Evaluation of Microstructure at Interfaces of Welded Joint Between Low Alloy Steel and Stainless Steel

K. RAVIKIRAN, G. DAS, SURANJIT KUMAR, P.K. SINGH, K. SIVAPRASAD, and M. GHOSH

In the present investigation, welded joints between low alloy steel and 304LN stainless steel were considered. The joints were fabricated using Inconel 152 and 182 as buttering materials. A thin layer of martensite, Type-I boundary, and Type-II boundary were revealed near the fusion boundary between low alloy steel and the buttering material. The polygonal grain structure and unmixed zone were found near the fusion boundary between the weld metal and austenitic stainless steel. During a uniaxial in situ deformation test in an SEM, it was revealed that crack initiation and propagation were through the buttering material. In this respect, the weld fabricated with the Inconel 152 buttering material exhibited better joint efficiency and a higher strain-hardening exponent than the joint produced with the Inconel 182 buttering material. The improved strength of the former joint was attributed to a fine grain structure, fine-scale distribution of complex carbide phases, and qualitatively high dislocation density within the Inconel 152 buttering material.

https://doi.org/10.1007/s11661-019-05192-2
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I. INTRODUCTION

WELDED components find wide application in the nuclear industry. For these assemblies, stringent safety is the prime requirement. The transition joint between stainless steel pipe and low alloy steel nozzle of the pressure vessel is one of such joints. SA 312 type 304LN stainless steel (304LN SS) has been used as a piping material owing to its excellent resistance to corrosion and embrittlement. Low alloy steel ASTM A508 Grade 3 Class I (LAS) has been used for the steam drum due to its adequate strength, toughness, and weldability.[1] Joints are produced with predefined parameters; however, these joints often fail during service exploitation due to degeneration/disbonding.[2]

A review was carried out by Lundin[3] on the welded assembly between low alloy steel and stainless steel. Physical and microstructural heterogeneities were reported for the joints. These occurred due to the variation in carbon concentration, differences in the coefficient of thermal expansion between virgin alloys, undesirable thermal stress, and phase transformation near the interface. Moreover, a composition variation across the interface was created.[4] These drawbacks were mitigated to some extent with the use of austenitic alloys as buttering material and weld metal. The feasibility of using Inconel and stainless steel as a filler alloy for joining LAS and stainless steel was studied by Sireesha et al.[5] They reported that Inconel became more effective in reducing thermal stresses and microstructural heterogeneity with respect to stainless steel for welded joints. Inconel alloys can also restrict carbon diffusion to a greater extent with respect to austenitic stainless steel.[6] The most commonly used nickel-base alloys are Inconel 82 and 182. However, joints fabricated with Inconel 82/182 exhibited susceptibility to stress corrosion cracking due to the presence of chromium.[7] On the other hand, it has been summarized by another group of researchers that welded joints fabricated with austenitic stainless steel exhibited superior mechanical properties at room temperature compared to joints produced with Inconel 182.[8]

Joining of austenitic stainless steel to low alloy steel using IN82/182 developed dendritic structure and segregation of secondary phases within dendritic arms. Signature of partial recrystallization and extensive grain boundary migration within weld metal were reported by Jang et al.[9] and Choi et al.[10] reported welding of Alloy 690 and A533 Gr.B using Inconel 152...
filler material. The presence of martensite, carbon-depleted region, and heat-affected zone (HAZ) near the fusion boundary were revealed. Brentrup and Dupont\cite{11} used dual wire feeders to join 2.25Cr-1Mo steel with three alloys, namely, IN800, IN82, and 347 SS. Their methodology helped in achieving a smooth transition across the bond line and reducing the sharp change in elemental concentration across the fusion boundary. In a different endeavor, the microstructural investigation was carried out on the SA508-309L/308L-316L welded joint.\cite{12} The presence of martensite, Type-I boundary, Type-II boundary, and alloy carbides was observed.\cite{13}

To reduce heterogeneities, the narrow gap welding of carbon steel and stainless steel was attempted. Nivas et al.\cite{14} joined stainless steel and low alloy steel with IN82 filler alloy by conventional V-groove and narrow gap welding. Narrow gap welding was also used for joining same base alloys using IN52 filler material.\cite{15} Carbon-depleted zone and Type-II boundary were absent in the same assembly. Few high-energy-density processes were also carried out to join dissimilar materials. Alloy 13CrMo44 was joined with AISI 347 by laser technology with and without filler material.\cite{16} Electron beam welding was performed for joining 2.25Cr-1Mo ferritic steel and 304L austenitic stainless steel.\cite{17} Three methods, namely, autogenous welding, welding with E308 austenitic filler wire, and welding with Inconel 82 filler wire, were adopted for fabrication of welds. The result confirmed that Inconel 82 might provide the best quality joint. The prospect of friction welding of different steels was also examined.\cite{18} However, the preceding welding method is not adopted widely over the conventional joining technique owing to economic limitations and lack of control over the process.

From the preceding discussion, it becomes evident that the role of IN82/182, 308L/309L has been explored to some extent for joining low alloy steel and austenitic stainless steel. In this respect, IN152 is a relatively newer alloy and has not been assessed thoroughly. The alloy exhibits adequate weldability and corrosion resistance.\cite{7, 19} Therefore, in the present endeavor, attempts have been made to draw a comparison between transition joints fabricated with Inconel 182 and Inconel 152 buttering materials. The correlation has been drawn between the structure and property of the assemblies to identify the preferable welding consumable.

### II. MATERIALS AND EXPERIMENTAL PROCEDURE

Low alloy steel and 304LN austenitic stainless steel were used for welding. Inconel 182 and Inconel 152 alloys (henceforth, IN182 and IN152, respectively) were chosen as both buttering material and weld metal. The chemical compositions of the alloys are collated in Table I. The bulk chemical compositions of low alloy steel and stainless steel were determined by spark emission spectroscopy (Bruker Q8 Magellan). For analysis, a rectangular polished block of ~ 20 mm² was taken and cleaned with acetone, and a spark was created. Readings were directly obtained over the screen. To obtain bulk chemical compositions of IN152 and IN182, fine turnings were dissolved in an acidic medium. Those samples were investigated in an inductively coupled plasma spectrometer (Thermo Scientific, ICAP 7600). Finally, a LECO gas analyzer was used to measure the carbon and sulfur content of all alloys.

The original dimensions of the pipe were ~ 324-mm o.d., ~ 200-mm length, and ~ 25-mm wall thickness (Figure 1a). During the fabrication of joints, buttering was done by shielded metal arc welding (SMAW). Root passes were carried out by gas tungsten arc welding (GTAW). Filling passes were performed by SMAW. Two sets of plain bevel joints were prepared with a bevel angle, root gap, and root face of ~ 37.5 deg., ~ 2.4 mm, and ~ 1.2 mm, respectively (Figure 1b); Joint 1 was fabricated with little change in dimensions keeping the same geometry with respect to Reference 8). Before depositing the weld metal, buttering was done on the LAS side using the SMAW–direct current reverse polarity method (Figure 1b). In this process, the consumable electrodes were ENiCrFe-3 (4-mm o.d.) and ENiCrFe-7 (4-mm o.d.) for Joint 1 (IN182) and Joint 2 (IN152), respectively. A preheating temperature of 373 K ± 10 K (100 °C ± 10 °C) and a maximum interpass temperature of ~ 383 K (~ 110 °C) were maintained. Postweld heat treatment (PWHT) was carried out after buttering at ~ 883 K (610 °C) for 90 minutes to relieve the internal stress. The final width of the buttering material was 4–5.5 mm. During GTAW of root passes, direct current straight polarity (DCSP) with 100 pct Ar gas shielding was employed. The nonconsumable electrode for this purpose was EWCe (2.4-mm o.d.). Filler rods were INNiCrFe-3 (1.6-mm o.d.) and INNiCrFe-7 (1.6-mm o.d.) for Joints 1 and 2, respectively. Following this, SMAW-DCSP was used for the

| Table I. Chemical Composition of Base Alloys and Welding Consumables |
|-------------------|---|---|---|---|---|---|---|---|---|---|
| Alloy             | C  | Mn | P  | S  | Si | Ni | Cr | Mo | Cu | Nb | Fe |
| LAS              | 0.20 | 1.2 | 0.010 | 0.001 | 0.001 | 0.80 | 0.20 | 0.50 | 0.03 | — | bal |
| 304LN            | 0.03 | 2.0 | 0.050 | 0.003 | 1.000 | 8.00 | 18.00 | — | 0.50 | 1.0 | 10.0 |
| IN182            | 0.10 | 5.0 | 0.030 | 0.020 | 1.000 | bal | 17.30 | — | 0.50 | 1.9 | 10.6 |
| IN152            | 0.04 | 3.0 | 0.003 | 0.001 | 0.500 | bal | 29.00 | 0.01 | 0.01 | — | bal |

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remaining filling passes with ENiCrFe-3 (4-mm o.d.) and ENiCrFe-7 (4-mm o.d.) electrodes for Joints 1 and 2, respectively (Figure 1(c)). The welding parameters are furnished in Table II.

Sampling has been done from the cross section of the welded joints for microscopic investigation (Figure 2(a)). Specimens were prepared by the conventional metallographic technique. Etching was done for the steel side with Nital (5 vol pct HNO₃ in alcohol) and for the remaining part of the sample with Glyceregia (15 mL glycerol, 10 mL HCl, and 5 mL HNO₃). Microstructural features were examined in an optical microscope (Leica DM 300) and scanning electron microscope (FEI Nova Nano SEM). The SEM was operated in secondary electron (SE) mode. Imaging was performed at an accelerating voltage of 15 kV. The distance between the detector and the sample surface was ~10 mm during the investigation. Statistical analysis of structural constituents was done using image analysis software (ImageJ Software). For image analysis, five different frames were selected randomly from various locations of the specimens.

Submicron structural characteristics near the fusion boundary between low alloy steel and buttering material (FB-I) were revealed by transmission electron microscopy (TEM). For this, a thin slice was cut from the cross section of the welded joint by the slow speed diamond saw (Buheler Isomet). The interface was at the center of the slice. The thickness of the slice was reduced to ~0.08 mm by grinding. Coupons of ~3.0 mm were obtained from the slice. Twin jet electropolishing was carried out (Fischione 120) to create a perforation around FB-I over the coupon. The electrolyte was a mixture (9:1) of ethanol and Perchloric acid. During

### Table II. Buttering and Welding Parameters for Fabricating the Joints Between Low Alloy Steel and Stainless Steel

<table>
<thead>
<tr>
<th>Welding Steps</th>
<th>Buttering/Welding with IN182 (Joint 1)</th>
<th>Buttering/Welding with IN152 (Joint 2)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Current (A)</td>
<td>Voltage (V)</td>
</tr>
<tr>
<td>Buttering (SMAW)</td>
<td>130 to 140</td>
<td>20 to 25</td>
</tr>
<tr>
<td>Root Passes (GTAW)</td>
<td>130 to 150</td>
<td>12 to 14</td>
</tr>
<tr>
<td>Fill Passes (SMAW)</td>
<td>95 to 100</td>
<td>20 to 22</td>
</tr>
</tbody>
</table>
polishing, temperature and voltage were maintained at 243 K (\(\sim 30^\circ C\)) and \(\sim 35\) V, respectively. The prepared samples were examined in an analytical TEM (Philips CM200) using energy-dispersive spectroscopy (EDS, EDAX Genesis).

X-ray diffraction (XRD) study was carried out near the fusion boundary between the weld metal and 304LN SS for identifying new phases (Bruker D8 Discover). The measurement range (2\(\theta\)), step size, current, and voltage were 30 to 110 deg, 0.02 deg/s, \(\sim 40\) mA, and \(\sim 40\) kV, respectively. The sample was cut parallel to FB-II, and the investigation was started from the weld metal side. The study was repeated and continued until the stainless steel region was reached after removing material by polishing. Thus, the diffraction study was performed 4 times for each welded joint. One representative indexed pattern has been furnished. The microhardness (Leica-VMHT) was determined per ASTM E384 across the weld centerline at a depth of \(\sim 2\) mm from the surface under \(\sim 50\)-gf load with 15-second dwelling time.

To examine the real-time structural changes, tensile testing was carried out on miniature specimens mounted over the deformation stage inside the SEM. The stage (2-kN load cell) was attached to an SEM (Hitachi S3400N). The location of sampling and the machined specimen are displayed in Figure 2(b). Specimens were polished from both sides to prepare a scratch-free mirror finish flat surface. The test was repeated for both welds at least 3 times to obtain reliable and reproducible data.

The crosshead speed during testing was \(\sim 0.05\) mm/min, and the data acquisition rate was \(\sim 500/s\). Images were captured from the gage area during testing. The load vs extension data of the tensile tests were used to obtain the yield strength, tensile strength, uniform strain, and breaking strain of the joints.

The strain-hardening behavior of the samples was studied using the relation \(\sigma = K\varepsilon_p^n\) [20], where \(K\) and \(n\) were the strength coefficient and work-hardening exponent, respectively. To determine \(K\) and \(n\), true stress–strain values (\(\sigma – \varepsilon\)) were calculated from the engineering stress–strain (\(s – \varepsilon\)) data considering the relations \(\sigma = s (1 + \varepsilon)\) and \(\varepsilon = \ln (1 + \varepsilon)\). In that case, the plastic strain was presented by \(\varepsilon_p = \varepsilon - \sigma/E\), where \(\varepsilon_p\) was the plastic component of the total strain (\(\varepsilon_t\)) and \(E\) was the elastic modulus. The true stress–true plastic strain plot produced a straight line when the logarithmic scale was considered. The slope of this straight line became the strain-hardening exponent (\(n\)). Broken tensile specimens were examined in the SEM for fractographic investigation.

### III. RESULTS

#### A. Microstructural Investigation

The microstructure of virgin low alloy steel contained tempered martensite. 304LN stainless steel consisted of equiaxed austenite grains (\(\sim 56\) \(\mu\)m) with annealing twins. After welding, a significant microstructural...
change occurred within the low alloy steel. Adjacent to FB-I, the grain size became coarse (Region I of Figures 3(a) and (b)). Away from this region, the presence of fine grain ferrite and unresolved pearlite colony was observed (Region II of Figures 3(a) and (b)).

Near FB-I, two different boundaries were present (Figures 3(c) and (d)). One of them was perpendicular to the fusion boundary and termed Type-I boundary. The other one was nearly parallel with the fusion boundary and termed Type-II boundary. The spacing between two successive Type-I boundaries was unequal. The Type-I boundary at some locations exhibited branching. The Type-II boundary was discontinuous in nature, which means at some locations it was completely absent.

Near FB-I, a deeply etched zone was found (Figure 4). This region primarily consisted of martensite. The presence of martensite was confirmed by selected area diffraction pattern (SADP) during TEM investigation (Figure 5).

Within the buttering material, the precipitation of the second phase was found near FB-I (Figures 6(a) through (d)).[21] The precipitation occurred predominantly along the boundary for Joint 1. On the other hand, the precipitation for Joint 2 was well distributed within the matrix of the buttering alloy. The precipitates were complex carbides and contained mainly Cr, Nb, and Fe. Precipitates were pre-existing within buttering alloys; they also formed during welding and subsequent PWHT due to the atomic migration of chemical species. Identification of these two different sets of second phases was beyond the scope of the present study.

The solidification microstructure of the buttering material changed from cellular (close to FB-I) to columnar dendritic (near the midsection) (Figures 7(a) and (b)). Both IN182 and IN152 weld metals were solidified in the austenitic phase near FB-I. Characteristics of the austenitic/ferritic (AF) and ferritic/austenitic (FA) modes of solidification were also prevalent for both joints (Figures 7(a) and (b)). Grain boundary migration was observed within the weld metal.

The Type-II fusion boundary between the weld metal and 304LN SS was sharp and did not reveal the formation of a new phase (Figures 7(c) and (d)). The absence of a new phase was confirmed by XRD study (Figure 8). The average grain size of 304LN SS adjacent to the interface was increased (Figures 7(c) and (d)). An unmixed zone of about ~45 µm width between the weld metal and 304LN SS was found near FB-II (Figures 7(c) and (d)). Quantitative microstructural information across welds is collated in Tables III and IV.

B. Microhardness Evaluation

The microhardnesses of low alloy steel and 304LN austenitic stainless steel were ~275 and ~260 VHN, respectively. The microhardness profile across fusion boundary-I depicted a peak, followed by a sharp fall (Figures 9(a) and (b)). The peak value in the case of Joint 2 was marginally higher than that of Joint 1. This peak (~320 VHN) was due to the presence of martensite.

![Fig. 3—Optical images of welded assemblies: (a) and (c) interface of joint with IN182 buttering and (b) and (d) interface of joint with IN152 buttering.](image-url)
close to FB-I. Within the buttering material, the sharp fall in hardness value (~210 VHN) was higher than the decrement observed within the LAS (~230 VHN). The microhardness profile exhibited fluctuation within the weld metal owing to the inherent heterogeneity of the solidification structure. Close to FB-II, the microhardness profile exhibited a small increment (~290 to 305 VHN) (Figures 9(c) and (d)).

Fig. 4—SEM images showing the presence of martensite near FB-I of weld fabricated with (a) IN182 buttering material and (b) IN152 buttering material.

Fig. 5—TEM images of martensitic lath along with its corresponding diffraction pattern near FB-I for different welded joints: (a) and (b) buttering with IN182 and (c) and (d) buttering with IN152.
C. In Situ Deformation Test and Fractography

The results of tensile tests are collated in Table V. Within the elastic limit, no change was observed over the gage section. After the yield point, faint shear bands appeared in the buttering material (Figures 10(a) and 11(a)). The low alloy steel side did not indicate any structural change. The crack was initiated from the open end of the rectangular specimen away from FB-I within the buttering material (Figures 10(b) and 11(b)). Crack propagation was through the buttering material for both joints (Figures 10(c) and 11(c)). Crack movement was almost parallel to the interface in the case of the joint with the IN182 buttering alloy. The crack path was in zig-zag fashion for the joint with IN152 buttering material. It can be inferred that the use of IN152 substantially improved the joint strength with respect to IN182 for welding LAS to 304LN SS. Breaking strain exhibited meager variation in this respect for these joints (Table V). IN152 weld displayed a higher average strain-hardening exponent in comparison to IN182 within the limit of yield strength (YS) and ultimate tensile strength (UTS). Fractography of both specimens divulged ductile dimple fracture (Figure 12).

IV. DISCUSSION

A. Microstructural Characterization

1. Region around FB-I

The microstructure of any region of a welded assembly depends on the thermal history. Region I corresponded to the HAZ of low alloy steel. Lu et al. [24] observed a small endothermic peak at \(~750°C\) for A508-3 steel during the thermoanalytical experiment. The peak corresponded to the transformation of bainite to austenite. During the heating cycle, the temperature of Region I exceeded \(A_{c3}\) and transformed to austenite. Thermal energy during welding and PWHT resulted in the significant diffusion of carbon from the LAS to the buttering material. This created a carbon-depleted region adjacent to FB-I, and the grains became coarse. [8] In this respect, diffusion also took place for Cr and Nb from the buttering material to the LAS down the concentration gradient. [8,14]

During welding, tempered martensite of low alloy steel transformed into a ferrite and cementite aggregate in Region II. The peak temperature was between \(A_{c3}\) and \(A_{c1}\) in this region. The coexistence of ferrite and austenite inhibited grain coarsening in this region. [8] At
the time of cooling, ferrite remained unaltered. However, austenite transformed to ferrite + pearlite/cementite. The width of this region was marginally higher for Joint 2 in comparison to Joint 1 (Table IV).

Complex alloy carbides consisting of Cr, Nb, and Fe were formed close to FB-I (Figures 6(b) and (d)). The qualitative number density of complex alloy carbides within IN152 was greater with respect to IN182. The size range of precipitate in the case of Joint 1 (~100 to 250 nm) was larger than Joint 2 (~50 to 100 nm). There are a number of contributing factors for this difference. The size of these carbides was dependent on available nucleation sites. Except boundaries, dislocations could also serve as the favorable locations for carbide...

Fig. 7—Microstructure of solidified material near FB-I and weld metal near FB-II for different assemblies buttered with (a) and (c) IN182 and (b) and (d) IN152.

Fig. 8—Characteristic X-ray spectrum adjacent to FB-II for a welded joint, buttering with IN182.
nucleation. The dislocation density close to FB-I was qualitatively higher in the case of IN152 with respect to IN182 (Figures 6(e) and (f)). Therefore, a limited number of nucleation sites for IN182 encouraged the growth of complex carbide precipitates. A qualitatively higher number of fine-scale carbides within IN152 with respect to IN182 might be the contribution of the higher amount of Cr and Nb content of IN152 than IN182.\[25\] The concentration difference of carbon between low alloy steel and IN152 was greater than the same between low alloy steel and IN182. Therefore, carbon migration from the LAS to the buttering material was greater for

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### Table III. Quantitative Analysis of Microstructural Characteristics

<table>
<thead>
<tr>
<th>Buttering Material</th>
<th>Width of Martensite (μm)</th>
<th>Width of Unmixed Zone (μm)</th>
<th>Distance of Type-II Boundary from FB-I (μm)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Range Average</td>
<td>Range Average</td>
<td>Range Average</td>
</tr>
<tr>
<td>IN152</td>
<td>5.5 to 25.8 16.1 ± 2</td>
<td>51.2 to 67.5 56.4 ± 3.7</td>
<td>11.8 to 19.2 15.7 ± 0.1</td>
</tr>
<tr>
<td>IN182</td>
<td>14.4 to 32.0 20.4 ± 0.6</td>
<td>36 to 63 53.4 ± 1.9</td>
<td>20.8 to 28.9 29.4 ± 1.9</td>
</tr>
</tbody>
</table>

### Table IV. Grain Size of Different Regions Across the Weld

<table>
<thead>
<tr>
<th>Buttering Material</th>
<th>Width of Fine Grain HAZ in LAS (μm)</th>
<th>Width of Coarse Grain HAZ in LAS (μm)</th>
<th>Grain Size at HAZ of 304LN SS (μm)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Range Average</td>
<td>Range Average</td>
<td>Range Average</td>
</tr>
<tr>
<td>IN152</td>
<td>1400 to 1732 1568 ± 135</td>
<td>373 to 642 539 ± 185</td>
<td>69.5 ± 6.0</td>
</tr>
<tr>
<td>IN182</td>
<td>962 to 1675 1421 ± 116</td>
<td>289 to 406 348 ± 55</td>
<td>78.7 ± 8.4</td>
</tr>
</tbody>
</table>

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**Fig. 9**—Microhardness distribution across fusion boundaries of welded joints fabricated with different buttering materials: (a) and (c) IN182 and (b) and (d) IN152.
Joint 2 than Joint 1. A higher degree of carbon migration also increased the width of the coarse grain region of Joint 2 with respect to Joint 1 (Table IV). Profuse carbon diffusion could be attributed to the bulk Cr content of the buttering material. The tendency of carbon to migrate from the low alloy steel side was enhanced with an increase in the Cr content of the buttering alloy.\[4\] Cr content was higher in IN152 with respect to IN182. This facilitated a greater extent of carbon migration from the LAS to the buttering material for Joint 2 than Joint 1. The effect of Cr on the diffusion characteristics of carbon was discussed by Rathod et al.\[5,6\]

Formation of martensite adjacent to FB-I can be justified with the help of the Schaeffler diagram. Using the diagram, the Cr$_{eq}$ and Ni$_{eq}$ ratios for the LAS and Inconel alloys can be obtained by the equations:

$$Cr_{eq} = pct \text{Cr} + \text{pct Mo} \times 1.5 + \text{pct Si} \times 0.5 + \text{pct Nb} \times 1$$

$$Ni_{eq} = \text{pct Ni} + 30 \times \text{pct C} + 0.5 \times \text{pct Mn} \times 3$$

The diagram was developed for predicting weld metal composition in austenitic stainless steels. It can also be applied for Ni-base alloys.\[9\] The Cr$_{eq}$ and Ni$_{eq}$ values of LAS were 0.7 and 7.4, respectively. The Cr$_{eq}$ and Ni$_{eq}$ values of IN182 were 15.0 and 74.8, respectively. The Cr$_{eq}$ and Ni$_{eq}$ values of IN152 were 30.2 and 56.8, respectively. These measured values were superimposed over the Schaeffler diagram and indicated the formation of martensite.\[4\] The differences in the chemical potential of the elements in the LAS and nickel-base alloys (IN182 and 152) along with the activation energy provided by PWHT encouraged elemental diffusion.\[26\]

The diffusion of elements from the buttering material to the LAS side moved the continuous cooling transformation diagram toward the right. It increased the hardenability of steel and the same was reported for SA508 Gr.3 Cl.2 alloy.\[27\] Ni has a considerable impact on $A_c_3$ and $M_e$ temperatures. The nickel content of ~25 wt pct in Fe-Ni alloy reduced $A_c_3$ and $M_e$ temperatures to ~873 K and ~373 K, respectively. Dupont and Kusko\[28\] have described the formation of martensite using the equation developed by Gooch\[29\] for predicting the $M_e$ temperature. Adjacent to fusion boundary-I, the $M_e$ temperature was raised. Normal air cooling triggered the formation of martensite. It was also demonstrated by them that the average width of the martensitic zone was nonuniform. As mentioned earlier, welding was carried out in multipasses and heterogeneity in heat input was unavoidable. The width of martensite for Joint 1 was marginally more compared to Joint 2 (Table III). Ni was higher in the case of IN182 (~ 64 wt pct) than IN152 (~ 54 wt pct). Being a strong austenite stabilizer, it increased the activity of carbon in austenite near FB-I. At the same time, strong carbide forming elements, such as Cr and Nb, reduced carbon activity adjacent to FB-I. Thus, comparing IN182 with respect to IN152, carbon activity was overall augmented and reflected in the higher average width of the martensitic zone.\[30\]

Near FB-I, two different boundaries were present. The average distances of Type-II boundary from FB-I were ~16 and ~30 μm (approximately) for Joints 1 and 2, respectively (Table III). Type-II boundary was an outcome of the change in the primary mode of solidification.\[31\] Nelson et al. suggested that the formation of Type-II boundary was associated with local phase transformation.\[32\] At the initial stage of solidification, it was $\delta$-$\gamma$ interface, so the epitaxial solidification was limited. Under this situation, heterogeneous nucleation of weld metal took place followed by the change of interface from $\delta$-$\gamma$ to $\gamma$-$\gamma$. Due to similar crystal structure, the boundary was highly mobile and required short-range diffusion to advance into several weld metal grains.

The mobility of the $\gamma$-$\gamma$ boundary was governed by heat input, temperature gradient, compositional change, and the state of stress/strain. Both joints were fabricated using the SMAW process except root passes; therefore, the difference in heat input and temperature gradient was negligible during buttering. However, adequate complex carbides within IN152 restricted the migration of Type-II boundary for Joint 2. Type-I boundary was formed due to the columnar grain growth of the parent alloy into the buttering material. The appearance of Type-I boundary was facilitated by the similar crystal structure of base and filler alloy.\[33\]

### 2. Weld metal

The microstructure of weld metal was controlled by heat input, cooling rate, distribution of alloying elements, and segregation coefficient ($K$). The segregation coefficient indicated the ratio between the concentration of a particular element in solid with respect to the concentration of the same element in liquid. For a nonequilibrium situation, such as dissimilar materials welding, $K$ becomes less than unity. The difference took place due to a change in the concentration of alloying elements, and heat input, temperature gradient, compositional changes, and state of stress/strain.

### Table V. Tensile Properties of Transition Joints and Virgin Alloys (at Room Temperature)

<table>
<thead>
<tr>
<th>Parameters</th>
<th>IN152 (MPa)</th>
<th>IN182 (MPa)</th>
<th>IN152 (Pct)</th>
<th>IN182 (Pct)</th>
</tr>
</thead>
<tbody>
<tr>
<td>YS (MPa)</td>
<td>420 ± 1</td>
<td>374 ± 3</td>
<td>~380</td>
<td>~310</td>
</tr>
<tr>
<td>UTS (MPa)</td>
<td>616 ± 17</td>
<td>534 ± 12</td>
<td>~625</td>
<td>~552</td>
</tr>
<tr>
<td>Breaking Strain (Pct)</td>
<td>18.3 ± 1.4</td>
<td>18.5 ± 0.5</td>
<td>~32.5</td>
<td>~30</td>
</tr>
<tr>
<td>Uniform Strain (Pct)</td>
<td>14.0 ± 0.5</td>
<td>15.7 ± 0.4</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>Strain-Hardening Exponent</td>
<td>0.34</td>
<td>0.22</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>Strength Coefficient ($K$)</td>
<td>1275</td>
<td>921</td>
<td>—</td>
<td>—</td>
</tr>
</tbody>
</table>

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elements. Elements, such as Fe, Ni, and Cr, have appreciable solubility in each other, and they do not segregate. This phenomenon was supported by the corresponding $k$ value, which was close to unity.\[34\] Elements, such as C and Nb, segregated during the solidification; however, carbon was an interstitial element and could be reverted. Niobium was a substitutional alloying element with high segregation tendency. The $k$ value of Nb was changed with Fe content. The segregation coefficient value for Nb ($k_{\text{Nb}}$) was decreased from 0.54 to 0.25 with an increment in Fe content from 2 to 47 wt pct.\[34\] Diffusion of Fe from the LAS to the Inconel side reduced the $k$ value and enhanced the tendency of segregation for Nb. Presence of Nb produced a strong constitutional supercooled region ahead of the solid–liquid interface.\[35\] This changed the microstructure from cellular to dendritic morphology for both joints away from FB-I. The constitutional supercooling was increased with the increase in the amount of Nb, which was dominant for IN152.

Both weld metals showed an AF/FA mode of solidification near the interface (Figures 7(a) and (b)). The compositional differences (Cr/Ni) at the fusion boundary and variations in the cooling rate affected the mode of solidification.\[36\] As mentioned earlier, due to fabrication through multipasses, compositional variation and differences in thermal cycles were unavoidable. In the AF mode, primary austenite was solidified with second-phase ferrite at the grain boundary. In the FA mode, primary ferrite solidified with austenite at the cell or dendritic walls.\[37\] The AF/FA mode of solidification was reported for joining LAS with Inconel.\[5, 8\] Away from the fusion boundary, bulk weld metal solidified in the AF mode.

Within weld metal, solidification grain boundaries (SGBs) and migrated grain boundaries (MGBs) were present. Both SGBs and MGBs were high-angle boundaries. SGBs migrated under the influence of external thermal energy.\[4\] Grain boundary migration was not extensive in IN152 as compared to IN182. A similar observation was described by Mouninot and Hanninen\[19\] during their work. They summarized that IN82 weld metal contained the denser dendritic structure and enriched Nb-rich precipitates owing to its higher niobium content than IN152.

3. Region around FB-II

Adjacent to fusion boundary-II, both weld metals (IN182 and 152) and 304LN SS were of the same crystal structure (fcc). 304LN SS acted as a substrate for nucleation during solidification. Though the compositional difference existed, the nucleation barrier was almost zero because of the complete wetting.\[38\] This resulted in epitaxial growth into the weld metal close to FB-II. Competitive growth, which was the outcome of the clash between the easy growth direction ($h100i$ for fcc) and the direction of maximum heat extraction, stifled the growth of some grains. This phenomenon triggered the formation of near-equiaxed grains in clusters at a few locations (Figures 7(c) and (d)).

An unmixed zone was present for both joints with negligible variation in width (~56 and ~53 μm for
Joints 1 and 2, respectively). The unmixed zone was formed owing to differences in melting point and chemical compositions between the weld metal and base alloy. During welding, a coexistence of molten pool and unmelted material occurred. The convection of heat became limited at this interface and the molten base alloy solidified back without significant mixing with the weld metal. An unmixed zone of nearly the same morphology was reported by Baeslack et al. for the interface between 304 stainless steel and 310/312 filler metals.

There was an HAZ within 304LN SS near FB-II. The HAZ of 304LN illustrated enhancement in grain size with respect to parent alloy (~ 79, ~ 69, and ~ 56 μm for IN182, IN152, and base alloy, respectively) (Table IV).

Grain coarsening in the HAZ of 304LN SS was due to the heat input during welding.

B. Microhardness

Microhardness distribution exhibited a peak (~ 315 and 350 HV for Joints 1 and 2, respectively) contiguous to FB-I (Figure 9). The peak could be ascribed due to the presence of martensite and complex alloy carbides. PWHT caused tempering of initially formed martensite and at the same time created fresh martensite further in the adjacent region owing to the diffusion of alloying elements across the interface. Within the LAS adjacent to FB-I, a valley was found (~ 230 HV). This decrement was due to the formation of the HAZ (Regions I and II)
in low alloy steel. Another decrement in microhardness was observed within the buttering material (~ 210 HV) near FB-I. This minimal value within the profile indicated the location of the weakest region across FB-I. This area was away from the martensitic region. Within the weld metal, the profile exhibited small variation due to the inherent heterogeneity of the solidification structure. Close to FB-II, the microhardness profile revealed an upward trend once again. The rise was because of the presence of unmixed zone. The strain might be one of the factors contributing to the increment in hardness near FB-II, as reported by Ming et al. A substantial amount of circumferential tensile residual stress was developed at the interface between IN82 weld metal and stainless steel for the assembly consisting of 2.25Cr-1Mo and 316 SS. The hardness profile displayed a downward trend within 304LN SS near FB-II owing to grain coarsening in the HAZ.

C. In Situ Uniaxial Deformation Test

Up to the yield point, there was no microstructural change over gage length. After the yield point, deformation bands appeared. These bands were the region of intense local shearing. Their number density was increased with the increment in strain (Figures 10(b) and 11(b)). Deformation and subsequent failure were confined within the buttering material. The crack was initiated from the open edge of the sample within the buttering material, which was away from fusion boundary-I (300 to 600 μm). During layer deposition of the buttering alloy, structural heterogeneity occurred. As an example, layer 1 contained Type-I boundary, Type-II boundaries, and martensite. For layer 2, these features were absent. A microstructural difference between the adjacent two layers became a potential location for crack initiation.

Kumar et al. showed that variation in tensile properties was related to the microstructural differences, inclusions, and heat input. In a different endeavor, the tensile property of alloy 690 weldment with I-82 and I-52 fillers was evaluated. It was revealed that the strength of I-52 weldment was low and the joint failed from the fusion zone. On the contrary, high strength was observed for the I-82 weldment and failure occurred from the base metal. It was concluded that I-82 weld metal had fine subgrains and small cellular spacing, which contributed to the increment in strength as well as elongation. In the present investigation, yield strength and tensile strength were higher for the joint with IN152 buttering material than IN182 buttering alloy (12 and 15 pct, respectively (Table V)). It was illustrated previously that near FB-I, the buttering alloy of Joint 2 contained a relatively fine distribution of complex alloy carbide in relation to Joint 1. The grain size of IN152 buttering material adjacent to FB-I was also small (IN152 ~ 45.9 μm and IN182 ~ 63.2 μm). Considering the fracture initiation and its movement, the average distance of the crack path from FB-I was ~ 400 to 500 and 600 to 800 μm for the joints with IN182 and IN152 buttering material, respectively. The average hardness of these regions was ~ 250 and ~ 290 VHN, respectively.

The relative high hardness of IN152 over IN182, therefore, was reflected in higher tensile properties of Joint 2 than Joint 1.

The strain-hardening exponent was higher for Joint 2 than Joint 1 (Table V). This phenomenon could be endorsed through the difference in the microstructure of the two buttering materials near FB-I. A large quantity of complex alloy carbide and small grain structure of IN152 were responsible for the greater strain-hardening exponent than that of IN182. Moreover, qualitatively, it was found that dislocation density was also higher in the case of Joint 2 than Joint 1 near fusion boundary-I (Figures 6(e) and (f)). Thus, the particle-dislocation, dislocation-dislocation, and dislocation-boundary interactions became more severe within the buttering material near FB-I of Joint 2 than Joint 1. The ultimate outcome became improvement in strength with the high strain-hardening exponent of Joint 2 in comparison to Joint 1.

V. CONCLUSIONS

Welding between SA508 Gr.3 Cl.1 low alloy steel and 304LN austenitic stainless steel was performed using IN152 and IN182 as buttering material and weld metal. The microstructural characterization was done, followed by in situ tensile testing. The findings of the investigation are illustrated subsequently.

1. Adjacent to the boundary between the low alloy steel and buttering material (fusion boundary-I), the low alloy steel contained a coarse grain denuded zone. This region was subsequently followed by a fine grain region away from FB-I. Both regions were wider for the joint fabricated using IN152 than that using IN182.

2. Near fusion boundary-I, Type-I boundary (perpendicular to fusion boundary-I), discrete Type-II boundary (parallel to fusion boundary-I), and discontinuous martensitic islands were formed. The average width of the martensitic cluster was greater for the joint fabricated using IN152 (~ 20.4 μm) than that using IN182 (~ 16.1 μm). The distance of the Type-II boundary from fusion boundary-I was higher in the case of the former with respect to the latter (~ 29.4 μm in IN152 and ~ 15.7 μm in IN182).

3. Complex alloy carbides were formed near fusion boundary-I. Their average size was less within IN152 than within IN182. Qualitatively, their number density was greater within IN152 with respect to IN182. The dislocation density within IN152 was greater than IN182 near fusion boundary-I.

4. Maximum hardness near fusion boundary-I was a result of martensite formation. A HAZ existed immediately before and after the peak, and the microhardness was dropped.

5. Local heterogeneity within the solidified buttering material acted as a nucleation site for crack initiation during in situ tensile testing. This crack was
ACKNOWLEDGMENTS

The authors are grateful to the director, CSIR-NML, for providing permission to use the infrastructural facilities to carry out the investigation. One of the authors thanks the professors at the National Institute of Technology, Tiruchirappalli, for their valuable suggestions.

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