Design, Fabrication, and Characterization of Graded Transition Joints

The susceptibility of hot cracking in the graded transition region is evaluated


ABSTRACT

Dissimilar welds between austenitic and ferritic steels suffer from premature failure driven by interfacial stresses and material degradation brought about by a mismatch in the coefficient of thermal expansion and carbon migration from ferritic steels to the interface, respectively. Trimetallic transition joints using graded composition between ferritic and austenitic alloys are considered a viable pathway to address this issue. However, hot cracking may occur when welding nickel alloys to stainless steel. This research attempts to reduce the hot cracking susceptibility of Inconel-82 alloys by functionally grading them with 316L. Optical and electron microscopy showed extensive cracking in the graded regions. Calculations using Scheil-Gulliver techniques attributed the cracking to the expansion in the solidification range of Inconel-82. To circumvent solidification cracking, another transition joint between SA 508 Grade 22 and SS 316L was designed and fabricated with coaxial powder-blown additive manufacturing using an SS 410-SS 316L grading. After fabrication, the joint was characterized using optical and scanning electron microscopy, wavelength dispersive spectroscopy, as well as electron backscattered and x-ray diffraction techniques. Characterization showed a successful transition joint with minor porosity. The measured composition gradients agreed with the designed composition gradients. This study showed that 12-Cr steels could potentially be used to fabricate transition joints without any hot cracking.

KEYWORDS

• Graded Transition Joint • Laser Direct Metal Deposition • Characterization • Electron Backscatter Diffraction • X-ray Diffraction

Introduction

Fossil fuel-fired and nuclear power plants use a combination of low-alloy Cr-Mo ferritic steel and high-alloy austenitic stainless steel piping. The cheaper ferritic steels are used in the lower temperature segments while the more expensive austenitic stainless steels are used in the higher temperature regions, thus requiring creep and corrosion resistance (Refs. 1–3).

This necessitates the use of dissimilar metal welds (DMWs) between these alloys (Ref. 3). However, DMWs bring about sharp changes in chemistry and coefficient of thermal expansion (CTE), which may lead to premature failure of components and facility shutdowns costing up to $850,000 per day (Ref. 3). The DMWs are characterized by two interfaces on either side of the fusion zone, one with austenitic alloy and another with ferritic alloy. In general, the interface that separates the ferritic steel and fusion is considered to undergo premature failure (Refs. 3–5). This is related to two phenomena discussed briefly in the following section.

Localized Thermal Stress

The large CTE mismatch between the 2.25Cr-1Mo and the stainless steels (typically 316 or 347H) leads to localized stress concentration that they nucleate with a spherical morphology but gradually acquire a lenticular shape (Refs. 4, 9, and 10). These type I carbides have been hypothesized to be the source for the nucleation of creep voids (Refs. 3, 4).

Following the nucleation of carbides, the chromium in the solution reduces, thus affecting the diffusivity of C. This creates a C-denuded ferritic side and a...
C-enriched interface. The C-denuded region, coupled with the stresses due to the CTE mismatch, can create a significantly weakened zone and act to reduce the creep life of the material. Accordingly, it may be concluded that the microstructural differences play a major role in determining the creep life of the dissimilar metal welds.

Existing Engineering Solutions: Trimetallic Joints

While the carbon migration can be reduced by using an Inconel filler metal for welding, it causes a semicontinuous network of carbides to develop at the interface (Ref. 11). The dissimilar welds between austenitic stainless and ferritic steels made using an Inconel filler metal fail in about one-third of their expected lifetime, while the Ni-based trimetallic transition joints fail in about half of their expected lifetime (Ref. 5). While using the trimetallic transition joint can increase the overall creep life, it is challenging to fabricate defect-free welds at the stainless steel Inconel interface.

Sireesha et al. performed detailed hot cracking studies on various filler metals to weld SS 316L to Alloy 800H to fabricate a trimetallic transition joint (Refs. 12, 13). The results showed that the filler 16-8-2 displayed the highest resistance to hot cracking, and Inconel-82 (In-82) displayed the least resistance (Ref. 13). The cracking and microfissuring was attributed to the formation of the Nb-rich Laves phase resulting in weld metal liquation. However, In-82 is the material of choice as a welding consumable due to the enhanced thermal stability, impact resistance, and tensile strength (Refs. 12, 13).

Role of Additive Manufacturing

Additive manufacturing is a promising candidate due to its ability to form functionally graded materials, allowing combinations of various different alloys (Ref. 14). Extensive work has been done by researchers at Lehigh University, Bethlehem, Pa., to develop functionally graded transition joints (GTJ) using the laser-directed energy deposition (Ref. 15) and dual-wire gas tungsten arc welding (GTAW) processes (Ref. 1). Both processes have shown reasonable promise to fabricate transition joints.

In graded transition joints, the metal matrix composition is to be gradually changed from a 100% ferritic to a 100% austenitic microstructure (Refs. 1–3, 5). Using this approach, prefabricated transition blocks with matching compositions to the low- and high-alloy steels on each respective end can be inserted between the ferritic and austenitic steel pipes. This can then be welded together with two similar welds. Though desired gradients can be obtained, hot cracking was a major concern and was reported to occur at the interfaces between the 2.25Cr-1Mo and Alloy 800 (Ref. 1) as well as in the stainless steel regions (Ref. 15) where the weld metal solidified by nonequilibrium austenitic mode.

The objectives of the current paper are as follows:

1. Evaluate the susceptibility of hot cracking in the graded transition region.
2. Develop a method to design, fabricate, and characterize a graded transition joint using the directed energy deposition.

Experimental Details

GTJ Fabrication with the Direct Metal Deposition Process

The powders (supplied by Carpenter Alloys) were manufactured by the gas atomization process, which led to a powder size range between 44–120 μm. The GTJ block was made using DMD-103D, a commercially available coaxial powder-blown laser (a diode laser operating at 910 nm) deposition technique at ORNL’s Manufacturing Demonstration Facility (MDF), Tenn. A laser power of 400 W, a constant laser travel speed of 600 mm/min with a powder feed rate of 5 g/min, and an
average build-layer thickness of 0.5 mm were used during the build. Prealloyed powders supplied by Carpenter Powder Products Inc. was used for the fabrication of the transition joints. Initially, In-82 to SS 316L reference builds were fabricated. Following this experiment, SS 410 to SS 316L transition joints were also fabricated after designing the compositional gradients. The composition in the graded region was optimized by controlling the powder flow from each hopper. A detailed explanation of the design methodologies is discussed later. The compositional gradients were achieved by gradual variation of the mass flow rates of the powders to arrive at the required grading ratios.

Compositional Design

Design methodology focuses on minimizing thermal stresses due to the CTE mismatch and the gradients in the carbon chemical potential. In this research paper, calculations focused only on SS 410 to 316L-graded transition joints. The carbon chemical potential was calculated using Thermo-Calc, a commercial software that uses the CALPHAD approach to model phase transformations in metals (Ref. 16).

To calculate the carbon chemical potential for the ferritic region, the matrix was assumed to be 100% BCC without any carbides, in agreement with the as-welded condition. For mixed BCC + FCC regions, the carbon chemical potential was calculated by setting the phase fractions of BCC and FCC to equilibrium values at 450°C. This may not have been a valid assumption because both welding and additive manufacturing leads to nonequilibrium solidification (Ref. 17). However, this assumption was considered the first step to develop a design methodology, due to lack of detailed kinetic models that consider nonequilibrium solidification (Ref. 18) as well as stability of the same for complex microstructural morphologies (Ref. 19).

After calculating the compositions that minimize the carbon chemical potential, the corresponding CTE mismatch was calculated using isoequilibrium contours developed by Elmer et al. (Ref. 20). These two calculations were done iteratively until a minimum gradient in the carbon chemical potential and CTE mismatch were obtained. The carbon-chemical potential gradient and the CTE values from the design are presented in Fig. 1.

Based on the design, the following deposition strategy was used to fabricate the SS 410 to SS 316L transition joint. Deposits were made on a SA 508 Grade 22 substrate. The first four layers were deposited as 100% SS 410 followed by two layers of 80% SS 410/20% SS 316L, 60% SS 410/40% SS 316L, 40% SS 410/60% SS 316L, and 20% SS 410/80% SS 316L each. These were followed by 15 layers of 100% SS 316L. A schematic illustration of the compositions used and the deposit is illustrated in Fig. 2A. A preheat of 300°C was maintained throughout the build to prevent the SS 410 steel from cracking.

Microstructure Characterization

The samples were sectioned using electric discharge machining. The specimens were then mounted in cold-setting epoxy and ground/polished using standard metallography techniques. The samples were given a colloidal silica (0.04 micron) final polish for the EBSD measurements. The samples were etched using a glyceregia solution. Optical micrographs were obtained using a Leica DMI 5000M optical microscope with a motorized and automated sample stage. Microhardness line scans were performed using a LECO TM103D microhardness tester with a 200-g load.

Elemental composition changes along the graded block were investigated using a JEOL JXA-8200X electron microprobe analyzer (EPMA) instrument equipped with five crystal focusing spectrometers for wavelength dispersive x-ray spectroscopy (WDS). Quantitative line scans were acquired utilizing a 15-kV accelerating voltage and an electron beam current of 100 nA. Additionally, elemental standards were used to compare experimentally acquired intensities to known standard intensities. A 200-μm step size was used for the scans, starting from the base plate and ending at the 100% 316L side of the deposit.

The XRD phase identification experiments were performed using a PANalytical X’Pert Pro diffractometer with Mo Kα radiation (λ = 0.709319 Å). Continuous 0–2θ scans were performed from nominally 17 to 42° 2θ. For these measurements, regions of interest along the build direction were isolated using zero-background plates to achieve spatial resolution. The sampling was done with finer steps (~2 mm) through the graded section to obtain the optimum balance between incident/diffracted intensity and spatial resolution in this zone.

For quantitative phase analyses, Rietveld refinements were performed.
on the obtained patterns using the high-score plus program from PANanalytical. To perform the electron backscatter diffraction (EBSD) analysis, a JEOL 6500F field emission gun-SEM (from JEOL USA Inc., Peabody, Mass.) equipped with an EDAX Apollo silicon drift detector and Hikari EBSD camera were used. The electron gun was set to an accelerating voltage of 20 kV and a tip current of 4 nA for collecting the EBSD data. The postprocessing of the collected data was performed using the TSL OIM Analysis software to obtain the phase and inverse pole figure (IPF) maps.

Results and Discussion

Rationalization of Cracking In-82 and SS 316L Reference Builds

As stated previously, the primary aim behind fabricating the In-82 to SS 316L transition joint was to evaluate the susceptibility of the microstructure to weld solidification cracking. The optical and electron microscopy shown in Fig. 3A–D displayed extensive cracking in the graded region. The EBSD shown in Fig. 3C displayed that the cracking occurred primarily along the crystallographic grain boundaries. This means that the cracks were solidification cracks (Refs. 21–23). The widely accepted theory for hot cracking/solidification cracking is that, during solidification, the liquid film always wets high-angle grain boundaries, embrittling them (Ref. 24).

This, coupled with the thermal stresses, resulted in cracking along the grain boundaries. Solidification cracking susceptibility is usually related to a large solidification temperature range. To evaluate the cracking susceptibility of various regions of the build, a composition analysis was performed using electron probe microanalysis in the cracked region adjacent to the cracks. Scheil solidification simulations were performed using Thermo-Calc to estimate the nonequilibrium solidification range as a function of compositions measured along the graded region. The results are presented in Fig. 4A–C.

Figure 4C clearly shows the increase in the solidification range with dilution of In-82 with Fe. This proves that significant changes to the composition of In-82 must be made to eliminate hot cracking in the build. For instance, EPRI has developed a new welding wire, EPRI P87, to fabricate Grade 91 to 347H steels by optimizing the hot cracking resistance with creep rupture properties (Ref. 25). In this work, we were restricted to searching for an alternate GTJ configuration that was resistant to weld solidification cracking.

Design and Evaluation of a New GTJ with 12Cr Steel as a Transition

Design Strategy. To minimize solidification cracking, we considered 12Cr martensitic stainless steel (SS 410) as a transition layer between 2.25Cr-1Mo and SS 316L. For comparison, typical elemental composition ranges of SS 316L, SS 410, and SA 508 Grade 22 are presented in Table 1.

The SS 410 was the material of choice since its Cr content is between the 2.25% Cr in the base Cr-Mo steel and 18% in the austenitic stainless steel. The material can provide an ex-
cellent compatibility with the CTE of the 2.25Cr-1Mo steel and SS 316L ma-
terial. The use of SS 410 steel is also advantageous since the cost benefits are significant compared to their austenitic counterparts (Ref. 26). The changes in chemical composition (Fe, Ni, and Cr) from the microprobe analyses are shown in Fig. 5. Although the trace elements Mo and Mn were measured, they were not presented due to the scatter in the data.

**Feasibility of Achieving Graded Composition by the DMD Process.** The experimentally measured elemental compositions of the SA 508 Grade 22 base plate, 100% SS 410, and 100% SS 316L zones in Fig. 5 were observed to be within the elemental composition limits of the commercial alloys when compared to the values in Table 1. Also presented in Fig. 5 are the expected average elemental compositions (weighted averages from Table 1 based on the grading design — Fig. 2A) overlaid (see dashed lines) on the experimentally measured compositions. Such a comparison was omitted for the trace elements (Mo and Mn) because only the maximum possible values were listed in the composition data sheets (Table 1).

Overall, a good agreement between the expected and measured compositions was observed. Such an agreement suggested that a good mixing be-

### Table 1 — Typical Compositions of the Alloys Used in This Work in wt-%

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>Si</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>N</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>SS 316L</td>
<td>0.03</td>
<td>2.00</td>
<td>0.045</td>
<td>0.03</td>
<td>0.75</td>
<td>16.0–</td>
<td>10.0–</td>
<td>2.0–</td>
<td>0.1 Bal.</td>
</tr>
<tr>
<td>(max)</td>
<td>(max)</td>
<td>(max)</td>
<td>(max)</td>
<td>(max)</td>
<td>(max)</td>
<td>18.0</td>
<td>14.0</td>
<td>3.0</td>
<td>(max)</td>
</tr>
<tr>
<td>SS 410</td>
<td>0.15</td>
<td>1.00</td>
<td>0.04</td>
<td>0.03</td>
<td>1</td>
<td>11.0–</td>
<td>—</td>
<td>—</td>
<td>— Bal.</td>
</tr>
<tr>
<td>(max)</td>
<td>(max)</td>
<td>(max)</td>
<td>(max)</td>
<td>(max)</td>
<td>(max)</td>
<td>13.5</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>SA-508 Gr 22</td>
<td>0.11–</td>
<td>0.3–</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>2.0–</td>
<td>—</td>
<td>0.9–</td>
<td>— Bal.</td>
</tr>
<tr>
<td>0.15</td>
<td>0.6</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>2.5</td>
<td>—</td>
<td>1.1</td>
<td></td>
</tr>
</tbody>
</table>
between the alloying powders, with close-to-expected ratios, were obtained for each layer within the build, and the desired compositions can be achieved through careful control of the powder flow rates.

Finally, there appeared to be a diluted zone between the base metal and the 100% SS 410 (with the composition gradually changing from the base metal to SS 410, indicated with black arrows in Fig. 5). However, this dilution was not quantified, and work was in progress as to how to calculate the dilution a priori.

The Schaeffler diagram (Ref. 27), a constitution diagram for stainless steel weld metals, is widely used to predict the final microstructure of welds based on elemental composition. By using the elemental compositions determined from the World Data System (WDS), the expected solidification microstructures at different sections of the graded joint can be estimated with the Schaeffler diagram.

For this purpose, the expected compositional changes in each zone were calculated based on weighted averages (based on the grading design — Fig. 2A) using the elemental compositions of 100% SS 410 and SS 316L as determined from the WDS analyses. The resultant average compositions, overlaid on the Schaeffler diagram, are presented in Fig. 6.

According to the diagram in Fig. 6, the 100% SS 410 and the 80% SS 410-20% SS 316L compositions remain in the M + F solidification regime, whereas the 60% SS 410-40% SS 316L and 40% SS 410-60% SS 316L fit into the A + M + F region. Finally, the 20% SS 410-80% SS 316L and the 100% SS 316L fall near the border of the A + F side, close to the fully austenitic region. It should also be emphasized that Fig. 6 is used as a guideline. Nonequilibrium solidification conditions taking place during additive manufacturing may yield different microstructures than shown on Fig. 4.

**Microstructure and Hardness in GTJs.** A montage image of the optical micrographs obtained from the graded block in its as-built state is presented in Fig. 7. Due to the unique etching characteristics of each zone (i.e., SA 508 Grade 22 base plate, 100% SS 410, SS 410-SS 316L grading, and 100% SS 316L), the different zones can be clearly distinguished in the micrograph (see arrows in Fig. 7).

The optical micrographs demonstrate a successful graded transition joint without any cracking. Minor porosity in the graded zone (highlighted with the yellow arrows on Fig. 7) was observed. Porosity formation in additively manufactured materials is currently a topic under research, and more fundamental research on the process parameter optimization is warranted to eliminate the porosity and fabricate a fully dense material (Ref. 14). The as-built structure can be further improved by eliminating the porosity either through optimizing the build parameters (Ref. 28) and/or performing a postbuild hot isostatic pressing treatment. Furthermore, the deposit exhibited a good bonding with the SA 508 Grade 22 base plate and between the layers. Finally, a heat-affected zone was observed in the base plate close to the weld interface line with the 100% SS 410 region.

Representative optical microscopy images were obtained from the base
plate, SS 410, grading, and SS 316L zones, which are presented in Fig. 8A–F, respectively. The base metal to the SS 410 interface is shown in Fig. 8B and is 100% martensitic, as expected. The WDS line scans showed significant dilution (Fig. 5), which increased the hardenability of this region.

The optical micrographs reveal that the 100% SS 410 zone showed a completely martensitic microstructure as shown in Fig. 8B. However, the Schaeffler diagram predicts a martensite + ferrite microstructure, meaning that the presence of delta ferrite can be expected. Hardness maps were generated to complement the optical microscopy results — Fig. 9. The unexpected hardness dropped in the transition region and indicated the possibility of the formation of delta ferrite in the microstructure. However, this is not supported by the microstructural analysis, so the reason for the drops in hardness warrants further investigation.

Discussion on Microstructure Evolution in GTJs. In 9–12% Cr steels, the preferred transformation path is normally as follows:

\[ L \rightarrow L + F \rightarrow F \rightarrow F + A \rightarrow A \rightarrow \text{Martensite} \] (Ref. 29).

The retention of delta ferrite (Fd) at room temperature occurs if it is significantly enriched in Cr and other ferrite-forming elements, thus stabilizing it at room temperature. The presence of delta ferrite in these martensitic steels was not desirable due to the embrittling effect (Ref. 26). This made it mandatory to investigate if residual delta ferrite was present in the steel after welding.

Along the transition region, the optical micrographs show a predominantly martensitic microstructure at locations where the dilution of SS 410 was less than 40% — Fig. 8C and D. Austenite pools began to show up at dilution levels above 40% and close to 20% — Fig. 8E and F.

The austenite phase was preferentially attacked by the etchant due to the lower chromium content in the austenite. This created a contrast where the austenite appeared darker. It was necessary to understand the percentage of phase fractions and compare it with the fractions predicted by the Schaeffler diagram. This was important because the calculations for CTE were based on the fact that the exact phase fractions could be predicted from the Schaeffler diagram. For instance, an increase in the percentage of ferrite could decrease CTE (Ref. 20) and provide a “diffusion highway” for carbon to migrate (Ref. 30). Hence, x-ray and electron backscatter diffraction measurements were performed to quantify the phase fractions and the distribution of the phases in the matrix. The results will be presented later in this paper.

The SS 316L zone showed a mixture of columnar and cellular dendritic microstructure without any delta ferrite. The absence of cracks was also surprising. It has often been suggested in the literature that stainless steels that solidify in the fully austenitic condition are more susceptible to hot cracking.
(Refs. 22, 31, and 32). During rapid solidification of stainless steel welds, a shift in the primary phase selection may occur depending on the thermal gradients and liquid solid interface velocities. Several theories based on partitionless solidification (Ref. 33) and Aziz solute trapping (Ref. 34) have been proposed to rationalize this change in the solidification mode. However, later research has shown that the shift in the solidification mode can be attributed to dendrite tip undercooling, which was proposed by Fukumoto and Kurz (Ref. 35) as well as shown experimentally by Babu et al. (Ref. 36).

Knowledge of the solidification mode of austenitic alloys is important from the standpoint of resistance to hot cracking during welding. Despite the fully austenitic mode of solidification, the reason for the high cracking resistance in this case was not fully understood. It could also be possible that the melt solidified in a ferritic delta mode, and the ferrite transformed to austenite via a massive transformation, resulting in a 100% austenitic microstructure (Ref. 32). However, such a transformation should not show any solidification subgrain boundaries. Additionally, the presence of solidification subgrain boundaries suggests otherwise (Ref. 32).

The dendritic solidification microstructure observed in the 100% SS 316L zone suggested that complete austenitic solidification took place in this zone. On the other hand, the changes in the substructure from cellular to columnar dendrites could be related to the degree of constitutional undercooling where formation of columnar dendrites was favored as the undercooling was increased (Ref. 37).

**Validation of Phase Evolution in GTJs.** The XRD patterns obtained from various isolated zones of the transition joint and the changes in the volume fractions of the FCC (austenite) and BCC (ferrite/martensite) phases as a function of position along the build direction are presented in Fig. 10A and B, respectively. Figure 10A shows the obtained diffraction patterns starting from the base plate, passing through the 100% SS 410 and the graded zone, ending at the 100% SS 316L zone. The total height of the transition joint was ~22 mm and the spatial positions from where the patterns were obtained are presented next to each pattern (see in red), with the zero-point corresponding to the bottom of the base plate. Here, such positions correspond to the midpoints of each isolated zone since the diffraction data was averaged over the measured zone. Additionally, only two regions from the 100% SS 316L were presented to save space because no major changes were recorded upwards of these zones. It should also be noted that due to having the same crystal structure (BCC) with similar lattice parameters, it was not possible to differentiate the ferrite and martensite phases, particularly in the transition zone from the SA 508 Grade 22 base plate to the 100% SS 410 zone.

In Fig. 10A, the strongest peaks for the FCC (austenite) and the BCC (ferrite/martensite) phases were observed to belong to the (111) and (110) reflections, respectively (see highlighted with arrows). Thus, these peaks were selected for pointing out the qualitative changes in the FCC and BCC phases. The XRD patterns revealed a fully ferritic structure in the SA508 Grade 22 base plate.

According to the Schaeffler diagram shown in Fig. 6, the SS 410-rich side of
the grading can be expected to solidify with a mixture of the ferrite and martensite phases. Therefore, in the subsequent patterns through grading the BCC peaks from the pure ferrite were replaced with the BCC peaks from the mixture of ferrite and martensite. Additionally, the presence of the FCC phase (austenite) started to be visible in the base plate to the grading transition zone. It then became more prominent further into the grading, eventually turning into the only present phase in the 100% SS 316L zone. The presence of austenite as the single phase in the 100% SS 316L zone also revealed that a fully austenitic solidification occurred without the presence of a ferritic phase as discussed in the microstructure analysis section.

Furthermore, it is important to note that broad BCC peaks were observed in the graded zone (indicated in Fig. 10A) and emphasized in Fig. 10C) as opposed to the sharper peaks in the base metal, which may suggest the presence of either closely overlapping peaks of ferrite and martensite phases or a lattice distortion, or both. Lattice distortion could indicate the formation of fresh martensite (i.e., solidification from liquid occurred as austenite transformed into martensite during cooling). Martensitic transformations occurred with an increased dislocation density and twinning, which can also cause significant lattice distortion/peak broadening.

Figure 10B shows the phase fraction data points (obtained from the Rietveld refinements), which correspond to the average weight fractions of the FCC and the BCC phases over the zones that the measurements were performed. Also shown in the graph are the borders of each measured zone (separated with dotted lines) as well as fits to the phase fraction data. The phase fraction showed a sigmoidal type of distribution with a relatively sharp change over a short distance, which corresponded to the graded zone, as expected. The midpoint of the graded zone should ideally correspond to the crossing point of the fits to the phase fractions (i.e., 50% of both FCC and BCC). Figure 10B shows that this crossing point corresponds to a distance of ~10 mm from the bottom of the base plate, which approximately corresponds to the midpoint of the graded zone based on optical observations.

The evolution of the lattice parameters of both the FCC and the BCC phases are presented in Fig. 11. It is interesting to note the increase in the lattice parameter of martensite and a decrease in the lattice parameter of the austenite in the graded regions compared to the regions that were 100% SS 410 and SS 316L, respectively.

At present, the reason for the increase in the lattice parameter of the martensite is not yet clear. The decrease in the lattice parameter of the austenite can be explained as follows. Martensite formation can induce a compressive residual stress due to the volume expansion associated with the transformation. This could be manifested as a reduction in the lattice parameter of austenite. Finally, the drop in the lattice parameter of the BCC phase near the 12-mm mark could be due to the delta ferrite formation. However, it should be noted that this decrease could also be due to the error of the fit since the BCC phase fraction in this zone was considerably low — Fig. 10B.

**Morphology and Crystallography of the GTJ Microstructure.** The EBSD measurements were performed to complement the optical microscopy investigations. EBSD was performed primarily to study the microstructures in the graded region. It may be recalled that one of the primary aims of fabricating the transition joint is to arrest the carbon migration. The idea behind using a BCC + FCC graded region is to prevent a continuous network of BCC in a FCC matrix. It is hypothesized that a discontinuous BCC phase will not provide a diffusion path for the carbon atoms to diffuse. The EBSD inverse pole figures and phase maps obtained from the base material, graded transition region, and the 100% SS 316L zone are shown in Figs. 12 and 13A–G, respectively.

The phase maps in Fig. 13 show the distribution of the FCC (austenite)
and the BCC (ferrite/martensite) phases in the material, which are represented with the green and red colors, respectively. Additionally, in agreement with the microscopy and XRD results, the structure in the 100% SS 316L zone was observed to be fully austenitic. Figures 12 and 13B reveal that the BCC (martensite/ferrite) phase formed a semicontinuous network inside a continuous FCC (austenite) matrix. This observation may suggest that the initial solidification took place in austenitic mode, and the martensite/ferrite phase was formed by a solid-state transformation during cooling.

Figure 12E–G shows that the microstructure was 100% martensitic at the dilution zone of the 2.25Cr-1Mo steel (SA 508 Grade 2 base plate) 100% SS 410, and 80% SS 410-20% SS 316L regions. We arrived at this conclusion based on the heavily dislocated structure of martensite that manifested itself on the image quality of the Kikuchi patterns (Ref. 38).

These observations were also found to be in agreement with the Schaeffler diagram (Fig. 6), which predicted a M + F type structure for these composition ranges. However, as the percentage of alloying with SS 316L increased, the fraction of austenite phases steadily increased — Fig. 13B–D. The 60% SS 410-40% SS 316L and the 40% SS 410-60% SS 316L agreed with the expected A + M + F solidification structure from the Schaeffler diagram, with a more dominant presence of the BCC phases (red) compared to the FCC phase (green). It is also interesting to note that the austenite and the martensite phases did not form a continuous matrix but were in discrete packets or locations.

Thus, it is hypothesized that the microstructure was more efficient in blocking carbon migration. However, more experiments involving postweld heat treatments are necessary to validate the hypothesis.

Finally, the microstructure observed in the 20% SS 410-80% SS 316L zone in Fig. 13B also appeared to agree with the expected A + F structure from the Schaeffler diagram (Fig. 6), with the more dominant presence of the FCC phase (green). However, at present, it is not clear whether the FCC phase in this region was martensite or ferrite (formed due to solidification in the FA mode).

The decrease in the lattice parameter may also support the fact that this could have been transformed ferrite and not martensite. To further understand if the ferrite is a result of solidification in the FA mode, the grain boundaries between the austenite and martensite must be analyzed. Nevertheless, the fact that the observed microstructures (Fig. 13) agreed with those expected from the Schaeffler diagram (Fig. 6), and that the martensite phase transformed from austenite, can strongly suggest that a chemical mixing between the two alloys occurred during deposition.

In the case of FCC to BCC transformations, there often exists a distinct orientation relationship (OR) between the parent and the product phases. A well-known commonality of the ORs is that the close packed planes tend to be parallel to minimize the interfacial energy (Ref. 39). The most common ORs in steel are the Kurdjimov-Sachs, Nishiyama Wasserman, and the Bain orientation relationships, though other less widely known ORs also exist, such as Greninger and Troiano (Ref. 39). The Kurdjumov-Sachs (K-S) OR requires the following condition between the two phases: {111} FCC//{110} BCC, <110> FCC//<111> BCC. However, slight deviations from this exact relation are also possible (Ref. 39). Figure 14 presents EBSD micrographs (obtained at 3500x) showing a more detailed view of the BCC phase inside the austenite grains.

To test the presence of a K-S OR, the phase boundaries satisfying the following: <111> FCC//<110> BCC, <110> FCC//<111> BCC relation within 5 deg were determined using the TSL OIM software and highlighted with blue on the phase map in Fig. 14. It shows that the base and product phase boundaries satisfy the K-S OR within a 5-deg tolerance. It can further be observed in Fig. 14 that not all of the austenite-martensite phase boundaries fit this relation. Several of these boundaries are pointed out with black arrows in the phase map. However, at this stage, it is not possible to clearly state if the BCC phase is delta ferrite or martensite. Further investigation is warranted in this area. Future work would aim at evaluating the mechanical properties of the transition joint in the as-deposited condition and after subjecting it to long-term aging to characterize the creep response.

**Implications of the Current Results**

As previously pointed out, the idea of a graded transition joint is very recent but not new. The major problem that has been reported in the literature is the dilution of Ni-based alloys with Fe leading to extensive solidification cracking. This paper proposes some design rules where SS 410 has been used instead of In-82 to prevent hot cracking. Based on the extensive characterization results, the present results indicate the feasibility of using the proposed design rules and current fabrication technique to fabricate a crack-free, functionally graded transition joint. The microstructure in the graded region did not have a continuous matrix but a discrete mix of a FCC and BCC structure. The diffusivity of C in BCC is orders of magnitude higher than that of FCC, resulting in a discontinuous network of BCC that could help prevent carbon migrating from the 214Cr-1Mo side to the austenitic side. However, before deployment of these joints, it is necessary to do the following:

1. Validate the hypothesis that a mixed microstructure leads to reduced carbon migration due to the removal of the highest for C migration by performing long-term aging treatments.
2. Evaluate the substitute SS 410 instead of In-82 since the effect of such a substitution on the creep rupture properties is not yet known.
3. Understand the unknown mechanism of strain partitioning between the FCC and BCC in the graded region during loading.

**Summary and Conclusions**

The conclusions from this work are as follows:

1. Hot cracking was a major problem while fabricating In-82 and SS 316L transition joints. Hence, 12-Cr steels were used and characterized instead of In-82.
2. The composition gradients across
the build measured using a wavelength dispersive spectroscopy (WDS) displayed good agreement with the build design, showing that good control of powder-mixing ratios was achieved.

3. Optical micrographs demonstrated that a good bonding was achieved between the base plate and the deposit as well as between the subsequent layers. While cracking was not observed, some level of porosity was detected, particularly in the graded zone.

4. Through scanning electron microscopy and electron backscatter diffraction measurements, it was deduced that the 100% SS 316L solidified in the fully austenitic mode. However, despite the fact that the austenite solidified by a primary austenite solidification, there was no cracking detected. The reason is under investigation.

5. Extensive peak broadening was observed in the FCC phases in the graded region, suggesting the formation of martensite. The volume expansion of the martensite resulted in a compressive residual stress in the austenite grains characterized by a decrease in the lattice spacing in the graded region.

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