Effect of Nitrogen on the Microstructure and Stress Corrosion Cracking of Stainless Steel Weld Metals

Nitrogen pickup from the atmosphere during welding can cause elimination of delta ferrite and an increase in the susceptibility to stress corrosion cracking

BY W. A. BAESLACK, III, W. F. SAVAGE AND D. J. DUQUETTE

ABSTRACT. The nitrogen content of conventionally-produced 18Cr-8Ni stainless steel weld metals can vary considerably. Variations in base and filler metal nitrogen contents and nitrogen pickup from the atmosphere during welding can result in weld metal nitrogen contents ranging from below 0.04 to above 0.3 wt-%. Unfortunately, aside from its influence on ferrite content, the importance of nitrogen content on 18Cr-8Ni weld metal properties has been for the most part undetermined.

Results of the present investigation have shown that nitrogen content in a range from approximately 0.04 to 0.25 wt-% influences both the mode of weld metal solidification and weld metal stress-corrosion cracking properties. At low nitrogen levels (below about 0.1 wt-%) weld metal solidification initiates as primary ferrite, with the final regions solidifying as an austenite-ferrite eutectic at the primary-ferrite cellular-dendrite boundaries. During subsequent cooling of the weldment, nearly all the ferrite is transformed to austenite by a diffusionless transformation, and ferrite is retained to room temperature at the cores of originally-solidified, primary-ferrite dendrites.

Increased nitrogen contents promote the primary solidification of austenite, with the final regions solidifying as an austenite-ferrite eutectic at primary-austenite cellular-dendrite and grain boundaries. As the weldment cools, some of this eutectic ferrite is transformed to austenite and a room-temperature microstructure exhibiting a discontinuous ferrite morphology results. At high nitrogen levels (greater than about 0.2 wt-%) an entirely austenitic room-temperature microstructure is produced. The absence of ferrite at austenite grain boundaries during weldment cooling allows considerable austenite grain growth.

Constant-extension rate tensile tests performed on Ferrite Numbers 6 (0.44 wt-% N), 3 (0.30 wt-% N), 1 (0.14 wt-% N), and 0 (0.24 wt-% N) for 18Cr-8Ni weld metals in boiling MgCl₂ solutions found that the stress-corrosion cracking resistance decreases with increases in nitrogen content and decreases in ferrite content. It was also determined that the stress-corrosion cracking morphologies are closely associated with weld metal microstructural characteristics. The susceptibilities of weld metals which solidify as primary ferrite (Ferrite Number 6) are similar to those of wholly-austenitic Type 304 base metal.

Stress-corrosion crack propagation in these weld metals occurs intergranularly at austenite-ferrite boundaries in the weld metal substructure. Ferrite Number 1 and 3 weld metals, which solidify as primary austenite, exhibit similar susceptibilities to stress-corrosion cracking which are somewhat greater than that observed with the Ferrite Number 6 weld metals. Crack propagation in these low-ferrite weld metals occurs by a mixed transgranular-intergranular mode. Wholly-austenitic weld metals are particularly susceptible to stress-corrosion cracking by a predominantly intergranular mode. The large grain size appears to promote rapid stress-corrosion crack propagation.

Introduction

Stress-corrosion cracking in austenitic stainless steels has been a problem of considerable importance to the nuclear power and chemical industries. Numerous studies have been performed to determine the characteristics of chloride-induced stress-corrosion cracking phenomena in both sensitized and unsensitized austenitic stainless steels. Although stress-corrosion cracking in these alloys has been extensively investigated, and well documented, several closely related areas remain poorly understood and highly controversial. One such area concerns the stress-corrosion cracking properties of duplex austeno-ferritic stainless steel weld metals.

Introductory studies completed at Rensselaer Polytechnic Institute have found that both the stress-corrosion cracking susceptibility and the failure mode of a duplex 18Cr-8Ni stainless steel weld metal may differ from the susceptibility and failure mode of an annealed 18Cr-8Ni base metal with the same chemical composition.¹ ² Results
of these studies have also shown the stress-corrosion cracking morphology in duplex 18Cr-8Ni weld metal to be closely associated with the weld metal microstructure and have indicated that this close association may be significant in determining stress-corrosion cracking resistance.

The importance of nitrogen content on the mechanical and corrosion properties of 18Cr-8Ni stainless steels has become increasingly apparent. Recent investigations have shown that the addition of nitrogen to 18Cr-8Ni stainless steels provides considerable strengthening, improves high-temperature creep properties, minimizes intergranular carbide precipitation, improves pitting and crevice corrosion resistance, and at levels greater than 0.04 wt-% does not appreciably influence stress-corrosion cracking susceptibility. Nearly all studies performed to date have involved annealed or cold-worked materials, with the influence of nitrogen on the characteristics of 18Cr-8Ni weld metals having thus far received little attention. This is unfortunate, since the nitrogen content of conventionally-produced 18Cr-8Ni weld metals may vary considerably, and differ greatly from that of the base metal.

Recent studies by Long and DeLong have suggested that nitrogen pickup during welding can result in weld metal nitrogen contents ranging from below 0.04 to above 0.3 wt-% and corresponding ferrite levels ranging from over Ferrite Number 0 to over Ferrite Number 12. Although their work has clearly demonstrated the importance of nitrogen in determining ferrite content, their approach was restricted to a quantitative and somewhat statistical viewpoint. The influence of nitrogen on the solidification and microstructural characteristics, upon which stress-corrosion cracking properties of 18Cr-8Ni stainless steels may directly depend, has yet to be thoroughly explored.

**Objectives**

The objectives of this investigation can be summarized as follows:

1. To study the influence of nitrogen content on the solidification and microstructural aspects of 18Cr-8Ni stainless steels.
2. To determine the importance of nitrogen-induced microstructural characteristics on weld metal stress-corrosion cracking properties.

**Experimental Procedures**

**Materials**

Autogenous gas tungsten-arc weldments produced on wholly-austenitic, hot-rolled Type 304 stainless steel sheet exhibited fusion zone ferrite levels of approximately Ferrite Number 6. The controlled addition of gaseous nitrogen to the normally inert shielding gas (high-purity argon) employed during welding resulted in the production of weldment fusion zones of Ferrite Numbers 0, 1 and 3. The parameters employed during weld metal fabrication are described in Table 1. The Type 304 base metal chemical composition and weld metal nitrogen contents are provided in Table 2.

A schematic illustration of the tensile specimen employed in constant extension rate testing is provided in Fig. 1. Weld metal specimens were machined from welded sheet materials such that the center of the gauge-section length was coincident with the center of the full-penetration, single-pass weldment length. The procedure assured that macrosegregation did not occur in the gage-section weld metal (resulting from solute transients in the initial and final regions of solidification); it also provided an all-weld metal gage section with a consistent ferrite content and morphology.

Subsequent to machining, the specimen gage section was dry ground to nominal size and wet ground to a 600 grit finish. The final grinding direction

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**Table 1—Chemical Compositions of Type 304 Stainless Steel Base Metal and Nitrogen Contents of Type 304 Weld Metals**

<table>
<thead>
<tr>
<th>Material designation</th>
<th>Ferrite Number</th>
<th>C</th>
<th>Cr</th>
<th>Ni</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Mo</th>
<th>Co</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>Type 304-BM</td>
<td>0</td>
<td>0.055</td>
<td>18.28</td>
<td>8.55</td>
<td>1.56</td>
<td>0.61</td>
<td>0.015</td>
<td>0.027</td>
<td>0.018</td>
<td>0.016</td>
<td>0.052</td>
</tr>
<tr>
<td>Type 304-WM</td>
<td>6</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.044</td>
</tr>
<tr>
<td>Type 304N1</td>
<td>3</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.10</td>
</tr>
<tr>
<td>Type 304N2</td>
<td>1</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.14</td>
</tr>
<tr>
<td>Type 304N3</td>
<td>0</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.24</td>
</tr>
</tbody>
</table>

*With the exception of nitrogen content, weld metal chemical compositions are not expected to differ significantly from that of the Type 304 base metal.*

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**Table 2—Welding Parameters and Procedures Employed in the Fabrication of Type 304 and Type 304(N) Weld Metals**

<table>
<thead>
<tr>
<th>Welding process:</th>
<th>Gas tungsten arc (GTAW)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weldment type:</td>
<td>Autogenous, full-penetration</td>
</tr>
<tr>
<td>Electrode type:</td>
<td>Thoriated tungsten</td>
</tr>
<tr>
<td>Electrode geometry:</td>
<td>0.125 in. (0.32 cm) diameter, ground conical tip, 90 deg included angle at tip</td>
</tr>
<tr>
<td>Welding fixture:</td>
<td>Copper backing and hold-down plates, full restraint.</td>
</tr>
<tr>
<td>Welding current:</td>
<td>275 amperes</td>
</tr>
<tr>
<td>Arc voltage:</td>
<td>12.5 volts, DCSP</td>
</tr>
<tr>
<td>Travel speed:</td>
<td>4 ipm (0.17 cm/sec)</td>
</tr>
<tr>
<td>Shielding gas:</td>
<td>Type 304-argon, 40 cfh</td>
</tr>
<tr>
<td>Type 304N1-argon, 40 cfh + Nitrogen, 1 cfh</td>
<td></td>
</tr>
<tr>
<td>Type 304N2-argon, 40 cfh + Nitrogen, 2 cfh</td>
<td></td>
</tr>
<tr>
<td>Type 304N3-argon, 40 cfh + Nitrogen, 4 cfh</td>
<td></td>
</tr>
</tbody>
</table>

*Argon and nitrogen were mixed in a three-stage mixing valve.*
Stress-Corrosion Cracking Testing

Constant extension rate tensile tests were performed on Ferrite Numbers 0, 1, 3, and 6 Type 304 weld metal specimens and on Type 304 base metal specimens (for comparative purposes) in 42 wt-% boiling magnesium chloride (309 °F, 154 °C) and in liquid paraffin (309 °F, 154 °C) at an extension rate of $5.08 \times 10^{-4} \text{ cm/min}$ (equivalent to an initial strain rate of $1.48 \times 10^{-5} \text{ s}^{-1}$).

Loads were applied at a constant extension rate by a conventional Instron testing machine and monitored continuously on a strip chart recorder. The pyrex test cell employed in stress-corrosion cracking studies is illustrated in Fig. 2. Solution temperature and concentration were maintained constant ($\pm 2 \text{ °F}$) by an immersion heater-reflux condenser system. MgCl$_2$ test solutions were prepared from reagent grade hydrous MgCl$_2$ crystals (MgCl$_2$· 6H$_2$O) and distilled water according to ASTM Standard G36-73.

Subsequent to testing, fractured specimens were analyzed using optical and scanning electron microscopy techniques.

**Results and Discussion**

Weld Metal Solidification and Structure

The equilibrium solidification sequence for an 18Cr-8Ni stainless steel alloy is illustrated in traverse A of Fig. 3, which shows a constant-iron section of the Fe-Cr-Ni equilibrium ternary diagram. Under equilibrium conditions, solidification occurs as primary ferrite. As the alloy cools below the solvus temperature, a solid-state ferrite-to-austenite transformation takes place.

**Case I**—equilibrium solidification; **Case II**—no solid diffusion; **Case III**—no diffusion in the solid, changes in composition of the liquid are by diffusion only and not by convection.

Mathematical solidification modeling using Case III solidification principles has been shown to be applicable to weld metal solidification. In Case III solidification, there is no diffusion in the solid and no mechanical mixing in the liquid. Concentration changes in the liquid occur by diffusion only.

$C_0 = \text{the nominal composition; } K_E = \text{the effective distribution coefficient ($= 1.41$ for Cr in Fe, $= 0.41$ for Ni in Fe).}$

As solidification progresses, the composition of the remaining liquid is shifted to the left, toward the three-phase eutectic region. During the initial transient portion, which is of the order of microns in extent, the solute rejected to the liquid establishes a diffusion gradient in the liquid such that the composition of the liquid at the moving liquid-solid interface approaches $C / k_e$. Once the concentration of the liquid interface reaches $C / k_e$, a dynamic equilibrium is established and solid of nominal composition, $C_0$, forms continuously until the terminal transient stage is entered.

The terminal transient stage begins where the diffusion gradient in advance of the moving solid-liquid interface either encounters a diffusion gradient associated with another moving solid-liquid interface or a stationary solid-liquid interface. During this final, or terminal transient another region of severe microsegregation occurs over an extremely small distance, typically, microns. The solute redistribution in this region is always such as to lower the effective solidus drastically. Solidification of the last-to-solidify liquid is ultimately completed in the three-phase region, and results in the formation of a small quantity of
divorced austenite-ferrite eutectic in the interdendritic regions. During subsequent rapid cooling of the weldment, the eutectic austenite consumes much of the eutectic and primary ferrite by a diffusionless, massive-type† transformation. The high chromium and low nickel contents of the initially-solidified ferrite, when coupled with the fast cooling rates present in weldments, can prevent the ferritic regions (which are compositionally identical to a ferritic stainless steel) from transforming to austenite on cooling and result in the retention of some ferrite down to room temperature at the cores of originally-solidified ferrite dendrites.

Scanning-transmission-electron microscopy studies of duplex 18Cr-8Ni weld metal have been recently performed by Lyman, et al.‡ Data obtained in these studies, which are illustrated in Fig. 4, clearly show an enrichment of chromium and a depletion of nickel in the delta ferrite retained down to room temperature. These results strongly support the Case III mode of 18Cr-8Ni weld metal solidification described above, and provide convincing evidence that the initial stages of 18Cr-8Ni weld metal solidification effectively control the room temperature microstructure.

Adding nitrogen to an 18Cr-8Ni alloy effectively shifts the Fe-Cr-Ni pseudo-binary diagram to the right. The powerful austenitizing effect of nitrogen (30 times greater than that of nickel) tends to suppress primary ferrite solidification and promote the primary solidification of austenite. At low nitrogen levels (below about 0.1 wt-% nitrogen)‡‡ commercial 18Cr-8Ni weld metal solidification occurs as primary ferrite in a manner similar to that observed in nitrogen-free 18Cr-8Ni alloys. Traverse B in Fig. 3 exemplifies the equilibrium solidification sequence for a commercial 18Cr-8Ni alloy containing approximately 0.075 wt-% nitrogen.

According to Case II solidification principles, the chromium content of the initially-solidified delta ferrite is lower than that of a nitrogen-free alloy. This reduction contributes to the retention of less ferrite at room temperature. The vermicular microstructure of the Ferrite Number 6 Type 304 weld metal (0.04 wt-% nitrogen) is illustrated in Fig. 5. Nearly-continuous ferrite is present as the cores of primary-solidified ferrite dendrites.

The presence of nitrogen in levels exceeding about 0.1 wt-% results in weld metal structures which solidify as primary austenite and exhibit lesser quantities of ferrite in a microstructure somewhat coarser than that observed with primary ferrite solidification. Traverse C in Fig. 3 exemplified this primary austenite mode of equilibrium solidification for a commercial 18Cr-8Ni alloy containing approximately 0.15 wt-% nitrogen. As in weld metals

†A massive reaction is one which occurs when the high and low temperature phases exhibit the same composition but different crystal structures.
‡In commercial 18Cr-8Ni alloys this value is dependent on the specific weld metal chemistry.
Table 3—Results of Constant Extension Rate Tests Performed on Type 304 Base and Weld Metal Specimens in Boiling MgCl₂ and in Liquid Paraffin at 309°F (154°C)

<table>
<thead>
<tr>
<th>Specimen designation</th>
<th>Ferrite number&lt;sup&gt;a&lt;/sup&gt;</th>
<th>Test environment</th>
<th>Maximum engineering stress,&lt;sup&gt;b&lt;/sup&gt; ksi</th>
<th>Elongation at failure, %</th>
<th>Yield strength,&lt;sup&gt;c&lt;/sup&gt; ksi</th>
</tr>
</thead>
<tbody>
<tr>
<td>Type 304-BM-LP</td>
<td>0</td>
<td>Liq. paraffin</td>
<td>72</td>
<td>35</td>
<td>33</td>
</tr>
<tr>
<td>Type 304-BM-MG</td>
<td>0</td>
<td>MgCl₂</td>
<td>34</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>Type 304-WM-LP</td>
<td>6</td>
<td>Liq. paraffin</td>
<td>68</td>
<td>34</td>
<td>29</td>
</tr>
<tr>
<td>Type 304-WM-MG</td>
<td>6</td>
<td>MgCl₂</td>
<td>36</td>
<td>—</td>
<td>—</td>
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<tr>
<td>Type 304N1-WM-LP</td>
<td>3</td>
<td>Liq. paraffin</td>
<td>71</td>
<td>34</td>
<td>30</td>
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<tr>
<td>Type 304N1-WM-MG</td>
<td>3</td>
<td>MgCl₂</td>
<td>32</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>Type 304N2-WM-LP</td>
<td>1</td>
<td>Liq. paraffin</td>
<td>71</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>Type 304N2-WM-MG</td>
<td>1</td>
<td>MgCl₂</td>
<td>30</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>Type 304N3-WM-LP</td>
<td>0</td>
<td>Liq. paraffin</td>
<td>70</td>
<td>41</td>
<td>30</td>
</tr>
<tr>
<td>Type 304N3-WM-MG</td>
<td>0</td>
<td>MgCl₂</td>
<td>24</td>
<td>—</td>
<td>—</td>
</tr>
</tbody>
</table>

<sup>a</sup>Measured with an Aminco Brenner Magne Gage.
<sup>b</sup>Average value obtained from multiple tests.
<sup>c</sup>Multiply ksi by 6.894757 to obtain MPa.
<sup>d</sup>Calculated from load-elongation curve at 0.2% strain offset.

which solidify as primary ferrite, non-equilibrium solidification associated with welding can permit the formation of an austenite-ferrite eutectic in the final regions of solidification. During subsequent weldment cooling, the austenite consumes most or all of the eutectic ferrite in a solid-state reaction.

Figure 6 illustrates the duplex fusion zone microstructures of the Ferrite Number 3 (0.14 wt-% nitrogen) and Ferrite Number 1 (0.11 wt-% nitrogen) Type 304(N) weld metals, respectively. Solidification in these weld metals occurs as primary austenite in a cellular-dendritic mode. Discontinuous eutectic ferrite at cellular-dendrite boundaries and in the last-to-solidify grain boundaries provides a "pinning" action which minimizes austenite grain growth during weldment cooling.

Figure 7 illustrates the microstructure of the Ferrite Number 0 (0.24 wt-% nitrogen) Type 304(N) weld metal, which solidifies in a mixed cellular/ cellular-dendritic mode. The transformation of eutectic ferrite to austenite immediately subsequent to weld metal solidification eliminates grain boundary pinning and allows considerable austenite grain growth after solidification is complete.

**Stress-Corrosion Cracking Properties**

Constant extension rate stress-corrosion cracking tests (sometimes described as constant strain rate or constant deflection rate tests) have been employed in both mechanistic and general susceptibility studies. Parkins<sup>19</sup> states that constant extension rate testing methods are advantageous in that numerous measurable and quantifiable parameters exist for classifying the stress-corrosion cracking susceptibilities of materials which exhibit similar mechanical properties when tested under neutral environmental conditions. Such parameters include percent reduction in area at failure, time to failure, percent elongation at failure, and maximum engineering stress.

In the present investigation, maximum engineering stress was employed as an indicator of stress-corrosion cracking susceptibility. The use of percent elongation at failure and time to failure were not employed because of the significant effects crack yawn (crack-opening displacement) has on these parameters, particularly in multiply-cracked specimens.<sup>20</sup>

Table 3 provides the data obtained from constant extension rate tests performed on base and weld metal specimens in liquid paraffin and boiling MgCl₂ at 309°F (154°C). The results of tests performed in the neutral, liquid paraffin environment indicate...
that the mechanical properties at 309 F (154 C) of Type 304 weld metals containing nitrogen (in a range from 0.04 to 0.24 wt-%) are very similar, and comparable to those of hot-rolled Type 304 base materials. Fracture of specimens tested in liquid paraffin occurs by a ductile, dimpled-rupture type mechanism.

Figures 8 and 9 illustrate the effects of ferrite level and nitrogen content, respectively, on the maximum engineering stress observed for specimens tested in boiling MgCl₂. The stress-corrosion cracking resistance of Type 304 weld metals decreases with increases in nitrogen content and decreases in ferrite level. Fracture of specimens tested in MgCl₂ solutions occurs by a combination of stress-corrosion cracking and ductile failure. Multiple stress-corrosion crack initiation and propagation are clearly observable on the fracture surface of as-tested specimens. These macroscopic fractographic characteristics are illustrated in Fig. 10 for a Ferrite Number 6 specimen.

The wholly-austenitic Type 304 base metal exhibits a stress-corrosion cracking resistance similar to that of the Ferrite Number 6 weld metal and considerably greater than that of the wholly-austenitic Type 304(N) weld metal. Stress-corrosion cracking in the base metal occurs by a mixed transgranular-intergranular mode. Transgranular stress-corrosion cracking appears to be preferred at the fracture surface edge, possibly near crack initiation sites, while intergranular cracking predominates near the center of the specimen.

Stress-corrosion cracking in the Ferrite Number 6 weld metal occurs predominantly at the semi-coherent interface between the nearly-continuous primary ferrite network and the austenite. Localized transgranular stress-corrosion cracking is sometimes in evidence; however, it is apparently restricted to regions near crack initiation sites and areas where the ferrite network is not entirely continuous.

The interphase-interface stress-corrosion cracking morphology is illustrated in the scanning-electron microscopy fractograph shown in Fig. 11A and in the optical micrograph shown in Fig. 11B. The cellular-dendritic nature of the duplex weld metal and the presence of ferrite on the fracture surface are clearly apparent. Thin layers of ferrite retained at the cores of primary ferrite cellular-dendrites appear to cover austenite “cells” in a
The observation of ferrite on the fracture surface supports the interphase-interface mode of stress-corrosion cracking in duplex Type 304 stainless steel weld metals originally reported by Duquette, Sherman, and Savage. The irregular fracture surface associated with the interphase-interface mode of stress-corrosion cracking indicates that microscopic crack directional deviations are numerous.

Stress-corrosion cracking susceptibilities of duplex weld metals which solidify as primary austenite are very similar, and are somewhat greater than that of the Ferrite Number 6 weld metal. The discontinuous ferrite morphology found in these Type 304(N) weld metals promotes stress-corrosion crack propagation by a mixed intergranular-transgranular mode. As in Type 304 base metal, transgranular cracking appears to dominate the early stages of crack growth. As cracks propagate into the specimen, localized crack propagation by an intergranular mode becomes increasingly apparent.

If sufficient ferrite is present at austenite grain boundaries, stress-corrosion cracks propagate preferentially along grain boundaries either by an interphase-interface mode at austenite-ferrite boundaries, or by an intergranular mode at austenite-austenite boundaries. The mixed transgranular-intergranular mode of cracking is illustrated for a Ferrite Number 3 weld metal in Figs. 12A and 12B. The SEM fractograph in Fig. 12C, which shows the specimen region in Fig. 12A at a higher magnification, suggests the presence of ferrite on the fracture surface. Ferrite appears to be located both along austenite grain boundaries and at cell boundaries which intersect grain boundaries. It is of importance to note that attack at austenite cellular-dendrite boundaries is not in evidence.

As the ferrite level approaches zero, the availability of an interphase-interface mode of stress-corrosion cracking at austenite grain boundaries is minimized, and transgranular, rather than intergranular, stress-corrosion cracking is promoted. The SEM fractograph and the optical micrograph shown in Fig. 13 illustrate the predominantly transgranular stress-corrosion cracking found in Ferrite Number 1 weld metal.

The stress-corrosion cracking susceptibility of wholly-austenitic weld metal is clearly greater than that of the wholly-austenitic base metal and the duplex weld metals. The SEM fractographs and optical micrograph illustrated in Fig. 1 indicate that stress-corrosion cracks propagate by a mixed transgranular-intergranular mode with the intergranular mode predominating. As in the FN-3 and FN-1 weld metals, attack along cellular-dendrite boundaries does not occur. The increased susceptibility of the wholly-austenitic weld metal is probably promoted by the large grain size and the presence of smooth, ferrite-free grain boundaries. Stress-corrosion crack propagation in this microstructure occurs with a minimal requirement for directional deviation.

Conclusions

The following conclusions pertain to the solidification and structure of 18Cr-8Ni weld metals containing nitrogen:

1. The nitrogen content of 18Cr-8Ni weld metals significantly influences both the mode of solidification and the quantity of ferrite retained in the room temperature microstructure.

2. Commercial 18Cr-8Ni weld metals containing less than approximately 0.1 wt-% nitrogen solidify as primary ferrite. Their room-temperature microstructures exhibit nearly-continuous ferrite at the cores of originally-solidified primary-ferrite dendrites.

3. Increased nitrogen contents promote weld metal solidification as primary austenite. Small quantities of discontinuous ferrite are retained at room temperature at austenite cellular-dendrite and grain boundaries. Nitrogen contents in excess of approximately 0.2 wt-% result in the production of commercial 18Cr-8Ni weld metals which are entirely austenitic. The absence of ferrite in the weld metal during weldment cooling allows considerable grain growth.

The following conclusions pertain to stress-corrosion cracking phenomena in 18Cr-8Ni weld metals containing nitrogen:
Stress-corrosion cracking resistance in MgCl₂ appears to decrease with increases in nitrogen content and decreases in ferrite level.

Stress-corrosion cracking morphologies observed in weld metals are closely associated with weld metal microstructural characteristics.

Weld metals which solidify as primary ferrite exhibit a stress-corrosion cracking resistance similar to that of wholly-austenitic Type 304 base metal. Stress-corrosion crack propagation occurs by an interphase/interface mode along the austenite-ferrite boundaries in the weld metal substructure.

Duplex weld metals which solidify as primary austenite exhibit similar stress-corrosion cracking susceptibilities. Stress-corrosion cracking occurs by a mixed transgranular-intergranular mode.

Wholly-austenitic weld metals are highly susceptible to stress-corrosion cracking by a predominantly intergranular mode. The large grain size promotes crack propagation.

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References


