MICROSTRUCTURAL AND SOLIDIFICATION CRACKING EVALUATION OF ELECTRON BEAM WELDS IN 304L

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ABSTRACT

Weld hot cracking of stainless steels is a major materials-related problem in the welding industry. This present investigation evaluates the crack susceptibility of highly-constrained EB welds made in materials whose DeLong ferrite potentials range from zero to nine FN. In addition, the effect of piece part strength level on cracking is examined.

This study has revealed that these deep penetration EB welds have regions that solidify as primary austenite, even when the DeLong ferrite potential is as high as 9 FN. This points out the critical role that solidification rate plays in the crack susceptibility of these highly restrained welds. In addition, 0 FN to 0 FN welds had primarily transverse cracks while 6 FN to 0 FN welds had primarily centerline cracks. Of particular interest is the observation that cracks still occur if a high ferrite (greater than 6 FN) component is welded to a zero FN component. Cracking is always associated with regions which solidify as primary austenite and these cracks occur because there are areas in the weld which do not mix. Thus it is not a recommended production practice to compensate for low ferrite in one piece part with high ferrite in its mate. Finally, it is shown that a DeLong FN threshold of 4 to prevent cracking in EB welds is not valid.

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OBJECTIVE

The objective of this study was to evaluate the relative effects of piece part delta ferrite potential and ultimate tensile strength on the cracking frequency and characteristics of a highly constrained, deep penetration EB weld. This weld is an autogeneous, quarter-inch deep, circular weld made in a variety of low-sulfur 304L stainless steels whose DeLong ferrite potentials range from zero to nine FN.

THEORY

The problem of hot cracking in austenitic stainless steel weld metal has been studied for over 40 years. The body of literature on this subject is considerable and is not fully reviewed here. Although the mechanism by which solidification hot-cracking occurs is not fully understood, several factors are known to be significant.

In order for hot cracking to occur, two conditions are necessary. One is a minimum level of tensile stress or restraint. The second is a crack susceptible microstructure. In welding, restraint occurs as the weld cools and the resultant shrinkage stress is always present. The magnitude of this restraint is a function of joint geometry and depth of weld penetration. This stress can also be the result of externally applied loads.

The factors which produce a crack susceptible microstructure have also been studied carefully. Chief among these is the effect of material chemistry. Elements such as sulfur, phosphorous, and silicon can form low-melting constituents during the last stages of solidification or reduce grain boundary surface tension, thereby producing hot cracks. Because austenite solidifies over a larger temperature range than ferrite, these tramp elements segregate more easily into the liquid ahead of an austenitic solidification front. The result is that, after the austenite has completely solidified, a thin film of lower melting composition remains at grain boundaries. As the weld cools further and contracts, this low melting, mushy material is unable to withstand the shrinkage stress and a crack is created (1 - 6). Other researchers have shown the same mechanism to influence hot cracking in alloy systems such as Inconel 600 and Hastelloy X (7, 8). This type of mechanism is most dominant in systems with high levels of impurities and where significant portions of the weld solidify as primary austenite.

The cooling and solidification rates control the solidification structure and extent of segregation. The welding process and procedure affect this, with the
electron beam welding process producing much higher solidification and cooling rates than the GTA process. Vitek and David have shown that high speed laser welds on 308 stainless steel can result in a primary austenite solidification mode when, in fact, this alloy normally solidifies as primary ferrite (9). These welds also displayed intercellular ferrite resulting from the last liquid to freeze, the result of continuous chromium enrichment in the liquid. No cracks were seen in these welds. David, et al, have also shown that the Schaeffler diagram does not adequately describe the effect of rapid solidification rates on the microstructure of stainless steel weld metal (10). According to this study, high speed welds resulted in fully austenitic structures in alloys which normally solidify as duplex austenite plus ferrite structures under conventional arc welding conditions. Although David saw no indication that fully ferritic steels could be made to solidify as primary austenite, the fact is that the Schaeffler diagram does not accurately predict the microstructures which result from rapid solidification.

Although a primary ferrite solidification mode is preferred and is more crack resistant than a primary austenite solidification mode, the reasons are not entirely clear. Hot cracking has been observed in austenitic stainless steels which solidify as primary ferrite and which have high ferrite potential (1, 11). In these cases, no phosphorus or sulfur microsegregation was identified. As a result, there must be more to hot cracking than just microsegregation. Hull has proposed that the difference in interfacial energies between the ferrite and austenite prevents the formation of continuous liquid boundary films resulting from segregation of alloy or impurity elements (12). This means that alloys which solidify as primary ferrite, part of which then transforms to austenite, then have a sufficient amount of this interface to be resistant to the propagation of cracks. Matsuda et. al. have suggested that primary ferrite solidification results in a microstructure with a "tortuous" crack path (13). This results from the peritectic/eutectic reaction in which liquid plus delta ferrite transforms to delta ferrite, after which this transforms to an austenite plus ferrite mixture. This ferrite/austenite mixture has a large amount of irregular boundary structure, which tends to impede crack growth. This would explain why fully austenitic solidification structures exhibit hot cracking even in the absence of detectable solute segregation.

There is some evidence which links voids or microporosity to hot cracking. Lippold has reported the presence of voids in the fully austenitic regions of 304L weld metal in deep penetration EB welds (11). In this case, no solute segregation was found; however, Lippold speculated that since nitrogen is a strong austenitizer, it could have promoted the nucleation and growth of an island of primary austenite in an otherwise primary ferrite solidification structure, particularly at fast solidification rates which can occur
in the center of some EB welds. The presence of a void in the primary austenite region suggests that sufficient gas, possibly nitrogen, had supersaturated that area of limited solubility. Also, Dixon has reported instances of microporosity in conjunction with hot cracks in fully austenitic weld metal (14). Both Lippold and Dixon speculate that microporosity could well be a precursor to hot cracks in these crack susceptible microstructures.

Olson et. al. have proposed that the crack behavior of cored structures and stress corrosion cracking can be related to a thermodynamic model based on the Cahn-Hilliard analysis which describes the degree to which a local surface energy is modified by the presence of a compositional gradient (15, 16). Both stress corrosion cracking and fatigue crack growth data were analyzed and shown to be related to surface energy terms which are strong functions of solute gradients. It is reasonable to think that this mechanism could also contribute to hot cracking models since cracks are often seen at boundaries between areas of primary austenite and primary ferrite solidification, even in alloys of minimal tramp element content. Certainly, the nickel and chromium concentration gradients at these boundaries are large. White has identified interfacial segregation, as opposed to second phase films, associated with hot cracks (17). White's analysis of this phenomena is consistent with the theory proposed by Olson and Matlock.

Clearly, hot cracking is a complex phenomena and all the models and theories described above must be considered to understand it fully. These concepts are a necessary background to understand the work presented herein.

BACKGROUND

At our plant, procurement specifications for "weldable" 304L base metal include numerous requirements which affect hot crack susceptibility. To reduce weld hot crack susceptibility, controls on the base metal chemistry are included, such as limiting the maximum sulfur to 30 ppm, phosphorous to 40 ppm, and silicon to 0.60 weight percent. The specifications also control the DeLong Cr equivalent and Ni equivalent to a predicted FN of 4 to 8 (using the DeLong equation).

Production experience since the late 1970s has shown continuing problems with centerline cracks in deep penetration, highly restrained electron beam welds. The cracks have always been associated with low ferrite materials. Attempts to reduce the cracking problem have all met with little or no success, such as changing the EB gun, using heat sinking, using circle
generation to better mix the weld pool, and changing parameters. The typical solution has been to use materials with adequate ferrite content.

**EXPERIMENTAL PROCEDURE**

In this investigation, the mockups which were utilized were circular disks called "lids" pressed into flat rings called "bodies." A schematic of this arrangement is shown in Figure 1. A circular electron beam weld was produced along the joint of this highly restrained configuration. This weld joint is a mockup of a production weld, but is also similar in design to the circular patch test.

Lids and bodies were made from various heats of 304L, the chemical compositions, DeLong FNs, and Suutala Cr/Ni ratios of which are listed in Table 1. Initial tests were performed on four combinations of annealed and hardened mockups (with ultimate tensile strengths of 85 ksi and 110 ksi). Welds were made between materials with identical FN as well as between materials of mismatched FN, i.e., bodies and lids machined from materials with different ferrite numbers.

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These welds were made on a high voltage 7.5 kW Leybold Heraeus Model 623 EB welder equipped with a CL167R gun with a ribbon filament. The welding parameters were 115 kV, 21 mA, 12 rpm (33 ipm) and a focus condition of 38 mA above sharp focus, meaning that the focal point of the beam was above the weld surface. The mockups were rotated under the beam with an eccentric table.

Each weld was cross sectioned at 8 locations and each transverse cross section was examined metallographically. The crack lengths were measured, and will be reported as average total crack length per weld. No nondestructive testing was performed on these welds. In fact, this joint geometry is not inspectable radiographically. The ferrite number of each weld cross section was measured with a Forster Meter.

EXPERIMENTAL RESULTS

Bar charts will be used to present the average total crack length per weld in inches for various combinations of strength and ferrite numbers. Figure 2 illustrates the results when lids and bodies were both produced from the same heat of zero FN material. When the body was machined from the low strength 85 ksi material, no cracking was observed, regardless of the lid strength level. However, increasing the body (outer component) strength to 110 ksi produced hot cracks. With a low strength lid, approximately 0.100 inch of cracking occurred. Increasing the lid strength level increased the average total crack length nearly four fold. All cracks were transverse (or perpendicular) to the welding direction.

Figure 3 shows the results when both lids and bodies were machined from 8.9 FN material. No cracking was observed under the same strength combinations as in the previous figure.

The worst condition for cracking is when both components are produced from zero FN material in the high strength condition of 110 ksi tensile strength. Increasing the body strength increased the total crack length and all cracks were transverse. Increasing the strength of the lid had a slightly smaller effect on average total crack length than did increasing the strength of the body. These results are consistent with the hoop stress equation which relates, in this case, the stress on a girth weld to the pressure, diameter and thickness of the geometry in question. Increasing the strength of the outer component (the body), which has a larger diameter, has more impact on cracking than increasing the strength of the inner component (the lid).
It is significant to note here that even the 8.9 FN material displayed some tendency to crack. Several microfissures of 0.001 to 0.002 inch in length were detected in the high strength combinations of this material. However, none of these would have been detected with non-destructive testing methods.

Historically, one approach to producing acceptable welds has been to mismatch material, using a high FN material to offset the lack of ferrite in its mate. This concept was also evaluated in this investigation. Welds were produced with the same zero and 8.9 FN materials in the 85 and 110 ksi tensile strength conditions, but matching low to high FN components. Figure 4 illustrates that even though the overall chemistry of the weld would predict a ferrite number of approximately 4.5, the welds were not crack free. When the inner component (the lid) was made of high ferrite material and the outer component was made from zero ferrite material, hot cracks were produced when at least one component was high strength. The amount of hot cracking increased with strength level, and again the strength level of the body had a greater impact. A very important observation from these welds is that the majority of the cracks, 73 % overall, were centerline cracks. From a stress standpoint, this is a worse condition than the horizontal, transverse cracks observed in the zero FN to zero FN welds. When the component ferrite numbers were reversed, cracking was again observed, but the extent was greatly reduced.

Figure 5 illustrates the types of cracking in the high strength mockups just discussed. The weld in Figure 5a was made between a zero FN lid and a zero FN body while the weld in Figure 5b was made between an 8.9 FN lid and a zero FN body. The zero FN to zero FN weld produced transverse cracks along primary austenite dendrite boundaries as the grains grew from the fusion lines toward the weld centerline. In the 8.9 FN to zero FN weld, the high FN material, which is on the left, produced primary ferrite solidification along the left side of the weld. The zero FN material is on the right and produced primary austenite solidification along the right side of the weld. No transverse cracks formed in the primary ferrite region on the left, but transverse cracks did form on the right side, in the region of primary austenite solidification. Long centerline cracks were also produced; these occurred where the grains growing from each side impinged on each other. Lippold found similar results and proposed that a rapid transition in solidification behavior corresponded to a change in growth direction along the weld centerline (1).

Figure 6 is a higher magnification photomicrograph of the centerline crack shown in Figure 5b. This reveals the difference in structure along the weld centerline. The centerline crack formed where the grains of primary
austenite solidification on the right impinged on the grains of primary ferrite solidification on the left. Notice the much finer structure in the primary ferrite region. Figure 7 is a scanning electron micrograph of the crack tip, which reveals the crack path to be along the centerline boundary between the primary ferrite and primary austenite grains.

These photographs indicate that the centerline cracks are produced along the boundary where a large gradient exists in both microstructure and composition. Incomplete mixing has occurred in the mismatched EB weld, resulting in primary ferrite solidification on one side and primary austenite on the other.

Figure 8 further illustrates incomplete mixing in EB welds made with mismatched FN materials. This weld, made between 0.6 FN and 8.4 FN materials, contained 90 percent primary austenite solidification, with some regions of primary ferrite solidification along the high ferrite side, as indicated in this figure. All combinations of low to high ferrite materials revealed extensive lack of mixing. Cracking was always observed to occur in the primary austenite region, or along the boundaries between primary austenite and primary ferrite grains. Solidification mode is obviously very important in determining hot crack resistance.

DISCUSSION

Traditionally, the DeLong Diagram has been used to calculate the ferrite content of austenitic stainless steel weld metal. Materials with FN of 4 or greater are generally accepted for gas tungsten arc welding and considered to be resistant to hot crack formation.

The seven materials used in these EB weld experiments along with four other heats of 304L were used to evaluate the effects of welding speed (that is cooling rate and solidification rate) on FN and solidification mode. The calculated DeLong ferrite numbers ranging from zero to 8.9 are shown along the left axis of Figure 9. Four sets of autogeneous gas tungsten arc welds were produced, increasing the travel speed from 5 to 20 inches per minute. The ferrite numbers produced by these conditions are shown, as measured with the Forster meter. Two different EB welding conditions were also used, the first with a focal point below the sample top surface and the second with the focal point above the top surface, which is the condition of all other EB welds previously reported. Schematic cross sections of these welds are illustrated along the bottom of the figure.
Figure 9 illustrates that the ferrite number decreases with increasing weld travel speed, and thus cooling rate. This is in agreement with the work of David and Vitek (9, 10). In the present case, all welds were also examined metallographically for solidification mode. Two curves have been drawn onto this figure representing the demarcation between different solidification modes. On the bottom is primary austenite solidification only, while on the top is primary ferrite solidification only. Between these two regions is an area where travel speeds produced both primary ferrite and primary austenite solidification. All EB welds within this region, and even one GTA weld made at 20 inches per minute, revealed grains of primary ferrite solidification and also grains of primary austenite solidification.

This figure reveals three key items. First, increasing travel speed reduces ferrite content. GTA welds made at high travel speeds may contain ferrite levels low enough that hot cracking could be a problem. Second, travel speed and cooling rate have a tremendous impact on solidification mode. Increasing travel speed causes a change from primary ferrite solidification to primary austenite solidification because of undercooling, even in GTA welds. Third, at extremely fast welding speeds, regions of the weld can remain unmixed and thereby allow primary ferrite solidification in some areas and primary austenite in others, where hot cracks could form. Dendrite tip undercooling also produces primary austenite solidification. This figure reinforces the work of David, Lippold, and others in that solidification rate must be an axis on a compositional diagram (1, 9 - 11).

Figure 10 illustrates the relationship between average total crack length for the electron beam welds and the DeLong predicted ferrite number. The crack length decreases with increasing ferrite number and the relationship is fairly accurate. However, above the traditional "threshold" ferrite number of 4 (for CTA welds), some cracking was still seen in the 5.2 FN material. A few microfissures of 0.001 to 0.002 inches in length were also observed in the 8.9 FN material. All cracks and microfissures observed in these experiments occurred in primary austenite regions. The threshold ferrite number of 4 is obviously not accurate for EB welds and other high speed welding processes.

Of significant interest is the observation that microstructural differences were noted in an EB weld joining lids and bodies both made from the 5.2 FN material. This cross section, shown in Figure 11, reveals the dark etching, primary ferrite solidification for most of the weld, with some areas of light etching primary austenite solidification. A centerline crack was located along a primary austenite solidification region where the cooling rate produced horizontally growing grains along the centerline. Grains on either side were produced by primary ferrite solidification. This reveals tremendous dendrite
tip undercooling along with lack of mixing and variations in cooling rate.

When the average total crack length is plotted against the Suutala Cr/Ni equivalent ratio (18), a curve similar to that with the DeLong ferrite number is produced and is shown in Figure 12. Cracking decreases with increasing Cr/Ni ratio. Below 1.52 Cr/Ni ratio, primary austenite solidification was observed, while above that value, a mixed solidification mode was seen. Crack-free welds were produced with Cr/Ni ratios of 1.70 and higher.

The most crack susceptible alloy contained a coarser primary austenite microstructure than the least susceptible primary austenite alloy, as illustrated by the photos on the left side of Figure 12. The difference in substructure size could be a key to understanding hot cracking. If the same amount of solute is available, its local concentration will be diminished if present in a finer microstructure. This is because the same amount of solute is distributed over a larger boundary area. This could explain the difference in crack potential between the two primary austenite solidification alloys illustrated. When primary ferrite solidification occurs, the microstructure is typically much finer, as illustrated by the photograph on the right, taken at 4 times the magnification of the other photographs. With the same amount of solute, the boundary area with primary ferrite solidification is dramatically increased, thus the concentration gradient along the boundary is reduced. This idea is further represented by the photomicrograph in Figure 13, showing the top centerline of an EB weld in a 5.2 FN material. The grain with primary austenite solidification has a much coarser structure than the grains with primary ferrite solidification. This may be an explanation as to the drastic differences in hot crack susceptibility between primary austenite and primary ferrite solidification conditions.

Another important item in hot crack resistance which is often overlooked is the type of ferrite. A weld which solidifies as primary austenite may have a measured FN of 1.8 while a weld which solidifies as primary ferrite may have a measured FN of 2.1. Although the ferrite contents are similar, only the region where ferrite formed first (primary ferrite) is resistant to hot cracking. When ferrite is present as a retained eutectic constituent along the cell boundaries, it does not improve crack resistance. Simply measuring the ferrite content of a weld may give a false impression that the weld is hot crack resistant. It is important that the ferrite be present as the primary constituent along the cell cores rather than at the cell boundaries.

Figure 14 plots average total crack length against the total impurity level of sulfur plus phosphorus plus boron. This shows that average total crack length increases with total impurity level, with a dramatic increase beyond
0.020 weight percent. Although the levels of these elements are quite low, there is a good relationship with hot cracking. This also agrees with data accumulated by Brooks, et al (19).

Figure 15 is a scanning electron micrograph showing a microfissure in an EB weld on the 8.9 FN material. This 0.001 inch long microfissure formed along the cell boundary in a small area of primary austenite solidification. Electron microprobe analysis revealed that the cell boundary ahead of the crack (location 12) was enriched in phosphorus 9 times and enriched in sulfur 5 times compared with the cell core (location 5). Segregation of these elements to the cell boundary enhanced hot crack formation.

SUMMARY

Numerous factors influence hot cracking. Solidification mode has been shown to be a major influence, with all cracks being associated with primary austenite solidification. Increasing travel speed and solidification rate shifts solidification mode to primary austenite and increases hot crack susceptibility. A finer microstructure increases the boundary area, reducing concentration gradients and thus hot crack susceptibility. Other factors which may influence hot cracking, especially in mismatched FN EB welds, include a mismatch in thermal expansion coefficients although Elmer showed this effect to be small (20), coalescence of microporosity, surface tension gradients, and chemical gradients.

Chemical segregation is also known to be a major player in hot cracking. Figure 16, a schematic taken from work done by Olson, Liu, and Edwards, illustrates how solute segregates to the interdendritic liquid to a maximum concentration of $c_0/k$ (21). When two dendrites approach each other, the final liquid has a composition equal to 2 times $c_0/k$. Tremendous chemical gradients thus exist along the interdendritic regions. If grains of primary austenite and primary ferrite solidification approach near the centerline, the chemical gradients are compounded. For example, for the case of primary austenite solidification, the distribution coefficient ($k$) for chromium is less than 1, meaning that the liquid has a higher chromium content than the solid. For primary ferrite solidification, the distribution coefficient for chromium is greater than 1 so chromium is depleted in the liquid. When the primary austenite and primary ferrite solidification fronts converge at the centerline, a liquid with an even larger compositional gradient of twice that experienced in the interdendritic spaces is formed. A maximum chemical gradient is produced along the weld centerline. Also, this liquid can have a
drastic gradient in liquid surface tension, possibly causing the liquid to flow out of the centerline regions, producing centerline hot cracks or microporosity. This large chemical gradient may drive tremendous diffusion of the solute elements, thereby producing porosity and possibly hot cracks.

The mechanism proposed here for the centerline cracking observed in the mismatched ferrite welds is that cracking occurs along the region of maximum surface tension gradient. This mechanism is supported and affected by other phenomena, including the fact that primary ferrite solidification cells effectively disperse tramp elements to reduce concentration gradients. Conversely, the primary austenite solidification cells have less cell boundary area and thus higher concentration gradients. This solidification mode promotes transverse cracking. Solidification rate affects both the microstructure type and size, with the finer structure, of either solidification mode, having lower hot crack susceptibility. Surface tension gradients are a function of both tramp elements and the major elements present in the stainless steel; mixing materials may produce dramatic gradients in composition and surface tension.
CONCLUSIONS

(1) Increasing component strength level increases the total crack length.

(2) The body strength level has a larger effect on total crack length than does the lid strength.

(3) Low to high FN welds cannot reliably be made crack free.

(4) Significant lack of mixing exists in mismatched FN EB welds to produce a microstructure which is crack susceptible because it contains regions which are the result of primary austenite solidification.

(5) Solidification rate is a critical variable affecting the threshold ferrite number for hot cracking.

(6) The DeLong FN threshold of 4 for hot crack resistance is not valid for EB welds.

(7) Solidification cell size controls the ability of the microstructure to effectively disperse tramp elements.

(8) Centerline cracking in mismatched FN welds is the result of a maximum surface tension gradient produced by a two-fold increase in concentration gradient in either Cr and Ni and/or in tramp elements.

ACKNOWLEDGMENTS

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REFERENCES


Figure 1. EB Weld Mockup Design

Figure 2 Relationship Between Hot Cracking and Strength Level for Welds Between 0 FN Lids and 0 FN Bodies.

Figure 3 Relationship Between Hot Cracking and Strength Level for Welds Between 8.9 FN Lids and 8.9 FN Bodies.
0.045 T 82 % CL
0.04 T 73 % of Cracks Are
0.035 i1 Centerline (CL)
0.03 i
Average o.o25
0.02 67 % CL
0.015 90 % CL _/
0.01
0.005

Lid - Body Strength Combinations (UTS in ksi)

Figure 4 Relationship Between Hot Cracking and Strength Level
for Welds Between 8.9 FN Lids and 0 FN Bodies.

Figure 5 Crack Morphology in EB Welds on 110 UTS Mockups.

a. Transverse Cracks 0 FN Lid to 0 FN Body
b. Centerline Cracks 8.9 FN Lid to 0 FN Body

0.1 mm
Figure 6 Centerline Crack at Primary Ferrite-Primary Austenite Interface

Figure 7 Scanning Electron Micrograph of Centerline Crack

Figure 8 Incomplete Mixing in EB Weld Between 0.6 FN Lid and 8.4 FN Body. Approximately 90% Primary Austenite.
Figure 9 Relationship Between Welding Speed (Cooling Rate) and Ferrite Number.

Figure 10 Relationship Between Crack Length and Ferrite Number.
Figure 11 Dendrite Tip Undercooling and Incomplete Mixing in EB Weld on 5.2 FN (Calculated) Material.

Figure 12 Relationship Between Crack Length and Suutala Cr/Ni Ratio.
Figure 13 Difference in Substructure Size Between Primary Austenite Grains and Primary Ferrite Grains in 5.2 FN Material.

Figure 14 Relationship Between Crack Length and Impurity Level.
Figure 15 Microfissure in Primary Austenite Solidification Region of 8.9 FN Material. Cell Boundary at Location 12 was Enriched in Phosphorus 9 Times and in Sulfur 5 Times over Cell Core at Location 5.

Figure 16 Compositional Profiles at Interdendritic Regions (21).
1) TITLE - My discussion this afternoon will present information on relationships between hot cracking, ferrite content, and Cr-Ni ratios in electron beam welds in 304L stainless steels.

2) STATEMENT OF PROBLEM - The problem being investigated is that weld hot cracks form during solidification of electron beam welds in 304L. This continues to be a major production problem with welding of austenitic stainless steels.

Numerous factors influence hot crack susceptibility.

Material chemistry has a tremendous impact, and elements such as S, P, and Si can form low melting constituents during the last stages of solidification or reduce grain boundary surface tension, thereby producing hot cracks.

The microstructure and solidification mode are important. In 304L, primary austenite solidification is more prone to hot cracking because of the wider solidification temperature range than during primary ferrite solidification.

The cooling and solidification rates control the solidification structure and extent of segregation. The welding process and procedure affect this, with the electron beam welding process producing much higher solidification and cooling rates than GTA and these are worse for hot cracking.

A certain level of tensile stress on the solidifying grain and cell boundaries is required for hot cracks to form. This can be simply thermal stresses or externally applied loads, all of which enhance hot crack susceptibility.

3) QUESTIONS - There are 5 basic questions I will attempt to answer today.

First:

What is the effect of component part strength on hot cracking?
Can 0 ferrite number to high ferrite number welds be made without formation of cracks?

What is the effect of solidification rate on the ferrite number threshold for cracking, specifically in electron beam welds?

Is the DeLong threshold of 4 ferrite number applicable to electron beam welds? and finally, What is the mechanism by which hot cracking occurs in electron beam welds?

4) At our plant, PROCUREMENT SPECIFICATIONS for "weldable" 304L base metal include numerous requirements which affect hot cracking. To reduce weld hot crack susceptibility, controls on the base metal chemistry are included (such as limiting the maximum sulfur to 30 ppm, P to 40 ppm, and Si to 0.60% ppm). The specifications also control the DeLong Cr equivalent and Ni equivalent to a predicted ferrite number of 4 to 12 (using the DeLong equation). However, the same specifications have base metal requirements which are "contradictory" to the weld requirements. If the weld ferrite content is above approximately 10 ferrite number, the base metal will contain ferrite stringers, which reduces toughness and also enhances sigma transformation.
Therefore, a maximum base metal ferrite content of 3% (Severn gage) is required. If ferrite is present in the base metal, steel mills could anneal the alloy to resolutionize the ferrite. However, another contradictory requirement is a maximum grain size in the base metal of ASTM 5. This restricts any annealing treatment which can be applied to reduce base metal ferrite content.

A compromise to these requirements is to produce the steel to a predicted weld ferrite number of 4 to 8 to provide acceptable weld ferrite content and hot crack resistance and also to reduce the base metal ferrite content. All of these are at the limit of stainless steel technology and these requirements are not easily met by steel mills.

5) DeLong Diagram. The specification chemistry limits define a large box on the DeLong Diagram, as shown. For weldability purposes, the ferrite content is desired to be in the 4 to 12 range, as illustrated, which reduces the size of the chemistry range. The compromise solution for both base metal and weld metal is a ferrite number of 4 to 8, which dramatically tightens up the chemistry range permissible.

6) BACKGROUND - Production experience since the late 1970s has shown continuing problems with centerline cracks in deep penetration, highly restrained electron beam welds. The cracks have always been associated with low ferrite materials. Attempts to reduce the cracking problem have all met with only limited success, such as changing the EB gun, using heat sinking, circle generation to better mix the weld pool, and changing parameters. The typical solution has been to use materials with adequate ferrite content.

7) LID-TO-BODY ELECTRON BEAM WELD MOCKUP - In this present investigation, mockups were utilized which consisted of a lid pressed into a body. A circular electron beam weld was produced along the joint of this highly restrained configuration. This weld joint is a mock up of a production weld, but is also similar in design to the circular patch test. Lids and bodies were made from various heats of 304L. Initial tests were performed on 4 combinations of annealed and hardened mockups (with ultimate tensile strengths of 85 ksi and 110 ksi). Various combinations of ferrite materials were also evaluated, including bodies and lids machined from the same material and also bodies and lids machined from materials with different ferrite numbers.

8) PROCEDURE - A high voltage 7.5 kW Hamilton-Standard EB welder was used with a CL167R gun and ribbon filament. The welding parameters are listed (115 kV, 21 mA, at 12 rpm [33 ipm]), the most important being that the electron beam was focused above the sample surface (above sharp by 38 mA). (A 2 sec. slope in and 3 sec. slope out with 8 sec. weld time were utilized). These welding conditions produced a narrow weld bead.

Each weld was cross sectioned at 8 locations and each transverse cross section was examined metallographically. The crack lengths were measured, and will be reported as average total crack length per weld. No nondestructive testing, such as radiography, was performed on any welds. The ferrite number of each weld cross section was measured with a Foerster Meter.
9) 0 FN LID - 0 FN BODY - Bar charts will be used to present the average total crack length per weld (in inches) for various combinations of strength and ferrite numbers. This graph illustrates the results when lids and bodies were both produced from the same heat of 0 ferrite number material. When the body was machined from the low strength 85 ksi material, no cracking was observed, regardless of the lid strength level of 85 ksi or 110 ksi. However, increasing the body (which is the outer component) strength to 110 ksi produced hot cracks. With a low strength lid, approximately 0.100 inch of cracking occurred. Increasing the lid strength level increased the average total crack length nearly four fold. All cracks were transverse (or perpendicular) to the welding direction.

10) 8.9 FN LID - 8.9 FN BODY - When both lids and bodies were machined from 8.9 ferrite number material, no cracking was observed under the same strength combinations as in the previous slide.

11) EFFECT OF STRENGTH LEVEL - The worst condition for cracking is when both components are produced from 0 ferrite number material in the high strength condition of 110 ksi tensile strength. Increasing the body strength increased the total crack length, and all cracks were transverse. Increasing the strength of the lid had a slightly smaller effect than that of the body. These results are consistent with the hoop stress equation which relates the stress to the pressure, diameter and thickness. Increasing the strength of the outer component (the body), which has a larger diameter, has more impact than increasing the strength of the inner component (the lid).

There is a minimal effect when both components are produced from 8.9 ferrite number material, regardless of strength level. Several microfissures of 0.001 to 0.002 inch in length were detected in the high strength combinations of this material. However, none of these would have been detected with non-destructive testing methods.

In the past, one approach to producing acceptable welds was to mismatch material. If high ferrite number components were welded to low ferrite number components, would the resultant weld be free of hot cracks?

12) To answer this question, welds were produced with the same 0 and 8.9 ferrite number materials in the 85 and 110 ksi tensile strength conditions, but matching low to high ferrite components. This graph illustrates that even though the overall chemistry of the weld would predict a ferrite number of approximately 4.5, the welds were not crack free. When the inner component (the lid) was made of high ferrite material and the outer component made from zero ferrite 304L, hot cracks were produced when at least one component was high strength. The amount of hot cracking increased with strength level, and again the strength level of the body had a greater impact.

A very important observation from these welds is that the majority of the cracks, 73% overall, were centerline cracks. From a stress standpoint this is a worse condition than the horizontal, transverse cracks observed in the 0 to 0 ferrite welds.

When the component ferrite numbers were reversed, cracking was again observed, but the extent was greatly reduced.
13) HIGH STRENGTH MOCKUPS - These photographs illustrate the types of cracking in high strength mockups just discussed. The weld on the left was made on low ferrite lid to low ferrite body (both of which had 0 ferrite number) while the weld on the right was made with a lid of a high ferrite number of 8.9 and a 0 ferrite number body. The 0 to 0 ferrite weld produced transverse cracks along dendrite boundaries as the grains grew from the fusion lines toward the weld centerline. The 8.9 to 0 ferrite weld produced some transverse cracks, but also long centerline cracks where the grains growing from each side impinged on each other. Lippold found similar results and proposed that a rapid transition in solidification behavior corresponded to a change in growth direction along the weld centerline.

From a stress standpoint these cracks are much worse than the transverse cracks.

14) CENTERLINE CRACKS IN MISMATCHED MATERIALS - These photographs are of the same high to low ferrite weld and illustrate lack of mixing. The high ferrite number material was on the left and produced primary ferrite solidification along the left side of the electron beam weld. The 0 ferrite material was on the right and produced primary austenite solidification along the right side of the weld.

No transverse cracks formed in the primary ferrite region on the left, but transverse cracks did form on the right side, in the region of primary austenite solidification.

A higher magnification photomicrograph of the centerline crack reveals the difference in structure along the weld centerline. The centerline crack formed where the grains of primary austenite solidification on the right impinged on the grains of primary ferrite solidification on the left. Notice the much finer structure in the primary ferrite region.

A scanning electron micrograph of the crack tip reveals the crack path to be along the centerline boundary between the primary ferrite and primary austenite grains.

These photographs indicate that the centerline cracks are produced along the boundary where a large gradient exists in microstructure and chemistry. Incomplete mixing has occurred in this mismatched electron beam weld, resulting in primary ferrite solidification on one side and primary austenite on the other.

15) EB WELDS IN MISMATCHED MATERIALS - These weld cross sections reveal incomplete mixing in electron beam welds made with mismatched materials. The 0 ferrite to 8.9 ferrite weld in this case is comprised of nearly 95% primary ferrite, with only a narrow region of primary austenite along the low ferrite side, as shown below (there were no centerline cracks in this cross section). Both welds made with 0.6 to 8.4 ferrite materials contained 85 to 90% primary austenite solidification, with some regions of primary ferrite solidification along the high ferrite side, as indicated in this bottom photomicrograph. All combinations of low to high ferrite materials revealed extensive lack-of-mixing.
Cracking was always observed to occur in the primary austenite region, or along the boundaries between primary austenite and primary ferrite grains. Solidification mode is very important in determining hot crack resistance.

16) EFFECTS OF WELDING SPEED ON FERRITE NUMBER - Traditionally, the DeLong Diagram has been used to calculate the ferrite content of austenitic stainless steel weld metal. Materials with ferrite numbers of 4 or greater are generally accepted for gas tungsten arc welding without hot crack formation.

The seven materials used in these electron beam weld experiments along with four other heats of 304L were used to evaluate the effects of welding speed (that is cooling rate and solidification rate) on ferrite number and solidification mode. The calculated DeLong ferrite numbers ranging from 0 to 8.9 are shown along the left axis. Four sets of autogenous gas tungsten arc welds were produced, increasing the travel speed from 5 to 20 inches per minute. The ferrite numbers produced by these conditions are illustrated, as measured with the Foerster meter. Two different electron beam welding conditions were also used, the first with a focal point below the sample top surface and the second with the focal point above the top surface, which is the condition of all other electron beam welds previously reported. Schematic cross sections of these welds are illustrated along the bottom of the figure.

This figure illustrates that the ferrite number decreases with increasing weld travel speed, and thus cooling rate. This is in agreement with the work of David and Vitek. In the present case, all welds were also examined metallographically for solidification mode. Two curves have been drawn onto this figure representing the demarcation between different solidification modes. On the bottom is primary austenite solidification only, while on the top is primary ferrite solidification only. Between these two regions is an area where welds revealed both primary ferrite and primary austenite solidification. All electron beam welds within this region, and even one GTA weld made at 20 inches per minute, revealed grains of primary ferrite solidification and also grains of primary austenite solidification.

This graph reveals three key items. First, increasing travel speed reduces ferrite content. GTA welds made at high travel speeds may contain ferrite levels low enough that hot cracking could be a problem. Second, travel speed and cooling rate have a tremendous impact on solidification mode. Increasing travel speed causes a change from primary ferrite solidification to primary austenite solidification because of undercooling, even in GTA welds. Third, at extremely fast welding speeds, regions of the weld can remain unmixed and thereby allow primary ferrite solidification in some areas and primary austenite in others, where hot cracks could form. Dendrite tip undercooling also produces primary austenite solidification. This figure reinforces the work of David, Lippold and others in that solidification rate must be an axis on a compositional diagram.

17) ELECTRON BEAM WELDS - Electron beam welds made on three materials with DeLong calculated ferrite numbers of 8.9, 8.4 and 0 are shown in this slide. Measurements of GTA welds made at 5 inches per minute on the same materials revealed the ferrite numbers shown. Notice that these numbers are lower than the DeLong calculated numbers and the center sample has a ferrite number just below the "threshold" of 4. The electron beam welds on the materials with
8.9 and 8.4 calculated ferrite numbers were crack free, but the 0 ferrite number material contained transverse cracks.

18) SOLIDIFICATION STRUCTURES - The substructures of these three welds reveals primary ferrite solidification for both the 8.9 and 8.4 ferrite materials, but primary austenite solidification for the 0 ferrite materials.

19) AVERAGE TOTAL CRACK LENGTH - This graph illustrates the relationship between average total crack length for the electron beam welds and DeLong predicted ferrite number. The crack length decreases with increasing ferrite number, and the relationship is fairly accurate. However, above the traditional "threshold" ferrite number of 4 for GTA welds, some cracking was observed in 5.2 ferrite number material. A few microfissures of 0.001 to 0.002 inches in length were also observed in the 8.9 ferrite number material. The threshold ferrite number of 4 may not be accurate for electron beam welds.

20) CENTERLINE CRACK IN 5.2 FN MATERIAL - When 5.2 FN material was welded to itself, differences in structure were also observed. This cross section reveals the dark etching, primary ferrite solidification for most of the weld, with some areas of light etching primary austenite solidification. A centerline crack was located along a primary austenite solidification region where the cooling rate produced horizontally growing grains along the centerline. Grains on either side were primary ferrite solidification. This reveals tremendous undercooling along with lack of mixing and variations in cooling rate. All cracks observed in these experiments occurred in primary austenite regions.

21) AVERAGE TOTAL CRACK LENGTH - When the average total crack length is plotted against the Suutala Cr/Ni equivalent ratio, a curve similar to that with the DeLong ferrite number is produced. Cracking decreases with increased Cr/Ni ratio. Below 1.52 Cr/Ni ratio, primary austenite solidification was observed, while above that value a mixed mode was seen. Crack-free welds were produced with Cr/Ni ratios of 1.70 and higher.

The most crack susceptible alloy contained a coarser primary austenite microstructure than the least susceptible primary austenite alloy, as illustrated by the photos on the left. The difference in substructure size could be a key in understanding hot cracking. If the same amount of solute is available, its concentration will be diminished if present in a finer microstructure. This is because the same amount of solute is distributed over a larger boundary volume. This could explain the difference in crack potential between the two primary austenite solidification alloys illustrated. When primary ferrite solidification occurs, the microstructure is typically much finer, as illustrated by the photograph on the right, taken at 4 times the magnification of the other photographs. With the same amount of solute, the boundary volume with primary ferrite solidification is dramatically increased, thus the concentration gradient along the boundary is reduced. This may be a plausible explanation as to the drastic differences in hot crack susceptibility between primary austenite and primary ferrite solidification conditions.

22) 5.3 FN MATERIAL - This idea is further represented by this photomicrograph along the top centerline of an EB weld in a 5.3 ferrite number material. The grains with primary austenite solidification have a
much coarser structure than the grains with primary ferrite solidification. Another important item in hot crack resistance which is often overlooked is the ferrite content. The primary austenite region has a measured ferrite content of 1.8 while the primary ferrite region has a measured ferrite number of 2.1. Although the ferrite contents are similar, only the region where ferrite formed first is resistant to hot cracking. When ferrite is present as a retained eutectic constituent along the cell boundaries, it does not improve crack resistance. Simply measuring the ferrite content of a weld may give a false impression that the weld is hot crack resistant. It is important that the ferrite be present as the primary constituent along the cell cores rather than at the cell boundaries.

23) AVERAGE TOTAL CRACK LENGTH - Average total crack length plotted against the total impurity level of sulfur plus phosphorus plus boron reveals an increasing relationship, with a dramatic increase beyond 0.020%. Although the levels of these elements are very low, there is a good relationship with hot cracking.

24) MICROFISSURE IN 8.9 FN 304L - This scanning electron micrograph reveals a microfissure in an electron beam weld on the 8.9 ferrite number 304L. This 0.001 in. long microfissure formed along the cell boundary in a small area of primary austenite solidification. Electron microprobe analysis revealed that the cell boundary ahead of the crack was enriched in phosphorus 9 times and enriched in sulfur 5 times compared with the cell core. Segregation of these elements to the cell boundary enhanced hot crack formation.

25) FACTORS THAT INFLUENCE CRACKING - Numerous factors influence hot cracking. Solidification mode has been shown to be a major influence, with all cracks being associated with primary austenite solidification. Increasing travel speed and solidification rate shifts solidification mode to primary austenite and increases hot crack susceptibility. A finer microstructure increases the boundary area, reducing concentration gradients and thus hot crack susceptibility. Other factors which may influence hot cracking, especially in mismatched ferrite electron beam welds, include a mismatch in thermal expansion coefficients (although Elmer showed this to have little effect). Coalescence of microporosity, surface tension gradients, and chemical gradients also may be major players in hot crack formation.

26) MICROPOROSITY IN WELD IN 0 FN MATERIAL - Some evidence of microporosity along cell boundaries, along with hot cracks, is shown here on the 0 ferrite material. Dixon proposed that hot cracks may be the coalescence of microporosity along the boundaries. Results of this work support that theory.

27) SEGREGATION - This schematic, taken from work by Olson, Liu and Edwards, illustrates how solute segregates to the interdendritic liquid to a maximum concentration of $c_0/k$. When two dendrites approach each other, the final liquid has a composition equal to 2 times $c_0/k$. Tremendous chemical gradients thus exist along the interdendritic regions.

Now, if grains of primary austenite and primary ferrite solidification approach near the centerline, the chemical gradients are compounded. For example, for the case of primary austenite solidification, the distribution coefficient ($k$) for chromium is less than 1, meaning that the liquid has a
higher chromium content than the solid. For primary ferrite solidification, the distribution coefficient for chromium is greater than 1 so chromium is depleted in the liquid.

When the primary austenite and primary ferrite solidification fronts converge at the centerline, a liquid with an even larger compositional gradient of twice that experienced in the interdendritic spaces is formed. A maximum chemical gradient is produced along the weld centerline. Also, this liquid can have a drastic gradient in liquid surface tension, causing the liquid to flow out of the centerline regions, producing centerline hot cracks.

This large chemical gradient may allow tremendous diffusion of the solute elements, thereby producing porosity and possibly hot cracks.

28) ENERGY - This figure (from Olson and others) illustrates the energy required to form a new surface, such as a crack. The reference energy state is for a homogeneous material. A large amount of energy is required to form a new surface in this material, as shown on the left. However, if the material has a compositional gradient present in it, the reference energy state is increased to that shown on the right. The increase in energy was proposed by Cahn and Hilliard as the formula shown, which includes the compositional gradient, \( dc/dx \), but as a squared term. Obviously, as the compositional gradient increases, the reference energy state increases significantly. Therefore, the energy required to form a new surface is drastically decreased, as shown in the top right.

Olson and others have shown that liquid metal embrittlement and stress corrosion cracks form where the surface tension gradient is a maximum. In the unmixed electron beam welds where a primary austenite solidification front impinges on a primary ferrite solidification front at the centerline, a maximum compositional and surface tension gradient exists where the liquids meet. This increases the reference energy state dramatically, thereby reducing the energy required to produce a new surface, or crack. Thus, maximum compositional and surface tension gradients may be the cause of hot cracking in unmixed, mismatched electron beam welds.

29) PROPOSED MECHANISM - The mechanism we propose for the centerline cracking observed in the mismatched ferrite welds is that cracking occurs along the region of maximum surface tension gradient.

Other proposals include the primary ferrite solidification grains effectively disperse tramp elements to reduce concentration gradients. Conversely, the primary austenite solidification has less cell boundary area and thus higher concentration gradients. This solidification mode promotes transverse cracking.

Solidification rate affects both the microstructure type and size, with the finer structure having lower hot crack susceptibility.

Surface tension gradients are a function of both tramp elements and the major elements present in the stainless steel and mixing materials may produce dramatic gradients in composition and surface tension.
30) CONCLUSIONS - The other conclusions from this investigation are that increasing component strength level increases the total crack length.

The body strength level has a larger effect on total crack length than does the lid strength.

Low to high ferrite welds cannot reliably be made crack free.

Solidification rate is a critical variable affecting the threshold ferrite number for cracking.

As a result, the DeLong ferrite number threshold of 4 is not valid for electron beam welds.

31) FUTURE WORK - Future work is needed to complete this investigation. Microprobe work must be continued to identify the elements associated with cracks and also the chemical gradients present.

The effects of the Cr/Ni ratio and the sulfur plus phosphorus plus boron relationships in conjunction with the various ferrite number equations must be evaluated. This should include a third dimension on all diagrams incorporating solidification rate.

Finally, other combinations of ferrite numbers must be evaluated. If 0 to 8 ferrite combinations produces centerline cracks, will reducing the difference between the materials help, such as with 3 to 7 or 5 to 9 ferrite numbers.

Thank you. Are there any questions.
Microstructural and Solidification Cracking Evaluation of Electron Beam Welds in 304L

P. L. Sturgill

R. D. Campbell, J. L. Henningsen

EB&G ROCKY FLATS
Statement of Problem

Hot cracks form during solidification of electron beam welds in 304L

Factors Influencing Hot Crack Susceptibility

1) Material Chemistry
2) Microstructure & Solidification Mode
3) Cooling & Solidification Rates
4) Stress Level
Questions

- What is the effect of component strength on hot cracking?
- Can 0 FN to 8.9 FN welds be made crack free?
- What is the effect of solidification rate on FN threshold for cracking?
- Is the DeLong threshold of 4 FN applicable to EB welds?
- What is the mechanism by which hot cracking occurs in EB welds?
Procurement Specifications for "Weldable" 304L

• To reduce weld hot crack susceptibility:
  - Control chemistry (S, P, Si)
  - Control DeLong CrEq & NiEq to predicted FN of 4 to 12

• "Contradictory" base metal requirements:
  - Maximum base metal ferrite content (3%)
  - Maximum base metal grain size

• Compromise: Produce to predicted FN of 4 to 8
  - Meets weld ferrite requirements
  - Reduces base metal ferrite stringers

• Limit of Stainless Steel Technology
DeLong Diagram

Cr Eq = %Cr + % Mo + 1.5 x % Si

Ni Eq = % Ni + 30 x % C + 30 x % N + 0.5 x % Mn

Austenite
Ferrite
Chemistry Limits
Compromise
Desired
Scheffler
A x M Line
Plus Ferrite
Background

- Centerline cracks seen in deep penetration, highly restrained EB welds
- Cracks associated with low ferrite materials
- Some techniques tried, but failed
  - change R-40 EB gun to R-32
  - use heat sinking
  - circle generation
  - change parameters
- Typical fix: use adequate ferrite material
Lid-To-Body Electron Beam Weld Mockup

Body

Lid

- 1.5"
- 0.866"
Procedure

- High voltage 7.5 kW EB welder
- CL167R gun with ribbon filament
  115kV  21 mA  33 ipm
  +38 mA above sharp focus
- Section each weld at eight locations
- Examine welds metallographically
- Measure crack lengths
- No NDT
0 FN Lid - 0 FN Body

Average Total Crack Length (in.)

All Cracks Are Transverse

Lid - Body Strength Combinations (UTS in ksi)
8.9 FN Lid - 8.9 FN Body

No Cracks Observed in These Welds

Average Total Crack Length (in.)

Lid - Body Strength Combinations (UTS in ksi)
Effect Of Strength Level

- Worst condition for cracking is 0 FN and 110 ksi UTS in both components
  - increasing body UTS increases TCL - all transverse
  - increasing lid UTS has slightly smaller effect
  - consistent with hoop stress equation: \( S = \frac{pd}{2t} \)
- Minimal effect when both components are 8.9 FN
8.9 FN Lid - 0 FN Body

73% of TCL is CL

Average Total Crack Length (in.)

85-85
110-85
85-110
110-110

Lid - Body Strength Combinations (UTS in ksi)
High Strength Mockups

Low Ferrite to Low Ferrite
Transverse Cracks
0 to 0

High Ferrite to Low Ferrite
Centerline Cracks
8.9 to 0

0.1 in.
Centerline Cracks in Mismatched Materials
8.9 FN Lid to 0 FN Body

Primary Ferrite | Primary Austenite | Primary Ferrite | Primary Austenite | Primary Ferrite | Primary Austenite

0.01 in | 0.01 in | 0.001 in
EB Welds in Mismatched Materials
Incomplete Mixing

0 FN to 8.9 FN
95% Primary Ferrite

0.6 FN to 8.4 FN
90% Primary Austenite

8.4 FN to 0.6 FN
85% Primary Austenite

Primary Austenite

Primary Ferrite

Primary Ferrite

Primary Austenite

Primary Ferrite

Primary Austenite

0.1 in.

0.001 in.
Solidification Structures

DeLong EN
Solidification Structure

Primary Ferrite
Primary Ferrite
Primary Austenite

0.001 in.
AVERAGE TOTAL CRACK LENGTH
Like-to-Like Ferrite Levels

Traditional "Threshold" for GTA Welds

Some Cracks

A Few Microfissures
Centerline Crack in 5.2 FN Material

Cracks only in Primary Austenite Region
AVERAGE TOTAL CRACK LENGTH

Like-to-Like Ferrite Levels

Primary Austenite
Primary Ferrite & Primary Austenite

Even Finer Microstructure
0.001 in.
(4X Other Photos)

Suutala Cr/Ni Ratio

Coarser Microstructure 0.01 in.

Finer Microstructure 0.01 in.
5.3 FN Material

Primary Austenite - Coarser Microstructure

Measured 1.8 FN

Primary Ferrite - Finer Microstructure

Measured 2.1 FN

0.001 in.
Microfissure in 8.9 FN 304L
Primary Austenite Solidification Region

0.001 in

Cell Boundary Ahead of Crack Compared with Cell Core:

Enriched in Phosphorus 9X  Enriched in Sulfur 5X
Factors That Influence Cracking

• Solidification Mode
• Solidification Rate
• Fine microstructure / increased boundary area
• Thermal expansion coefficient mismatch
• Coalescence of microporosity
• Surface tension gradient
• Chemical gradient
Microporosity in Weld in 0 FN Material

Hot Cracks and Porosity
Along Dendrite Boundaries
New surface

Specific surface energy \( \gamma_{\text{hom}} \)

Energy

Specific surface energy \( \gamma_{\text{hmr}} \)

Heterogeneous material reference energy state

Homogeneous material reference energy state

\[
2 N v \int \left[ G \left( \frac{dc}{dx} \right)^2 \right] dx
\]
Proposed Mechanism

- Cracking occurs at maximum surface tension gradient
  - primary ferrite solidification effectively disperses tramp elements to reduce gradient
  - primary austenite solidification has less cell boundary area: promotes transverse cracks
  - solidification rate affects both microstructure type, size, and fineness
- Surface tension gradient a function of both tramp elements and Cr, Ni, Mn
Conclusions

- Increasing component strength level increases TCL
- Body UTS level has a larger effect on TCL than lid UTS
- 0 FN to 8.9 FN welds cannot reliably be made crack free
- Solidification rate is a critical variable affecting the threshold FN for cracking
- DeLong FN threshold of 4 is NOT VALID for EB welds
Future Work

- Continue microprobe work to identify
  - elements associated with cracks
  - chemical gradient
- Separate effects of Cr/Ni ratio and S+P+B
- Examine 3 FN to 7 FN and 5 FN to 9 FN welds