WELDING METALLURGY OF ALLOY HR-160

The Weldability of an Advanced Corrosion Resistant Alloy is Investigated and Linked to Microstructural Development in the Fusion Zone

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ABSTRACT

The solidification behavior and resultant solidification cracking susceptibility of autogenous gas tungsten arc fusion welds in alloy HR-160 was investigated by Varestraint testing, differential thermal analysis, and various microstructural characterization techniques. The alloy exhibited a liquidus temperature of 1387 °C and initiated solidification by a primary L → γ reaction in which Ni, Si, and Ti segregated to the interdendritic liquid and Co segregated to the γ dendrite cores. Chromium exhibited no preference for segregation to the solid or liquid phase during solidification. Solidification terminated at ≈ 1162 °C by a eutectic-type L → [γ + (Ni,Co)₁₀(Ti,Cr)₆Si₇] reaction. The (Ni,Co)₁₀(Ti,Cr)₆Si₇ phase is found to be analogous to the G phase which forms in the Ni-Ti-Si and Co-Ti-Si ternary systems, and similarities are found to exist between the solidification behavior of this commercial multicomponent alloy and the simple Ni-Si and Ni-Ti binary systems. Reasonable agreement is obtained between the calculated and measured volume percent of the [γ + (Ni,Co)₁₀(Ti,Cr)₆Si₇] eutectic-type constituent with the Scheil equation using experimentally determined k values for Si and Ti from electron microprobe data. The alloy exhibited a very high susceptibility to solidification cracking in the Varestraint test. This is attributed to a large solidification temperature range of 225 °C and the presence of 2 to 5 vol% solute rich interdendritic liquid which preferentially wets the grain boundaries and interdendritic regions.
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INTRODUCTION

Alloy HR-160 is a Ni-Co-Cr alloy with the addition of \( \approx 2.7 \) wt\% silicon that was designed for good corrosion resistance in sulfidizing atmospheres. Previous work [Refs. 1,2] has shown this alloy can form a protective oxide scale even in aggressive sulfidizing environments and exhibits corrosion rates significantly lower than other austenitic alloys such as HR-120, Alloy 800, Alloy 556, and Type 347 stainless steel. The ability to form a protective oxide scale in such environments with low oxygen partial pressures makes this alloy an attractive candidate for several industrial applications. Examples include coal gasification and fossil-fired boilers operating with low NO\(_x\) burners where enhanced sulfidation corrosion can occur [Ref. 3]. In such applications, components are often joined by fusion welding processes. In this case, microstructural development and solidification cracking susceptibility in the fusion zone need to be considered as part of the overall alloy selection process.

Solidification cracking susceptibility is known [Refs. 4,5] to be controlled mainly by the solidification temperature range and amount/distribution of solute rich liquid that exists at the terminal stages of solidification. The solidification temperature range and amount terminal liquid, in turn, are controlled by the microsegregation behavior during solidification. The degree of microsegregation in the final weld is also of considerable importance with regard to possible effects on corrosion performance. Since corrosion resistance is highly composition-dependent, the extent of local variations in composition within the dendritic substructure due to microsegregation should be considered. However, no information exists on the microsegregation behavior and weldability of this relatively new alloy. Thus, the objective of this research is to develop an understanding of the solidification behavior and resultant weldability of alloy HR-160.
EXPERIMENTAL PROCEDURE

Table 1 lists the chemical composition of the HR-160 alloy investigated in this work. The composition of several other alloys, which were used for Varestraint tests as a basis for comparison, are also listed in Table 1. The alloys were received in the mill-annealed condition with a 3.2 mm thickness. Varestraint specimens of 165 mm length x 25 mm width were prepared for autogenous weldability evaluations. Varestraint tests [Refs.6,7] were conducted over an applied strain range of 0.34% to 3.6%. The gas tungsten arc welding (GTAW) process was used with the following parameters: 100 A, 8.5 V, 3 mm/s travel speed, 3.2 mm diameter 2% thoriated tungsten electrode with a 60° tip angle, and argon shielding gas. The maximum crack length (MCL) was used as the indicator of cracking susceptibility and was measured at 100X magnification using a light optical microscope interfaced with a quantitative image analysis system.

The HR-160 alloy was further investigated by differential thermal analysis (DTA) and various microstructural characterization techniques. The DTA was conducted on a Netzsch STA 409 thermal analyzer using an argon cover gas and an alumina crucible. The sample size was = 0.65 gram and pure Ni was used as the reference material. Three tests were conducted to ensure reproducibility of the observed peaks. The samples were heated to 1400 °C at a rate of 5 °C/min to determine the solidus and liquidus temperatures. The samples were then cooled at 20 °C/min through the solidification temperature range to measure secondary reaction temperatures associated with non-equilibrium solidification. The DTA system was calibrated by determining the melting temperature of pure Ni.

Specimens were removed from the HR-160 Varestraint and DTA samples for microstructural characterization by light optical microscopy (LOM), scanning electron microscopy (SEM), electron probe microanalysis (EPMA), and diffraction analysis via back-scattered electron Kikuchi patterns (BEKP). The
Varestraint samples were mounted in planar view to examine the solidification cracks which intersected the sample surface. All samples were polished to a 0.04 μm finish using colloidal silica and electrolytically etched at = 3 volts in a 10% chromic acid/90% water solution. SEM was conducted at 10 kV in the secondary electron mode on carbon-coated samples using a JEOL 6300 field emission gun scanning electron microscope. The area fraction of eutectic-type constituent in the welds was determined on a LECO 2001 quantitative image analysis (QIA) system. Area fraction was assumed to be equal to volume fraction. Localized composition gradients due to dendritic microsegregation were quantitatively measured on an autogenous HR-160 weld (prepared under same parameters used for Varestraint testing) using EPMA. This was accomplished by marking small areas of parallel dendrites on polished and etched samples with a precision microscope scribe. The EPMA samples were then re-polished to a 0.04 μm finish and carbon coated. EPMA was conducted on a JEOL 733 microprobe equipped with four wavelength dispersive spectrometers. The microprobe was operated at an accelerating voltage of 15 kV and beam current of 20 nA. The Kα lines were used for all elements. Raw data were reduced to weight percentages using a ZAF algorithm.

Back-scattered electron Kickuchi patterns, also known as electron back-scattering patterns, were collected using a JEOL 6400 scanning electron microscope and a charge coupled device (CCD)-based camera system. The camera design has been described previously [Ref. 8]. Patterns were obtained from samples in the etched condition using a 20 kV accelerating voltage, 10 nA beam current, and 70° specimen tilt. The patterns were collected by stopping the electron beam on the feature or area of interest. The CCD camera acquisition time is controlled by automatic blanking of the SEM electron beam. Typical exposure times for this study were in the range of 2 to 10 seconds. The raw patterns were corrected for the background intensity using a flat-fielding procedure [Ref. 8]. Qualitative chemistry information was obtained
through energy dispersive x-ray spectrometry (EDS), and the crystallographic information was automatically extracted from the patterns using software developed at Sandia National Laboratories. The crystallographic information along with the qualitative chemistry information was used to search the International Center for Diffraction Data (ICDD) Powder Diffraction File (PDF). Phase identification was confirmed by overlay of a calculated pattern on the collected pattern to demonstrate the agreement between the two.
RESULTS

Figure 1 shows the Varestraint results. As a basis for comparison, data are also shown for 304 stainless steel, 310 stainless steel, and alloy IN718 evaluated under identical conditions. For the 304 stainless steel composition evaluated here (Table 1), the 1992-WRC diagram indicates this alloy has a primary ferrite solidification mode. As a result, the 304 SS exhibits excellent cracking resistance. The solidification cracking susceptibility of the 310 is higher than 304 (larger MCL) due to primary austenite solidification mode. These differences have been documented in detail in prior work [e.g., Ref. 9]. Alloy IN718 has been shown [Ref. 10] to be quite susceptible to fusion zone cracking due to the formation of a γ/Laves eutectic-type constituent which forms at ≈ 1190 °C and widens the solidification temperature range. By comparison, HR-160 exhibits the highest solidification cracking susceptibility of all the alloys evaluated.

Figure 2a shows a typical DTA on-heating curve. During heating, the sample exhibits a solidus temperature of 1322 °C and reaches the liquidus temperature at 1387 °C (equilibrium melting range of 65 °C). Figure 2b shows two typical DTA on-cooling curves. During solidification, the alloy exhibits a large exothermic peak at 1378 °C (9 °C undercooling), which is associated with formation of the primary austenite dendrites. Solidification is completed at ≈ 1162 °C by a second small, but reproducible, terminal exothermic reaction.

Figure 3 shows photomicrographs of a eutectic-type constituent observed in the Varestraint sample. As shown by the low magnification LOM photomicrograph in Figure 3a, the eutectic-type constituent was often distributed preferentially along grain boundary and interdendritic regions (see arrows). The DTA samples exhibited characteristics similar to the welds, where a eutectic-type constituent formed in the grain boundary and interdendritic areas - Figure 4. The eutectic spacing was coarser in the DTA samples and
the eutectic morphology was not as well developed as in the welds because of the slower cooling rate of the DTA sample. Quantitative image analysis revealed that 2 to 5 vol. % of this constituent formed in the weld metal.

Figure 5b shows a raw BEKP collected from the second phase region in the DTA sample, and the area used for analysis is indicated in Figure 5a. EDS of this phase showed that it contained Ni, Si, Co, Ti and Cr in order of decreasing composition, and the BEKP shows that this phase is cubic. A search of the ICDD PDF database showed that there were no entries in the database that contained all of the elements listed. However, there are three phases with identical crystal structures (face centered cubic Fm3m) and nearly identical lattice parameters that could form a solid solution to produce the phase that is present in the HR-160 DTA sample. These phases have the form of \( A_{16}B_6Si_7 \), where the A elements are Ni and Co and the B elements are Cr and Ti. The phases and their lattice parameters, as listed in the database, are shown in Table 2. It should be noted that there is another possible match to the pattern. The \( M_{23}C_6 \) carbide has a lattice parameter of 1.066 nm and is face centered cubic (Fm3m). However, the second phase in the DTA sample was analyzed by EDS using a thin-window detector and wavelength dispersive spectrometry (WDS) in order to identify the presence of carbon, and no carbon peak was observed in the EDS spectrum or the WDS analysis. In addition, the \( M_{23}C_6 \) is usually Cr-rich and the phase in the HR-160 is not. Figure 6a shows the pattern from Figure 5b overlaid with the simulation based on Ni\(_{16}Ti_6Si_7\) and there is excellent agreement between the simulation and the pattern collected from the secondary phase in the HR-160 DTA sample.

Figure 7a shows a high magnification SEM photomicrograph of the eutectic-type constituent which formed in the Varestraint sample, and the patterns collected from the two phases indicated are shown in Figures 7b and 7c. EDS of the area shown in Figure 7b indicated it was similar in composition to the
matrix, but with a slightly elevated Si content. Analysis of the BEKP shows that this is simply a face centered cubic crystal structure consistent with austenite. Figure 6b shows the pattern from Figure 7b overlaid with the simulation based on γ-Ni, and there is excellent agreement between the simulation and the pattern collected from the matrix phase within the eutectic-type constituent. The finer lamellar phase was more difficult to collect patterns from due to the small size, and the pattern shown in Figure 7c is a bit diffuse as a result, but is still usable. Analysis of the pattern and comparison to the database resulted in a similar identification to that in the DTA sample. This is demonstrated in Figure 6c, which shows the simulation based on the Ni\textsubscript{16}Ti\textsubscript{6}Si\textsubscript{7} structure overlaid on the collected pattern. As with the second phase in the DTA sample there is excellent agreement. Comparison of the pattern shown in Figure 6a with that shown in Figure 6c demonstrates that there are many similarities between the two patterns collected from the DTA and the Varestraint samples, indicating that they are from the same phase. Thus, this phase is identified as (Ni,Co)\textsubscript{16}(Ti,Cr)\textsubscript{6}Si\textsubscript{7}, and the eutectic-type constituent is identified as γ(Ni,Co)\textsubscript{16}(Ti,Cr)\textsubscript{x}Si\textsubscript{y}.

Figure 8 shows an EPMA trace conducted across several dendrites in the HR-160 autogenous weld. A typical example of a dendrite core location is noted in the EPMA composition trace. The dendrite cores are depleted in Ni, Si, and Ti. Cobalt segregates preferentially to the dendrite cores, while Cr shows little preference for segregation to either the core or interdendritic areas. The EPMA data can be used with simple solidification concepts to quantify the segregation potential of each alloying element. It is known that the diffusion of substitutional alloying elements in austenite during solidification conditions typical of fusion welding is negligible [Ref. 11]. Under these conditions, solute redistribution can be described reasonably well with the Scheil equation [Ref. 12]

\[ \frac{C_x^*}{kC_o[1 - f_x]^k - 1} \]  

(1)
where \( C_* \) is the solid composition at the solid/liquid interface, \( C_o \) is the nominal composition, \( k \) is the equilibrium distribution coefficient defined as \( C_i/C_l \) (where \( C_i \) is the solid composition and \( C_l \) is the liquid composition), and \( f_s \) is the fraction solid. Equation (1) assumes that equilibrium is maintained at the solid/liquid interface, there is negligible diffusion in the solid, perfect mixing in the liquid, and that dendrite tip undercooling is insignificant. At the start of solidification, \( f_s = 0 \), and \( k = C_{core}/C_o \) where \( C_{core} \) is the composition at the dendrite core where the first solid forms from liquid at the nominal concentration. The dendrite core compositions, nominal compositions, and \( k \) values for the major alloying elements are summarized in Table 3. Elements with lower \( k \) values exhibit a stronger tendency for microsegregation.

The very fine scale of the eutectic-type constituent in the welding metal makes it possible to measure the average "eutectic" composition via EPMA [Refs. 10, 11]. When making these measurements, it must be assumed that the data correction factors for the average eutectic composition is similar to those for the individual \( \gamma \) and \((\text{Ni},\text{Co})_6(\text{Ti},\text{Cr})_6\text{Si}_7\) phases within the eutectic. This assumption does not introduce significant errors in this study due to the relatively small differences in atomic numbers, with the Si data having the largest error. When no other eutectic-type constituents form during solidification, as observed here, this composition is equivalent to the composition of the liquid at the start of the eutectic-type transformation. The results of these measurements, averaged from four different locations, are summarized in Table 4.
DISCUSSION

Solidification

The DTA on-cooling results exhibit a large exothermic peak that is associated with formation of the primary austenite dendrites. A single terminal reaction peak is observed at 1162 °C, and both the DTA and weld microstructures exhibit a single eutectic-type $\gamma/(\text{Ni,Co})_{16}(\text{Ti,Cr})_6\text{Si}_7$ constituent. Thus, the terminal reaction peak at 1162 °C can be linked to the $L - [\gamma + (\text{Ni,Co})_{16}(\text{Ti,Cr})_6\text{Si}_7]$ eutectic-type reaction. These observations indicate that relatively minor additions of Si and Ti in alloy HR-160 have a significant influence on microstructural development during solidification. Although the secondary phase within the eutectic-type constituent is rather complex, the microstructural morphology of alloy HR-160 is similar to that expected in a simple eutectic system where single phase cored dendrites form prior to a two phase, interdendritic eutectic constituent. It is interesting to compare phase formation in this complex multi-component alloy to that in the simple binary Ni-Si and Ni-Ti systems and ternary Ni-Ti-Si and Co-Ti-Si systems.

The similarities in solidification characteristics between the Ni-Si and Ni-Ti [Ref. 14] binary systems and alloy HR-160 are summarized in Table 5. The Ni-Si and Ni-Ti systems each exhibit a terminal eutectic reaction consisting of $L - (\gamma + \text{Ni}_x\text{X})$, where X is Ti or Si. In alloy HR-160, both Si and Ti react with the solute-rich eutectic liquid to form a ternary-type compound $\text{Ni}_{16}\text{Ti}_6\text{Si}_7$. This phase is very similar to the G phase that forms in both the Ni-Ti-Si [Ref. 14] and Co-Ti-Si systems [Ref. 15]. In those ternary systems, the G phase has the stoichiometric composition of $\text{A}_{16}\text{Ti}_6\text{Si}_7$, where A = Ni or Co. Thus, in alloy HR-160, Co and Ni behave similarly as do Cr and Ti to form the $(\text{Ni,Co})_{16}(\text{Ti,Cr})_6\text{Si}_7$ phase. The temperature of the $L - (\gamma + \text{Ni}_x\text{Si})$ reaction (1143 °C) in the Ni-Si binary system is very similar to that observed in alloy HR-160 (1162 °C). Of course, in the simple Ni-Si system the reaction occurs at a single temperature and composition. By contrast, the $L - [\gamma + (\text{Ni,Co})_{16}(\text{Ti,Cr})_6\text{Si}_7]$ reaction in the multi-
component HR-160 alloy may occur over a range of temperature and composition. The combined (Ti + Si) \("eutectic\) composition of 12.4 wt\% is intermediate to the Ni-Si (11 wt\% Si) and Ni-Ti systems (14 wt\% Ti). Assuming that the equilibrium distribution coefficient, \(k\), is constant during solidification, then the maximum solid solubility of each element in the austenite of alloy HR-160, \(C_{\text{max}}\), can be estimated by \(C_{\text{max}} = kC_e\), where \(C_e\) is the eutectic composition. These values are also included in Table 5. The combined maximum solid solubility of Si and Ti in the austenite matrix, \(C_{\text{max}} = 7.8\) wt\% (Si + Ti) is slightly lower than for either the Ni-Si \((C_{\text{max}} = 8.2\) wt\% Si) or Ni-Ti system \((C_{\text{max}} = 11.6\) wt\% Ti). This can probably be attributed to the presence of other alloying elements (Co and Cr) which are also dissolved in the austenite and lower the solubility of Si and Ti. The \(k\) value for Si in the Ni-Si system, \(k_{\text{Si}} = 0.75\), is very close to that observed in alloy HR-160, \(k_{\text{Si}} = 0.71\). The \(k\) value for Ti in alloy HR-160, \(k_{\text{Ti}} = 0.44\), is appreciably lower than that for the Ni-Ti system, \(k_{\text{Ti}} = 0.83\). This difference may be linked to the reduced solubility. Any factor that reduces the solubility of Ti in austenite will also lead to increased segregation to the liquid phase and a concomitantly lower value of \(k\).

With this background, the solidification behavior of alloy HR-160 can be understood. Solidification initiates by formation of primary austenite dendrites which reject Si and Ti to the liquid. As solidification proceeds, the austenite and liquid each gradually become enriched in these elements until the maximum solid solubility in austenite is reached at 1162 °C. At this point the remaining liquid transforms to austenite and the \((\text{Ni,Co})_6(\text{Ti,Cr})_2\text{Si}_7\) phase by a eutectic-type reaction. The consistency among the alloy parameters determined here, which include the nominal alloy composition (Table 1), \(k\) values determined from EPMA (Table 3), the \("eutectic\) composition (Table 4), and quantitative metallography can be checked by calculating the amount of eutectic-type constituent, \(f_e\), which is expected to form at the terminal stage of solidification via the Scheil equation.
Here, $C_e$ is the eutectic composition and $C_o$ is the nominal alloy composition. Taking Si and Ti as the "solutions" and using the values in Table 1 for $C_o$, Table 3 for $k$, and Table 4 for $C_e$ the value of $f_e$ is calculated to be 1.6 wt% (using Si) and 2.5 wt% (using Ti). Results of the calculations using values for each solute should have identical results. The differences can be attributed to errors in the EPMA analysis used to determine $k$ and $C_e$. In addition, $k$ may vary during solidification, but here it is treated as a constant. However, the results obtained using the values for $k_{Ti}$ and $k_{Si}$ are similar and are in reasonable agreement with the experimental value of 2 to 5 vol.% determined from quantitative metallography. In addition, this calculation assumes that the densities between the austenite and $(Ni,Co)_6(Ti, Cr)_6Si_7$ phases are similar so that the weight fraction of eutectic calculated by equation (2) can be directly compared to the volume fraction of eutectic measured by quantitative image analysis. The density of the $(Ni,Co)_6(Ti, Cr)_6Si_7$ phase was estimated from the diffraction patterns at 6.7 g/cm$^3$, which is slightly lower than the density of $\gamma$, $\approx$ 8.9 g/cm$^3$. Thus, the calculated weight fraction values should be slightly lower than the measured volume fractions values, which is consistent with the comparison.

**Weldability**

Solidification cracking susceptibility is controlled mainly by the solidification temperature range ($\Delta T_s$) and amount/distribution of solute rich liquid that exists at the terminal stages of solidification. $\Delta T_s$ controls the size of the crack susceptible (solid + liquid) mushy zone while the amount and distribution of terminal solute-rich liquid controls the ease with which a crack can propagate through the low temperature
region of the mushy zone. When the solid/liquid interfacial surface tension is low, the terminal liquid easily wets the grain boundary and interdendritic regions. In this case, it is difficult to develop solid/solid boundaries to support the residual shrinkage strain and cracking susceptibility is high. Under these conditions, solidification cracking susceptibility can typically be correlated directly with the solidification temperature range [Ref. 5]. When the liquid does not wet the boundaries, it forms a spherical-type morphology and does not significantly interfere with the formation of solid/solid boundaries. This type of morphology is less detrimental, and cracking susceptibility can generally not be correlated directly with $\Delta T_s$ because cracks do not extend completely to the edge of the low temperature side of the mushy zone [Ref. 5]. When large amounts of eutectic liquid forms, typically above 10 vol. %, back-filling can occur and cracking susceptibility is typically less severe [Ref. 16].

The solidification temperature range of the fusion zone is most appropriately given by the difference between the liquidus temperature measured via the on-heating DTA scan (1387 °C) and the eutectic-type temperature measured on cooling (1162 °C). The on-heating liquidus is utilized because it contains no element of undercooling and thus closely approximates the epitaxial growth condition in the fusion zone, where solid forms directly from the pre-existing base metal grains at the fusion line. The DTA results indicate that the solidification temperature range for alloy HR-160 is 225°C. This solidification temperature range is quite large in comparison to other Ni base alloys [Refs 5,10,18]. In addition, examination of the solidification microstructures in the Varestraint samples (Figure 3) shows that the eutectic-type constituent is preferentially distributed along the interdendritic and grain boundary regions. This indicates that the terminal solute-rich liquid (which transformed to the $\gamma/(\text{Ni,Co})_6(\text{Ti,Cr})_3\text{Si}$ eutectic-type constituent when solidification was complete) was also distributed preferentially along the grain boundary and interdendritic regions. In addition, the amount of terminal liquid (2 to 5 vol.%) probably is too low for any appreciably
crack healing by back-filling [Ref. 16]. Based on these observations, crack propagation through most or all of the (solid + liquid) mushy zone can be expected and the maximum crack length should be controlled primarily by the solidification temperature range.

Figure 9 shows a plot of MCL as a function of the solidification temperature range for a wide range of experimental Ni base superalloys [Ref. 5] (evaluated under similar conditions) compared to alloys HR-160 and IN718 (from Figure 1). The solidification temperature range for IN718 was previously determined by Knorovosky at el. at 165 °C [Ref. 10]. The data reveals the direct trend between MCL and solidification temperature range, independent of alloy composition. These results indicate that the poor weldability of alloy HR-160 can be attributed to a wide solidification temperature range and the formation of a small amount of solute rich liquid which preferentially wets the grain boundary and interdendritic regions.
CONCLUSIONS

The solidification behavior and resultant solidification cracking susceptibility of autogenous gas tungsten arc fusion welds in alloy HR-160 was investigated by Varestraint testing, differential thermal analysis, and various microstructural characterization techniques. The following conclusions can be drawn from this work:

1. Solidification of alloy HR-160 is initiated by a primary L - γ reaction, and solidification completes at 1162 °C by a eutectic-type L - [γ + (Ni,Co)₁₀(Ti,Cr)₆Si₇] reaction. The (Ni,Co)₁₀(Ti,Cr)₆Si₇ phase is similar to the ternary compounds which form in the Ni-Ti-Si and Co-Ti-Si ternary systems that have the form A₁₆B₆Si₇. For HR-160, the A elements are Ni and Co while the B elements are Ti and Cr.

2. During the primary stage of solidification, Ni, Si, and Ti segregate to the interdendritic liquid and Co segregates to the γ dendrite cores. Chromium exhibits no preference for segregation to the solid or liquid phase during solidification.

3. Reasonable agreement is obtained between the calculated and measured volume percent of the [γ + (Ni,Co)₁₀(Ti,Cr)₆Si₇] eutectic-type constituent with the Scheil equation using experimentally determined k values for Si and Ti from electron microprobe data.

4. The high susceptibility to solidification cracking of alloy HR-160 can be attributed to a large solidification temperature range of 225 °C and the presence of 2 to 5 vol% solute rich interdendritic liquid which preferentially wets the grain boundaries and interdendritic regions.
Acknowledgments

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References


Table 1. Alloy compositions. All values in weight percent.

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<th>Element</th>
<th>304 Stainless Steel</th>
<th>310 Stainless Steel</th>
<th>Inconel 718</th>
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<td>Ni</td>
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<td>19.36</td>
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Table 2. Lattice parameters of various ternary phases.

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<td>Ni&lt;sub&gt;16&lt;/sub&gt;Cr&lt;sub&gt;6&lt;/sub&gt;Si&lt;sub&gt;7&lt;/sub&gt;</td>
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Table 3. Nominal compositions, dendrite core composition, and equilibrium distribution coefficients for elements in alloy HR-160. Composition values given in weight percent.

<table>
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<tr>
<th>Element</th>
<th>Nominal Composition</th>
<th>Core Composition</th>
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<td>1.08</td>
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<tr>
<td>Cr</td>
<td>27.64</td>
<td>28.0</td>
<td>1.01</td>
</tr>
<tr>
<td>Si</td>
<td>2.60</td>
<td>1.85</td>
<td>0.71</td>
</tr>
<tr>
<td>Ti</td>
<td>0.48</td>
<td>0.21</td>
<td>0.44</td>
</tr>
</tbody>
</table>

Table 4. EPMA analysis of eutectic-type constituent in weld metal of alloy HR-160. Average of four measurements. All values in weight percent.

<table>
<thead>
<tr>
<th>Element</th>
<th>Concentration, wt%</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni</td>
<td>44.2 ± 0.2</td>
</tr>
<tr>
<td>Co</td>
<td>20.2 ± 0.4</td>
</tr>
<tr>
<td>Cr</td>
<td>21.8 ± 0.8</td>
</tr>
<tr>
<td>Si</td>
<td>8.6 ± 0.2</td>
</tr>
<tr>
<td>Ti</td>
<td>3.8 ± 0.3</td>
</tr>
</tbody>
</table>

Table 5. Comparison of solidification characteristics in Ni-Si, Ni-Ti, and HR-160.

<table>
<thead>
<tr>
<th>System</th>
<th>Eutectic Reaction</th>
<th>Eutectic Temp.</th>
<th>Eutectic Comp.</th>
<th>C_{max}</th>
<th>k</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni-Si *</td>
<td>L - γ + Ni_3Si</td>
<td>1143 °C</td>
<td>11 wt% Si</td>
<td>8.2 wt% Si</td>
<td>0.75</td>
</tr>
<tr>
<td>Ni-Ti *</td>
<td>L - γ + Ni_3Ti</td>
<td>1304 °C</td>
<td>14 wt% Ti</td>
<td>11.6 wt% Ti</td>
<td>0.83</td>
</tr>
<tr>
<td>HR-160</td>
<td>L - [γ + (Ni,Co)_6(Ti,Cr)_6Si_7]</td>
<td>1162 °C</td>
<td>12.4 wt% (Si +Ti)</td>
<td>7.8 wt% (Si+Ti)**</td>
<td>k_{Si}=0.71 k_{Co}=0.44</td>
</tr>
</tbody>
</table>

* - Reference 13
** - Estimated by C_{max} = kC_e
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