Stress Corrosion Behavior of Stainless Steel Welds in High Temperature Water Containing Chlorides

In tests with three different grades of AISI 308, 309 and 316 stainless steels, a Type 316L stainless steel surfacing weld exhibits the lowest susceptibility to stress corrosion cracking in aerated 316 C (601 F) water containing chlorides.

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ABSTRACT. The effects of delta ferrite, carbon content and postweld heat treatment on the stress corrosion susceptibility of AISI Types 308, 309 and 316 stainless steel surfacing welds in high temperature water have been investigated. For each type of stainless steel, the initial ferrite level was controlled at three predetermined levels in the range 1-10%, and the carbon content was varied so as to cover the high and the low ends of the range normally encountered for the steel. Bent beam specimens machined from the overlays (i.e., surfacing welds) were tested with and without postweld heat treatments of 1, 15 and 30 hours at 608 C (1126 F).

Results show the stress corrosion susceptibility of the steel to be relatively independent of the initial ferrite content. The carbon content on the other hand, appears to exert a more significant influence, with the higher carbon grades always exhibiting higher susceptibility than the lower carbon grades, for a given type of stainless steel. The effect of postweld heat treatment at 608 C (1126 F) was found to depend on the duration of the heat treatment.

The maximum susceptibility for cracking was encountered in samples postweld treated for intermediate periods of time in the range of 1-15 hours, while the minimum susceptibility to cracking was observed in samples heat treated for 30 hours and in those in the as-welded condition. Crack morphologies were found to be interdendritic, intergranular or transgranular, depending upon the specific combination of carbon level, ferrite level and postweld heat treatment in a given sample. Selective characterization of the samples by optical and electron microscopy has been completed; the results indicate that the stress corrosion behavior of the various samples can be reasonably correlated with the microstructural parameters.

Introduction

Austenitic stainless steel surfacing welds are widely used as the lining of pressure vessels in the chemical, petrochemical and the nuclear power industries. Although these surfacing welds possess excellent resistance to general corrosion in the environments to which they are normally exposed, selective forms of corrosion such as pitting, intergranular corrosion and stress corrosion can still occur, resulting in failure of the cladding if the microstructure has not been carefully optimized. Two of the metallurgical variables deemed to be important in the context of selective corrosion are the amount of delta ferrite in the weld, and the degree of “sensitization” of the grain boundaries.

In general, to avoid hot cracking during welding, controlled amounts of ferrite are deliberately introduced, by adjusting the alloy composition. The weld metals are also postweld heat treated to relieve the residual stresses set up in the metal during welding. The postweld heat treatments can sometimes lead to undesirable microstructural and chemical changes in the local area of the grain boundaries, resulting in an increased susceptibility of the weld to intergranular attack. This phenomenon is commonly referred to as sensitization.

A search of the published literature on the corrosion of stainless steels shows that, while there is an abundance of information pertaining to wrought stainless steels, information relating to weldments is meager. Even the few studies that have been conducted deal primarily with intergranular corrosion. With the exception of a recent study by Ishihara, Shimizu and Ito, there has been no detailed study concerning the effects of such variables as carbon level, ferrite level, and postweld heat treatment parameters on the stress corrosion behavior of stainless steel welds. This lack of information prompted the study reported here.

Experimental Procedure

Specimens

Surfacing welds from three different AISI grades of steel—namely, Types 308, 309 and 316—were included in this study. In order to evaluate the effects of carbon content and ferrite content, each steel weld was investigated at three different compositions, resulting in a total of nine welds. Chemical analyses of the welds are given in Table 1.

The as-received surfacing weld
carbon steel. Details of the weld surfacing procedure, as supplied by weld on the base metal, which was arc process was used to deposit the specimen block is illustrated in Fig. 1.

The sheet was then cut into specimens that were approximately 6 mm (0.24 in.) wide and 8.9 mm (0.35 in.) long as shown in Fig. 2.

Following machining, selected specimens were exposed to a simulated postweld heat treatment at 608 C (1126 F) 1½, 15 or 30 hours (h). Heat treating was conducted in an argon atmosphere followed by furnace cooling to room temperature. The heat treated and the as-welded specimens were ground with 600 grade emery paper to provide a smooth and uniform surface finish to all the specimens. These specimens were further degreased in benzene and alcohol, rinsed in water, dried and then used for stress corrosion tests.

**Stress Corrosion Tests**

Specimens for stress corrosion tests were stressed in the 4-point loading device shown in Fig. 3. The specimens were subjected to a maximum uniform deflection of about 18 mm (¾ in.), resulting in a nominal circumferential strain of 10% at the apex. Since the objective of this study is only to characterize the stress corrosion behavior of the steels, the use of the commonly employed U-bend samples was ruled out, for fear that it might lead to premature cracking during bending, especially in the low ferrite steels. In fact, several samples stressed similarly to the stress corrosion samples were examined metallographically at x300, and it was ascertained that no fissuring or cracking occurred in any of the samples prior to the stress corrosion tests.

The stress corrosion tests were conducted in high temperature water containing chlorides in an autoclave with continuous recirculation of the test solution. Test solutions containing 100 or 10 ppm concentration of chloride ion were prepared from sodium chloride and demineralized high purity water. The tests were run at 316 C (601 F) for a duration of 200 h under an undeated condition.

At the end of the stress corrosion test, equal lengths of arc (about 35 mm, i.e., 1½ in.) were cut from each sample, mounted and polished for metallographic examination. The number of cracks, and the lengths of the cracks were determined. The samples were then etched to reveal the preferred path for each crack.

The crack morphology in each sample was determined by examination of the cracks at high magnification (X300) using the optical microscope and the scanning electron microscope. In samples where more than one type of crack morphology was observed, their relative proportions were estimated. The dominant morphology was then classified as the "major" mode and the others were classified as the "minor" modes.

**Microstructural Examination**

To determine whether postweld heat treatments led to a decay of ferrite content, specimens in the as-welded and postwelded conditions were evaluated for ferrite content by several different techniques. Among these were optical metallography, use of magnagage, Severn gage and the foster ferrite meter and measurement of saturation magnetization.

Although differences in ferrite levels as large as 2.5% were observed from

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**Table 1—Chemical Composition of Stainless Steel Welds, Wt-%**

<table>
<thead>
<tr>
<th>AISI grade of steel</th>
<th>Heat identification</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>N</th>
<th>Co</th>
<th>Ferrite Number(a)</th>
<th>Ferrite content(b)</th>
</tr>
</thead>
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<tr>
<td>308</td>
<td>8372</td>
<td>0.044</td>
<td>18.20</td>
<td>10.50</td>
<td>2.36</td>
<td>0.99</td>
<td>0.008</td>
<td>0.013</td>
<td>0.028</td>
<td>1.6</td>
<td>7.9</td>
<td>4.6</td>
<td>3.7</td>
</tr>
<tr>
<td>8149</td>
<td>0.045</td>
<td>19.60</td>
<td>9.12</td>
<td>1.85</td>
<td>1.05</td>
<td>0.008</td>
<td>0.010</td>
<td>0.016</td>
<td>0.044</td>
<td>11.4</td>
<td>11.4</td>
<td>3.3</td>
<td>3.3</td>
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<tr>
<td>8222</td>
<td>0.074</td>
<td>19.00</td>
<td>9.20</td>
<td>2.25</td>
<td>0.99</td>
<td>0.010</td>
<td>0.010</td>
<td>0.016</td>
<td>0.030</td>
<td>4.6</td>
<td>3.7</td>
<td>8.5</td>
<td>8.5</td>
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<tr>
<td>309</td>
<td>8361</td>
<td>0.039</td>
<td>22.60</td>
<td>13.70</td>
<td>2.21</td>
<td>0.88</td>
<td>0.010</td>
<td>0.021</td>
<td>0.050</td>
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<td>8.5</td>
<td>3.7</td>
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<tr>
<td>8223A</td>
<td>0.075</td>
<td>2.32</td>
<td>12.48</td>
<td>1.95</td>
<td>1.22</td>
<td>0.005</td>
<td>0.028</td>
<td>0.18</td>
<td>0.030</td>
<td>11.4</td>
<td>11.4</td>
<td>3.7</td>
<td>3.7</td>
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<tr>
<td>8335</td>
<td>0.116</td>
<td>13.07</td>
<td>2.13</td>
<td>0.67</td>
<td>0.007</td>
<td>0.015</td>
<td>0.015</td>
<td>0.059</td>
<td>Not reported</td>
<td>Not reported</td>
<td>2.5</td>
<td>2.5</td>
<td></td>
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<tr>
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<td>8166</td>
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<td>19.00</td>
<td>13.40</td>
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<td>0.44</td>
<td>0.012</td>
<td>2.13</td>
<td>0.059</td>
<td>5.1</td>
<td>5.1</td>
<td>4.8</td>
<td>4.8</td>
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<tr>
<td>8167</td>
<td>0.028</td>
<td>19.00</td>
<td>12.60</td>
<td>1.82</td>
<td>0.26</td>
<td>0.011</td>
<td>0.018</td>
<td>2.63</td>
<td>0.030</td>
<td>5.8</td>
<td>5.8</td>
<td>4.8</td>
<td>4.8</td>
</tr>
<tr>
<td>8223</td>
<td>0.042</td>
<td>18.65</td>
<td>12.68</td>
<td>1.76</td>
<td>0.91</td>
<td>0.011</td>
<td>0.018</td>
<td>2.63</td>
<td>0.030</td>
<td>5.8</td>
<td>5.8</td>
<td>4.8</td>
<td>4.8</td>
</tr>
</tbody>
</table>

(a)Measured by Westinghouse using a saturation magnetization technique.
(b)Supplied by Arcos Corporation based on Magnagage measurements.

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*Fig. 1—Simulated surfacing weld specimen*

*Fig. 2—Specimen blank for stress corrosion test*

*Fig. 3—View of jig for stressing samples*
one technique to another, each technique was consistent in itself and the overall trends indicated by different techniques were the same. The saturation magnetization was found to be the most reproducible and reliable. Hence only the ferrite values determined by this technique are used in discussing the results. However, corresponding values of ferrite number determined as per American Welding Society Standard practice are also listed in Table 1 for the convenience of the Welding Journal readers.

The distribution of ferrite in the various samples was characterized by light microscopic examination of conventionally mounted and polished metallographic sections etched in a 10% Br₂-CH₃OH solution. The distribution of carbides was determined selectively on some samples by electron microscopic examination of direct carbon extraction replicas, prepared from metallographic samples. To determine the weight fractions of carbide, some specimens were partially dissolved in 10% HCl-CH₃OH at low current densities, and the insoluble carbide residues were collected on millipore and weighed. The residues were then analyzed by standard X-ray powder diffraction techniques.

**Results**

**Microstructure**

Examination of surfacing weld samples in the optical microscope showed the microstructure to consist of large austenite grains within which the ferrite phase was distributed as shown in Fig. 4. The distribution of ferrite was found to vary with the amount of ferrite initially present in the weld.

In samples initially containing a small amount of ferrite, the ferrite phase was present discontinuously, as discreet islands decorating the dendritic structure. As the ferrite content increased, the ferrite islands started to link up to form increasingly continuous chains. This is illustrated in Fig. 5.

Postweld heat treatment of the welds resulted in a decay of the ferrite content in all the cases. Figures 6-8 depict the decay of ferrite content of the surfacing welds as a function of the duration of postweld heat treatment at 608 C (1126 F). The most dramatic decay of ferrite is observed in the case of the Type 309 stainless steel.
In general, the ferrite content drops rapidly at first, but levels off to a constant low value for times in excess of about 15 h.

Examination of carbide extraction replicas showed that, in the as-welded condition, large rounded carbides were present in a uniform distribution in the austenite phase. The ferrite phase was relatively free of carbides. Postweld heat treatment caused a precipitation of fine carbides, the distribution of which varied with the ferrite content of the weld. In welds containing large amounts of ferrite, carbide precipitation resulting from postweld heat treatment (PWHT) occurred preferentially along the dendritic ferrite network, as illustrated in Fig. 9. In welds containing small amounts of ferrite, precipitation of carbides occurred both within the islands of ferrite and along the austenite-austenite grain boundary network. An example of carbide precipitation along austenite-austenite grain boundaries is shown in Fig. 10.

Results from bulk carbide extractions (see Table 2) showed that in all the samples examined some carbide precipitation was already present in the as-welded condition. Postweld heat treatments resulted in considerable additional precipitation in Type 309 stainless steel surfacing welds. In Types 308 and 316 stainless steel surfacing welds, however, no additional carbide precipitation due to PWHT was detectible. By X-ray diffraction analysis the carbide residues were identified to be principally M$_{23}$C$_6$ type chromium rich carbides.

**Stress Corrosion Tests**

Results from metallographic evaluations of the stress corrosion samples tested in water containing 100 ppm chloride are summarized in Table 3; the numbers as well as the average size of small and large cracks are listed. For this analysis, cracks less than 0.5 mm (0.02 in.) long are defined as small cracks.

Comparison between samples can be made on the basis of the number or the size of cracks. It can be seen from Table 3 that the relative ranking of samples in terms of performance is the same regardless of which parameter is used as a basis for comparison. The size of cracks and the number of cracks go hand in hand. A convenient way to compare samples is therefore in terms of the total length of all the cracks in a given sample. This parameter is included in Table 3 and will henceforth be used as the index of stress corrosion susceptibility.

Figure 11 depicts the variation of SCC susceptibility as a function of the duration of postweld heat treatment for selected heats of Types 308 and 309 stainless steel. Interestingly, the maximum susceptibility is observed for intermediate times of heat treatment, while the susceptibility for cracking is minimal in the as-welded condition and following extended PWHT for 30 h. Only in two out of the nine welds was any cracking observed in the as-
welded condition. PWHT for 30 h eliminated the cracking completely with the exception of one surfacing weld. A low carbon grade of Type 316 stainless steel surfacing weld metal showed the least susceptibility to crack, while a high carbon grade of Type 308 stainless steel surfacing weld metal showed the greatest susceptibility to crack.

To evaluate the effect of lowering the chloride level on SCC susceptibility and to verify the trends in SCC behavior observed at the 100 ppm chloride level, a second stress corrosion test was conducted at a chloride level of 10 ppm in the water. Results from this test (Table 4) show that the lower chloride level resulted in a considerable reduction in cracking of Types 309 and 316 surfacing welds; the degree of attack in Type 308 surfacing welds was relatively the same as in 100 ppm chloride. The worst weld was once again found to be the high carbon grade of Type 308 stainless steel; earlier observations regarding the maximum attack occurring in samples heat treated for intermediate times were also confirmed.

Three different crack morphologies were observed in the stress corrosion samples:

1. Transgranular, wherein the crack path straddled across the dendrite boundaries and the austenite boundaries.
2. Intergranular, wherein the crack path was along austenite grain boundaries.
3. Interdendritic, wherein the crack path followed the dendritic pattern of ferrite distribution.

Figure 12 illustrates the three different crack morphologies:

Table 3—Results of Stress Corrosion Tests at 316 C (601 F) in Aerated Water Containing 100 ppm NaCl

<table>
<thead>
<tr>
<th>AISI Grade of Steel</th>
<th>PWHT at 608 C (1126 F), h</th>
<th>Number of Small Cracks</th>
<th>Average Depth of Small Cracks, Microns</th>
<th>Number of Large Cracks</th>
<th>Average Depth of Large Cracks, Microns</th>
<th>Total Length of Cracks, Microns</th>
<th>Crack Morphology</th>
</tr>
</thead>
<tbody>
<tr>
<td>308</td>
<td>8149</td>
<td>NIL</td>
<td>NIL</td>
<td>NIL</td>
<td>NIL</td>
<td>0</td>
<td>TG + IG</td>
</tr>
<tr>
<td>309</td>
<td>8149</td>
<td>NIL</td>
<td>NIL</td>
<td>NIL</td>
<td>NIL</td>
<td>0</td>
<td>TG + IG</td>
</tr>
<tr>
<td>316</td>
<td>8149</td>
<td>NIL</td>
<td>NIL</td>
<td>NIL</td>
<td>NIL</td>
<td>0</td>
<td>TG + IG</td>
</tr>
</tbody>
</table>

Table 2—Variation of the Amount of Carbide Precipitates With PWHT as Determined by Electrolytic Extraction

<table>
<thead>
<tr>
<th>AISI Grade of Steel</th>
<th>PWHT Duration, h</th>
<th>Extraction Residue, wt-%</th>
</tr>
</thead>
<tbody>
<tr>
<td>308</td>
<td>8149</td>
<td>AW</td>
</tr>
<tr>
<td>309</td>
<td>8149</td>
<td>AW</td>
</tr>
<tr>
<td>316</td>
<td>8149</td>
<td>AW</td>
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</tbody>
</table>

**Table 2—Variation of the Amount of Carbide Precipitates With PWHT as Determined by Electrolytic Extraction**

**Table 3—Results of Stress Corrosion Tests at 316 C (601 F) in Aerated Water Containing 100 ppm NaCl**

**Table 4—Results of Stress Corrosion Tests at 308 C (601 F) in Aerated Water Containing 100 ppm NaCl**
ent crack morphologies observed. Most specimens contained a mixture of two of the above three crack morphologies. Based on the relative proportion of the different modes, each mode is classified as the “major” or the “minor” mode. Results from such an evaluation are included in Tables 3 and 4, and lead to the following general observations:

1. Cracking in the Type 316 stainless steel surfacing welds was always transgranular; in Types 308 and 309 stainless steel surfacing welds the transgranular mode was the predominant mode in the as-welded sample.

2. With postweld heat treatment interdendritic and intergranular cracking also emerged as major modes; the interdendritic mode was associated with high ferrite welds, while the intergranular mode was associated with the low ferrite welds.

3. In tests at the 10 ppm chloride level, the degree of interdendritic attack appeared to have increased in most of the samples, compared to the tests at 100 ppm chloride level.

In view of the fact that many of the samples tested contained no cracks the observations regarding crack morphologies are not sufficiently extensive to permit quantitative and systematic characterization. The conclusions drawn here are therefore only tentative and suggestive of broad trends.

Discussion

Based on the metallographic results provided in Tables 3 and 4, the various surfacing welds can be ranked in terms of susceptibility to SCC. Such a ranking is shown in Table 5. In arriving at this ranking, the maximum total length of cracks observed in the samples regardless of the PWHT is used as the index of SCC for the surfacing weld, since this parameter is indicative of the “worst case” performance for the surfacing weld. Based on Table 5, the low carbon grades of Types 316 and 309 stainless (316L and 309L) steels can be clearly concluded to be the least susceptible to SCC in both 100 ppm and 10 ppm chloride tests. For each grade of stainless steel, the susceptibility to SCC goes up with increasing carbon level. In the 10 ppm chloride test, susceptibility to SCC declines with Types 316, 309 and 308 stainless steels in that order.

Within the range of ferrite levels investigated, there does not appear to be any correlation between the initial ferrite content and the stress corrosion susceptibility of the cladding. For instance, in Type 308 stainless steels, the cladding with 7.9% ferrite cracks more readily than the cladding containing only 1.58% ferrite, even though the carbon contents are about the same. Similarly among Type 309 stainless steels, cladding 8335 containing only 3.5% ferrite has a cracking susceptibility similar to cladding 8223A containing 9.8% ferrite. Among the Type 316 stainless steels the cladding 8166 has better resistance to SCC than the other two heats despite a lower ferrite content.

The lack of correlation between ferrite content and SCC susceptibility is in disagreement with the findings of other investigators6,7 who found that

### Table 4—Results of Stress Corrosion Tests at 316 C (601 F) in Aerated Water Containing 10 ppm NaCl

<table>
<thead>
<tr>
<th>AISI grade of steel</th>
<th>Heat identification</th>
<th>Specimen identification</th>
<th>Time of PWHT at 608 C (1126 F), h</th>
<th>Number of small cracks***</th>
<th>Average length of small cracks, microns</th>
<th>Number of large cracks***</th>
<th>Average length of large cracks, microns</th>
<th>Total length of cracks, microns</th>
<th>Crack morphology</th>
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<tbody>
<tr>
<td>308</td>
<td>8372</td>
<td>50A</td>
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<td>NIL</td>
<td>-</td>
<td>NIL</td>
<td>-</td>
<td>0</td>
<td>TG + ID</td>
</tr>
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<td></td>
<td>50B</td>
<td>1½</td>
<td>NIL</td>
<td>NIL</td>
<td>1</td>
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<td>2200</td>
<td>1600</td>
<td>TG + ID</td>
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<td>53A</td>
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<td>NIL</td>
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<td>300</td>
<td>NIL</td>
<td>600</td>
<td>600</td>
<td>1200</td>
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<td></td>
<td>8B</td>
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<td>4</td>
<td>1600</td>
<td>NIL</td>
<td>1750</td>
<td>1750</td>
<td>1200</td>
<td>ID + TG</td>
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<tr>
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<td>1½</td>
<td>300</td>
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<td>600</td>
<td>600</td>
<td>1200</td>
<td>ID + TG</td>
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<td>NIL</td>
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<td></td>
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<td>30</td>
<td>NIL</td>
<td>5</td>
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<td>309</td>
<td>8223A</td>
<td>66B</td>
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<td>30</td>
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<td>0</td>
<td>0</td>
<td>0</td>
<td>TG</td>
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***No cracks were observed in any of the other Type 309 or 316 stainless steel surfacing welds.
ferrite was beneficial in reducing SCC. It is believed that the lack of correlation between ferrite content and SCC susceptibility observed in this study may be due to the narrow range of ferrite levels investigated and due to the variable carbon content of the different surfacing welds. It is entirely conceivable that the effect of ferrite has been masked by a more dominant influence of carbon level on SCC susceptibility.

The effects of PWHT on cracking susceptibility as well as on the crack morphology deserve further discussion. Specifically, the "peak" in SCC susceptibility observed at intermediate durations of PWHT and the variability in SCC attack in terms of its interdendritic or intergranular nature must be explained. In order to do this one must consider the various microstructural and macrostructural changes that may be caused by PWHT. The more important among these changes are:

1. Relief of residual stresses.
2. Decrease of ferrite content.
3. Increase in dissolved carbon due to precipitation of carbides.
4. Sensitization of interfaces.

It is certainly possible that the surfacing welds contained residual stresses in the as-received condition. A large portion of these stresses, however, may be relieved during the various cutting and machining operations involved in preparing test specimens. Furthermore, the effect of ferrite on SCC susceptibility is well known. The dissolution of ferrite due to PWHT has been shown to be beneficial in reducing SCC. It is believed that the lack of correlation between ferrite content and SCC susceptibility observed in this study may be due to the narrow range of ferrite levels investigated and due to the variable carbon content of the different surfacing welds. It is entirely conceivable that the effect of ferrite has been masked by a more dominant influence of carbon level on SCC susceptibility.

In previous investigations where ferrite was found to be beneficial from a stress corrosion point of view, the beneficial effects were attributed to a keying action of ferrite, wherein a propagating crack was impeded by the ferrite-austenite interface boundary. It is not clear from literature whether such impeding action could be maintained even after decomposition of ferrite into carbides. In this study no correlation was observed between the initial ferrite content and the SCC susceptibility. Furthermore, the "peaking" behavior of SCC susceptibility with the duration of PWHT was independent of the degree of decomposition of ferrite due to PWHT. Peaks in SCC susceptibility at intermediate durations of PWHT were observed even in welds where no appreciable decay of the ferrite had occurred.

Ishihara et al. claimed that carbon in solution was favorable to SCC and that decreasing the amount of dissolved carbon by carbide precipitation was a major reason for the beneficial effect of PWHT. In the present investigation, some changes in the dissolved carbon level are indicated by the metallographic results. The dissolved carbon hypothesis, however, cannot explain why a peak in SCC susceptibility should occur at intermediate durations of PWHT.

Table 5—Ranking of Various Stainless Steel Surfacing Welds Based on Their Maximum Cracking Susceptibility in 316 C (601 F) Water

<table>
<thead>
<tr>
<th>AISI grade of steel</th>
<th>Heat</th>
<th>Initial ferrite content, %</th>
<th>C in steel, wt-%</th>
<th>Total crack length (max), microns</th>
<th>Crack morphology</th>
<th>Overall rank</th>
<th>Results in 100 ppm Cl-</th>
</tr>
</thead>
<tbody>
<tr>
<td>308L</td>
<td>8166</td>
<td>2.5</td>
<td>0.016</td>
<td>0</td>
<td>TG</td>
<td>1</td>
<td>0</td>
</tr>
<tr>
<td>309L</td>
<td>8361</td>
<td>8.5</td>
<td>0.039</td>
<td>600</td>
<td>TG</td>
<td>2</td>
<td>0</td>
</tr>
<tr>
<td>316</td>
<td>8167</td>
<td>5.1</td>
<td>0.028</td>
<td>2500</td>
<td>TG</td>
<td>3</td>
<td>0</td>
</tr>
<tr>
<td>308</td>
<td>8372</td>
<td>1.6</td>
<td>0.044</td>
<td>3120</td>
<td>TG + IG</td>
<td>4</td>
<td>0</td>
</tr>
<tr>
<td>309</td>
<td>8149</td>
<td>7.9</td>
<td>0.058</td>
<td>5500</td>
<td>TG + ID</td>
<td>5</td>
<td>0</td>
</tr>
<tr>
<td>309</td>
<td>8335</td>
<td>3.5</td>
<td>0.116</td>
<td>8100</td>
<td>TG + IG</td>
<td>6</td>
<td>0</td>
</tr>
<tr>
<td>309</td>
<td>8233A</td>
<td>9.8</td>
<td>0.075</td>
<td>9750</td>
<td>TG + ID</td>
<td>7</td>
<td>0</td>
</tr>
<tr>
<td>316</td>
<td>8223</td>
<td>4.8</td>
<td>0.042</td>
<td>14450</td>
<td>TG</td>
<td>8</td>
<td>0</td>
</tr>
<tr>
<td>308</td>
<td>8222</td>
<td>3.8</td>
<td>0.074</td>
<td>36000</td>
<td>IG + TG</td>
<td>9</td>
<td>0</td>
</tr>
</tbody>
</table>

Note: TG, IG and ID denote Transgranular, Intergranular and Interdendritic crack morphology, respectively.
The only possible phenomenon that can explain a bell-shaped curve of SCC susceptibility vs. time of PWHT, by causing highly localized and subtle structural and compositional changes often undetectable, is the phenomenon of sensitization. The term sensitization generally refers to the susceptibility of austenitic materials to intergranular corrosion resulting from the combined effects of precipitation of \( M_23C_6 \) type carbides at grain boundaries and other localized chemical changes concomitant to the precipitation. Formation of \( M_23C_6 \) type carbide precipitates results in a narrow region of material depleted in chromium, adjacent to the grain boundaries and renders the grain boundaries susceptible to attack by corrosive agents.

There is considerable controversy as to the actual mechanism of intergranular corrosion. Direct attack of the carbides as well as attack of the region around the carbides have been observed, depending on the corrosive agent.\(^5\)\(^6\) Chemical attack of the chromium depleted region\(^5\)\(^6\) as well as electrochemical attack of the matrix itself\(^6\) (carbide acting noble) have been suggested as possible mechanisms. Segregation of impurity elements such as P and S to grain boundaries during the sensitization heat treatment in the range 450-800°C has also been suggested as a reason for accelerated grain boundary attack.\(^5\)\(^6\)

An interesting aspect of the sensitization phenomenon is that, with prolonged exposure, the sensitization is reversed. Change in carbide morphology and replenishment of the chromium depleted layer have been suggested as the mechanisms of desensitization. It is not the intent of this study to resolve the mechanisms of sensitization and desensitization in detail. The only reason for the foregoing description of the phenomenon is to serve as a basis for explaining the observed “peak” in SCC susceptibility at intermediate PWHT, and the crack morphologies in different surfacing weld samples.

Results from this study as well as from others show that precipitation of \( M_23C_6 \) carbides occurs first at the ferrite-austenite boundaries in duplex steels. In fact, there is evidence that these boundaries would get preferentially saturated and eventual precipitation of carbides at austenite-austenite boundaries may occur only if residual carbon in solution is available at all for such precipitation.\(^5\)\(^6\) It is clear therefore that the degree of sensitization of dendrite boundaries (i.e., ferrite-austenite) vs. that of grain boundaries (austenite-austenite) in weldments would depend on the ferrite content of the weld and the carbon content of the weld.

In low ferrite surfacing welds (i.e., 8372, 8222, and 8335) although sensitization of the dendrite boundaries may occur during PWHT, the ferrite phase is distributed discontinuously and does not provide a continuous path for crack propagation. Further, sufficient carbon may be available to sensitize the austenite grain boundaries, since the ferrite-austenite dendrite boundary area is small. The austenite grain boundaries, when sensitized by PWHT can provide an easy continuous path for crack propagation. Hence intergranular fracture along the austenite boundaries can occur readily in the low ferrite stainless steel welds after PWHT, in clear agreement with the results shown in Table 2. The transition from transgranular to intergranular crack morphology following PWHT is illustrated in Fig. 13.

In the sensitized condition after PWHT, the high ferrite surfacing welds (i.e., 8149, 8361 and 823A) can have a different crack morphology compared to the low ferrite surfacing welds, due to two reasons:

1. The amount of ferrite is sufficiently large to take up all the carbon as carbides at the dendrite boundaries, so that sensitization of the austenite grain boundaries is unlikely to occur.

2. The ferrite is distributed more continuously as a chain and can provide a continuous path for crack propagation; hence in high ferrite alloys cracks will tend to be interdendritic rather than intergranular following sensitization.

This rationale is in agreement with the result shown in Table 2. A close look at Fig. 12C reveals how a crack propagates along an interdendritic boundary in clear preference to an adjoining grain boundary in a high ferrite weld.

Figure 14 illustrates the transition from a completely transgranular crack morphology to a mixed transgranular-interdendritic crack morphology due to PWHT. Ferrite levels included in this study are not sufficiently high to cause exclusively interdendritic cracks and most of the cracks are partly interdendritic and partly transgranular. Ferrite levels in excess of 10% may be needed before an exclusively interdendritic crack morphology is achieved.

Elimination of cracking after 30 h of PWHT is attributed to a desensitization process consisting of a replenishment of the chromium depleted regions near the boundaries. Estimations by Devine show that in Type 308 stainless steels where sensitization occurs at ferrite-austenite boundaries, the width of the chromium-depleted zone would be sufficiently small and that the diffusion of chromium would be sufficiently rapid to cause replenishment of the depleted layer in short times at 600°C (1112°F). On the other hand, in low ferrite alloys where sensi-
tization occurs by carbide precipitation at austenite boundaries, both sensitization and desensitization would be moved to longer times. This is in qualitative agreement with the observations made in this study that, in weld 8222 where intergranular fracture along constant boundaries is a major mode of fracture, the peak susceptibility occurs at 15 h and the intergranular fracture susceptibility persists even after 30 h of PWHT.

While the arguments relating to sensitization and desensitization can be applied reasonably well to explain the SCC behavior of Types 308 and 309 stainless steel surfacing welds, they cannot account for the behavior of Type 316 stainless steel welds, which crack completely transgranularly even after PWHT. The reason for the observed peak in SCC susceptibility at intermediate PWHT times is not clear at this time for the case of Type 316 stainless steel surfacing welds.

Conclusions

Surfacing weld samples prepared from AISI 308, 309 and 316 stainless steels have been evaluated for stress corrosion susceptibility in aerated 316 C (601 F) water containing chlorides. Based on the results of the evaluation the following conclusions have been drawn:

1. In each grade, susceptibility to SCC decreases with decreasing carbon content of the weld. Among the nine surfacing welds investigated the lowest susceptibility to SCC was exhibited by a Type 316L stainless steel surfacing weld, closely followed by Type 309L.

2. Highest susceptibility to SCC is observed in samples postweld heat treated for intermediate durations in the range of 1-15 h at 608 C (1126 F). In the as-welded condition and in the condition following prolonged heat treatments for about 30 h the SCC susceptibility is minimal.

3. SCC morphologies vary with the ferrite content, carbon content and the duration of PWHT at 608 C (1126 F). In the as-welded condition cracking is generally transgranular. With PWHT the cracks become increasingly interendritic in the high ferrite welds and intergranular in the low ferrite Types 308 and 309 stainless steel welds. The observed crack morphologies can be explained in terms of the sensitization phenomenon for Type 308 and 309 stainless steel welds.

4. There is no clear-cut correlation between the initial ferrite content of the weld and its performance in stress corrosion.

5. Reducing the chloride content of aerated water from 100 ppm to 10 ppm results in a marked reduction in the SCC susceptibility of Type 309 and 316 stainless steel welds, but there is no effect on the SCC susceptibility of Type 308 stainless steel welds. The reasons for this are not clear.

Acknowledgments

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References


Appendix: Welding Parameters for Submerged ARC Process

1. Base Material: ASTM A-533, Grade B, Q & T Plate.
2. Filler Metal: ASTM A-371, Type ER316L, 3/4 in. (1.59 mm) diameter, heat no. 7-55350.
3. Flux: First layer—Arcosite 642CDG; balance—Arcosite N82.
4. Electrical characteristics: DCSP.
5. Preheat: 95 C (203 F).
8. Voltage: 34-36 V.
9. Travel: 1/4 to 1/2 in. (6.35-12.7 mm) per ft. (3.05-6.10 m).
11. Bead width: 2.86 cm (1.13 in.)
12. Head (pitch) index: 0.15-0.35 in.
13. Oscillation width: 1/8 in. (0.32 cm).

Correction . . .

See Figs. 10 and 11 on Page 43-s in the February 1979 Welding Research Supplement. As shown on Page 43-s in February, Figs. 10 and 11 were in error but are reproduced correctly in the two opposite columns. The article of which they are part is "Cryogenic Toughness of SMA Austenitic Stainless Steel Weld Metals: Part II—Role of Nitrogen" by E. R. Szumachowski and H. F. Reid.

Fig. 10—Lateral expansion of Type 24-16-15 weld metals if tested at -320 F (~196 C)

Fig. 11—Lateral expansion of Type 18-16-2-15 weld metals if tested at -320 F (~196 C)