A Functionally Graded Joint between P91 Steel and AISI 316L SS

BY A. KULKARNI, D. K. DWIVEDI, AND M. VASUDEVAN

ABSTRACT
A novel way of developing a functionally graded (FG) joint between a dissimilar metal combination of P91 steel and AISI 316L stainless steel by fusion welding has been presented. Four interlayers of both base metals of different widths were stacked between the two base metals, and three overlapping passes of autogenous activated gas tungsten arc welding (A-GTAW) were performed. Due to the self-grading effect and dilution from the interlayers and base metals, a step-wise compositional and microstructural gradient was obtained in the FG weld (FGW) joint. The performance of the FGW joint was compared with a conventional A-GTAW (CA-GTAW) joint fabricated without interlayers through the study of metallurgical and mechanical properties. The beneficial effect of the FGW joint in retarding carbon diffusion has been discussed based on metallography, microhardness testing, and thermodynamic analysis. The finite element analysis performed to study thermal stress distribution in the weld joints indicated lower thermal stress levels in the FGW joint compared to the CA-GTAW joint. The thermal cycling studies showed a better thermomechanical fatigue life from the FGW joint than the CA-GTAW joint.

KEYWORDS
• Dissimilar Metal Weld • Functionally Graded Joint
• Activated Gas Tungsten Arc Welding Joint
• Thermal Cycling
• Coefficient of Thermal Expansion

Introduction
Dissimilar metal weld (DMW) joints between ferritic and austenitic stainless steels are regularly employed in thermal and nuclear power plants (Ref. 1). Differences in the coefficient of thermal expansion (CTE) and chemical composition of base metals and weld zones give rise to problems, such as thermal stresses and carbon migration, which in turn result in the failures of DMW joints from the ferritic steel side (Ref. 2). Many attempts have been made to improve the life of ferritic-austenitic DMWs, such as the use of nickel-based fillers/buttering to reduce carbon migration and the use of a transition piece of Alloy 800 or Inconel® 600 to reduce the CTE mismatch (Refs. 3, 4). Recently, functionally graded (FG) joints have been developed to facilitate a smooth transition in the chemical composition and mechanical properties between ferritic steels and austenitic stainless steels (SSs) (Refs. 5–10). Farren et al. (Ref. 5) fabricated a FG material (FGM) between low-alloy steel and Alloy 800 developed by a directed-energy deposition method effectively controls carbon migration in comparison with a conventional weld joint. However, these additive manufacturing approaches require base materials to be in the powder form. Additionally, a large number of process parameters is required to be optimized to avoid defects, such as porosity, lack of adherence, and undissolved powder (Refs. 11, 12). Fusion welding has also been successfully employed to develop FGMs between ferritic and austenitic steels, but only on a few occasions (Refs. 8, 9). Brentrup et al. (Ref. 8) developed FG transition joints between T22 ferritic steel and austenitic SS by gas tungsten arc welding (GTAW) with double-wire feeding. A significant reduction in both carbon migration and peak thermal stress was achieved with the FG joint (Ref. 9). However, there is no literature that presents a FG joint between dissimilar metals developed by an activated GTAW (A-GTAW) process. A-GTAW has been found to be beneficial in reducing carbon diffusion in dissimilar steel joints (Ref. 13). A-GTAW uses a layer of activated flux coating prior to welding to improve the depth of penetration (Ref. 14). Being an autogenous process in nature, the DMW made by A-GTAW exhibits a composition intermediary to that of two base metals, and a step-wise compositional gradient is naturally achieved from one base metal to the other (Ref. 15). This phenomenon, referred to as self-grading, suggests that, due to the dilution effects from both base metals, a gradual compositional variation can be obtained even if two alloys are simply welded without filler or deposited over each other (Refs. 12, 16). Moreover, Kulkarni et al. (Ref. 17) showed the composi-

https://doi.org/10.29391/2021.100.024
tion and microstructure of the fusion zone in the A-GTAW joint could be modified in a favorable manner with the use of Incoloy® 800/Inconel 600 interlayers. Therefore, dilu-
tions from interlayers and the base metal can be utilized fa-
vorably to achieve functional grading.

In this work, a FG weld (FGW) joint between P91 steel and 316L SS was developed by three overlapping autoge-
nous A-GTAW passes using multiple interlayers of both base
metals of different widths. The metallurgical and mechani-
cal properties of the FGW joint were compared with a
conventional A-GTAW (CA-GTAW) joint. The results of the
thermal cycling tests on the CA-GTAW and FGW joints were
discussed with CTE measurements and finite element (FE)
simulations. The effectiveness of the FGW joint in retarding
carbon migration was also studied.

Experimental Procedure

The CA-GTAW joint without the addition of any interlay-
ers was fabricated between 8-mm-thick plates of dissimilar
steel coupled with P91 steel (8.3%Cr-0.9%Mo-0.3%Ni-
0.4%Mn-0.2%Si-Fe) and AISI 316L SS (16.3%Cr-2.1%Mo-
10%Ni-1.3%Mn-0.2%Si-Fe) — Fig. 1A. To fabricate the
FGW joint, interlayers of 5 and 1 mm thicknesses were
sandwiched between two base metals (Fig. 1B). Three over-
lapping autogenous weld passes were made by following the
welding sequence given in Fig. 1C. For the finalization of the
interlayer sizes, combinations of small and large inserts of
narrow (0.5 and 3 mm), medium (1 and 5 mm), and broad
sizes (1.5 and 7 mm) were used. The interlayers of medium-
sized widths (1 and 5 mm) were finalized for characteriza-
tion due to the complete mixing of interlayers and base met-
als, decent overlapping of the weld zones, and no require-
ment of torch offsetting for the FGW joint development.

Different flux mixtures used for different welding passes
for the CA-GTAW and FGW joints are provided in Table 1.
The fluxes were finalized after multiple iterations and earli-
er research experience, which implied that multi-component
flux mixtures rich in SiO2 and TiO2 work well for stainless
steel welds, and MoO3-rich flux mixtures give better results
for A-GTAW joints of ferritic steels (Refs. 13–15). After each
welding pass in the FGW joint, the top surface of the plates
was polished by a 100-grade emery paper and cleaned for
the application of the activated flux for the next pass. Dur-
during welding, a current of 230 A, a speed of 80 mm/min, an
arc voltage of 15 and 16 V , and an argon gas flow rate of 10
L/min were used for all weld passes in the CA-GTAW and
FGW joints, which resulted in a net heat input of 1.87
kJ/mm (at 70% efficiency).

The extracted specimens from the weldments were pol-
ished with emery paper, followed by cloth and etching with
Vilella’s reagent and aqua regia. The weldments were charac-
terized using an optical microscopy (OM), a scanning elec-
tron microscopy (SEM) coupled with an energy dispersive

![Fig. 1 — Schematics of: A — CA-GTAW joint preparation; B — FGW joint preparation; C — welding sequence in the FGW joint; D —
specimen for tensile test; E — fixture for thermal cycling test; F — thermal cycle for FE simulation.]

<table>
<thead>
<tr>
<th>Welding Pass</th>
<th>Combination</th>
<th>Flux Composition (wt-%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CA-GTAW joint</td>
<td>P91–316L</td>
<td>30% SiO2 + 35% TiO2 + 25% MoO3 + 10% CuO</td>
</tr>
<tr>
<td>FGW-1</td>
<td>316L–P91–316L</td>
<td>60% MoO3 + 20% TiO2 + 20% Cr2O3</td>
</tr>
<tr>
<td>FGW-2</td>
<td>FGW-1–P91</td>
<td></td>
</tr>
<tr>
<td>FGW-3</td>
<td>FGW-2–316L–P91</td>
<td></td>
</tr>
</tbody>
</table>

New-kulkarni-supp-202061.qxp_Layout 1  7/8/21  5:46 PM  Page 270
spectroscopy (EDS), an electron probe microanalyzer (EPMA), an optical emission spectroscopy (OES), and a transmission electron microscopy (TEM). CTE measurements were done on a Linseis DIL L75 dilatometer. The CTE measurements were done from 25˚ to 700˚C with constant heating and cooling rates of 5˚C/s and a hold time of 15 min at 700˚C. Microhardness measurements were carried out by applying a 300-g load and 10-s dwell time. Tensile testing was carried out in an as-welded condition on the universal testing machine. Subsize specimens of 25 mm gauge length and 5 mm thickness were prepared per the ASTM A370, Standard Test Methods and Definitions for Mechanical Testing of Steel Products, standard (Fig. 1D). For assessment of carbon migration, both weld joints were postweld heat treated at 750˚C for 25 h and subsequently air cooled, followed by microhardness and microstructural studies. Thermodynamic simulations of carbon migration were performed on the Thermo-Calc Software diffusion module (DICTRA module) using TCFE and MOBFE databases.

Klüe et al. (Ref. 18) proposed a simple test for the evaluation of thermomechanical fatigue behavior of dissimilar metal joints. Prior to the thermal cycling test, specimens of 90 × 20 × 3 mm³ were extracted by electro-discharge machining (EDM). The samples were subjected to a postweld heat treatment (PWHT) of 750˚C for 2 h. These machined specimens were subjected to three-point bend loading (Fig. 1E). The prestress (σb) was determined by Equation 1, where t is the specimen thickness, y is the displacement, and L is the distance between outer supports.

\[
\sigma_b = \frac{6Ety}{L^2}
\]  

(1)

The prestress levels of 200 and 400 MPa were applied. Thermal cycling was carried out between 25˚ and 625˚C. All four specimens were loaded simultaneously in the preheated induction furnace kept at 625˚C and held for 1 h, and subsequently quenched in water. After every ten cycles, specimens were examined to detect cracking. Cracks that developed at the interfaces were later studied with an optical microscope. A FE simulation was carried out with the ABAQUS FEA software. A 2D axisymmetric model was developed, which resembled a transition DMW joint between pipes with an outer diameter of 355.5- and 8-mm wall thickness (Ref. 2). The materials were assumed to exhibit elastic behavior during thermal cycling. Additionally, the weld joints were assumed to be stress free at 0˚C. The residual stresses after welding were ignored. The model was meshed with quadrilateral axisymmetric stress elements (CAX4R). The thermal cycle used for the FE simulation is shown in Fig. 1F.

Results and Discussion

Weld Bead Geometry

The macrographs showing weld bead geometries of CA-GTAW and FGW joints are shown in Fig. 2. Through-thickness depth of penetration was achieved by the CA-GTAW joint as well as in FGW joint passes. The through-thickness penetration in both weld joints was due to the reversal of the Marangoni convection, which was inflicted by the addition of oxygen into the weld pool through decomposition of oxide fluxes during welding (Ref. 13). In the CA-GTAW joint, the width at the top side was 10.3 mm and the depth-to-width ratio was 0.77 (Fig. 2A). On the other hand, the three overlapping welding passes in the FGW joint were evident in the macrograph (Fig. 2B).

| Table 2 — Elemental Compositions of Different Weld Zones as Obtained from OES |
|---------------------------------|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|
| Element (wt-%)                  | Cr  | Mo  | Si  | Nb  | Ni  | Mn  | C   | Fe  | Cr_eq | Ni_eq |
| P91 steel                       | 8.33| 0.94| 0.20| 0.10| 0.33| 0.37| 0.10| Bal.| 9.6   | 3.5   |
| AISI 316L                       | 16.33| 2.17| 0.21| 0.25| 10.03| 1.32| 0.03| Bal.| 18.9  | 11.6  |
| CA-GTAW joint weld zone         | 11.97| 1.42| 0.20| 0.12| 4.13 | 0.90| 0.06| Bol.| 13.7  | 6.6   |
| FGW-1                           | 15.52| 2.01| 0.23| 0.18| 8.53 | 1.18| 0.04| Bol.| 17.9  | 10.8  |
| FGW-2                           | 12.17| 1.52| 0.18| 0.15| 4.63 | 0.83| 0.07| Bol.| 14.1  | 7.1   |
| FGW-3                           | 10.26| 1.24| 0.22| 0.11| 2.27 | 0.55| 0.08| Bol.| 11.9  | 4.9   |

Fig. 2 — Macrographs showing weld bead geometries: A — CA-GTAW joint; B — FGW joint.
FGW joint widths at the top and bottom sides were found to be 22.5 and 19.6 mm, respectively. No defects, such as porosities or solidification cracks, were revealed in the macrograph of both weld joints.

**Chemical Composition Analysis**

The concentrations of alloying elements obtained from the OES are summarized in Table 2. From the measured chemical compositions, chromium equivalent (Creq = %Cr + %Mo + 0.5 × %Nb + 1.5 × %Si) and nickel equivalent (Nieq = %Ni + 30 × %C + 0.5 × %Mn) were calculated. The fusion zone of the CA-GTAW joint exhibited a chemical composition intermediate to that of both base metals due to self-grading (Creq = 13.7, Nieq = 6.6). Figure 3 shows the EPMA measurements of major alloying element (Fe, Cr, Ni, Mo, and Mn) concentrations along the transverse section of the CA-GTAW and FGW joints. The CA-GTAW joint showed a steep elemental concentration gradient across both base metal and weld zone interfaces (Fig. 3A).

On the other hand, the FGW joint showed a step-wise variation in chemical composition between two base metals with a gradual change in the chemistry across all of the interfaces (Fig. 3B). The first welding pass (FGW-1) was made between two 316L SS substrates and a P91 steel interlayer. Therefore, in FGW-1, the dilution (approximately 12%) from the P91 steel interlayer reduced the Creq and Nieq as compared to the 316L SS base metal. Afterward, a second welding pass (FGW-2) was made between the FGW-1 and 5-mm-wide P91 steel strip, which showed Creq = 14.1 and Nieq = 7.1. The Creq and Nieq were further lowered in the third weld (FGW-3) due to the dilution from the P91 steel base metal, 1-mm-wide 316L interlayer, and overlapped FGW-2. In the FGW joint, the self-grading effect from base metals, interlayers, and overlapped weld zones was utilized favorably to mitigate sudden changes in the elemental composition across the interfaces.

**Microstructure Evolution**

The P91 steel base metal showed a tempered martensitic structure with fine distribution of carbides, while the 316L SS showed an equiaxed austenitic structure. A detailed discussion on the base metal microstructure is available (Ref. 17). Lath martensite encroded within prior austenitic grain boundaries (PAGBs) was present in the CA-GTAW joint weld zone — Fig. 4A. In the FGW joint, the first weld zone (FGW-1) showed an austenitic-ferritic microstructure (Fig. 4B). A bright-colored ferrite phase was observed in the dark-colored austenitic matrix. The second weld zone (FGW-2) revealed a martensitic-austenitic structure (Fig. 4C). The austenite phase was present at the dendritic grain boundaries, while martensite was observed within the dendritic cores. The third weld (FGW-3) exhibited a predominant lath martensitic structure encroded within PAGBs (Fig. 4D).

The heterogeneity in microstructures in the FGW joint can be related to the martensitic start (Ms) and martensitic finish (Mf) temperatures. The Ms temperatures for different zones were calculated for all of the zones (Table 3) (Ref. 19). In the FGW-1, the Ms temperature was below room temperature. Therefore, it was expected that there was no driving force in the form of a Gibbs-free energy difference for the nucleation of martensite. However, delta ferrite was retained at room

<table>
<thead>
<tr>
<th>Zone</th>
<th>P91 Steel</th>
<th>AISI 316L</th>
<th>CA-GTAW Joint Weld Zone</th>
<th>FGW-1</th>
<th>FGW-2</th>
<th>FGW-3</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ms(°C)</td>
<td>540 – 497C – 6.3Mn – 46.6Mo – 10.8Cr – 36.3Ni (Ref. 25)</td>
<td>342</td>
<td>-124</td>
<td>159</td>
<td>-72</td>
<td>130</td>
</tr>
</tbody>
</table>
temperature due to the presence of the high amount of ferrite-promoting elements, such as Cr and Mo, in the weld zone. This increases the stability of delta ferrite, which resists transformation to austenite during cooling (Ref. 20). As a result, a dual-phase, austenitic-ferritic microstructure was observed in the FGW-1. In the FGW-2 and FGW-3, Ms temperatures were well above the room temperature. Therefore, these zones were expected to exhibit a predominantly martensitic structure. However, some amount of austenite was retained due to the incomplete transformation of austenite to martensite, i.e., Mf temperature below the room temperature (Ref. 6).

**Microhardness Variation**

Figure 5 shows a Vickers microhardness variation for the CA-GTAW and FGW joints in as-welded and postweld heat-treated conditions. The base metals exhibited a hardness of 268 ± 3 HV (P91 steel) and 209 ± 4 HV (AISI 316L), respectively. The CA-GTAW joint weld zone showed an average hardness value of 441 ± 11 HV, which was reduced to 375 ± 6 HV after PWHT (Fig. 5A). In the FGW joint the mean hardness values in the as-welded condition were 212 ± 11 HV (FGW-1), 388 ± 9 HV (FGW-2), and 432 ± 15 HV (FGW-3). After PWHT, hardness values were found to be 219 ± 7 HV for FGW-1, 342 ± 4 HV for FGW-2, and 327 ± 7 HV for FGW-3 (Fig. 5B). The P91 steel heat-affected zone (HAZ) in the as-welded condition showed an increase in hardness compared to the base metal due to the dissolution of carbides during heating and its subsequent transformation to untempered high-carbon martensite on cooling (Ref. 21). The reduction in hardness levels after PWHT was due to the tempering of the martensitic structure in the weld zones and HAZ (Ref. 22).

A steep hardness gradient of 232 HV (as-welded) and 164 HV (PWHT) was observed at the AISI 316L-weld zone interface of the CA-GTAW joint due to the occurrence of a predominant martensitic structure in the weld zone and a soft austenitic structure in the AISI 316L base metal. On the other hand, in the FGW joint, the hardness gradient was detected at the FGW-1 and FGW-2 interfaces due to a soft austenitic-ferritic structure in the FGW-1 and a hard martensitic-austenitic structure in the FGW-2. The hardness gradients at the FGW-1 and FGW-2 interfaces were 176HV (as-welded) and 123 HV (PWHT), which were lower than the gradients observed in the CA-GTAW joint.

**Tensile Testing**

Tensile testing results showed an overmatching strength
in both the CA-GTAW and FGW joints in the as-welded condition. However, the ductility in both weld joints was compromised (Table 4). The CA-GTAW joint specimen failed from the AISI 316L base metal near the weld zone interface — Fig. 6A. On the other hand, the FGW joint specimen failed from the FGW-1 and FGW-2 interfaces.

The failures from the interfaces were attributed to the strain concentration during tensile loading caused by the steep hardness gradient across the interfaces. This resulted in the deformation of only the softer side of the interface (316L SS side), while the harder part (weld zone and P91 steel) restrained the plastic flow during tensile testing of the CA-GTAW joint. Similarly, in the FGW joint, the strength mismatch between the weaker austenitic-ferritic region (FGW-1) and the stronger martensitic-austenitic region (FGW-2) resulted in interfacial failures. The fractography of tensile-tested specimens showed cleavage facets and microvoids, which suggested a quasi-cleavage fracture mode in both the CA-GTAW (Fig. 6B) and FGW joints (Fig. 6C).

**Carbon Migration Analysis**

A qualitative assessment of the severity of carbon migration was done after subjecting the weld joints for PWHT at 750°C for 25 h through hardness measurements and OM. The ferritic steel (P91 steel)–weld zone interface is more

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Yield Strength (MPa)</th>
<th>Ultimate Tensile Strength (MPa)</th>
<th>Elongation (%)</th>
<th>Failure Location</th>
</tr>
</thead>
<tbody>
<tr>
<td>P91 steel</td>
<td>725</td>
<td>829</td>
<td>16.74</td>
<td>–</td>
</tr>
<tr>
<td>AISI 316L</td>
<td>325</td>
<td>610</td>
<td>77.43</td>
<td>–</td>
</tr>
<tr>
<td>CA-GTAW joint</td>
<td>350</td>
<td>665</td>
<td>29.28</td>
<td>AISI 316L–weld zone interface</td>
</tr>
<tr>
<td>FGW joint</td>
<td>360</td>
<td>673</td>
<td>23.97</td>
<td>FGW-1–FGW-2 interface</td>
</tr>
</tbody>
</table>

**Fig. 5** — Microhardness variation along with the observed microstructure in the respective zones (M-martensite, A-austenite, and F-ferrite): A — CA-GTAW joint; B — FGW joint.

**Fig. 6** — Results of tensile tests: A — Stress-strain curves for weld joints and base metals as well as photographs of failed weld joints; B — fractograph of a tensile-tested specimen for the CA-GTAW joint; C — fractograph of a tensile-tested specimen for the FGW joint.
susceptible to carbon migration (Ref. 4). The micrographs of the interfaces that are prone to carbon migration in CA-GTAW and FGW joints are shown in Figs. 7 and 8. In the CA-GTAW joint, a clear, dark-etched band was evident between the P91 steel HAZ and the weld zone (Fig. 7A). The TEM analysis of the carbon-enriched zone (CEZ) showed a presence of M23C6-type carbides the size of 254 ± 94 nm (Fig. 7C, D). The average CEZ width was 39.5 ± 9 μm. The CEZ was formed at the expense of the diffused carbon from the P91 steel side. Consequently, the P91 steel side got depleted in carbon and formed a carbon-depleted zone (CDZ), and it was also indirectly indicated by the hardness depletion (Fig. 7B). The average hardness of the CDZ and CEZ was found to be 205 ± 2 and 335 ± 9 HV, respectively, whereas the P91 steel base metal and CA-GTAW joint weld zone exhibited a hardness of 229 ± 4 and 313 ± 2 HV, respectively, after the heat treatment (750˚C/25 h). The formation of densely populated carbides increased the hardness of the CEZ, while depletion in carbon from the martensitic matrix reduced the CDZ hardness (Ref. 23).

On the other hand, there was no visible dark-etched band across any of the FGW joint interfaces, which suggests no formation in the CEZ (Fig. 8A–C). Additionally, there was no drastic increase or reduction in hardness values across the interfaces (Fig. 8D–F). On the P91 steel side near the interface, the average hardness values before and after PWHT (750˚C/25 h) were 241 ± 4 and 239 ± 4 HV, respectively. On the other hand, the FGW-3 zone near the interface exhibited a hardness of 311 ± 2 and 318 ± 6 HV before and after PWHT (750˚C/25 h), respectively. Similarly, no significant hardness gradients due to the diffusion of carbon were observed at other interfaces of the FGW joint. These observations suggested that the FGW joint was beneficial in alleviating carbon diffusion across the interfaces in long-term service. It is a significant improvement in the performance of the FGW joint over the CA-GTAW joint.

The reduction in the tendency of carbon diffusion can be explained with the help of carbon potential and diffusivity. Carbon potential, or carbon activity, depends on the content of carbon- and carbide-forming elements, e.g., Cr, Mo, and Si. The carbide-forming elements reduce carbon potential by forming stable carbides (Ref. 24). Therefore, the P91 steel...
base metal, containing a lower amount of carbide-forming elements than all of the weld zones, exhibited the highest carbon potential (Ref. 25). A smooth variation in concentration of carbide-forming elements reduced carbon-diffusion tendency in the FGW joint. The driving force for carbon migration, i.e., carbon potential gradient for various interfaces, was calculated using Thermo-Calc Software (Table 5).

The potential carbon gradient was lower across the P91 steel and FGW-3 interface in the FGW joint (7588 J/mol) compared to the P91 steel–weld zone interface in the CA-GTAW joint (13,931 J/mol). Although carbon potential gradient values increased at the other interfaces of the FGW joint, the diffusion coefficients were lower due to the presence of austenite. An increase in the austenitic content reduces the carbon diffusivity due to the close-packed face-centered cubic (FCC) structure of austenite, which impedes the diffusion of interstitial carbon atoms (Ref. 17).

CTE Measurements

The CTE for base metals and different weld zones increases with an increase in temperature — Fig. 9. In the temperature range of 25˚–700˚C, the average CTE of P91 steel and AISI 316L were found to be 11.8 and 17.9 µm/m/˚C, respectively. Yin and Faulkner (Ref. 26) attributed the CTE differences to the crystal structures of ferritic/martensitic P91 steel (body-centered cubic [BCC]) and austenitic 316L SS (FCC) and the bond energy between their atoms. The mean CTE of the CA-GTAW joint weld zone was 14.3 µm/m/˚C (Fig. 9A). On the other hand, the FGW-1, FGW-2, and FGW-3 of the FGW joint showed an average CTE of 17.3, 15.4, and 13.3 µm/m/˚C, respectively, which resulted in a gradual change in CTE values from one side to the other (Fig. 9B). Woo et al. (Ref. 10) showed that reduction in the dilution of stainless steel reduces the CTE in ferritic-austenitic FGM, which is in agreement with the observations made in the current study.

The CTE variation can be related to the modifications in the microstructure and chemical composition in different weld zones. In the FGW-1, the retention of the delta ferrite in the austenitic matrix resulted in the overall CTE reduction compared to the 316L SS base metal. Elmer et al. (Ref. 27) reported that the CTE value of the austenitic SS depends upon its elemental composition, type of phases, and phase fraction. However, the type of phases has the highest effect on CTE values. The CTE of a composite system can be predicted using the Thomas theory, as given by Equation 2 (Ref. 28).

\[
\ln \alpha_w = f_A \ln \alpha_A + f_F \ln \alpha_F
\]  

(2)

In Equation 2, \(\alpha_w\) is the CTE for FGW-1, \(\alpha_A\) is the CTE for fully austenitic material (18.4 µm/m/˚C), \(\alpha_F\) is the CTE for retained delta ferrite (15 µm/m/˚C), and \(f_A\) (85%) and \(f_F\) (15%) are the volume fractions of austenite and ferrite, respectively.
(15%) are phase fractions of austenite and delta ferrite in the FGW-1, respectively. Using this equation, the predicted value (17.8 μm/m/˚C) of the CTE for the FGW-1 was in line with the observed CTE (17.3 μm/m/˚C). Similarly, in the FGW-2, the presence of retained austenite (f_A ~ 20%) in the martensitic matrix structure resulted in the CTE value intermediary of base metals. In the FGW-3 and CA-GTAW joint weld zone, there was an increase in the concentration of Cr, Ni, and Mo as compared to P91 steel. According to Wu et al. (Ref. 29), the increase in concentrations of Cr, Ni, and Mo increases the volume fraction of retained austenite in the martensitic matrix through the reduction of Ms temperature. The theoretically estimated contents of retained austenite (γ(%) = 36.412 – 0.126 × Ms(˚C)) in the FGW-3 and CA-GTAW joint weld zone were 6 and 16%, respectively. The increase in retained austenite content increased the CTE of these zones. Additionally, a reduction in C content in the FGW-3 and CA-GTAW joint weld zone compared to the P91 steel could have increased the CTE, as observed by Yin and Faulkner (Ref. 26). Moreover, Hidnert (Ref. 30) reported that an increase in Ni content between 0 and 10% in Fe-Cr-Ni steels increases the CTE. Therefore, it can be deduced that modifications in phases and chemistry altered the CTE of weld zones in CA-GTAW and FGW joints.

**Thermal Cycling Tests**

In the present study, the presence of cracks was revealed in the CA-GTAW joint after 290 cycles at 25˚–625˚C/400 MPa (Table 6). On the other hand, the FGW joint under similar conditions revealed cracks after 360 cycles at 400 MPa. For thermal cycling at the prestress level of 200 MPa, both the CA-GTAW and FGW joints showed no failure even after 400 thermal cycles.

In earlier studies performed by other researchers, the interfacial cracks were observed in ferritic (P22 steel)-austenitic SS (316L) DMW made by a Type 309 SS filler after 64 cycles at 138 MPa/593˚C and 2300 h of high-temperature exposure (Ref. 18). However, interfacial failures from the ferritic P22 steel side were mitigated by the introduction of an Alloy 800 transition piece (Ref. 31). In these studies, crack initiation was attributed to the metallurgical degradation of ferritic steel and interfacial thermal stresses. In another work, no crack was observed at the P91 steel-Inconel 82 filler-Alloy 800 joint at 270 MPa prestress level after 125 cycles with an exposure of 2750 h at 625°C (Ref. 32). In the current study, although the high-temperature exposure time was lower (1-h hold time per cycle), the number of cycles required for crack initiation was higher for the much higher prestress level.

In the CA-GTAW joint, the crack was observed at the interface of the weld zone and AISI 316L SS — Fig. 10A. On the other hand, in the FGW joint, the interface of the FGW-2 was the location of crack initiation (Fig. 10B). However, the locations of failures were contrary to the earlier studies conducted by many researchers, which observed that the ferritic steel–weld zone interface was the failure region in the ferritic-austenitic DMWs (Refs. 2, 4, 31). It should be noted that, in previous studies, the DMW joints were developed using austenitic (Type 309 SS) (CTE

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Prestress Level</th>
<th>Whether Crack Observed after 400 Cycles</th>
<th>Number of Cycles for Crack Initiation</th>
<th>Crack Location</th>
</tr>
</thead>
<tbody>
<tr>
<td>CA-GTAW joint</td>
<td>200 MPa</td>
<td>No</td>
<td>–</td>
<td>–</td>
</tr>
<tr>
<td>FGW joint</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>CA-GTAW joint</td>
<td>400 MPa</td>
<td>Yes</td>
<td>290</td>
<td>Weld zone—316L interface</td>
</tr>
<tr>
<td>FGW joint</td>
<td></td>
<td>Yes</td>
<td>360</td>
<td>FGW-1–FGW-2 interface</td>
</tr>
</tbody>
</table>

**Fig. 10 — Optical micrographs of crack initiation at interfaces: A — CA-GTAW joint; B — FGW joint.**
~ 18 μm/m/˚C) or nickel-based (Inconel 82/182) (CTE ~ 15.5 μm/m/˚C) filler metals, which have a much higher CTE than ferritic steels (Ref. 2). Therefore, the CTE mismatch in weld joints with these fillers was higher on the ferritic steel side interface than that of the austenitic SS side. In the current study, the presence of martensitic weld zones having a CTE value closer to the P91 steel leads to lower CTE mismatch and reduction in thermal stresses at the P91 steel–weld zone interfaces. The observations are in line with the earlier efforts made to develop ferritic-austenitic DMWs using fillers that exhibit similar CTE values to ferritic steels, such as martensitic steel filler metal (HFS-3) and nickel-based fillers (HFS-6 and EPRI P87), which in turn reduce thermal stresses across the ferritic steel–weld zone interface (Refs. 33–35).

Moreover, thermal exposure time at high temperature in the current study was lower as compared to other studies for severe metallurgical degradation, such as carbon diffusion and oxide notch formation at the ferritic steel side. However, the PWHT for 25 h at 750˚C used during the study of carbon migration corresponds to approximately 23,800 h thermal exposure at 625˚C calculated based on the Larson-Miller parameter equation ($P = T(C + \log t)$; $C = 20$) (Ref. 6). These observations confirmed a significant improvement in the joint performance of the FGW joint than the CA-GTAW joint by reducing the severity of carbon migration and increasing thermomechanical fatigue life.

**FE Simulations**

Thermal stresses developed at the interfaces in the DMW joint significantly affect its life and performance. A FE analysis was conducted to analyze the stresses developed during thermal cycling. It can be seen that a 37% reduction in peak thermal stresses was achieved in the FGW joint as stresses at 625˚C were lower (294 MPa) than that of the CA-GTAW joint (467 MPa) — Fig. 11. The maximum stress location in the CA-GTAW joint was the AISI 316L–weld zone interface (Fig. 11A). On the other hand, the interface between FGW-1 and FGW-2 was the location of peak stress in the FGW joint (Fig. 11B). Additionally, higher stresses were located at the toe of the weld joint in both cases, i.e., at the outer periphery in the 2D axisymmetric model, rather than the root side of the weld joint. The predicted locations of peak stress by the FE analysis were in line with crack initiation locations in the thermal cycling tests.

The reduction in peak thermal stress is attributed to the CTE mismatch ($\Delta\alpha$). Eiasazadeh et al. (Ref. 36) reported that CTE mismatch plays a nontrivial role in the magnitude of residual stresses in ferritic-austenitic DMWs. Woo et al. (Ref. 37) postulated a linear relationship between the CTE mismatch and normalized residual stresses in ferritic-austenitic DMWs. A simplified approximation for thermal stress ($\sigma_{th}$) in a material having Elastic modulus (E) and Poisson’s ratio ($\nu$) due to the CTE mismatch ($\Delta\alpha$) in the bimetallic configuration after a change in temperature ($\Delta T$) can be given by Equation 3 (Ref. 38).

$$\sigma_{th} = \frac{E}{1-\nu} \times \Delta\alpha \times \Delta T$$

The estimated values of peak thermal stresses for a temperature rise of 625˚C were 463 MPa (CA-GTAW joint) and 318 MPa (FGW joint), respectively, which were quite comparable with the predicted values by the FE simulation. Therefore, it can be deduced that reduction in the CTE differences across the interfaces is vital in reducing thermal stresses and improving the service life of the DMW between steels and stainless steels.
Effect of Interlayer Widths in FGW Joint Development

With an increase in interlayer width, the dilution level from the interlayer melting increases. The issues, such as incomplete interlayer melting, arise at higher interlayer widths. The interlayers of a narrow size interlayer combination (0.5 and 3 mm) resulted in more overlapping between the welds, which in turn reduced the width of the fusion zone. The width of the fusion zone increased with a broad size interlayer combination. To achieve a sound FGW joint with a narrow size interlayer combination, a torch offsetting up to 2 mm is required toward the side where an earlier welding pass was made.

A step-wise chemical compositional gradient can be achieved with narrow, medium, and broad size interlayer combinations. However, the narrow size interlayer combination resulted in a higher Cr\textsubscript{eq} and Ni\textsubscript{eq} in the weld zones of the FGW joint than in those made with broad size interlayers. This was attributed to the higher overlapped area and dilution from the weld zone developed in the earlier welding pass in case of a narrow size interlayer combination. Additionally, in the first welding pass of the FGW joint with a narrow size interlayer combination, a lower dilution of 0.5-mm-wide P91 steel interlayer compared to the AISI 316L resulted in the higher Cr\textsubscript{eq} and Ni\textsubscript{eq} in the FGW-1. Therefore, the subsequent passes were produced with higher Cr\textsubscript{eq} and Ni\textsubscript{eq} due to the self-grading effect. No significant differences were observed in the type of phases present in the fusion zones with different sized interlayers. However, the carbon potential gradient at the P91 steel–FGW-3 interface was lower for the FGW joint with broader interlayers.

Conclusion

In this work, a novel way of developing a FGW joint between a dissimilar steel combination of P91 steel and AISI 316LSS with the use of multiple interlayers and three overlapping A-GTAW passes was presented. The performance of the FGW joint was compared with the CA-GTAW joint. The following conclusions can be drawn from this work.

1. A step-wise functional grading in chemical composition and microstructure was achieved in the FGW joint from the P91 steel side to the AISI 316L side by employing a novel welding procedure involving multiple interlayers.
2. During tensile testing, both the CA-GTAW and FGW joints exhibited overlapping strengths.
3. The severity of carbon migration was significantly reduced in the FGW joint.
4. During thermal cycling tests, the FGW joint exhibited a longer life compared to the CA-GTAW joint.
5. The 316LSS–weld zone interface of the CA-GTAW joint and FGW-1 and FGW-2 interfaces of the FGW joint were the location of failures during both tensile testing and thermal cycling tests.

Acknowledgements

The authors would like to acknowledge the support of the Board of Research in Nuclear Sciences, Department of Atomic Energy, Government of India, for funding the research project titled “Dissimilar Steel Welding by Activated Flux GTAW” under the scheme-sponsored research project sanction No. 36(2)/14/70/2014-BRNS/10416. The authors would also like to thank Prof. Vikram V. Dabhade (MMED, IIT-Roorkee) and Harshal Kulkarni for their support in carrying out dialatometric analysis.

References


ANUP KULKARNI (anupkulkarni@me.iitr.ac.in) and DHEERENDRA KUMAR DWIVEDI are with the Department of Mechanical and Industrial Engineering, Indian Institute of Technology, Roorkee, India. M. VASUDEVAN is with the Materials Development and Technology Division, Metallurgy and Materials Group, Indira Gandhi Centre for Atomic Research, India.