considering the thermo-mechanical conditions experienced in the HAZ during spot-Varestraint testing.

Metallographic and fractographic analysis of the crack morphology and surface of spot-Varestraint specimens by Ernst et al. (Ref. 58) concluded that intergranular cracking resulted from the presence of thin liquid films at grain boundaries. Comparing these observations with the microstructures and data obtained from the hot-ductility tests indicates that cracks should propagate across a region within which the material exhibits nil-ductility, i.e., from the T$_L$ to the NDT as defined by the hot-ductility tests. It is of importance to note that in an actual weldment a liquation crack may be expected to propagate from the fusion line to a point corresponding to the DRT, depending on when sufficient strain is accumulated. In the present study, however, the use of a very short cooling time (0.05 sec) prior to straining precluded the formation and presence of a HAZ microstructures generated in on-cooling hot ductility test specimens. Actually, during spot-Varestraint testing, straining and cracking occurred when the HAZ was
Figure 23. Comparison of test results for Incolloys 903 and 909 from the spot-Varestraint and hot-ductility tests, using the rate and amount of ductility recovery criterion in interpreting hot-ductility test results.

Figure 24. Comparison of test results for Incolloys 903 and 909 from the spot-Varestraint and hot-ductility tests, using the NST-DRT$_{NST}$ criterion in interpreting hot-ductility test results.
experiencing peak temperatures rather than as this region was cooling. Consequently, cracks would only be expected to propagate to the NDT temperature versus the lower DRT temperature associated with on-cooling behavior. Based on the discussion above, it follows that the maximum crack length obtained from the spot-Varestraint test should correlate most closely with the temperature between the $T_L$ and NDT. As shown in Table 6 and Figure 24, this temperature range was about twice as wide in Incoloy 909 as in 903 (262°C versus 153°C), which correlates well with the greater maximum open + backfilled crack lengths in Incoloy 909 versus Incoloy 903 (1.3 mm versus 0.75 mm), assuming that both materials exhibited the same temperature gradient with the same welding parameters.

It should be noted that the true width of the nil-ductility region in the HAZ should be measured from the spot-Varestraint test only under a saturated strain condition (Table 2). However, the effects of strain on the crack length for these two alloys are almost parallel (Ref. 58) and would not be expected to influence the above correlation.

One of the "conventional" cracking indices for hot-ductility testing, the NST-DRT$_{NST}$, can be related to the period of time during which the ductility at a point in the HAZ experiencing a peak temperature of the NST remains negligible. The longer this time period, the more strain can be accumulated during weld cooling and the higher the likelihood of HAZ liquation cracking. The use of the different peak temperatures in on-cooling hot-ductility tests allows evaluation at different locations in the HAZ. The results of this study confirmed that the rate and amount of ductility recovery decreases with an increase in peak temperature (Figure 17).
Table 6. Pertinent hot ductility and spot-Varestraint test results of Incoloy 903 and 909.

<table>
<thead>
<tr>
<th></th>
<th>Incoloy 903</th>
<th>Incoloy 909</th>
</tr>
</thead>
<tbody>
<tr>
<td>$T_L$ (Ref. 134)</td>
<td>1393°C</td>
<td>1432°C</td>
</tr>
<tr>
<td>NST</td>
<td>1340°C</td>
<td>1269°C</td>
</tr>
<tr>
<td>NDT</td>
<td>1240°C</td>
<td>1150°C</td>
</tr>
<tr>
<td>$DRT_{NST}$</td>
<td>1050°C</td>
<td>1100°C</td>
</tr>
<tr>
<td>$T_L$ - NST</td>
<td>53°C</td>
<td>164°C</td>
</tr>
<tr>
<td>Width of A-B in Figure 22</td>
<td>0.17 mm</td>
<td>0.57 mm</td>
</tr>
<tr>
<td>NST - NDT</td>
<td>100°C</td>
<td>118°C</td>
</tr>
<tr>
<td>$T_L$ - NDT</td>
<td>153°C</td>
<td>282°C</td>
</tr>
<tr>
<td>MCL</td>
<td>0.73 mm</td>
<td>1.30 mm</td>
</tr>
<tr>
<td>NST - $DRT_{NST}$</td>
<td>290°C</td>
<td>168°C</td>
</tr>
</tbody>
</table>

Using this analysis, it is clear that the conventional cracking indices of the hot-ductility and spot-Varestraint tests evaluate two different aspects of weld HAZ liquation cracking susceptibility. The MCL, as characterized by the spot-Varestraint test, represents the width of a crack susceptible region in the HAZ. In contrast, the temperature range between NST and $DRT_{NST}$ as determined by the hot-ductility test depicts the time period for the restoration of HAZ ductility on cooling at a point in the HAZ where experiences a peak temperature equal to the NST.
2.6 Conclusions

The following conclusions can be drawn from the preliminary study.

1. The cracking and loss of high-temperature ductility of Incoloy 903 and 909 result from the presence of thin, low-melting grain boundary liquid films.

2. Incoloy 909 exhibits a lower NST and NDT than Incoloy 903 but recovers ductility more rapidly.

3. The difference in the temperature range of NDT-DRT_{NST} between Incoloy 903 and 909 was attributed to the difference in the liquating phases, with Nb-rich carbides in the former and Laves phases in the latter.

4. The conventional cracking indices for hot ductility and spot-Varestraint tests evaluate two different aspects of HAZ liquation cracking susceptibility. The temperature range between the NST and DRT_{NST} as determined by the hot-ductility test depicts the time period for the restoration of HAZ ductility on cooling at a point in the HAZ which experiences a peak temperature equal to the NST. The maximum crack length as characterized by the spot-Varestraint test represents the width of a crack susceptible region in the HAZ.

5. The maximum crack length (open + backfilled) in the spot-Varestraint test correlates best with the temperature range between the liquidus and nil-ductility temperature in the hot-ductility test.
CHAPTER III

OBJECTIVES

Currently, weldability testing techniques are not capable of unambiguously quantifying HAZ liquation cracking susceptibility. As such, this investigation has been undertaken to further study and extend the methodology developed in the preliminary study. The research program was designed to develop the physical relationship between weldability test results and material properties. The development of this relationship leads to a new methodology which quantifies a material-specific parameter for predicting HAZ liquation cracking susceptibility. The development of this methodology is envisioned to expand the understanding of HAZ liquation cracking mechanisms and the criterion for designing a weldability testing technique.

Specifically, the program objectives were:

1. study a physical relationship among weldability test results, cracking theories and material properties;
2. develop a generic methodology and define a material-specific parameter, based on the hot-ductility, the longitudinal- and spot-Varestraint test results, that quantifies HAZ liquation cracking susceptibility;
3. investigate the correlations among these three techniques;
4. re-evaluate the current methodologies for interpretation of the hot-ductility, spot- and longitudinal-Varelstraint tests in light of the new methodology which is proposed;

5. study and characterize the metallurgical phenomena occurring in the crack susceptible region during weldability testing.
CHAPTER IV
EXPERIMENTAL DESIGN

4.1 Materials

A-286 and Type 310 stainless steels were investigated in this study. A-286 is a precipitation hardened iron-base alloy designed for applications requiring moderately high strength to 700°C and oxidation resistance to about 815°C. The excellent room and elevated temperature strengths exhibited by this alloy result from the presence of Ni, Ti and Al, which promote formation of the γ' Ni₃(Al,Ti) strengthening phase. This alloy is one of the most popular high temperature alloys used for jet engine and gas turbine applications due to its highly desirable combination of properties.

Extensive studies in the past have shown this alloy to be susceptible to both weld solidification and HAZ liquation cracking (Refs. 37-38, 45, 62). Vagi et al (Ref. 45) associated HAZ liquation cracking with the formation of a Fe₃Ti Laves phase. They proposed that grain boundary liquation by the Fe-Fe₃Ti eutectic (1310°C) allows the grains to be separated by thermally-induced strain. Blum et al (Ref. 37) agreed with Vagi et al (Ref. 45), but they also proposed that grain boundary precipitation of a continuous film of TiC and intragranular precipitation of γ' inhibits deformation during cooling, which causes base metal cracks to propagate and relieve the weld stresses. Later works by Brooks (Ref. 62) indicated that the loss of hot ductility above 1150°C
was attributed to the presence of boron. He suggested that the constitutional liqation of borides caused grain boundary liqation and resulted in cracking.

Type 310 is a fully austenitic stainless designed for applications requiring high corrosion resistant. Due to the fully austenitic nature of the resolidified weld metal, this alloy suffers from weld solidification cracking as well as HAZ liqation cracking (Refs. 23-25, 27-28, 33-34), although it is not so susceptible as A-286. Comparing the HAZ liqation cracking susceptibility of this alloy with other austenitic stainless steels, Kujanpää et al (Ref. 23) and Morishige et al (Ref. 24) indicated that Type 310 is more susceptible than Type 316, 347, 321, 304 and 309. A fundamental investigation of the HAZ liqation cracking mechanism of this alloy by Tamura et al (Ref. 34) showed that Cr and Ni in the solute-rich zones in the matrix were swept up and assimilated into the migrating grain boundaries during weld thermal cycling. A subsequent eutectic reaction involving Cr and Ni at the grain boundaries caused a depression in the effective solidus.

The chemical compositions of A-286 and Type 310 stainless steels investigated in this study are listed in Table 7. Both materials were in the form of 6.50 mm (0.25 in) thick plate. The base metal microstructures of both materials are shown in Figure 25. A-286 exhibited a recrystallized austenitic structure with a grain size of ASTM No. 9 and contained a bimodal distribution of titanium carbides and/or carbonitrides. Type 310 was fully austenitic with a grain size of ASTM No. 6. The compositions reported in Table 7 are within the specification limits for each material. Both alloys exhibited relatively low levels of impurities, particularly with respect to sulfur and phosphorus contents.
Table 7. Chemical Compositions of A-286 and Type 310 stainless steels investigated in this study (wt%).

<table>
<thead>
<tr>
<th>Element</th>
<th>A-286</th>
<th>Type 310</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>0.054</td>
<td>0.083</td>
</tr>
<tr>
<td>Mn</td>
<td>0.19</td>
<td>1.42</td>
</tr>
<tr>
<td>P</td>
<td>0.016</td>
<td>0.024</td>
</tr>
<tr>
<td>S</td>
<td>0.009</td>
<td>0.004</td>
</tr>
<tr>
<td>B</td>
<td>0.006</td>
<td>N/A</td>
</tr>
<tr>
<td>Si</td>
<td>0.25</td>
<td>0.46</td>
</tr>
<tr>
<td>Ni</td>
<td>24.22</td>
<td>18.72</td>
</tr>
<tr>
<td>Cr</td>
<td>14.58</td>
<td>24.75</td>
</tr>
<tr>
<td>Mo</td>
<td>1.32</td>
<td>0.27</td>
</tr>
<tr>
<td>Cu</td>
<td>0.10</td>
<td>0.32</td>
</tr>
<tr>
<td>Al</td>
<td>0.34</td>
<td>0.062</td>
</tr>
<tr>
<td>Ti</td>
<td>2.29</td>
<td>0.014</td>
</tr>
<tr>
<td>V</td>
<td>0.22</td>
<td>0.054</td>
</tr>
<tr>
<td>Fe</td>
<td>bal.</td>
<td>bal.</td>
</tr>
</tbody>
</table>

4.2 Hot-Ductility Testing

Hot-ductility tests were performed under an argon atmosphere using a Gleeble™ 1000 system. Standard specimens 6.35 mm in diameter and 100 mm long (0.25 by 4 inches) were machined from the plate. A specimen freespan (jaw spacing) of 25 mm (1 in.) was used throughout the investigation. The NST was determined by heating samples at a rate of 111°C/sec (200°F/sec) under a static load sufficient to overcome the frictional force of the fixture, approximately 10 Kg (22 lb). On-heating tests were conducted by heating samples to the peak temperature in 12.15 seconds.
Figure 25. Base metal microstructures of A-286 and Type 310 stainless steels.
and pulling them to failure at a rate of 5 cm/sec (2 in/sec). The on-heating time to peak temperature was based on the NST test. On-cooling tests were performed after heating to the NST in 12.15 seconds and cooling to the desired temperature at 50°C/sec. This cooling rate was the maximum achievable with the materials and specimen freespan utilized. On-cooling test samples were also pulled to failure at 5 cm/sec (2 in/sec). This rapid stroke rate was selected to minimize the change in specimen temperature during stretching. Sample ductility, in terms of reduction-in-area, was subsequently measured using a vernier caliper. The conditions for hot ductility testing are summarized in Table 8.

Table 8. Conditions for hot-ductility testing of A-286 and Type 310 stainless steels.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Specification</th>
</tr>
</thead>
<tbody>
<tr>
<td>Heating Time</td>
<td>12.15 sec</td>
</tr>
<tr>
<td>Holding Time at peak temp</td>
<td>0.03 sec</td>
</tr>
<tr>
<td>Cooling Rate</td>
<td>50°C/sec (90°F/sec)</td>
</tr>
<tr>
<td>Holding Time at test temp</td>
<td>0.03 sec</td>
</tr>
<tr>
<td>Stroke Rate</td>
<td>5 cm/sec (2 in/sec)</td>
</tr>
<tr>
<td>Sample Freespan</td>
<td>25 mm (1 in)</td>
</tr>
<tr>
<td>Atmosphere</td>
<td>Argon</td>
</tr>
</tbody>
</table>

4.3 Longitudinal-Varestraint Testing

Longitudinal-Varestraint tests were performed only on A-286 stainless steel at strains ranging from 0.5% to 7%. Three samples were tested at each strain level on a subscale Varestraint test unit using specimens with dimensions of 170 X 25 X 6.35 mm (6.5 by 1 by 0.25 in) (Figure 11). Two auxiliary bending bars were located on each side of the sample to ensure that the specimen fully conformed with the surface of the die block. The conditions for longitudinal-Varestraint testing are listed in Table 9. The
rapid strain rate (approximately 13%) was selected to minimize the change in specimen temperature in the HAZ during bending. A 7.5 mm weld bead width was obtained with these welding parameters.

A binocular microscope with 40X magnification was utilized to identify the locations of each crack tip relative to the instantaneous weld pool center along the fusion boundary. These crack tip locations and the fusion line circumscribed a cracked HAZ region. The maximum width of the cracked HAZ was designated as the maximum crack length (MCL), which is the distance between the crack tip of the longest crack and the fusion line projecting in a direction perpendicular to the fusion line.


<table>
<thead>
<tr>
<th>Condition</th>
<th>Specification</th>
</tr>
</thead>
<tbody>
<tr>
<td>Current</td>
<td>190 amps</td>
</tr>
<tr>
<td>Voltage</td>
<td>12 volts</td>
</tr>
<tr>
<td>Travel Speed</td>
<td>2.54 mm/sec</td>
</tr>
<tr>
<td>Augmented strain</td>
<td>0.5% - 7%</td>
</tr>
<tr>
<td>Shielding Gas</td>
<td>Argon, 30 CFH</td>
</tr>
<tr>
<td>Electrode</td>
<td>W-2%ThO₂, 2.4 mm (3/32 in) dia., 60° included angle</td>
</tr>
<tr>
<td>Air Pressure</td>
<td>0.55 MPa (80 psi)</td>
</tr>
<tr>
<td>Strain Rate</td>
<td>approximately 13%</td>
</tr>
</tbody>
</table>

4.4 Spot-Varestraint Testing

Spot-Varestraint tests were performed on a Tigamajig-type test unit using specimens with dimensions of 140 X 25 X 6.35 mm (5.5 by 1 by 0.25 in) (Figure 12). Slots (rather than holes) at both ends of the sample allowed samples to be accurately located and fixtured, and also minimized any axial tensile force during specimen
bending. The conditions for spot-Varestraint testing are listed in Table 9. The criterion for determining the welding parameters was to obtain a cooling rate which was comparable with that used during on-cooling hot-ductility testing. For the parameters selected, the average cooling rate from 1350°C to 1200°C was about 84°C/sec for A-286 and 112°C/sec for Type 310. The average cooling rate at a larger temperature range, from 1350°C to 1000°C, was about 60°C/sec for A-286 and 70°C/sec for Type 310. The rapid strain rate (approximately 13%) was also selected to minimize the change in specimen temperature during bending.

The spot-Varestraint tests were conducted in two stages. The first set of tests, called the on-heating spot-Varestraint tests, was aimed at determining the cracking behavior on weld heating. This was achieved by performing the tests with no cooling time (Table 2). Thus, the HAZ was on-heating when it was straining. The on-heating spot-Varestraint tests were performed from 0.5% to 7% augmented strain with three samples for each strain level. The second set of spot-Varestraint experiments, called the on-cooling spot-Varestraint tests, was designed to investigate the cracking behavior of the material on weld cooling. This was accomplished by performing the tests above the saturated strain level (Table 2) with variable cooling times. The cooling time was varied by adjusting the time between arc extinction and sample bending. Cooling times ranging from 0 to 4.5 seconds were utilized for A-286 and 0 to 0.45 seconds for Type 310.

Because cooling time is a critical parameter in this test, it is precisely calibrated and monitored throughout the test. The actual cooling time is the period between the instant at which the welding current drops to essentially zero and the time at which the die block contacts the specimen (Figure 26). A setup schematically shown in Figure
26 was devised to monitor this actual cooling time. The recorder in the oscilloscope was triggered by a signal from the spot-Varestraint test controller. This signal also turned off the welding arc. Both the signal from the power supply showing the decay of the welding arc, and the signal resulting from the electrical contact between the die block and the specimen were stored in the oscilloscope. The time period between these two signals represented the actual cooling time.

It is also important to point out that for Type 310, the weld pool became irregular for weld times longer than about 20 seconds. In order to produce a uniform circular weld pool, an electro-magnetic coil, as shown in Figure 27, was custom-built and attached to the weld torch as shown in Figure 28. The size of the coil is 25 mm in height with 35 mm I.D. and 50 mm O.D. (1 in. H X 1.375 in I.D. X 2 in. O.D.). There were 280 turns of gage 19 wire in this coil, with 4.0 amps of alternate current driving the coil to generate a strong magnetic field around the weld pool for both A-286 and Type 310 alloys. With this experimental setup, it was found that uniform circular weld pools could be easily obtained. Typical spot welds for these two alloys are shown in Figure 29.

Quantitative cracking data were obtained from as-tested samples by measuring the length of the longest crack between the fusion line and crack tip projecting in a direction perpendicular to the fusion boundary using a binocular microscope under 40X magnification. Thus, the maximum crack length (MCL) reported here represents the distance between the isotherms at the crack tip and the fusion line, rather than the actual length of the observed crack.
Figure 26. Schematic illustration of the setup for cooling time measurement.

Figure 27. The electro-magnetic coil.
Figure 28. Experimental setup for the spot-Varestraint test.

Table 10. Conditions for spot-Varestraint testing of A-286 and Type 310 stainless steels.

<table>
<thead>
<tr>
<th></th>
<th>A-286</th>
<th>Type 310</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weld Time</td>
<td>35 sec</td>
<td>Weld Time : 35 sec</td>
</tr>
<tr>
<td>Current</td>
<td>96 amps</td>
<td>Current : 110 amps</td>
</tr>
<tr>
<td>Voltage</td>
<td>16 volts</td>
<td>Voltage : 12 volts</td>
</tr>
<tr>
<td>Weld Size</td>
<td>11.5 mm</td>
<td>Weld Size : 12 mm</td>
</tr>
<tr>
<td>Shielding Gas</td>
<td>Argon, 20 CFH</td>
<td>Shielding Gas : Argon, 20 CFH</td>
</tr>
<tr>
<td>Air Pressure</td>
<td>0.55 MPa (80 psi)</td>
<td>Air Pressure : 0.55 MPa (80 psi)</td>
</tr>
<tr>
<td>Strain Rate</td>
<td>approximately 13%</td>
<td>Strain Rate : approximately 13%</td>
</tr>
<tr>
<td>Augmented Strain</td>
<td>0.5% - 5%</td>
<td>Augmented Strain : 1% - 7%</td>
</tr>
<tr>
<td>Cooling Time</td>
<td>0 sec - 4.5 sec</td>
<td>Cooling Time : 0 sec - 0.45 sec</td>
</tr>
<tr>
<td>Electrode</td>
<td>W-2%ThO₂, 2.4 mm (3/32 in)</td>
<td>60° included angle</td>
</tr>
</tbody>
</table>

(60° included angle)
Figure 29. Typical spot-Varestraint samples showing a uniform circular spot weld.

(A) A-286

(B) Type 310
4.5 Thermal Cycle Measurement

The thermal cycles experienced during both the spot- and longitudinal-Varestraint testing were measured using fine wire S-type (Pt - Pt 10%Rh, 0.2 mm in diameter) and K-type (Chromel and Alumel, 0.25 mm in diameter) thermocouples. These thermocouples were percussion welded at the bottom of 1.6 mm (1/16 in) holes which were drilled from the bottom of specimens to within about 0.3 mm from the surface on which the weld was placed, as shown in Figure 30. The temperature was recorded using a Labtech Notebook software package (Ref. 135) at a sampling rate of 20 Hz on a IBM 386 compatible personal computer. With this experimental setup, the thermal cycles at different locations from the fusion line into the HAZ were obtained. The peak temperatures and cooling rates were determined and temperature gradients approximated.

Figure 30. Schematic illustration of the setup for temperature measurement.
4.6 Metallurgical Characterization

Representative longitudinal- and spot-Varestraint test specimens were bent flat before sectioning. Samples were metallographically prepared by removing about 0.2 mm thick material from the top surface and polishing through 0.05 micron alumina. Hot-ductility specimens were ground down to the center section of the sample and then polishing through 0.05 micron alumina. Both A-286 and Type 310 stainless steels were etched with a mixed acid solution comprised of equal parts of concentrated nitric, hydrochloric and acetic acids. Microstructures were characterized using an optical microscope at magnifications up to 1000X. Cracks in the spot-Varestraint samples were carefully opened. The crack surface of the spot-Varestraint test samples and the fracture surface of the hot-ductility samples were examined at magnifications up to 2000X using an SEM with an accelerating voltage of 25 kV.
CHAPTER V
DEVELOPMENT OF THE CRACK SUSCEPTIBLE REGION

5.1 introduction

Liquation cracking has been the subject of considerable investigation since the 1940's. Most effort has been aimed at developing more weldable materials by controlling composition and microstructure using weldability testing techniques. However, as reviewed in Chapter 1, the lack of a physical relationship between weldability test results and cracking susceptibility has resulted in difficulties in the interpretation of test results.

In this chapter, a new methodology for HAZ liquation cracking is proposed to elucidate the relationship between weldability test results and HAZ liquation cracking susceptibility and clarify the interpretation of the hot-ductility, the spot- and longitudinal-Varestraint test results. The methodology quantifies a thermal crack susceptible region (CSR) (Table 2) in the HAZ in which liquation cracking may occur. This thermal CSR is a material-specific parameter and represents a true quantification of weldability.

The thermal CSR was theoretically constructed based on the ductility of the material during welding as obtained from the hot-ductility test and the criteria assumed in the development of liquation cracking theories. This theoretical hypothesis was subsequently demonstrated to be inherently material-specific using the longitudinal- and spot-Varestraint tests. The correlation between these three weldability test results
is clarified by determining the ductility of the material within the thermal CSR as obtained with the longitudinal- and spot-Varestraint tests and by directly comparing the magnitude of the thermal CSR determined from the three techniques. This chapter describes the procedure for the development of the thermal CSR and the methodologies of data interpretation of the three tests. The conventional methods of data interpretation of these three tests are also critically reviewed.

5.2 Theoretical Hypothesis of the Thermal Crack Susceptible Region

Metallurgically, HAZ liquation cracking is associated with the occurrence of grain boundary liquation. In the past, most efforts in studying HAZ liquation cracking mechanisms have been directed towards characterizing the evolution of liquid in the HAZ. However, the mere presence of liquid films at grain boundaries is not sufficient to induce a liquation crack. In order to cause cracking, it is essential that the crack susceptible microstructure be subjected to a sufficient tensile strain (or stress). During welding, the tensile strain (or stress) does not generally develop until the weld begins to cool (Ref. 64). As a result, liquation cracking occurs during the solidification of the liquid films. The cooling cycle of the liquid films in the HAZ is similar to the final stages of weld solidification, although the origin of the liquid films and the microstructural boundaries may be different from those developed during weld solidification. Consequently, to a first approximation, the criteria that govern weld solidification cracking can be adopted to explain the solidification of liquid and resultant cracking in the HAZ.

Many mechanisms have been proposed to describe weld solidification cracking. Among these, the shrinkage-brittleness theory (Refs. 85-88), the strain theory (Refs.
the Generalized theory (Refs. 75, 92-96), and the technological strength theory (Refs. 97-98) are the best known, and have been reviewed in chapter 1. Although these theories differ in their approach to cracking mechanics, there is general agreement that cracking occurs in a discrete temperature envelope called the “Brittle Temperature Range (BTR)” (Table 2). Metallurgically, the BTR describes a range between the temperature where liquid is confined within the solidification structure to that where the boundary liquid is partially or completely solidified and the material recovers its ductility. Mechanically, the BTR represents the regime over which the ductility of material is essentially zero and, thus, susceptible to cracking.

As reviewed in chapter 1, the ductility of the material during a weld thermal cycle can be determined using the hot-ductility test. According to this test (Figure 13), a material loses its ductility when the temperature reaches the NDT on heating and recovers ductility at the DRT on cooling. The hot-ductility test results indicate that there are two temperature ranges within which the material exhibits negligible ductility. These are the on-heating nil-ductility temperature range which is the temperature regime between the NDT and T1, and the on-cooling nil-ductility temperature range which is the temperature range between a peak temperature (Tp) and the corresponding DRT. While the on-heating nil-ductility temperature range is constant, the on-cooling nil-ductility temperature range is a function of the peak temperature. Results from the preliminary study (Chapter 2) showed that different peak temperatures resulted in a variation in the corresponding DRT. In general, as the peak temperature is increased above the NDT, the DRT decreases, resulting in a net increase in the on-cooling nil-ductility temperature range. From a physical sense, different peak temperatures for the on-cooling test represent different locations in the HAZ. Thus, a
peak temperature equal to the liquidus \((T_L)\) would represent a point at the weld fusion line, while a peak temperature equal to the NST would represent a point in the HAZ which experiences a thermal cycle with a peak temperature equal to the NST. Further away from the fusion line the peak temperature decreases rapidly and the corresponding DRT increases.

Based on this discussion, it follows that a nil-ductility region (NDR) (Table 2) in the HAZ could be constructed on a temperature field during welding, as illustrated in Figure 31. Assuming that the weld is progressing from right to left at an instant with the weld center located at point O, HDAG represents the solid/liquid interface of the weld pool at a temperature equal to the alloy liquidus. The dashed lines represent the isotherms of different instantaneous temperatures. The fusion line is represented by AA'. Point A represents the instantaneous liquidus temperature. Point B represents the point at which the thermal cycle experiences a peak temperature equal to the NDT. CB represents the isotherm of the NDT. Temperatures along the line AB, the transition between on-heating and on-cooling, would then represent various peak temperatures used to determine on-cooling ductility. For each set of the on-cooling hot-ductility tests, a DRT can be determined that is specific to a given peak temperature along AB. This DRT can then be located at the intersection of the DRT isotherm and a line from the point of the peak temperature parallel to the fusion line toward the cooling direction, because the DRT and the peak temperature are in the same thermal cycle. For example, Point E and E' are in the same thermal cycle with a peak temperature equal to the NST. This thermal cycle achieves its peak temperature at point E and falls off along line EE' reaching the DRT_{NST} isotherm at point E'. Based on the on-cooling hot ductility data, line EE' represents a regime in which the material exhibits
negligible ductility. A similar process can be used to determine the entire DRT curve BA', as shown in Figure 31. Thus, the thermal envelope of the NDR is defined by the NDT isotherm, line CB, and the various DRT’s, line BA', for the on-heating and on-cooling behavior, respectively. This NDR represents an area in which the material exhibits negligible ductility. According to the criteria assumed in the development of the liquation cracking mechanisms, which state that cracking results from the localized loss of ductility, this NDR should be equivalent to an area in the HAZ which is susceptible to liquation cracking. By definition, it is the thermal CSR (Table 2).

Hot ductility test data is not sufficient, however, to absolutely define the whole thermal CSR in Figure 31. Since on-cooling hot ductility tests are very difficult to perform using peak temperatures above the NST, the value for the DRT’s between point E' and A' in Figure 31 cannot be obtained using the conventional hot-ductility test.

Figure 31. Theoretical hypothesis of the thermal crack susceptible region.
5.2.1 Hot-Ductility Test Results

In this study, A-286 and Type 310 stainless steels were investigated to verify the proposed hypothesis. The on-heating and on-cooling hot ductility curves for the two alloys are presented in Figures 32. The NST for both alloys was approximately 1350°C (2462°F). The on-heating ductility of the A-286 decreased rapidly above 1150°C (2102°F) and approached zero at 1200°C (2192°F). On-cooling from the NST, the ductility of this alloy did not recover significantly until it had been cooled to below 1050°C (1922°F). Thus, for A-286, the NDT is 1200°C and the DRT$_{NST}$ is 1050°C. Type 310 stainless steel exhibited similar on-heating ductility behavior with a NDT of 1325°C (2417°F). However, ductility recovery was much more rapid as evidenced by a DRT$_{NST}$ of 1325°C (2417°F). The hot-ductility results (Table 11) indicate that the on-heating nil-ductility temperature ranges (T$_L$ - NDT) for A-286 and Type 310 are 222°C and 61°C, and the BTR at a point which experiences a peak temperature equal to the NST (NST - DRT$_{NST}$) are 300°C and 25°C, respectively.

Table 11. Hot-ductility test results of A-286 and Type 310 stainless steels.

<table>
<thead>
<tr>
<th></th>
<th>A-286</th>
<th>Type 310</th>
</tr>
</thead>
<tbody>
<tr>
<td>T$_L$*</td>
<td>1422°C</td>
<td>1386°C</td>
</tr>
<tr>
<td>NDT</td>
<td>1200°C</td>
<td>1325°C</td>
</tr>
<tr>
<td>NST</td>
<td>1350°C</td>
<td>1350°C</td>
</tr>
<tr>
<td>DRT$_{NST}$</td>
<td>1050°C</td>
<td>1325°C</td>
</tr>
<tr>
<td>T$_L$ - NDT</td>
<td>222°C</td>
<td>61°C</td>
</tr>
<tr>
<td>NST - DRT$_{NST}$</td>
<td>300°C</td>
<td>25°C</td>
</tr>
</tbody>
</table>

* T$_L$ was determined as the peak temperature experienced at the fusion line during spot-Varestraint testing.
Figure 32. Hot-ductility test results of A-286 and Type 310 stainless steels.
5.3 Experimental Verification of the Thermal Crack Susceptible Region

The thermal CSR is material-specific. It is inherently associated with any of the conventional fusion welding processes. This region can be revealed by applying a sufficient tensile strain on a test material, while it is experiencing a fusion welding process. This can be effectively achieved using the spot- and longitudinal-Varestraint tests.

5.3.1 Longitudinal-Varestraint Test

5.3.1.1 Longitudinal-Varestraint Test Results

The longitudinal-Varestraint tests were performed only on A-286 with augmented strain levels ranging from 0.5% to 7%. The locations of each crack tip in the HAZ of the as-tested specimen surface were identified relative to the instantaneous weld center along the fusion boundary. The profile of these crack tips and the fusion line circumscribed a cracked HAZ region. It was found that the size of the cracked HAZ region expanded for the strain level ranging from 0.5% to 3% as shown in Figure 33. Above a saturated strain (Table 2) level of 3%, the size and shape of this cracked HAZ region is essentially constant. This "saturated" cracked HAZ region was previously defined as a crack susceptible region (CSR) (Table 2). As shown in Figure 34, the CSR of A-286 expanded on weld heating and shrank upon cooling. The maximum width of the CSR occurs at the transition between on-heating and on-cooling regions.

The maximum widths of the cracked HAZ region at different strain levels, as defined by the maximum HAZ crack length (MCL), are shown in Figure 35. The MCL increased as augmented strain increased from 0.5% up to 3%, with a threshold strain (Table 2) of about 0.6%. Above 3% strain the MCL was essentially constant at 0.80 mm from the fusion line.
Figure 33. The cracked HAZ region for A-286 at different strain levels obtained from the longitudinal-Varestraint test.

Figure 34. The crack susceptible region (CSR) of A-286 obtained from the longitudinal-Varestraint test.
Figure 35. The maximum crack length at different augmented strain levels for A-286 obtained from the longitudinal-Varestraint test.

5.3.1.2 Development of the Thermal CSR Based on the Longitudinal-Varestraint Test Results

The size of the CSR shown in Figure 34 depends on the material tested and the welding parameters employed during testing. In order to directly compare the CSR among different materials, the welding process variables must be isolated from the material factor. One way to achieve this is to normalize the test results with respect to the thermal conditions the HAZ experienced during testing. This, in fact, provides a material-specific parameter, termed the "thermal" CSR, which uniquely describes the CSR in terms of temperature.

The thermal conditions of the HAZ during testing of A-286 were obtained using implanted thermocouple wires. The procedure for the temperature measurement is
described in chapter 4. A typical weld thermal cycle in the HAZ is shown in Figure 36. The temperature reached a peak in about 4.0 seconds and then decreased rapidly. The average cooling rate from a peak temperature of 1350°C to 1000°C for A-286 was about 67°C/sec. Twenty seven thermal cycles at different locations in the HAZ were obtained. The peak temperatures are shown in Figure 37 with an approximate temperature gradient of 286°C/mm.

In conjunction with the thermal cycle data, the spatial CSR (Figure 34) can be translated into the thermal CSR as illustrated in Figure 38. The weld is progressing from right to left at an instant when the solid/liquid interface intercepts the fusion line at point A. Point A, thus, represents the instantaneous liquidus temperature (T_l). The temperature at the fusion line after the weld pool has passed is represented by AA'.

![Graph](image_url)

**Figure 36.** A typical HAZ thermal cycle for longitudinal-Varestraint testing of A-286.
Figure 37. Peak temperatures at different locations in the HAZ during longitudinal-Varestraint testing.

Figure 38. Schematic illustration of thermal CSR obtained from the longitudinal-Varestraint test.
Point B represents the crack tip temperature at the transition between on-heating (left of AB) and on-cooling regions (right of AB). The temperature at Point B ($T_B$) can be derived by the equation:

\[ T_B = T_L - (MCL \times G_T) \]  

(5.1)

where, $G_T$ is the temperature gradient during longitudinal-Varestraint testing.

Temperatures along the line AB would then represent various peak temperatures ($T_P$) the HAZ experienced during testing. Line A"BA' represents the profile of crack tip temperatures. The crack tip temperature ($T_{tp}$) can be experimentally derived by the following equation using a point along line AB with the same distance from the fusion line as a reference.

\[ T_{tp} = T_P - (CR_T \times CHL_{OC/OC} / V_T) \]  

(5.2)

where, $T_P$ = the peak temperature of a thermal cycle which the point of interest experienced;

$CHL_{OC/OC}$ = the length measured along the fusion boundary between the crack tip and the point which experiences the peak temperature of the same thermal cycle (Table 2). This length is designated as $CHL_{OH}$ for the crack tip located in the on-heating portion, and $CHL_{OC}$ for the on-cooling portion (Figure 34);

$CR_T$ = the average cooling (or heating) rate over $CHL_{OC}$ (or $CHL_{OH}$);

$V_T$ = welding speed during longitudinal-Varestraint testing.

For example, Point E and E' are in the same thermal cycle with a peak temperature equal to $T_E$, as they are equidistant from the fusion line. This thermal cycle achieves its peak temperature at point E and falls off along line EE'. Cracking is observed over line EE'. Thus, $CHL_{OC}$ is represented by the length EE'. The time required ($t_{EE}$) for the
temperature to fall from point E to point E' is the distance between these two points (EE') divided by the welding speed during testing (V_T), t_{EE'} = EE' / V_T. The temperature in the thermal cycle reaches point E' (T_{E'}) at a time period t_{EE'}, after it has achieved the peak. Or, mathematically, T_{E'} = T_E - (CR_T \times EE' / V). CR_T is the average cooling rate from the peak temperature T_E over a time period of t_{EE'}. Pertinent thermal CSR data for A-286 obtained from the longitudinal-Varestraint test are summarized in Table 12.

Table 12. Pertinent longitudinal-Varestraint test results of A-286.

<table>
<thead>
<tr>
<th></th>
<th>A-286</th>
</tr>
</thead>
<tbody>
<tr>
<td>MCL (at transition between on-heating and on-cooling)</td>
<td>0.80 mm</td>
</tr>
<tr>
<td>CHL_{OC} at NST</td>
<td>11.5 mm</td>
</tr>
<tr>
<td>CHL_{OC} at T_L</td>
<td>14.5 mm</td>
</tr>
<tr>
<td>G_T</td>
<td>286°C/mm</td>
</tr>
<tr>
<td>Average CR_T at NST from NST over CHL_{OC} at NST</td>
<td>66°C/sec</td>
</tr>
<tr>
<td>Average CR_T at T_L from T_L over CHL_{OC} at T_L</td>
<td>69°C/sec</td>
</tr>
</tbody>
</table>

5.3.2 Spot-Varestraint Test

5.3.2.1 Spot-Varestraint Test Results

The spot-Varestraint tests were conducted in two stages. The first set of tests, called the on-heating spot-Varestraint tests, was aimed at determining the cracking behavior of a material during weld heating. This was achieved by performing the tests with a cooling time equal to 0 second (Table 2) (i.e. specimens were deformed immediately after arc extinction). The on-heating spot-Varestraint tests were performed from 0.5% to 7% augmented strain with three samples for each strain level. The
second set of spot-Varestraint experiments, called the on-cooling spot-Varestraint tests, was designed to investigate the cracking behavior of the material on weld cooling. This was accomplished by performing the tests above the saturated strain (Table 2) level with variable cooling times. The cooling time was varied by adjusting the time period between arc extinction and sample bending.

The on-heating spot-Varestraint test results for both A-286 and Type 310 alloys are shown in Figure 39. Because there was no time for the HAZ to cool before cracking occurred, the maximum HAZ crack length (MCL) represented the width of the cracked HAZ under a specific strain level when the HAZ was at peak temperature. For Type 310 stainless steel, above a threshold strain (Table 2) level of 1%, the MCL increased as augmented strain increased up to 3%. Above 3% strain, designated the saturated strain (Table 2), the MCL was essentially constant at 0.39 mm from the fusion line.

A threshold strain could not be determined for A-286, since significant cracking was observed at the lowest level of augmented strain (0.5%). MCL increased slightly between 0.5 and 3.0% and was essentially constant above 3.0%. The maximum width of the cracked HAZ for A-286, as defined by the MCL in the saturated strain region, was 1.93 mm.

Note that above a saturated strain level, the width of the cracked HAZ during weld heating is essentially constant. As defined previously, the width of the cracked HAZ at saturated strain level represents the extent of the CSR during heating. The on-cooling spot-Varestraint tests were performed with a saturated strain of 5% for both alloys to characterize the extent of the CSR during weld cooling. Results from these tests show the on-cooling portion of the CSR. As shown in Figure 40, both alloys...
Figure 39. On-heating spot-Varestraint test results of A-286 and Type 310 showing the width of the cracked HAZ at different augmented strain levels.

Figure 40. On-cooling spot-Varestraint test results of A-286 and Type 310 showing the on-cooling portion of the CSR.
behaved in a similar manner, in that the width of the CSR shrank as cooling time increased. For A-286, it disappeared after a cooling time of 4.13 seconds, while for Type 310 this region persisted for only 0.41 seconds.

5.3.2.2 Development of the Thermal CSR Based on the Spot-Varestraint Test

Results

The methodologies utilized to develop the thermal CSR based on the spot-Varestraint test results are basically the same as that for the longitudinal-Varestraint test. The thermal CSR can be developed by combining the spatial CSR (Figure 40) and the thermal conditions experienced in the HAZ during testing. The thermal conditions in the HAZ during testing were determined using the implanted thermocouple technique as described in chapter 4.

A typical HAZ thermal cycle for spot-Varestraint test is shown in Figure 41. The heating rate is rapid in the early stage, and slows down when the temperature approaches the peak. After arc extinction, the temperature decreases rapidly and the cooling rate slows down as the temperature decreases. In comparison with the HAZ thermal cycle for a continuous weld (Figure 36), the spot weld exhibits a slower heating rate and a faster cooling rate. At a given point in the HAZ, the difference in heating rate and cooling rate is attributed to preheating before a weld pool reaches and postheating after the weld pool has passed in a continuous weld. It is difficult to obtain similar heating and cooling rates for both test conditions by adjusting welding parameters. Twenty-two and thirty different HAZ thermal cycles were obtained for Type 310 and A-286, respectively. The peak temperatures are shown in Figure 42 with an approximate gradient of 101°C/mm for A-286 and 156°C/mm for Type 310.
Figure 41. A typical HAZ thermal cycle during spot-Varestraint testing.

The peak temperature at the fusion line was determined to be 1422°C for A-286 and 1386°C for Type 310. These temperatures are believed to be the liquidus \( T_L \). These data are very close to the published data (1400°C - 1450°C for Type 310 and 1370°C - 1430°C for A-286) (Ref. 136).

In conjunction with the thermal cycle data in the HAZ, Figure 40 can be translated to represent the on-cooling portion of the thermal CSR, as schematically shown in Figure 43. Point A represents the instantaneous liquidus temperature \( T_L \). Point B represents the crack tip temperature determined with no cooling time \( t_c = 0 \) sec. The temperature at Point B \( T_B \) can be derived following the equation:

\[
T_B = T_L - (MCL \times G_r)
\]

(5.3)

where, \( G_r \) is the temperature gradient during spot-Varestraint testing, and MCL is the maximum HAZ crack length performed with no cooling time. Temperature along the
Figure 42. Peak temperatures at different locations in the HAZ for A-286 and Type 310 during spot-Varestraint testing.
line AB represents different peak temperatures achieved in the HAZ. Since the cooling time is equivalent to the period after weld pool has passed in a continuous weld. Line AA' would then represents various temperatures at the fusion line after arc extinction. Point A' locates at the greatest cooling time above which cracks does not propagate across the fusion line into the HAZ. The temperature at the fusion line \( T_{FL} \) after arc extinction can be derived from the cooling time \( t_c \) following the equation:

\[
T_{FL} = T_L - (CR_T \times t_c)
\] (5.4)

where, \( CR_T \) is the average cooling rate at the fusion line over a period of \( t_c \). Line BA' represents crack tip temperatures at various cooling times. These crack tip temperatures \( T_{tip} \) can be derived using a point along line AB with the same distance from the fusion line as a reference and following the equation:

\[
T_{tip} = T_P - (t_c \times CR_T)
\] (5.5)

where,
- \( T_P \) = the peak temperature of a HAZ thermal cycle which the point of interest experienced during spot-Varestraint testing;
- \( t_c \) = the cooling time over which cracking is observed at the point of interest;
- \( CR_T \) = the average cooling rate of a thermal cycle with a peak temperature of \( T_P \) from the peak temperature over a time period of \( t_c \).

For example, Point E and E' are in the same thermal cycle with a peak temperature equal to \( T_E \), as they exhibit equidistant from the fusion line. Cracking at Point E is observed over a cooling time period of \( t_E \). Thus, the temperature in the thermal cycle reaches Point E' \( (T_E) \) at a time period \( t_E \) after it has achieved the peak.

Mathematically, \( T_{E'} = T_E - (CR_T \times t_E) \), where, \( CR_T \) is the average cooling rate from \( T_E \) over
a period of $t_c$. Pertinent thermal CSR data obtained from the spot-Varestraint test are summarized in Table 13.

![Figure 43](image)

Figure 43. Schematic illustration of the on-cooling portion of the thermal CSR obtained from the spot-Varestraint test.

<table>
<thead>
<tr>
<th></th>
<th>A-286</th>
<th>Type 310</th>
</tr>
</thead>
<tbody>
<tr>
<td>MCL ($t_c = 0$)</td>
<td>1.93 mm</td>
<td>0.39 mm</td>
</tr>
<tr>
<td>$t_c$ at NST</td>
<td>4.10 sec</td>
<td>0.20 sec</td>
</tr>
<tr>
<td>$t_c$ at $T_L$</td>
<td>4.13 sec</td>
<td>0.41 sec</td>
</tr>
<tr>
<td>$G_T$</td>
<td>101°C/mm</td>
<td>156°C/mm</td>
</tr>
<tr>
<td>Average $CR_T$ at NST from NST over $t_c$ at NST, ($E$ - $E'$ in Fig. 43)</td>
<td>68°C/sec</td>
<td>140°C/sec</td>
</tr>
<tr>
<td>Average $CR_T$ at $T_L$ from $T_L$ over $t_c$ at $T_L$, ($A$ - $A'$ in Fig. 43)</td>
<td>94°C/sec</td>
<td>220°C/sec</td>
</tr>
</tbody>
</table>
5.3.3 Discussion

There are two approaches to experimentally verify the theoretical hypothesis. One method is to characterize the ductility of the material in the thermal CSR obtained from the longitudinal- and spot-Varestraint tests. The other approach is to directly compare the extent of the thermal CSR obtained from the three methods.

5.3.3.1 Characterization of the Ductility of the Material in the Thermal CSR Obtained from the Longitudinal- and Spot-Varestraint Tests

One of the important criteria assumed in the development of liqutation theories is that cracking results from the localized loss of ductility. In order to prevent liqutation cracks from propagating, the material must exhibit enough ductility to accommodate the thermally-induced and/or externally applied strain. Based on this criterion, to a first approximation, it would be appropriate to assume that the material at the crack tip exhibits localized ductility with the magnitude equal to the applied augmented strain during both the longitudinal- and spot-Varestraint testing. Thus, in conjunction with the thermal cycle data, Figures 35 and 39 can be translated to physically show the localized ductility of the material at different temperatures during weld heating as shown in Figure 44 for the longitudinal-Varestraint test and Figure 45 for the spot-Varestraint test. In these figures, the localized ductility is analogous to the augmented strain. The temperature of the material represents the temperature at the crack tip \( T_{tip} \) during testing, which is translated from the MCL following the equation,

\[
T_{tip} = T_L - (MCL \times G_T)
\]

Thus, results from the on-heating portion of the longitudinal-Varestraint test (Figure 44) showed that A-286 exhibited substantial localized ductility (greater than 7% in this
case) for temperatures below 1194°C. As the temperature increased beyond 1194°C, the localized ductility dropped dramatically to 3%, and further decreased to 0.5% at T_L. In contrast, results from the on-heating spot-Varestraint tests (Figure 45) indicated that A-286 exhibited considerable localized ductility (greater than 5% in this case) for temperatures below 1227°C. The localized ductility dropped substantially to 3% as the temperature was beyond 1227°C. With a further increase in the temperature, the localized ductility decreased gradually to 0.5% at a temperature of 1250°C, and A-286 exhibited essentially no ductility (less than 0.5%) for the temperature range between 1250°C and T_L. On the other hand, Type 310 (Figure 45) exhibited greater than 7% localized ductility for temperatures below 1325°C. The localized ductility dropped rapidly to 3% as the temperature exceeded 1325°C, and further decreased to approximately 1% at T_L.

Note that there is a transition temperature, 1194°C or 1227°C for A-286 and 1325°C for Type 310, above which the localized ductility of the material dramatically drops to essentially zero. As noted earlier, the temperature range between T_L and the transition temperature defines the extent of the on-heating thermal CSR. In order to determine this transition temperature, the saturated augmented strain has to be employed during both the longitudinal- and spot-Varestraint testing. Obviously, the saturated augmented strain is one of the important parameters for the two tests. The importance of this parameter, however, has been neglected by most investigators since these two tests were introduced.

It is interesting that the transition temperatures for A-286 defined with the longitudinal- and spot-Varestraint tests are essentially the same, if reasonable experimental error is taken into account. However, the localized ductility in the on-
Figure 44. The localized ductility in the on-heating thermal CSR of A-286 obtained from the longitudinal-Varestraint test.

Figure 45. The localized ductility in the on-heating thermal CSR of A-286 and Type 310 obtained from the spot-Varestraint test.
heating thermal CSR is inconsistent. This discrepancy can be rationalized by considering the different strain distribution on the specimen surface during testing.

As shown in Figure 46, the HAZ liquation cracks are parallel to each other in the longitudinal-Varestraint test. Because the applied augmented strain is theoretically calculated based on the elongation of the top surface of the specimen, the opening of the cracks would accommodate some amount of the applied strain. The as-tested specimen surface clearly showed that the opening of the crack is wider at the trailing edge of the weld pool and becomes narrower toward the welding direction (left part of Figure 46). It implies that the strain distribution is uneven with a greater strain at the trailing edge of the weld pool. Thus, the actual augmented strain for the crack located at the transition between on-heating and on-cooling regions would be lower than the calculated strain. In contrast, the HAZ liquation cracks are radially oriented relative to the weld center in the spot-Varestraint test, as shown in Figure 47. The crack with a maximum length is always the widest and located perpendicular to the longitudinal direction of the specimen, indicating that it received the greatest amount of localized strain. The actual augmented strain is, thus, greater than the calculated strain. Based on this discussion, it follows that the difference in the localized strain in the longitudinal- and spot-Varestraint results in the difference in the experimentally determined localized ductility in the on-heating thermal CSR from these two tests, with a greater localized ductility in the former.

It was observed that both the longitudinal- and spot-Varestraint test specimens did not conform perfectly to the surface of the die block. Specimens in both tests tended to kink slightly at the highest augmented strain level (7%). The kinking of the specimen may thus result in a higher applied strain than that calculated. However, it
did not affect the ductility transition temperature (Figures 44 and 45), since it was determined at a saturated strain level of 3% for both alloys. As shown in Figures 44 and 45, even with a higher level of augmented strain (>3%), the transition temperature does not change and, thus, specimen kinking did not influence the results.

The localized ductility in the on-cooling portion of the thermal CSR determined with the longitudinal- and spot-Varestraint was not evaluated in this study. However it is definitely less than 5% for the two alloys investigated, since 5% augmented strain was employed for the on-cooling spot-Varestraint tests, and the CSR was obtained for A-286 at 3% augmented strain from the longitudinal-Varestraint test. Thus, the thermal CSR shown in Figures 38 and 43 represents a regime within which the material (both A-286 and Type 310) exhibits less than 5% ductility (or perhaps 3%, since 3% is the saturated strain for both alloys in both tests).

Considering the complexity of the strain field during the longitudinal- and spot-Varestraint testing and different thermal and mechanical conditions for hot-ductility testing, it would be appropriate to equate this very low-ductility thermal CSR to the nil-ductility region (NDR) as determined from the hot-ductility test. The equivalence between the CSR and the NDR provides experimental evidence for the criterion assumed in the development of liquation cracking theories, which states that liquation cracking results from the localized loss of ductility.

5.3.3.2 Correlation of Test Results Obtained from the Three Methods

Two basic elements with metallurgical significance can be visualized in the development of the thermal CSR. These are the extent of the thermal CSR during weld heating and the BTR on cooling. The correlation between the theoretical
Figure 46. A micrograph showing the HAZ liquation cracks in A-286 on a metallographically prepared specimen surface from the longitudinal-Varestraint test (specimen was bent flat during the sample preparation).

Figure 47. A micrograph showing the HAZ liquation cracks in A-286 on a metallographically prepared specimen surface from the spot-Varestraint test (specimen was bent flat during the sample preparation).
hypothesis and the experimental results can be demonstrated by comparing the magnitude of these two elements. As listed in Table 14, very good correlation was obtained for A-286 and Type 310 stainless steels from the three tests.

The development of this correlation also clarifies the interpretation of results from the three tests. Physically, the temperature range between the $T_L$ and NDT represents the extent of the thermal CSR during weld heating. This temperature range correlates with the temperature range over which the MCL occurs at the transition between on-heating and on-cooling regions in the longitudinal-Varestraint test performed with saturated strain. It also correlates with the temperature range over which the MCL occurs in the spot-Varestraint test performed with saturated strain at no cooling time. The temperature range between a peak temperature and the corresponding DRT is the temperature range in a cooling cycle within which the material exhibits negligible ductility, previously defined as the BTR. This temperature range in turn correlates with the cooling time ($t_c$) over which cracking is observed in the

Table 14. Correlations among the hot-ductility, the spot- and longitudinal-Varestraint test results.

<table>
<thead>
<tr>
<th></th>
<th>Hot-Ductility Test</th>
<th>Spot-Varestraint Test</th>
<th>Longitudinal-Varestraint Test</th>
</tr>
</thead>
<tbody>
<tr>
<td>extent of CSR on heating, A-286</td>
<td>$T_L$ - NDT = 1422-1200 = 222°C</td>
<td>$MCL \times G_T = 1.93 \times 10^1 = 195°C$</td>
<td>$MCL \times G_T = 0.80 \times 286 = 229°C$</td>
</tr>
<tr>
<td>extent of CSR on heating, Type 310</td>
<td>$T_L$ - NDT = 1386-1325 = 61°C</td>
<td>$MCL \times G_T = 0.39 \times 156 = 60°C$</td>
<td>Not Determined</td>
</tr>
<tr>
<td>BTR at NST A-286</td>
<td>NST - DRT$_{NST}$ = 1350-1050 = 300°C</td>
<td>$t_c \times CR_T = 4.10 \times 68 = 279°C$</td>
<td>CR$<em>T$ X CHL$</em>{OC}$ / $V_T$ = 66X11.5/2.54 = 299°C</td>
</tr>
<tr>
<td>BTR at NST Type 310</td>
<td>NST - DRT$_{NST}$ = 1350-1325 = 25°C</td>
<td>$t_c \times CR_T = 0.20 \times 140 = 28°C$</td>
<td>Not Determined</td>
</tr>
</tbody>
</table>
spot-Varestraint test, and the on-cooling portion of the cracked HAZ length (CHL_{oc}) in the longitudinal-Varestraint test. Despite the difference in testing techniques, the correlation obtained for both A-286 and Type 310 is excellent, reinforcing the complementary nature of these techniques for predicting HAZ liquation cracking. This good correlation also provides evidence that the thermal CSR is a material-specific parameter, and it is a true quantification of HAZ liquation cracking susceptibility.

5.4 Application of the Thermal Crack Susceptible Region to Actual Welding Conditions

Although the thermal CSR is developed using weldability testing techniques, this thermal CSR is a material-specific parameter and can be applied to any of the conventional fusion welding processes. The application of the thermal CSR in a continuous weld can be clearly demonstrated with the longitudinal-Varestraint test as illustrated in Figures 34 and 46. Table 15 summarizes the procedures for determining the thermal CSR using the hot-ductility, the longitudinal- and spot-Varestraint tests. It is of importance to note that the shape and size of the thermal CSR reflects the material factor only and is independent of any process or restraint factors. This thermal CSR allows comparison of cracking susceptibility among different materials. Using this methodology, the thermal CSR for A-286 and Type 310 have been determined, as shown in Figure 48. It is clear that A-286 exhibited a larger BTR and correspondingly wider thermal CSR than Type 310, suggesting that A-286 would be more sensitive to HAZ liquation cracking than Type 310 under the same process and restraint conditions.
In applying this material-specific envelope to actual welding conditions, the spatial extent of the CSR travelling with the weld pool can be determined as a function of the welding parameters, which determine the thermal conditions in the HAZ. Table 16 provides a transformation matrix for determining the magnitude of the thermal and spatial CSR. As a result, the thermal envelope determined via hot ductility, spot- and longitudinal-Varestraint weldability tests can be directly applied to actual welding situations if the thermal gradient \((G_w)\) and cooling rate \((CR_w)\) can be measured or derived empirically. The contribution of the welding process/parameter factor in cracking susceptibility can also be visualized from the construction of the spatial CSR.

Table 15. Development of the thermal CSR.

<table>
<thead>
<tr>
<th>Point Line</th>
<th>Theoretical Quantity</th>
<th>Hot Ductility Test</th>
<th>Spot-Varestraint Test</th>
<th>Longitudinal-Varestraint Test</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>(T_L)</td>
<td>-</td>
<td>(T_p) at fusion line</td>
<td>(T_p) at fusion line</td>
</tr>
<tr>
<td>CB</td>
<td>NDT</td>
<td>NDT</td>
<td>(T_L) - ((MCL \times G_T))</td>
<td>(T_L) - ((MCL \times G_T))</td>
</tr>
<tr>
<td>E</td>
<td>NST</td>
<td>NST</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>E'</td>
<td>(DRT_{NST})</td>
<td>(DRT_{NST})</td>
<td>(NST - (t_c,\text{NST} \times CR_T))</td>
<td>(T_L - CR_T \times CHL_{OC,NST} / V_T)</td>
</tr>
<tr>
<td>BE'</td>
<td>(DRT's)</td>
<td>(DRT's)</td>
<td>(T_{ip}) as function of (t_c)</td>
<td>(T_{ip}) as function of (CHL_{OC})</td>
</tr>
<tr>
<td>E'A'</td>
<td>(DRT's)</td>
<td>-</td>
<td>(T_{ip}) as function of (t_c)</td>
<td>(T_{ip}) as function of (CHL_{OC})</td>
</tr>
<tr>
<td>A'</td>
<td>(DRT_{TL})</td>
<td>-</td>
<td>(T_L - (t_c,\text{FL} \times CR_T))</td>
<td>(T_L - CR_T \times CHL_{OC,TL} / V_T)</td>
</tr>
</tbody>
</table>
Table 16. Transformation matrix for the derivation of the thermal and spatial CSR.

A. The Hot-Ductility Test

<table>
<thead>
<tr>
<th>Width of CSR</th>
<th>Test Result</th>
<th>Thermal CSR</th>
<th>Spatial CSR</th>
</tr>
</thead>
<tbody>
<tr>
<td>Width of CSR</td>
<td>$T_L$ - NDT</td>
<td>$T_L$ - NDT</td>
<td>$(T_L - NDT) / G_w$</td>
</tr>
<tr>
<td>BTR</td>
<td>$T_p$ - DRT</td>
<td>$T_p$ - DRT</td>
<td>$(T_p - DRT) \times V_w / C_{R_w}$</td>
</tr>
</tbody>
</table>

B. The Longitudinal-Varestraint Test

<table>
<thead>
<tr>
<th>Width of CSR</th>
<th>Test Result</th>
<th>Thermal CSR</th>
<th>Spatial CSR</th>
</tr>
</thead>
<tbody>
<tr>
<td>Width of CSR</td>
<td>MCL</td>
<td>MCL $\times G_T$</td>
<td>MCL $\times G_T / G_w$</td>
</tr>
<tr>
<td>BTR</td>
<td>CHL_{OC}</td>
<td>CHL_{OC} $\times CR_T / V_T$</td>
<td>CHL_{OC} $\times CR_T / V_T \times V_w / C_{R_w}$</td>
</tr>
</tbody>
</table>

C. The Spot-Varestraint Test

<table>
<thead>
<tr>
<th>Width of CSR</th>
<th>Test Result</th>
<th>Thermal CSR</th>
<th>Spatial CSR</th>
</tr>
</thead>
<tbody>
<tr>
<td>Width of CSR</td>
<td>MCL</td>
<td>MCL $\times G_T$</td>
<td>MCL $\times G_T / G_w$</td>
</tr>
<tr>
<td>BTR</td>
<td>$t_c$</td>
<td>$t_c \times CR_T$</td>
<td>$t_c \times CR_T \times V_w / C_{R_w}$</td>
</tr>
</tbody>
</table>

5.5 Discussion on Conventional Methodologies

The development of this methodology has provided a means to link together weldability test results, cracking theories and material properties, and elucidated the interpretation of the hot-ductility, the spot- and longitudinal-Varestraint test results. The utility of conventional methodologies can also be evaluated based on this approach as discussed below.
Figure 48. Thermal CSR of A-286 and Type 310 stainless steels.
5.5.1 Hot-Ductility Test

The construction of the NDR using the hot-ductility test was first demonstrated by Duvall et al (Refs. 8-9). Unfortunately, they could not develop a relationship between the NDR and HAZ liquation cracking susceptibility. Based on a rough calculation of the weld size, they concluded that HAZ liquation cracks propagated outside of the NDR and that the amount and rate of ductility recovery were the most significant factors governing liquation cracking susceptibility. Despite the high levels of strain used in this investigation during spot- and longitudinal-Varestraint testing, cracks were never observed to propagate outside of the NDR. Thus, the importance of ductility recovery below the DRT, as suggested by Duvall et al, appears to be negligible based on the results from both A-286 and Type 310 stainless steel.

The NST-DRT temperature range has been utilized extensively as a quantitative cracking index (Refs. 36, 39, 44, 123). This index is restrictive in that it only represents the BTR at a specific location in the HAZ. Since the highest peak temperature which can be easily achieved in the hot-ductility test is in the vicinity of the NST, this approach only provides information for a point in the HAZ that is removed from the fusion boundary. When considering that HAZ liquation cracking is normally observed in the region immediately adjacent to the fusion boundary, this cracking index may not be representative of actual behavior or predictive of liquation cracking susceptibility. This is particularly true if the difference between $T_L$ and the NST is large. This problem is further exacerbated when peak temperatures below the NST are used to develop on-cooling hot ductility behavior, as is the case in many of the published reports which use hot-ductility data to predict HAZ liquation cracking susceptibility.
The hot strength criterion proposed in 1960's (Refs. 13, 123-124) is commonly utilized to rationalize the inconsistency between the hot-ductility results and cracking observed in the field experience. However, considering the fact that all materials exhibit extremely low strength in the temperature range in which cracking occurs, the elastic strain induced by this hot strength would be negligible. In addition, most alloys exhibit a critical level of threshold strain for cracking in the spot- and longitudinal-Varestraint testing (for example, 1% for Type 310 in Figure 39). The observation of significant threshold strain to cause cracking indicates that the materials have the ability to resist cracking even under a substantial applied stress. This suggests, therefore, that it is the hot ductility rather than hot strength which dominates equation cracking.

5.5.2 Spot-Varestraint Test

The conventional cracking indices for the spot-Varestraint test, MCL or TCL at an arbitrary augmented strain level and an arbitrary cooling time, are even more complicated from a metallurgical standpoint. The complexity results from both inadequate test parameters and improper interpretation. First, the crack length does not provide a material-specific quantification; rather, it represents a specific material/process/restraint combination. Second, the saturated strain level is material dependent. An arbitrary augmented strain below the saturated strain level may only reveal part of the CSR. Third, the cooling time has proved to be the parameter which governs the HAZ temperature during testing and, thus, the on-cooling behavior of material. Tests performed with an arbitrary, fixed cooling time will only reveal a specific width of the CSR as shown in Figure 40. This specific width of the CSR would
neither correlate with the on-cooling behavior obtained from the hot-ductility test nor reflect the cracking behavior of the material during actual welding. Finally, the similarity of terminologies, such as MCL and TCL, utilized in the spot-, longitudinal- Varestraint and Transvarestraint tests has led to improper interpretation of test results. Physically, the MCL in all these three tests characterizes a region over which cracking can occur. However, the thermal history in this cracked region along the crack length in longitudinal- and spot-Varestraint test is different. Yet in the Transvarestraint, it is the same. This difference has resulted in the discrepancy between width of the CSR in the spot-Varestraint test and the BTR in the Transvarestraint test (Ref. 102).

5.5.3 Longitudinal-Varestraint Test

The discussion on the conventional cracking indices for the longitudinal-Varestraint test, TCL or MCL, is basically the same as in the spot-Varestraint test, except that TCL is the cumulative length of all cracks with the same strain level at different thermal conditions in the longitudinal-Varestraint test, but it is the cumulative length of all cracks with the same thermal conditions at different strain levels in the spot-Varestraint test. Both indices represent results for a specific material/process/restraint combination. The magnitude of these indices would vary if different welding parameters or augmented strain are employed. The CHL criterion, proposed by Lundin et al (Ref. 40), can be normalized with the thermal conditions experienced during testing to obtain a BTR, if the tests are performed with a saturated strain and only the on-cooling portion of the CHL (CHL_∞) is considered, as illustrated in Figure 34 and Table 14. However, it is very unlikely that the CHL_∞ at a specific location in the HAZ can be precisely determined without the development of the entire CSR.
5.6 Closure

Although the approach described here has rationalized the debate in the interpretation of the hot-ductility, the longitudinal- and spot-Varestraint test results, and has great utility for providing a unified methodology for defining a material-specific temperature regime within which HAZ liquation cracking occurs, some inherent problems still exist. First, the hot-ductility data obtained in this study does not provide a comprehensive measurement of the ductility within the BTR. Although the spot- and longitudinal-Varestraint tests have shown their ability to characterize the ductility within the CSR, using the procedure described in this report, these procedures may not be appropriate from a practical point of view because of the large number of samples required (> 30) for the spot-Varestraint test and complicated HAZ thermal cycle measurement required for both tests. However, the ductility within the BTR has proved to be a critical factor in terms of predicting cracking susceptibility (Refs. 97-98).

Second, the strain rates in these three tests are very rapid compared with the rate of accumulation of strain in actual fabrication. The CSR may be strain rate dependent, such that significant variation in the rate of strain application, either among various welding processes or relative to weldability testing, may influence the size and/or shape of the CSR. In addition, the effect of liquid "healing" was not accounted for the methodology proposed in this report. This effect may play a very important role in the actual fabrication conditions. These issues must be considered and quantified to allow the broader application of the methodology proposed for predicting HAZ liquation cracking susceptibility using weldability data.
5.7 Conclusions

1. A new methodology has been developed for quantifying HAZ liquation cracking susceptibility. This methodology quantifies a thermal crack susceptible region (CSR) in the HAZ. This is the region in which HAZ liquation cracking may occur. It is a material-specific parameter and represents a true quantification of HAZ liquation cracking susceptibility. This region can be determined using any of the hot-ductility, the spot- or longitudinal-Varestraint tests.

2. The thermal crack susceptible region can be described by two basic elements: the extent of the crack susceptible region during weld heating and the brittle temperature range on cooling. The temperature range between the liquidus ($T_l$) and nil-ductility temperature (NDT) in the hot-ductility test represents the extent of the on-heating thermal crack susceptible region. This temperature range correlates with the temperature range over which the maximum crack length occurs at the transition between on-heating and on-cooling regions in the longitudinal-Varestraint test performed with the saturated strain. It also correlates with the temperature range over which the maximum crack length occurs in the spot-Varestraint test performed with saturated strain at no cooling time after arc extinction.

3. The temperature range between the peak temperature ($T_p$) and the corresponding ductility recovery temperature (DRT) as obtained from the hot-ductility test represents the brittle temperature range (BTR). This temperature range in turn correlates with the cooling time over which cracking is observed in the spot-Varestraint test, and the on-cooling portion of the cracked HAZ length in the longitudinal-Varestraint test.
4. The localized ductility within the thermal crack susceptible region is essentially zero. These results provide experimental evidence supporting the criterion assumed in the development of liquation cracking theories, which states that liquation cracking results from the localized loss of material ductility.

5. The importance of the saturated augmented strain in the both the longitudinal- and spot-Varestraint tests was evaluated. It was found that the saturated strain defines the transition temperature above which the ductility of the material drops to essentially zero.

6. The effects of the cooling time on test results in the spot-Varestraint test have been developed. It was found that the crack susceptible region on weld cooling could be characterized by varying the cooling times during testing.

7. For alloy A-286, the extent of the on-heating thermal crack susceptible region is 222°C. The brittle temperature range at the location of the nil-strength temperature is 300°C. The brittle temperature range at the fusion line is 387°C.

8. For Type 310, the extent of the on-heating thermal crack susceptible region is 61°C. The brittle temperature range at the location of the nil-strength temperature is 25°C. The brittle temperature range at the fusion line is 90°C.
CHAPTER VI
METALLURGICAL CHARACTERIZATION OF THE CRACK SUSCEPTIBLE REGION

6.1 Introduction

In chapter 5, a methodology that characterizes a crack susceptible region (CSR) for quantifying HAZ liquation cracking susceptibility was developed. The thermal CSR is a material-specific parameter and can be applied to any of the conventional fusion welding processes. Since the CSR defines the region in which HAZ liquation cracking may occur, the metallurgical phenomena occurring in this region also provides important insight as to HAZ liquation cracking mechanisms.

In this chapter, results of metallographic and fractographic analyses on the hot-ductility, longitudinal- and spot-Varestraint test specimens of the four alloys investigated in this study: Incoloy 903, Incoloy 909, A-286 and Type 310 are presented. The metallurgical phenomena occurring in the CSR are discussed based on these metallographic and fractographic results. This metallurgical characterization provides valuable information in understanding the evolution of liquid films and resultant cracking in the HAZ.

6.2 Literature Review on Metallurgical Phenomena Related to Liquation Cracking

The fracture mechanisms of a material at elevated temperature have been subjects of extensive investigation over the past 50 years (Refs. 9, 15, 123-124, 137-
It is known that the ductility falls off abruptly as the temperature approaches the bulk melting point. Sherby (Ref. 137) proposed that this fall-off is due to local melting at impurity regions, usually at grain boundaries. Williams and Singer (Ref. 138) ascribed the fracture mechanism to the presence of a grain boundary liquid film. Williams (Ref. 123) speculated that under high tensile strain rate conditions, the NDT is reached at the temperature where melting first occurred while the NST was the temperature where extensive grain boundary wetting occurred. Yeniscavich (Ref. 15) postulated that liquation started at the maximum ductility temperature, and that the strength dropped to zero when the liquid film covered the entire grain boundary area. This model was criticized by Owczarski (Ref. 139) who proposed that a shift in failure mode occurred at the temperature of the downturn of ductility and that melting subsequently occurred when the NDT temperature was exceeded. Duvall et al. (Ref. 9) found experimental evidence in Ni-base superalloys for a shift in fracture mode from ductile rupture to intergranular at the temperature of maximum ductility but did not report liquation to occur in this region. Weiss et al. (Ref. 124) studied IN600 and related the severe loss of ductility and strength above the maximum ductility temperature to partial melting. Partial melting resulted from the constitutional liquation of second phase particles. They also related the NST to the presence of a molten volume fraction of a few percent (=3%). The above review indicates that most investigators agree that the loss of ductility at elevated temperatures is due to the presence of liquid films at the grain boundaries. However, the temperature at which liquation initiates and the metallurgical behavior at the NDT and NST are still subjects to controversy. These debates can be rationalized based on the results of the metallurgical characterization of the CSR.
6.3 On-Heating Test Results

6.3.1 Spot-Varestraint Test

6.3.1.1 A-286 and Type 310 Stainless Steels

The HAZ microstructure of A-286 is shown in Figure 49 with the peak temperatures experienced during testing. Two distinct regions in the cracked HAZ can be clearly visualized. Immediately adjacent to the fusion line, region A-B experienced appreciable intergranular and intragranular liquation. This region is termed as a "localized melting region" in this report. For the on-heating spot-Varestraint tests (cooling time = 0 sec), no cracking occurred in this region. Further away from the fusion line, there is another region B-C in which on-heating liquation cracking occur, but intragranular localized melting was not observed. This region is referred to as a "liquated region" in this report. The transition temperature between these two regions approximates the NST as shown in Figure 49. Although intergranular liquation is not visible in the liquated region under an optical microscope, the liquation-related intergranular rupture on the on-heating crack surface (Figure 50) confirms that liquation indeed occurred intergranularly. Notice that on-heating liquation cracking only occurred over the temperature range between the NST and NDT. For the purpose of convenience, the crack tip closer to the weld center is called a "high temperature tip", and the crack tip away from the weld center is called a "low temperature tip". The fracture surface of high temperature tip exhibited protrusions on the grain faces and probably represented traces of the last-to-solidify liquid (Ref. 140), indicating extensive grain boundary liquation during thermal cycling. Close to the low temperature tip, grain faces become more distinct and protrusions were not observed, suggesting that a very thin liquid film was present during testing.
Figure 49. HAZ microstructure of A-286 stainless steel resulting from spot-Varestraint testing.

The HAZ microstructure of Type 310 stainless steel is shown in Figure 51 with peak temperatures experienced during testing. The localized melting region is not apparent due to the small amount of liquid present. However, this region indeed exists and can be observed at higher magnification (about 400X). Similar to A-286, on-heating HAZ liquation cracking stopped at the NDT, but distinct microstructural changes at the NST were not apparent. The fracture surface of a crack produced at a 0 second cooling time is shown in Figure 52. The entire crack surface exhibited liquation-related intergranular fracture.
Figure 50. Fracture surface of a crack of A-286 produced from spot-Varestraint testing. (Strain = 5%, \( t_c = 0 \) seconds).
Figure 51. HAZ microstructure of Type 310 stainless steel resulting from spot-Varestraint testing.

Figure 52. Fracture surface of a crack of Type 310 produced from spot-Varestraint testing. (Strain = 5%, \( t_c = 0 \) seconds).
6.3.1.2 **Incoloy 903 and 909**

HAZ microstructures of Incoloy 903 and 909 are shown in Figure 22. Similar to A-286, the localized melting region and the liquated region can be clearly observed. On-heating liquation cracking ($t_c = 0$ seconds) only occurred in the liquated region. Although detailed temperature measurements were not performed, the NST as a transition temperature was also supported by metallographic observations. As shown in Table 6, region A-B is about three times as wide in Incoloy 909 as in 903 (0.57 mm versus 0.17 mm), which correlates well with the greater $T_L$ - NST temperature range in Incoloy 909 (164°C versus 53°C), assuming that both materials exhibited the same temperature gradient with the same welding parameters.

Although fractographic analyses on spot-V-restraint test samples of these two alloys were not performed in this study, a recent study by Ernst *et al* (Ref. 58) confirmed that the fracture surface of Incoloy 903 spot-V-restraint specimens exhibited a distinct intergranular appearance. Careful examination of these surfaces showed a fine, wavy pattern on the grain faces, and evidence of melting along grain boundaries and at triple points, suggesting a liquation-related fracture mechanism.

6.3.2 **Hot-Ductility Test**

The metallurgical behavior of the on-heating hot-ductility testing for the four alloys tested can be divided into three stages. For the specimens tested at temperatures up to the on-heating maximum ductility temperature, the fracture surfaces exhibited a ductile dimple morphology as shown in Figures 53-55, indicating a solid-state mechanical failure. As the temperature increased above the maximum ductility temperature, the ductile dimple morphology disappeared and a "spongy" appearance
emerged as shown in Figure 56. This fracture mechanism is predominantly intergranular, with the spongy morphology superimposed on the grain faces. This fracture morphology occurs over the temperature range from the maximum ductility temperature to just below the NDT, as shown in Figures 57-60. The transition of fracture path from ductile rupture to spongy intergranular suggests that grain boundaries become preferred sites for void nucleation above the maximum ductility temperature. Near the NDT, the spongy morphology was less apparent and eventually disappeared at the NDT as shown in Figures 61, 63, 65 and 67. Although traces of liquid on the grain faces at the NDT (Figures 61, 63, 65 and 67) were not obvious, the featureless grain faces suggested that very thin liquid films were present at the grain boundary during testing. Evidence of intergranular liquid films could not be observed on the metallographically prepared sample surfaces as shown in Figures 62, 64, 66 and 68. This suggests that the liquid present at high temperature is "reabsorbed" into the surrounding structure on-cooling. Above the NDT, the fracture surfaces were featureless intergranular and the grain surfaces became more rounded due to a thicker liquid film as the temperature increased (Figures 20, 69-70). The terrace-like appearance in Figure 20 showed the presence of liquid films at grain boundaries before fracture. Near the NST, both extensive grain boundary liquation and intragranular localized melting occurred as evidenced by the widened grain boundaries and traces of resolidified liquid pockets (Figures 16, 71-72). Based on the results of fractographic analysis, it is hypothesis that liquation begins along grain boundaries just above the maximum ductility temperature. Continuous grain boundary liquid films form at the NDT and become thicker as the temperature increases. The onset of liquation
Figure 53. Fracture surface of Incoloy 903 from an on-heating hot-ductility test at 982°C. (Reduction-in-Area = 79%, maximum ductility temperature = 1160°C, NDT = 1240°C).

Figure 54. Fracture surface of Incoloy 909 from an on-heating hot-ductility test at 1060°C. (Reduction-in-Area = 80%, maximum ductility temperature = 1125°C, NDT = 1150°C).
Figure 55. Fracture surface of A-286 from an on-heating hot-ductility test at 1100°C. (Reduction-in-Area = 94%, maximum ductility temperature = 1140°C, NDT = 1200°C).

Figure 56. Fracture surface of Type 310 from an on-heating hot-ductility test at 1309°C. (Reduction-in-Area = 91%, maximum ductility temperature = 1300°C, NDT = 1325°C).
Figure 57. Fracture surface of Incoloy 903 from an on-heating hot-ductility test at 1200°C. (Reduction-In-Area = 56%, maximum ductility temperature = 1160°C, NDT = 1240°C).

Figure 58. Fracture surface of Incoloy 909 from an on-heating hot-ductility test at 1132°C. (Reduction-in-Area = 44%, maximum ductility temperature = 1125°C, NDT = 1150°C).
Figure 59. Fracture surface of A-286 from an on-heating hot-ductility test at 1188°C. (Reduction-in-Area = 28%, maximum ductility temperature = 1140°C, NDT = 1200°C).

Figure 60. Fracture surface of Type 310 from an on-heating hot-ductility test at 1322°C. (Reduction-in-Area = 48%, maximum ductility temperature = 1300°C, NDT = 1325°C).
Figure 61. Fracture surface of Incoloy 903 from an on-heating hot-ductility test at the NDT.

Figure 62. Microstructure of Incoloy 903 at the NDT obtained from the hot-ductility test.
Figure 63. Fracture surface of Incoloy 909 from an on-heating hot-ductility test at the NDT.

Figure 64. Microstructure of Incoloy 909 at the NDT obtained from the hot-ductility test.
Figure 65. Fracture surface of A-286 from an on-heating hot-ductility test at the NDT.

Figure 66. Microstructure of A-286 at the NDT obtained from the hot-ductility test.
Figure 67. Fracture surface of Type 310 from an on-heating hot-ductility test at the NDT.

Figure 68. Microstructure of Type 310 at the NDT obtained from the hot-ductility test.
Figure 69. Fracture surface of A-286 from an on-heating hot-ductility test at 1300°C. (NDT = 1200°C, NST = 1350°C).

Figure 70. Fracture surface of Type 310 from an on-heating hot-ductility test at 1345°C. (NDT = 1325°C, NST = 1350°C).
Figure 71. Microstructure of A-286 at the NST obtained from the hot-ductility test.

Figure 72. Microstructure of Type 310 at the NST obtained from the hot-ductility test.
is also supported by the model proposed by Weiss et al (Ref. 124). The fracture surface from the on-heating hot-ductility tests can be summarized in Figure 73.

6.4 On-Cooling Test Results

6.4.1 Hot-Ductility Test

For the on-cooling hot-ductility tests performed on the four alloys, the on-cooling hot ductility curve and associated fracture behavior can be generally divided into two types depending upon the amount of liquid present at the peak temperature employed for the on-cooling tests.

![Figure 73. Schematic illustration showing the characteristics of fracture surface from an on-heating hot-ductility test.](image-url)
6.4.1.1 For A Peak Temperature at Which Extensive Grain Boundary Liquation Occurs

For the on-cooling hot-ductility tests from a peak temperature equal to the NST of Incoloy 903, Incoloy 909 and A-286, extensive grain boundary liquation occurs at the peak temperature as shown in Figures 18, 71 and 72. In this case, ductility gradually recovered below the DRT and the amount of ductility recovery was low. As shown in Figures 17 and 32, the on-cooling hot ductility remained below 20% over 100°C below the DRT. In this case, the fracture surface exhibited a featureless intergranular morphology between the peak temperature and the DRT. Grain faces became more distinct as the temperature decreased. A distinct change in fracture mechanism at the DRT was not apparent. Rather, the fracture behavior changed gradually on-cooling below the DRT. A very small fraction of solid-state rupture appeared at the DRT as shown in Figures 74-76. Below the DRT, the fraction of solid-state rupture increased as the temperature decreased as shown in Figures 77-79. Below the maximum on-cooling hot ductility temperature, the majority of the fracture surface was solid-state mechanical failure, although some remnants of liquation-related intergranular fracture still remained as shown in Figures 80-82.

6.4.1.2 For A Peak Temperature at Which Only A Limited Amount of Grain Boundary Liquation Occurs

For the on-cooling hot-ductility tests from a peak temperature equal to the NST of Type 310 and from a peak temperature midway between the NST and NDT of Incoloy 903 and 909, only a limited amount of grain boundary liquation occurs at the peak temperature as shown in Figure 19 and Figure 72. In this case, the on-cooling
hot ductility rapidly recovered below the DRT as shown in Figures 17 and 32. A distinct change in fracture behavior was apparent at the DRT. For the temperature range between the peak temperature and the DRT, the fracture surfaces exhibited a featureless intergranular morphology with the grain faces becoming more distinct as the temperature decreased. The fracture surfaces near the DRT are shown in Figures 83-85. Below the DRT, two distinct types of fracture surface could be observed. For the temperature range between the DRT and the maximum on-cooling ductility temperature, the fracture surfaces exhibited an intergranular spongy morphology as shown in Figures 86 and 87. The fracture mechanism subsequently evolved to solid-state ductile rupture at the on-cooling maximum ductility temperature as shown in Figures 88 and 89. It is interesting that the appearance of the fracture surfaces in this case is basically a reversal of the on-heating tests. The characteristics of fracture surface of the on-cooling hot-ductility tests are summarized in Figure 90.

Figure 74. Fracture surface of Incoloy 903 from an on-cooling hot-ductility test at the DRT$_{NST}$(1050°C). (Reduction-in-area = 7%, $T_p$ = NST = 1340°C).
Figure 75. Fracture surface of Incoloy 909 from an on-cooling hot-ductility test at the DRT_{NST} (1110°C). (Reduction-in-area = 1%, T_p = NST = 1268°C).

Figure 76. Fracture surface of A-286 from an on-cooling hot-ductility test at the DRT_{NST} (1050°C). (Reduction-in-area = 3%, T_p = NST = 1350°C).
Figure 77. Fracture surface of Incoloy 903 from an on-cooling hot-ductility test at 1000°C from a peak temperature equal to the NST. (Reduction-in-area = 22%, $\text{DRT}_{\text{NST}} = 1050°C$).

Figure 78. Fracture surface of Incoloy 909 from an on-cooling hot-ductility test at 1075°C from a peak temperature equal to the NST. (Reduction-in-area = 7%, $\text{DRT}_{\text{NST}} = 1110°C$).
Figure 79. Fracture surface of A-286 from an on-cooling hot-ductility test at 978°C from a peak temperature equal to the NST. (Reduction-in-area = 21%, DRT_{NST} = 1050°C).

Figure 80. Fracture surface of Incoloy 903 from an on-cooling hot-ductility test at 950°C from a peak temperature equal to the NST. (Reduction-in-area = 15%, DRT_{NST} = 1050°C).
Figure 81. Fracture surface of Incoloy 909 from an on-cooling hot-ductility test at 950°C from a peak temperature equal to the NST. (Reduction-in-area = 15%, DRT$_{NST}$ = 1110°C).

Figure 82. Fracture surface of A-286 from an on-cooling hot-ductility test at 915°C from a peak temperature equal to the NST. (Reduction-in-area = 34%, DRT$_{NST}$ = 1050°C).
Figure 83. Fracture surface of Incoloy 903 from an on-cooling hot-ductility test at the DRT (1075°C) from a peak temperature midway between the NDT and NST. (Reduction-in-area = 8%, $T_p = 1290°C$, NDT = 1240°C, NST = 1340°C).

Figure 84. Fracture surface of Incoloy 909 from an on-cooling hot-ductility test at the DRT (1140°C) from a peak temperature midway between the NDT and NST. (Reduction-in-area = 4%, $T_p = 1219°C$, NDT = 1150°C, NST = 1268°C).
Figure 85. Fracture surface of Type 310 from an on-cooling hot-ductility test at the DRT\textsubscript{NST} (1325°C). (Reduction-in-area = 0%, NST = 1350°C).

Figure 86. Fracture surface of Incoloy 909 from an on-cooling hot-ductility test at 1132°C from a peak temperature midway between the NDT and NST. (Reduction-in-area = 33%, DRT = 1140°C, maximum on-cooling ductility temperature = below 1050°C).
Figure 87. Fracture surface of Type 310 from an on-cooling hot-ductility test at 1322°C from a peak temperature equal to the NST. (Reduction-in-area = 29%, DRT_{NST} = 1325°C, maximum on-cooling ductility temperature ~ 1260°C).

Figure 88. Fracture surface of Incoloy 903 from an on-cooling hot-ductility test at 1050°C from a peak temperature midway between the NDT and NST. (Reduction-in-area = 72%, T_p = 1290°C, NDT = 1240°C, NST = 1340°C).
6.4.2 Spot-Varestraint Test

For each set of on-cooling hot-ductility tests, a DRT corresponding to the peak temperature employed can be determined. As shown in Figure 31, this DRT only represents one data point on the on-cooling CSR envelope. In order to comprehensively investigate the metallurgical behavior at the boundary of the CSR, it is easier to examine the on-cooling spot-Varestraint test specimens, since one useful crack can be produced from each specimen and the temperature at the low temperature tip of the crack also represents one data point on the on-cooling CSR envelop.
(A) For a peak temperature at which extensive grain boundary liquation occurs.

(B) For a peak temperature at which only a limited amount of grain boundary liquation occurs.

Figure 90. Schematic illustration showing the characteristics of fracture surface from on-cooling hot-ductility tests.
6.4.2.1 A-286 Stainless Steel

As the cooling time increased above 0 second, the crack shifts toward the weld center. For a cooling time equal to 0.2 seconds, the high temperature tip moved into the localized melting region. As shown in Figure 91, solidification substructure at the grain boundaries was clearly visible. The widened grain boundaries (Figure 91) suggested that extensive grain boundary liqation occurred during testing. Above 0.25 seconds, the high temperature tip moved into the fusion zone. Correspondingly, the low temperature tip moved closer to the original fusion boundary. Figure 92 shows a series of fracture surfaces of the low temperature tip at different cooling times. As shown in Figure 40, the low temperature tip locates the boundary of the on-cooling CSR. For the cooling time from 0 to 3 seconds, the low temperature tip exhibited an intergranular spongy morphology. Above 3 seconds cooling time, the spongy structure gradually disappeared. The spongy morphology was not visible at 3.9 seconds cooling time. As shown in Figure 40, the temperature at the low temperature tip of 3.9 seconds cooling time is about the $DRT_{N\sigma}$ obtained from the hot-ductility test. Although the temperatures at the boundary of the on-cooling CSR (DRT's) varied depending on the corresponding peak temperature, the material essentially exhibited a similar fracture surface (Figure 92), indicating a similar metallurgical phenomena occurring at these DRT's.
Figure 91. Optical microstructure and fracture surface at the high temperature tip of a crack of A-286 produced from the spot-Varestraint test. (Strain = 5%, Cooling Time = 0.2 second).
Fig. 92. Fracture surface of the low temperature tip of A-286 cracks produced from the spot-Varestraint test for the cooling times ranging from 0 to 3.9 seconds. (Strain = 5%).
Figure 92 continues

(C) Cooling Time = 1.0 seconds

(D) Cooling Time = 2.0 seconds
Figure 92 continues

(E) Cooling Time = 3.0 seconds

(F) Cooling Time = 3.9 seconds
6.4.2.2  **Type 310 Stainless Steel**

Similar to A-286, the high temperature tip of Type 310 moved into the fusion zone as cooling time increased. The fracture surface near the fusion boundary is shown in Figure 93. Because of the small amount of liquid present in the PMZ, the fusion line can be clearly observed. The HAZ grain faces are distinct and no evidence of protrusions on grain faces in contrast to Figure 50. The fracture surfaces near the low temperature tip at different cooling times are shown in Figure 94. All the surfaces exhibited an intergranular spongy morphology. Again, the temperature at the low-temperature tip represents a DRT on the on-cooling thermal CSR. Although the DRT's are different (Figure 48), the fracture surfaces are similar, indicating similar metallurgical phenomena occurring in these DRT's. Note that there was not any evidence of solid-state rupture on the fracture surfaces.

![Fracture surface of a crack of Type 310 produced from spot-Varestraint test showing the structure at the fusion line. (Strain = 5%, Cooling Time = 0.3 seconds).](image)

Figure 93. Fracture surface of a crack of Type 310 produced from spot-Varestraint test showing the structure at the fusion line. (Strain = 5%, Cooling Time = 0.3 seconds).
Figure 94. Fracture surface of the low temperature tip of Type 310 cracks produced from the spot-Varestraint test for the cooling times ranging from 0 to 0.4 seconds. (Strain = 5%).
Figure 94 continues

(C) Cooling Time = 0.2 seconds.

(D) Cooling Time = 0.3 seconds
6.5 Discussion

6.5.1 The Evolution of Liquid in the Crack Susceptible Region

According to a simple binary phase diagram, with an increase in temperature, an alloy transforms from a solid phase through a solid + liquid phase to a complete liquid phase at the liquidus temperature. Although the liquidus is essentially constant, the temperature at which liqutation begins depends on the heating rate, and the location at which liqutation initiates depends on the localized composition and structure, as described in chapter 1. To a first approximation, the bulk material can be considered to be composed of two parts, namely, grain boundaries and interior of grains. Generally, grain boundaries exhibit a higher defect density and contain higher levels of impurities and/or solutes (Ref. 74). From a free energy viewpoint, melting or
liquation will always occur first at grain boundaries (Ref. 141). The presence of impurities and/or solutes further lowers the local melting temperature (Ref. 141). Thus, liquration or melting of an alloy can be discussed on two different scales: a macroscopic scale (bulk material) and a microscopic scale (grain boundary). The temperature at which liquration initiates at the grain boundary is also the lowest liquration temperature of the bulk material. However, as discussed above, grain boundary may completely melt at a temperature lower than the liquidus of the bulk material.

Considering the liquration sequence of the grain boundary, it is hypothesized that the grain boundary also undergoes partial melting before it completely liquates. Liquid first forms as an isolated pocket and the fraction of the liquated area in a grain boundary increases as the temperature increases. The formation of the liquid decreases the localized ductility dramatically. The portion of the grain boundary which remains solid provides the major path for the load to transfer from one grain to another. As the temperature increases, the fraction of liquated boundary increases and fracture is thus initiates at the grain boundary. The drop in ductility due to partial melting of grain boundary is supported by the hot-ductility test results. As shown in Figures 17 and 32, the on-heating hot ductility of the material falls off above the maximum ductility temperature and approaches zero at the NDT. Fractographic analysis in this temperature range shows an intergranular spongy morphology, as presented in Figures 57-60. The protrusions of the spongy structure would correlate with the solid portion of the grain boundary. As temperature approaches the NDT, the fraction of protrusions decreases and grain faces become featureless at the NDT, as shown in Figures 61, 63, 65 and 67. Based on this discussion, it follows that liquation
of grain boundaries and the bulk material begins slightly above the maximum ductility temperature. The grain boundary is completely liquated at the NDT. The bulk material remains partially melted until the temperature reaches the alloy liquidus, at which point the entire grain melts.

Above the NDT, it is postulated that all the grain surfaces are covered by a thin layer of liquid and that the thickness of the liquid film increases with an increase in temperature as suggested by Figures 20, 69 and 70. Although this liquid film cannot accommodate any strain, the bulk material still exhibits a certain level of strength due to the surface tension between the liquid film and the solid grain substrate. The strength of a material in the presence of a thin liquid film at the grain boundaries has been extensively discussed by Weiss et al (Ref. 123). According to them, the strength of a material (σ) can be calculated according to:

$$\sigma = 2\gamma/h$$  \hspace{1cm} (6.1)

where, \(\gamma\) is the surface tension between the liquid film and the solid grain, and \(h\) is the thickness of the liquid film. As the temperature increases, the thickness of the liquid film increases and surface tension decreases, resulting in a net decrease in strength. At some limiting temperature, the strength of the material drops to essentially zero. By definition this is the NST.

6.5.2 The Solidification of Liquid Films and Resultant Cracking In the Crack Susceptible Region

After the weld pool passes, liquid films at the grain boundaries in the HAZ solidify. It is hypothesized that solidification proceeds with a reduction of the liquid film
by epitaxial growth on the solid substrate. Solid-solid contact begins when the liquid film is too thin to separate the solid grains.

On-cooling from a peak temperature above the NDT, the thickness of the liquid film decreases as temperature decreases. The solidification process eventually reaches a state at which the liquid films are confined to the grain boundaries and solid-solid contact between grains begins. As the temperature further decreases, the fraction of solid formed increases and finally, grain boundary liquid films completely solidify. Thus, the solidification process in this case is essentially a reversal of the liquation process. It is necessary that there is a temperature range over which solid and liquid phase coexist at the grain boundaries. The magnitude of this temperature range depends on the solidification temperature range of the last-to-solidify constituents, which is directly related to the amount and composition of liquid present at the peak temperature.

Results from this study indicated that the temperature range over which solid and liquid coexist can be generally divided into two classifications. For the peak temperature at which only a limited amount of liquid is present (1290°C for Incoloy 903, 1219°C for Incoloy 909 and NST for Type 310 in hot-ductility test), the temperature range over which the film solidifies is very small. The ductility of the material recovers rapidly in this temperature range as shown in Figures 17 and 32. This small temperature range correlates well with that between the DRT and the maximum on-cooling ductility temperature (Figures 17 and 32). The fracture surface of specimens fractured in this temperature range exhibited an intergranular spongy morphology (Figures 86 and 87), which is essentially the same structure as that which occurs in the on-heating grain boundary partial melting temperature range (Figures 53-56). It is
important to note that the fracture surface at the low temperature tip of cracks with cooling times ranging from 0 to 3.0 seconds in A-286 and all cracks in Type 310 produced from the spot-Varestraint test also exhibited the same structure as shown in Figure 94 and (A) - (E) in Figure 92. As discussed in chapter 5, the temperature at the low temperature tips in the spot-Varestraint test corresponds to the DRT's in the hot-ductility test.

For the peak temperature at which extensive grain boundary liquation occurs (NST in Incoloy 903, Incoloy 909 and A-286), the solidification temperature range is large as shown in Tables 6 and 14. As a result, ductility recovers gradually within this larger solidification temperature range and the magnitude of ductility is low even after the liquid films have completely solidified (Figures 17 and 32). The fracture surface of specimens fractured in this solidification temperature range did not exhibit spongy morphology, but rather showed a flat grain surface with some evidence of solid-solid bridging as shown in Figures 77-79 and (F) in Figure 92. The discussion on the solidification of liquid films and resultant cracking in the CSR is summarized in Figure 96.

Based on the above discussion, it is clear that metallurgically, the NDT is the temperature on-heating at which grain boundaries completely liquate, and the DRT is the temperature on-cooling at which the liquid is confined to a thin layer of film at the grain boundary and solid grain contact begins. To be more precise, liquation cracks may propagate at a temperature slightly lower than the NDT on weld heating and slightly lower than the DRT's during cooling. The actual crack tip temperature may depend on the applied strain. However, the difference between the actual crack tip temperature and the NDT or DRT's is probably negligible. Based on this metallurgical
characterization, the CSR developed in this study is equivalent to the PMZ during welding.

One of the important findings from this study is that the boundary of the on-cooling thermal CSR is the profile of the DRT's and the DRT is the effective solidus temperature. It varies with the distance from the fusion line. As the distance from the fusion line decreases, the peak temperature of a weld cycle increases and the corresponding DRT decreases. This difference in the DRT's can be rationalized by considering the compositions of the last-to-solidify liquid at the grain boundaries. For Type 310, no eutectic formed in the resolidified grain boundaries as shown in Figure 51. As the peak temperature was higher (closer to the fusion line), greater amount of liquid present at the grain boundaries, probably resulting in a higher level of solutes and/or impurities. The effective solidus temperature of the last-to-solidify liquid was thus decreased. Further study in this area is required to clarify this issue. For A-286, two distinct regions in the HAZ were observed as shown in Figure 49 and discussed previously. In the liquated region, no evidence of eutectic formation was observed in the grain boundaries. The variation in the DRT's in this region would be the same as in Type 310. The DRT in this region would be higher than the Laves/γ eutectic temperature at the same impurity level. In the localized melting region, formation of Laves phase in the grain boundaries was observed. Although the Laves/γ eutectic temperature is fixed, the difference in the DRT's was observed. This difference probably depends on the different impurity level. With a greater amount of liquid at the higher peak temperature (closer to the fusion line) may involve higher level of impurities to partition to the last-to-solidify liquid and result in a lower DRT.
Figure 95. A hypothesis for the evolution of liquid and resultant cracking in the crack susceptible region.
6.6 Comparison of Liquation Behavior Between A-286 and Type 310

The review in chapter 1 on HAZ liquation cracking mechanisms indicated that there are three major mechanisms contributing to the formation of liquid in the PMZ. These three mechanisms are: the partial melting hypothesis, the penetration mechanism and the segregation mechanism. The partial melting hypothesis is applied between the temperature range of liquidus and the NST as discussed in section 6.3.1. The penetration mechanism operates from liquidus temperature to the eutectic temperature of the alloy system (Ref. 71). The segregation mechanism is effective between liquidus and the effective solidus determined by local grain boundary melting (Refs. 22, 34). It is important to note that liquation would occur once the partial melting hypothesis or the penetration mechanism are involved. However, the segregation mechanism only provides a model for solute and/or impurity elements to segregate to the grain boundaries. Occurrence of liquation still depends on the composition and the local temperature.

Based on the results of metallurgical characterization in this study and the published literature (Refs. 34, 37-38, 45, 62). It can be concluded that the three aforementioned mechanisms are involved in A-286, but only the partial melting hypothesis and the segregation mechanism are involved in Type 310. The difference in the amount of liquid formed in the PMZ in these two alloys was attributed to the contribution of the penetration mechanism. The constitution liquation of titanium carbides and/or carbonitrides in A-286 results in a much greater amount of liquid and the formation of Laves/γ eutectic resulting in a larger BTR as evidenced in Figure 48.
6.7 Comparison of Microstructures Resulted from Different Tests

Although the thermal CSR determined from the three tests is essentially identical, a difference in microstructures from different tests was also noted. Because the thermal cycles (Figures 36, 41, Table 9) employed for the three tests are very different, another thermal simulation was performed on the Gleeble with a heating thermal cycle similar to that employed in the spot-Varestraint tests. In this thermal simulation, the specimens were heated linearly to 1200°C in 17 seconds and reached the peak temperature in 35 seconds. After the peak has reached, the specimen cooled immediately with a cooling rate of about 50°C/sec. The resultant microstructures at the NDT and NST for A-286 are shown in Figures 96 and 97, respectively, along with the microstructures at the corresponding temperatures obtained from the three tests. It was found that the grain size at the NDT and NST is different with the largest grain size resulting from the thermal simulation described above and the smallest from the hot-ductility test. A comparison of HAZ microstructure of A-286 from the spot- and longitudinal-Varestraint tests is provided in Figure 98. Difference in grain size is apparent.

In addition to the difference in thermal cycle (heating rate and cooling rate), the temperature gradient is another factor which affects the microstructure. As shown in Figures 37 and 42, the temperature gradient in the HAZ of a spot weld is much shallower than that in a continuous weld. Although the temperature gradient in the hot-ductility test specimens investigated in this study were not exactly measured, the temperature distribution on a 304 stainless steel test sample at a peak temperature of 985°C was monitored for reference as shown in Figure 90. It is apparent that the hot-ductility specimens experienced a very shallow temperature gradient. As pointed by
Dolby et al. (Ref. 142), the difference in temperature gradient may result in a difference in microstructure even though the thermal cycle is identical.

The thermal conditions determine the metallurgical reactions in the HAZ during weldability testing. During weld heating, the heating rate is the principal factor governing the evolution of liquid as discussed in Chapter 1. While for the on-cooling portion, the cooling rate affects the redistribution of solute and/or impurity elements during solidification, thus, controls the amount and type of constituent formed during cooling. Although detailed metallurgical characterization (using TEM, AEM for instance) was not performed in this study, no significant difference in the grain boundary microstructure, except for the grain size, at the corresponding temperature among these three tests was observed. It is hypothesized that within a certain range of the heating rate and cooling rate, the thermal cycle determines the rate at which a metallurgical condition (the amount and type of constituents) at the grain boundaries in the HAZ is achieved. It does not significantly alter the metallurgical condition at the grain boundaries at the corresponding temperature. For example, the thermal CSR of A-286 and Type 310 determined via high speed laser or electron beam welding process may be different from that presented in this study. Yet it is likely to be the same for conventional fusion welding processes.
Figure 96. Microstructure of A-286 at the NDT (1200°C) obtained from different thermal cycles.
Figure 97. Microstructure of A-286 at the NST (1350°C) obtained from different thermal cycles.
Figure 98. HAZ microstructure of A-286 obtained from the spot- and longitudinal-Varestraint tests.
6.8 Conclusions

1. During weld heating, grain boundary liquation begins just above the maximum on-heating ductility temperature. The presence of grain boundary liquid results in an abrupt drop in ductility. Grain boundaries are completely liquid at the nil-ductility temperature, at which point the liquid film cannot accommodate any strain and the ductility drops to zero. With a further increase in temperature, the thickness of the grain boundary liquid films increases.

2. The fracture mechanism shifts from ductile rupture to intergranular above the maximum on-heating ductility temperature. Over the temperature range within which grain boundaries exhibit partial melting, the fracture surface exhibits an intergranular spongy structure. This morphology is attributed to the presence of
isolated liquated regions at the grain boundary. Above the nil-ductility temperature, the fracture surface is always featureless intergranular.

3. Optical microstructures in the on-heating crack susceptible range from the spot- and longitudinal-Varestraint tests can be generally divided into two regions. Directly adjacent to the fusion line, a localized melting region experiences extensive intergranular and intragranular melting or liquation. On-heating liquation cracking does not occur in this region. Further away from the fusion line, there is a liquated region within which intragranular localized melting does not occur. Liquation occurs intergranularly although it is generally not visible under an optical microscope. On-heating liquation cracks occur in this region.

4. During weld cooling, solidification of the grain boundary liquid film proceeds with a reduction of the liquid film by epitaxial growth on the solid substrate. Solid-solid contact begins at the ductility recovery temperature. Fraction of solid contact increases as the temperature decreases. Grain boundary liquid films completely solidify below the maximum on-cooling ductility temperature.

5. From the peak temperature to the corresponding ductility recovery temperature during weld cooling, the fracture surface morphology is featureless intergranular. The fracture surface in the temperature range within which a liquid and a solid phase coexist at the grain boundary depends on the amount of grain boundary liquid present at the peak temperature. For a peak temperature at which extensive grain boundary liquation occurs, the fracture surface exhibits a mixture of liquation related intergranular fracture and evidence of solid-state rupture. For a peak temperature at which only limited amount of grain boundary liquation occurs, the fracture surface exhibits an
intergranular spongy structure. The fracture mechanism shifts from liqution-related intergranular to complete solid-state rupture at the maximum on-cooling ductility temperature.

6. The crack susceptible region developed in this study is equivalent to the partial melting zone in the HAZ during welding.
CHAPTER VII
SUMMARY

The methodologies for quantifying the susceptibility of a material to heat-affected zone (HAZ) liqation cracking and the mechanisms by which liqation cracking occurs in the HAZ have been the subjects of considerable investigations over the last several decades. The present study was thus undertaken to develop a testing methodology which describes the physical relationship among weldability test results, cracking theories and material properties. It also clarifies the methods of data interpretation of the hot-ductility, the spot- and longitudinal-Varestraint tests, and the mechanisms by which liqation develops and results in cracking in the HAZ.

Comparison of the HAZ LIquation Cracking Susceptibility Between Incoloy 903 and 909

A preliminary research program was aimed at comparing the HAZ liqation cracking susceptibility of Incoloy 903 and 909 using the conventional spot-Varestraint and hot-ductility methods. Results indicated that the cracking and loss of high-temperature ductility of Incoloy 903 and 909 resulted from the presence of thin, low-melting grain boundary liquid films. Incoloy 909 exhibited a lower nil-strength temperature (NST) and nil-ductility temperature (NDT) than Incoloy 903 but recovered ductility more rapidly. The difference in the temperature range of NDT-DRT (ductility
recovery temperature) between incoloy 903 and 909 was attributed to the difference in the liquating phases, with Nb-rich carbides in the former and Laves phases in the latter. Most importantly, hot-ductility test results were inconsistent with predictions of cracking susceptibility obtained with the spot-Varestraint test, with a higher susceptibility of Incoloy 903 in the former and Incoloy 909 in the later, if both test results were interpreted using "conventional" techniques. This discrepancy was rationalized using newly developed testing methods for the two tests. With these new methodologies, it was found that the conventional cracking indices for hot ductility and spot-Varestraint tests were evaluating two different aspects of HAZ liquation cracking susceptibility. The temperature range between the NST and corresponding DRT as determined by the hot-ductility test depicts the time period for the restoration of HAZ ductility on cooling at a point in the HAZ which experiences a peak temperature equal to the NST. The maximum crack length (MCL) as characterized by the spot-Varestraint test represents the width of a crack susceptible region in the HAZ. The MCL was found to correlate well with the temperature range between the liquidus (T_L) and NDT in the hot-ductility test.

Development of the Thermal Crack Susceptible Region in the HAZ

The new methodologies developed in the preliminary study were further investigated using modified spot- and longitudinal-Varestraint (mini-Varestraint) tests performed on A-286 and type 310 stainless steels. Results from this research led to the development of a theoretical model which describes a thermal crack susceptible region (CSR) in the HAZ based on the ductility of a material during welding. This CSR was determined from the hot-ductility test, and the criteria assumed in the
development of liquation cracking theories. The CSR defines a thermal envelope in
the HAZ in which liquation cracking may occur. It expands on weld heating and
shrinks upon cooling, and takes on a steady state shape surrounding the weld pool in
a continuous weld.

The concept of the CSR was subsequently verified with the modified
longitudinal- and spot-Varestraint tests, which were developed in this study. The
longitudinal-Varestraint tests were performed over a range of augmented strain. The
location of each crack tip in the HAZ was determined along the fusion boundary
relative to the instantaneous weld center. The profile of crack tips and the fusion line
circumscribed a region in which HAZ liquation cracking occurred. The size of this
cracked HAZ region expanded with an increase in augmented strain and remained
constant above a saturated strain. The cracked HAZ region determined with a
saturated strain was defined as the CSR. Incorporating the CSR results with thermal
conditions experienced in the HAZ during longitudinal-Varestraint testing, the thermal
CSR was determined.

The spot-Varestraint tests were performed in two stages. On-heating spot-
Varestraint tests were initially performed using no cooling time (or bending delay time)
after arc extinction to determine the saturated strain and the extent of on-heating CSR.
Subsequently, on-cooling spot-Varestraint tests were performed at a saturated strain
level over a range of cooling times to determine the on-cooling CSR. Incorporating
these results with thermal conditions experienced in the HAZ during spot-Varestraint
testing, the entire thermal CSR was determined.

The experimental verification of the theoretical model was achieved by
characterizing the localized ductility of the material in the thermal CSR determined
from the longitudinal- and spot-Varestraint tests, and directly comparing the extent of the thermal CSR obtained from the three methods. The thermal CSR is composed of two basic elements: the extent of the CSR during weld heating and the brittle temperature range (BTR) upon cooling. The correlation between the theoretical model and the experimental results by comparing the magnitude of these two elements obtained for A-286 and Type 310 stainless steels was excellent, reinforcing the complementary nature of these techniques for predicting HAZ liquation cracking susceptibility. This good correlation also provides experimental evidence for the criteria assumed in the development of liquation cracking theories, which state that liquation cracking results from the localized loss of ductility, and confirms that the thermal CSR is a material-specific parameter. This parameter is a true quantification of HAZ liquation cracking susceptibility.

**Correlation Among Results Obtained from the Hot-Ductility, Spot- and Longitudinal-Varestraint Tests**

The complete correlation among results obtained from the three aforementioned techniques can be demonstrated by comparing the magnitude of the two basic elements for the development of the thermal CSR. The temperature range between the $T_L$ and NDT in the hot-ductility test represents the extent of the on-heating thermal CSR. This temperature range correlates with the temperature range over which the MCL occurs at the transition between on-heating and on-cooling in the longitudinal-Varestraint test performed with the saturated strain. It also correlates with the temperature range over which the MCL occurs in the spot-Varestraint test performed with saturated strain at no cooling time after arc extinction. The
temperature range between the peak temperature and the corresponding DRT as obtained from the hot-ductility test represents the BTR. This temperature range in turn correlates with the cooling time over which cracking is observed in the spot-Varestraint test, and the on-cooling portion of the cracked HAZ length in the longitudinal-Varestraint test.

The Evolution of Liquid and Resultant Cracking in the Crack Susceptible Region

Metallographic and fractographic analyses were performed on the specimens of the three tests for the four alloys investigated in this study: Incoloy 903, Incoloy 909, A-286 and Type 310. Results from these analyses indicated that during weld heating, grain boundary liquation begins just above the maximum on-heating ductility temperature. For all the alloys tested, the presence of grain boundary liquid resulted in an abrupt drop in ductility. Grain boundaries were hypothesized to completely liquate at the NDT, since a liquid film cannot accommodate any strain and the ductility drops to zero. Further increase in temperature approaching the NST, results in an increase in the thickness of the grain boundary liquid films.

During weld cooling, solidification of the grain boundary liquid films proceeds initially with a reduction of liquid film by epitaxial growth on solid substrate. Solid grain contact begins at the DRT. As solid grain contact develops at the grain boundaries, the material recovers its ductility. The ductility continuously recovers in the temperature range over which a solid and a liquid phase coexist at the grain boundaries. Grain boundary liquid films completely solidify below the maximum on-cooling ductility temperature.
Fractographic Maps of the Crack Susceptible Region

Fractographic analysis indicated that fracture mechanism shifts from ductile rupture to intergranular above the maximum on-heating ductility temperature on weld heating. Over the temperature range within which grain boundaries partially melt, the fracture surface exhibited an intergranular spongy morphology. The evolution of this structure was attributed to the presence of isolated liquid pockets at the grain boundaries. As the temperature increased, the spongy morphology diminished and eventually disappeared near the NDT. Above the NDT, the fracture surface was always featureless intergranular. The grain surfaces became more rounded as temperature increased. During weld cooling from a peak temperature to the corresponding DRT, the fracture surface was featureless intergranular. The fracture surface in the temperature range within which a liquid and a solid phase coexist at the grain boundaries depends on the amount of grain boundary liquid present at the peak temperature. For a peak temperature at which extensive grain boundary liquation occurs, the fracture surface exhibits a mixture of liquation-related intergranular fracture and evidence of solid-state rupture. For a peak temperature at which only limited amount of grain boundary liquation occurs, the fracture surface exhibits an intergranular spongy structure. The fracture mechanism shifts from intergranular to solid-state mechanical failure below the maximum on-cooling ductility temperature.

Formation of the Partially Melted Zone

The microstructure in the on-heating crack susceptible range can be generally divided into two regions. Directly adjacent to the fusion line, a localized melting region experiences extensive intergranular and intragranular melting or liquation. On-
heating liqation cracking does not occur in this region. Further away from the fusion line, there is a liquated region within which intragranular localized melting does not occur. Liquation occurs intergranularly although it is generally not visible under an optical microscope. On-heating liqation cracks produced from the spot- and longitudinal-Varestraint tests only occur in this region.

Results of metallurgical characterization confirmed that the CSR developed in this study is equivalent to the partially melted zone (PMZ) during welding. For A-286, the extent of the on-heating PMZ is 222°C. At a location in the HAZ which experiences the peak temperature equals the NST (1350°C), liquid films persists 300°C below the NST during weld cooling, and the liquid films persists 387°C below the liquidus for a point at the fusion line (1422°C). For Type 310, the extent of the on-heating PMZ is 61°C. At a location in the HAZ which experiences the peak temperature equals the NST (1350°C), liquid films persists 25°C below the NST during weld cooling, and the liquid films persists 90°C below the liquidus for a point at the fusion line (1386°C).
CHAPTER VIII
FUTURE WORK

This research is a fundamental study of weldability testing techniques as related to the development of a CSR in the HAZ. Based on the results of this study, many research projects could be undertaken to benefit material producers or the manufacturing industry concerning about liquation cracking. Several interesting subjects are highlighted below.

Further Evaluation of the Thermal Crack Susceptible Region Criterion

As discussed in this dissertation, the size of the thermal crack susceptible region may depend on test parameters such as strain rate. It would be useful to evaluate the effects of test conditions on the CSR. Some important parameters are stroke rate, heating rate and cooling rate for hot-ductility test, and strain rate and welding parameters for spot- and longitudinal-V-restraint tests.

Results from this study show that the methodology works very well for the alloys investigated: Incoloy 903, Incoloy 909, A-286 and Type 310. In order to ensure that the methodology proposed is generic, it would be necessary to apply this methodology to other families of alloys such as aluminum- and nickel-based alloys.
Development of a Weldability Index for Quantifying HAZ Liquefaction Cracking

Susceptibility

The crack susceptible region in the HAZ identified in this study defines a region in which HAZ liquefaction cracking may occur. Although this region is a material-specific parameter and is very important from a liquefaction cracking standpoint, its determination is time consuming and requires moderate technical skills. In addition, the probability with which liquefaction cracking occurs in this region cannot be precisely predicted based on the size of this region. Consequently, the crack susceptible region would not become an attractive weldability index for industry people.

According to Prokhorov's technological strength theory and Borland's Generalized theory, a good weldability index should take the major factors which affect liquefaction cracking susceptibility into account. The four major factors of these are (1) the brittle temperature range, (2) ductility within the brittle temperature range, (3) backfilling effects and (4) the coefficient of thermal expansion. In order to combine these four factors into one index, the easiest way could be to characterize the critical accumulation rate of strain at which liquefaction cracking occurs while the material is recovering its ductility during weld cooling. This can be achieved by letting the recovery of material ductility compete with the accumulation of strain during weld cooling. The strain can be a simple, externally applied mechanical strain with a constant application rate and using an experimental setup with which the thermally-induced strain is very small and is negligible. With this concept, a critical application rate of strain above which cracking occurs can be obtained. Normalizing this critical application rate with the thermal conditions experienced in the HAZ during welding, a material-specific parameter representing a specific amount of ductility recovery of a
material during weld cooling could be obtained. Subtracting the coefficient of thermal expansion from this normalized critical deformation rate, the specific amount of ductility recovery available for thermally-induced strain can be determined. This material-specific cracking index would correlate best with the probability with which liquation cracking occurs during welding.

The development of this cracking index relies on a machine which can apply a range of constant rates of strain while a material is subjecting to weld thermal cycling. This machine is currently under development by the author at Edison Welding Institute.

Development of the Complete Crack Susceptible Region in a Weldment

The methodologies utilized for determining the crack susceptible region in the HAZ can also be applied to the fusion zone. Thus, a complete crack susceptible region in a weldment can be determined. This methodology can also be utilized to compare the susceptibility of a material to fusion zone, HAZ and weld metal HAZ liquation cracking.

Application of the Concept of Using a Material-Specific Parameter as a Weldability Index in Predicting Liquation Cracking in Actual Welding Conditions

In this study, a concept of using a material-specific parameter as a weldability index was introduced in the development of the HAZ crack susceptible region. This concept would be useful in predicting liquation cracking in actual welding conditions. Basically, the occurrence of liquation cracking in actual welding conditions is a result of a specific material, process and restraint combination. A very weldable
material may crack under a severe restraint condition and alternatively, a cracking susceptible material may be successfully welded under a slight restraint condition. Thus, it is suggested that two steps can be taken in the prediction of liquation cracking in actual welding conditions. The first step is to characterize the three specific factors. For example, the material factor can be represented by the cracking index described in the above section. Process factor would include the welding process employed and the welding parameters utilized. Specifically, the process factor would affect the temperature distribution in the weldment during welding. Restraint factor would involve the thickness of the material, the type of joint design and the fixturing conditions. The second step is to develop a physical relationship between cracking and the three factors mentioned. It would be possible to predict the occurrence of liquation cracking in actual welding fabrications if the two tasks can be well developed.
REFERENCES


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