Characterization of Heat-Affected Zone Cracking in Austenitic Stainless Steel Welds

Conditions that initiate HAZ cracking are counteracted by the formation of primary ferrite

V. P. KUJANPÄÄ, S. A. DAVID AND C. L. WHITE

ABSTRACT. Heat-affected zone (HAZ) cracking in austenitic stainless steel welds was investigated using pure laboratory and commercial heats. The formation of HAZ cracks was found to be associated with solute segregation to grain boundaries, mode of solidification, and level of purity. These observations indicate that HAZ cracking is strongly influenced by interfacial energy considerations.

Introduction

The problem of hot cracking has been known to occur during welding of austenitic stainless steels. Two manifestations of hot cracks during welding of austenitic stainless steels are solidification cracking and heat-affected zone cracking. Numerous studies during the last 30 years have addressed this subject (Refs. 1-6). Most of these investigations have concentrated on solidification cracking.

The history of HAZ cracks is different from that of solidification cracks (Refs. 7-9). The HAZ in these steels normally consists of homogenized base metal (with the exceptions of multipass welds and repair welds), which is exposed to a weld thermal cycle. The peak temperature experienced in the HAZ decreases with distance from the fusion line. The phase changes, if any, are mostly solid state. In addition, temperatures and cooling rates experienced by the HAZ are lower than those experienced by the fusion zone (Refs. 10, 11). Many kinds of HAZ cracks have been observed (Ref. 12). In the partially melted zone, liquation cracks may occur. In the regions of the HAZ subjected to lower peak temperatures, there is a possibility of ductility-dip cracks. Both of these types of cracks can occur in the HAZ of a single-pass weld or in the earlier passes of a multipass weld.

Solidification cracking in austenitic stainless steel welds using both high-purity model ternary alloys and commercial alloys was investigated in an earlier work (Ref. 13). In the present study, HAZ cracking in these same austenitic stainless steels has been characterized, and the mechanisms of its formation are discussed.

Experimental Procedure

Materials selected for the investigation were austenitic stainless steels. Both laboratory and commercial heats, whose compositions are listed in Table 1, were used. Alloy compositions were chosen to...
investigate both the effect of impurities and the mode of solidification on the HAZ cracking susceptibility of austenitic stainless steel welds. Commercial heats were of types AISI 304, 309, 310, and 316 and one commercial non-AISI type containing 20% Cr, 25% Ni, 6% Mo, and 0.2% N, designated AL6XN. Laboratory heats A to E are corresponding ternary Fe-Cr-Ni alloys. For the laboratory heats, zone-refined Fe, Ni, and Cr were used. A detailed description of the preparation of laboratory heats and samples for the spot Varestraint test can be found in the earlier work (Ref. 13).

The spot Varestraint test was used to study the HAZ cracking susceptibility of the welds (Ref. 15). In the test, the specimen is subjected to a gas tungsten arc spot weld thermal cycle for a sufficient time to establish approximately steady-state thermal conditions. As the arc current is interrupted, an augmented strain is applied to the specimen by bending. Depending on the delay between extinguishing the arc and the start of bending, either solidification cracks or HAZ cracks can be formed. An approximate delay time of 0.3 s was used for HAZ cracking. The tangential strain on the top surface of the specimen, ε, is a function of specimen thickness, t, and bending radius, R, and can be calculated from:

\[ \epsilon = \frac{t}{2R} \]

Specimens from each heat were tested at strains from 0.9 to 4.9%. The other variables were kept constant and are shown in Table 2. In this study, the individual crack lengths in the HAZ are summed up and the sum is designated total crack length (TCL) at a given strain. Optical microscopy, scanning electron microscopy (SEM), electron probe microanalysis (EPMA), and Auger electron spectroscopy (AES) were used to characterize the microstructures and HAZ cracks. AES specimens containing completely enclosed HAZ cracks were sectioned from selected specimens. In order to avoid contamination of the crack surfaces, and resulting artifacts in the Auger spectra, it is important that the cracks analyzed using AES not be connected to the external surface. The details of special specimen preparation techniques for the various analyses are described elsewhere (Ref. 13). Ferrite numbers of the welds were measured using a Magne-Cage calibrated according to the AWS procedure A4.2-74.

### Table 1—Compositions of Materials Tested (wt-%)

<table>
<thead>
<tr>
<th>Heat</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>Cu</th>
<th>N</th>
<th>Ti</th>
<th>Nb</th>
<th>Cr_{eq}/N_{eq} (a)</th>
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<tbody>
<tr>
<td>A</td>
<td>0.004</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.005</td>
<td>0.007</td>
<td>20.0</td>
<td>16.1</td>
<td>0.02</td>
<td>0.01</td>
<td>0.001</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>1.24</td>
</tr>
<tr>
<td>B</td>
<td>0.006</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.005</td>
<td>0.006</td>
<td>19.8</td>
<td>10.9</td>
<td>0.01</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>1.80</td>
</tr>
<tr>
<td>C</td>
<td>0.005</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.005</td>
<td>0.008</td>
<td>26.0</td>
<td>25.4</td>
<td>0.01</td>
<td>0.02</td>
<td>0.001</td>
<td>0.01</td>
<td>&lt;0.01</td>
<td>1.02</td>
</tr>
<tr>
<td>D</td>
<td>0.003</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.005</td>
<td>0.007</td>
<td>25.7</td>
<td>20.0</td>
<td>0.01</td>
<td>0.02</td>
<td>0.001</td>
<td>0.01</td>
<td>&lt;0.01</td>
<td>1.28</td>
</tr>
<tr>
<td>E</td>
<td>0.006</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.005</td>
<td>0.007</td>
<td>25.7</td>
<td>15.3</td>
<td>0.01</td>
<td>0.01</td>
<td>0.003</td>
<td>&lt;0.01</td>
<td>0.01</td>
<td>1.67</td>
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<tr>
<td>304</td>
<td>0.079</td>
<td>0.45</td>
<td>1.67</td>
<td>0.026</td>
<td>0.015</td>
<td>18.6</td>
<td>8.5</td>
<td>0.13</td>
<td>0.17</td>
<td>0.004</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>1.77</td>
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<tr>
<td>309</td>
<td>0.020</td>
<td>0.42</td>
<td>2.01</td>
<td>0.015</td>
<td>0.008</td>
<td>23.2</td>
<td>14.2</td>
<td>0.05</td>
<td>0.04</td>
<td>0.029</td>
<td>0.01</td>
<td>&lt;0.01</td>
<td>1.52</td>
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<tr>
<td>310</td>
<td>0.036</td>
<td>0.45</td>
<td>0.96</td>
<td>0.019</td>
<td>0.009</td>
<td>25.7</td>
<td>19.1</td>
<td>0.15</td>
<td>0.25</td>
<td>0.023</td>
<td>0.02</td>
<td>&lt;0.01</td>
<td>1.28</td>
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<tr>
<td>316</td>
<td>0.039</td>
<td>0.29</td>
<td>1.94</td>
<td>0.026</td>
<td>0.012</td>
<td>16.8</td>
<td>11.5</td>
<td>2.33</td>
<td>0.36</td>
<td>0.021</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>1.52</td>
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<tr>
<td>AL6XN</td>
<td>0.022</td>
<td>0.43</td>
<td>1.54</td>
<td>0.026</td>
<td>0.007</td>
<td>20.7</td>
<td>24.8</td>
<td>6.24</td>
<td>0.07</td>
<td>0.21</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>1.04</td>
</tr>
</tbody>
</table>

(a) Calculated according to Hammar and Svensson (Ref. 14).

### Table 2—Parameters Used in Spot Varestraint Tests

| Method   | Gas tungsten arc       |
| Position | Flat                    |
| Polarity | DCEN                    |
| Current  | 75 A                    |
| Arc length | 3.2 mm (1/8 in.)       |
| Electrode | EWTh-2, diam. 2.4 mm (9% in.), cone angle 10 deg |
| Arc time  | 23 s                    |
| Shielding gas | 99.99% Ar, flow rate 14 L/min (30 cfh) |
| Applied strain | 0.9, 2.9, or 4.9%       |

### Table 3—Solidification Modes and Ferrite Contents of the Welds of Different Materials

<table>
<thead>
<tr>
<th>Material</th>
<th>Solidification Mode (b)</th>
<th>Ferrite Number (c)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>A</td>
<td>0.0</td>
</tr>
<tr>
<td>B</td>
<td>FA</td>
<td>3.1-3.9</td>
</tr>
<tr>
<td>C</td>
<td>FA</td>
<td>5.2-5.5</td>
</tr>
<tr>
<td>D</td>
<td>AF</td>
<td>1.5-1.7</td>
</tr>
<tr>
<td>E</td>
<td>FA</td>
<td>26-27</td>
</tr>
<tr>
<td>304</td>
<td>FA</td>
<td>2.9-3.1</td>
</tr>
<tr>
<td>309</td>
<td>FA</td>
<td>12-14</td>
</tr>
<tr>
<td>310</td>
<td>AF</td>
<td>1.2-1.4</td>
</tr>
<tr>
<td>316</td>
<td>FA (AF) (d)</td>
<td>2.6-2.8</td>
</tr>
<tr>
<td>AL6XN</td>
<td>A</td>
<td>0.0</td>
</tr>
</tbody>
</table>

(b) Measured by Magne-Cage.
(c) Ten to thirty percent of the grains in the welds had solidified in AF mode.
(d) A = single-phase austenitic; AF = austenitic-ferritic; FA = ferritic-austenitic.

### Results and Discussion

#### Microstructure

The solidification mode of each weld was assessed on the basis of its microstructure. Table 3 summarizes the modes of solidification observed in various alloy systems used in this investigation. Alloys A, C, D, 310, and AL6XN were found to exhibit a primary austenitic mode of solidification, and Alloys B, E, 304, and 309 exhibited a primary ferritic mode of solidification, as shown in Figs. 1A and 1B, respectively. The Type 316 weld had solidified predominantly in the primary ferritic mode; however, some regions were determined to have solidified in a primary austenitic mode. The measured ferrite numbers on various welds are tabulated in Table 3. Solidification substructure of the welds was found to be mostly cellular or cellular dendritic, as shown in Fig. 2. However, a small fraction of the weld along the fusion line contained evidences of planar growth. This observation was most prevalent in the pure laboratory heats.

Except for Alloys 309 and E, which contained a small fraction of ferrite even after extensive solution heat treatment, the base metal microstructure for all of the alloys was fully austenitic. The microstructure in the vicinity of the weld for all of the alloys investigated contained typical base metal microstructure. However, a small region very close to the fusion line showed evidence of grain growth. A careful examination of the HAZ microstructure also revealed a narrow region close to the fusion zone where the grain boundaries had apparently been coated with liquid. These grain boundaries, along with the cracks, are shown in Fig. 3. Some of these cracks appear to have been backfilled with residual liquid. This partially liquated region was clearly wider in the commercial alloy welds than in the high-purity laboratory alloy welds. Almost all of the liquated grain boundaries extended into the fusion zone. Figure 4 is a SEM fractograph of the crack surface that was in contact with the liquid pool. One end
of a surface (end that was in contact with the liquid pool) shows clear evidence of the presence of liquid, while the other is a clean, intergranularly fractured surface.

Susceptibility to HAZ Cracking

Figure 5 shows the dependence of the total HAZ crack length on the applied strain. It shows that the level of cracking increases with the applied strain. The minimum strain required for cracking in 316, 310, and AL6XN is 1% or less; for Alloys 304, C, and D, it is less than 3%; and it is over 5% for Alloys A, B, E, and 309. Figure 5 also shows the correlation between HAZ cracking and the solidification mode of the weld metal. When comparing the commercial heats, the welds of AL6XN and 310, which solidify in a primary austenitic mode, clearly have a higher susceptibility to HAZ cracking than the welds of Heats 304 and 309 solidifying in a primary ferritic mode. Alloy 316 welds, in which about 10 to 30% of grains solidified as primary austenite and the remaining portion as primary ferrite, also showed a higher HAZ cracking susceptibility than the welds solidifying entirely in a primary ferritic mode. The laboratory heats were also generally less susceptible to HAZ cracking than their corresponding commercial heats. However, among the laboratory heats, the welds of primary austenitic solidification (in Alloys A, C, and D) showed some tendency for cracking, while the welds solidifying in a primary ferritic solidification mode (in Alloys B and E) showed no evidence of cracking. The role of impurities (total S and P) and the solidification mode in the HAZ cracking tendency is shown in Fig. 6. Here, total S and P content is plotted as a function of the \( \text{Cr}_{eq}/\text{Ni}_{eq} \) ratio, where \( \text{Cr}_{eq} \) and \( \text{Ni}_{eq} \) are chromium and nickel equivalents, respectively, for various alloys investigated, with the boundary between primary austenitic and ferritic solidification (Ref. 16). The results are very similar to observations made of solidification crack-
ing. However, the role of ferrite in suppressing HAZ cracking will be discussed later in this paper.

Figure 7 shows the relationship between ferrite number (FN) of the weld and the susceptibility to HAZ cracking at an applied strain level of 4.9%. It shows that for ferrite levels >3 FN, the cracking tendency is low. The commercial heats show a higher tendency for HAZ cracking than the laboratory heats. The cracks in all the specimens were intergranular and oriented normal to the fusion line.

EPMA and AES Results

The composition of the metal close to the crack tip was investigated using EPMA in order to study the backfilling of HAZ cracks. Step scan profiles for the elements nickel and chromium in a Type 310 austenitic stainless steel weld are shown in Fig. 8. The region close to the crack tip was found to be enriched in nickel and chromium. The solidification of Type 310 stainless steel seems to promote nickel and chromium enrichment of the liquid during solidification. Thus, the liquid backfilling the crack and the subsequent solid that forms in this region are enriched in nickel and chromium. However, it should be noted that the chromium enrichment of the liquid backfilling the crack is not as pronounced as that of nickel. This solute build up ahead of the crack tip is an evidence of backfilling.

The elemental distribution on the HAZ crack surfaces was investigated using AES. Figures 9 and 10 show typical results for the Alloy 310. Scanning electron micrographs of the HAZ crack surfaces exposed inside the AES system and the distribution of Fe, Cr, Ni, and S (Auger maps) over these fields of view are shown in Fig. 9. Both elemental maps (Fig. 9) and Auger spectra (Fig. 10) show that the HAZ crack surfaces were enriched mainly in sulfur. The Auger spectrum taken on the HAZ crack surface of Alloy 310 (Curve A) in Fig. 10 revealed a much larger Auger peak from S than did the region of ductile tearing (Curve D).

Sputter etching and reanalysis of the surfaces in Fig. 10 revealed that the sulfur-enriched layers on the crack surfaces were on the order of 2 nm thick. The sulfur content on the top layer was estimated to be 20 to 30 at.-%, which is about 2000 times the bulk sulfur composition for the alloy. Absence of O and C in the spectra indicates that the analyzed areas did not contain any contamination layers.
Fig. 7 – Relationship between the total crack length and the room temperature ferrite number for the welds (4.9% applied strain).

Fig. 8 – Step scanning profiles for chromium and nickel in the backfilled regions of the weld.

Fig. 9 – AES studies on weld HAZ crack in Alloy 310 opened inside the Auger electron microscope. A and B – SEM profile of the surface studied; C through F – Fe, S, Cr, and Ni AES maps, respectively.
Although comparison of impurity segregation in alloys exhibiting both high and low crack susceptibility would be desirable in order to demonstrate a correlation between cracking susceptibility and segregation behavior, the low density and small size of cracks in the less susceptible alloys precluded this comparison. We should note, however, that studies of similar austenitic steels provide a strong link between the segregation of impurities such as sulfur and the intergranular failure of these alloys at elevated temperatures. The absence of any clear correlation between bulk sulfur levels and cracking tendency may not be too significant, because there can be significant uncertainty in the bulk analysis at these levels and because the intensity of segregation at any bulk level can be strongly affected by alloying elements such as Ni, Cr, and Mn.

Based on these observations, we conclude that there is strong circumstantial evidence that impurities such as sulfur play a role in HAZ cracking. The role of the segregation could be to alter the intrinsic cracking resistance of the alloys or to alter its interaction with the molten weld metal. Such segregation-induced effects could derive from the well-established relationships between segregation and interfacial energies or from the less well understood relationships between segregation and interfacial kinetics or mass transport.

**Initiation of HAZ Cracks**

The HAZ cracking discussed previously occurs in the region of the HAZ closest to the fusion zone, which experiences highest peak temperatures in the range of 1300° to 1450°C (2372° to 2642°F). These cracks are always intergranular and always occur in the presence of a liquid film along the grain boundaries. Two possible sources may contribute to the formation of a liquid film along the grain boundaries in the HAZ during welding: 1) depression of solidus temperature at the grain boundary due to the segregation of low-melting elements, such as S and P, and 2) the ease with which liquid can penetrate from the fusion zone and wet the grain boundaries.

In the present investigation there is clear evidence to support either of the above mechanisms. The results of this study strongly imply that impurities play a significant role in the HAZ cracking susceptibility—Fig. 9. The laboratory heats with overall low impurity concentrations exhibited lower tendency for HAZ cracking than did the commercial heats. AES results on the commercial heats (Type 310) show that sulfur is the predominant impurity on the cracked HAZ grain boundaries. The sulfur content on the top few atom layers of the cracks was about 20 to 30 at.-%. This translates to an estimated sulfur enrichment of 2000 to 20000 times the bulk sulfur composition of the alloy. Sulfur levels of this magnitude would suppress the solidus several hundred degrees, thus promoting liquid film formation along the grain boundaries in the HAZ. The combination of the presence of liquid film along the grain boundaries and the superimposed thermal stresses could lead to eventual HAZ cracking. Although the influence of other solute segregates on the HAZ cracking was not investigated extensively in this particular study, it seems that minor elements such as P, C, and Si also may play a significant role in the HAZ cracking susceptibility of austenitic stainless steels.

The second possible source of a liquid film depends on interfacial energy considerations. It is generally accepted that the interfacial tension of γ-γ, or δ-δ, boundaries will be greater than that of a δ-γ interphase boundary (Refs. 17-19). Hence, a γ-γ boundary in contact with liquid will have a lower liquid dihedral angle at the γ-γ boundary (promoting wetting of the γ-γ boundaries) during the weld thermal cycle. As discussed previously, this wetting in turn could promote HAZ cracking during welding.

An interesting observation of this study is the influence of grain boundary ferrite on the HAZ cracking susceptibility. The presence or absence of grain boundary ferrite has been found to play a significant role in the HAZ cracking. Although the base metal is fully austenitic, the HAZ microstructure may contain a narrow band close to the fusion zone wherein ferrite precipitation occurs along the grain boundaries. Figure 11 shows such precipitation during autogenous welding of Type 316 stainless steel observed by Goodwin (Ref. 20). The ease with which the ferrite precipitation can occur...

![Fig. 10 — AES spectra for the surface in Fig. 9. A and D—Points A and B as fractured, respectively; B and C—Point A after 2- and 7-min sputtering, respectively.](image1)

![Fig. 11 — Precipitation of ferrite along the HAZ grain boundaries formed during a weld thermal cycle.](image2)
depends on the mode of freezing. Most austenitic stainless steels that freeze in the primary ferritic mode tend to form grain boundary ferrite easily. Accordingly, Alloys 309, 304, 316, and the laboratory Heats B and E that form grain boundary ferrite have been found to show a lower tendency for HAZ cracking. However, it should be noted that all of the high-purity laboratory heats, whether they have a tendency to form grain boundary ferrite or not, show a lower tendency for HAZ cracking. The formation of grain boundary ferrite can be easily explained using a pseudo-binary section of the Fe-Cr-Ni (Ref. 21) system shown in Fig. 12. For a given alloy composition $C_0$, although the equilibrium structure (in the wrought state) of the alloy is fully austenitic, during the weld thermal cycle the portion of the HAZ that experiences temperatures in excess of $T_0$ will form ferrite along the grain boundaries. The ease with which the ferrite forms depends on time at temperature (controlled by welding variables) and cooling rates.

The formation of grain boundary ferrite may influence both mechanisms of liquid film formation discussed previously. First, the formation of grain boundary ferrite may promote a scavenging effect for a large percentage of the grain boundary segregates, in particular, sulfur. Sulfur, having higher solubility in $\delta$-ferrite than in austenite, might preferentially partition to the ferrite, thus devoicing the grain boundaries close to the fusion zone of sulfur. This may reduce the likelihood of liquid film formation caused by suppression of the solidus along the grain boundaries, in turn reducing the HAZ cracking tendency. Second, the grain boundary ferrite may play a significant role in altering the interfacial energy considerations discussed previously, as follows. In the absence of any grain boundary ferrite, since the $\gamma-\gamma$ grain boundary interfacial tension is higher than $\gamma-\delta$ boundary tension, based on the interfacial energy considerations, liquid film may penetrate deep into the grain boundaries and provide a susceptible microstructure for HAZ cracking. This is shown schematically in Fig. 13A. However, in the presence of grain boundary $\delta$-ferrite, the creation of the $\delta-\gamma$ boundary ($\delta-\gamma$ boundary interfacial tension being low) will circumvent the penetration of liquid deep into the grain boundaries close to the fusion zone—Fig. 13B. This in turn may decrease the propensity for HAZ cracking.

**Summary**

Five ternary Fe-Cr-Ni alloys and five commercial austenitic stainless steel counterparts were evaluated by the spot Varestraint test in order to study the formation mechanism of HAZ cracking. The results allow the following conclusions to be made. The welds solidifying in a primary austenitic mode showed a higher susceptibility to HAZ cracking than the welds solidifying in a primary ferritic mode. This distinction was not very clear in the laboratory heats because they all exhibited a lower tendency for HAZ cracking. Also, the laboratory heats were less susceptible to HAZ cracking than their corresponding commercial heats.

Significant sulfur concentration was observed on the HAZ crack surfaces of commercial alloys highly susceptible to HAZ cracking. However, a direct comparison of sulfur segregation in alloys exhibiting low cracking susceptibility was not possible because of the low density and small size of the cracks.

The results indicate that both possible depression of solidus temperature at the...
grain boundary due to the segregation of low-melting solute such as S and the ease with which liquid from the fusion zone can penetrate grain boundaries may play a significant role in initiating HAZ cracking. Formation of primary ferrite significantly reduces the chance of either of those factors inducing HAZ cracking.

Acknowledgments

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References

20. Goodwin, G. M. Private communication, Oak Ridge National Laboratory, Oak Ridge, Tenn.

Weldability of Steels—Fourth Edition
By Prof. R. D. Stout

This well-known book on the welding metallurgy of steels has been extensively revised by Prof. Stout, Dean Emeritus of the Graduate School, Lehigh University. The fully indexed, 450-page, cloth-bound volume, which contains numerous illustrations and tables, was published in April 1987. The price of Weldability of Steels—Fourth Edition is $40.00 per copy, plus $5.00 for postage and handling. Orders should be sent with payment to the Welding Research Council, Suite 1301, 345 E. 47th St., New York, NY 10017.