Influence of intermetallic precipitation on metallurgical, mechanical and pitting behavior of AISI 2205 duplex stainless steel welded joints

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Abstract
Post weld thermal aging of DSS (duplex stainless steel) welds resulted into complex precipitation that comprised of intermetallic carbides, oxides and sigma-phase in different zones of the welds. Maximum ferrite dissolution was observed in the weld metal followed by the HAZ (heat affected zone) and the base metal. Sigma phase precipitated preferentially along $\gamma/\delta$ (austenite/ferrite) interfaces besides the formation of secondary austenite ($\gamma_2$). Precipitated joints showed an increase in microhardness and yield strength, whereas their UTS (ultimate tensile strength) did not alter much. Aging resulted into a significant loss of tensile ductility (percentage elongation) which showed a degradation of 15.27% and 62.46% for the base metal and the weld metal respectively. The base metal’s impact toughness did not show any appreciable loss, but a severe degradation of around 96% occurred in the impact toughness of the aged weld metal. Fatigue crack growth behavior showed that weld metal possessed superior fatigue strength than the base metal, which upon aging reduced drastically to 48.24% and 28.44% respectively. Pitting corrosion studies showed that weld metal exhibited superior pitting resistance than the base metal, and both of them suffered a loss of pitting resistance due to precipitation.

1. Introduction

Duplex stainless steels (DSS), owing to their superior performance than the conventional austenitic stainless steels, find extensive use in environments involving operating conditions like chemical, petrochemical, nuclear, marine, fertilizer, food industries etc. DSSs possess higher strength, toughness and high resistance to corrosion than any other steel. This combination of properties can be attributed to their dual phase structure which comprises of nearly equal volume of austenite ($\gamma$) as well as ferrite ($\delta$) phases [1–3]. Since their development continuous research efforts have been made for their improvement via understanding the role of various alloying elements. The main research focus includes the wrought products of DSS besides which limited number of works have been reported on the weld products also. Important industrial applications of DSS and that too in harsh environments continue to pose a serious challenge from the beginning itself in terms of maintaining an optimum phase balance which gets disturbed due to welding. In view of this microstructural imbalance, their mechanical as well as corrosion performance get significantly affected due to which the engineering potential of the DSS does not get fully utilized. Several important microstructural studies on DSSs have been reported in the literature. Intermetallic precipitations in these steels have been investigated by various researchers [4–8]. In particular, $\sigma$-phase increases with aging time and maximum volume of it occurs at 850 °C [9]. In a study on pulsed-laser welded DSS a significant reduction in $\sigma$-phase has been reported in the fusion zone at the leading side which makes the weld heating control necessary in these steels [10]. Effect of prolonged temperature exposure on pitting corrosion has been reported to be initiated at the ferrite phase selectively which along with the presence of sigma makes DSS prone to pitting [11]. $\sigma$-phase has a deleterious effect on the impact toughness.
and the pitting potential of annealed DSS [12]. A similar study on corrosion fatigue of a DSS in high temperature water shows that with increased aging time the hardness and modulus of ferrite increases whereas impact energy and corrosion resistance decrease [13]. Welding results into dissolution of phases in DSS, this shows that welding thermal cycle exerts a decisive influence on their toughness properties [14]. Further, welding thermal cycle leads to different microstructures in different zones of the weld due to which ferrite grain coarsening as well as intragranular and secondary austenite formation occurs in the ferrite grains [15]. Fatigue crack propagation is influenced by low (375 °C) as well as high (800 °C) embrittlement temperatures [16].

Based upon the review of the published research it is observed that for DSS, although their impact toughness, microstructural studies, sensitization studies, fatigue studies etc have been reported, however interdependence of metallurgical issues with that of mechanical as well as corrosion behavior of DSS welds has not been reported much. Thus to gain further understanding about these issues, the present work was undertaken, where post weld thermally aged DSS welds were studied for their metallurgical, tensile, impact toughness, fatigue and pitting corrosion behavior.

2. Experimental details

2.1. Materials combination and welding procedure

AISI 2205 grade, commonly known as DSS, was selected as the base material in the present investigation welded with E2209 coated electrode (AWS 5.4) for shielded metal arc welding (SMAW) process and ER2209 TIG welding wire (AWS 5.9) as the compatible fillers for welding, and the composition of both these materials is shown in Table 1.

The base plates were cut into suitable size (600 mm × 75 mm × 10 mm each) from the hot rolled plate and the edges were prepared formulating a double-V groove configuration as shown in Figure 1(a). After pre-cleaning the edges of the base plates to make them free from any source of contamination like dirt, oil, grease, oxide scales. Butt welds were fabricated using the welding conditions as shown in Table 2, corresponding to each weld pass along with heat input used during each weld pass. Welding procedures used for fabricating the butt welded joints comprised of firstly, giving the root weld pass (1) using GTAW process, followed by the middle weld pass (2 and 3 respectively) and then the cover pass (4 and 5 respectively) on each side of the symmetric joint using SMAW process. An interpass temperature of around 150 °C was maintained during welding and due care was taken for interpass cleaning between subsequent passes.

After extraction from different locations the specimens were subjected to two post weld thermal aging conditions as T1 (350 °C/30 min) and T2 (850 °C/30 min) in a furnace followed by cooling in still air.

2.2. Metallurgical studies of welds

2.2.1. Ferrite studies

Metallographically prepared specimens were subjected to ferrite measurements using an instrument Ferritescope (Model: M30, Make: Fisher International, Germany). These measurements were taken at different locations of the weldments and the average values taken from each location were recorded.

2.2.2. Microstructural studies

Weld specimens were extracted from the welded joints and were prepared for metallographic examination using standard polishing procedures which involved successive grinding using different emery grades followed by polishing using diamond paste. Etching was done by immersing the specimens in Murakami’s reagent (comprising of 1 gm. K2S2O5 + 15 ml HCl + 85 ml distilled H2O), followed by cleaning of the etched surface using acetone. The surfaces so prepared were viewed under optical microscope as well as SEM coupled with EDS for capturing micrographs from different regions of the welds for studying different microstructural phases. EDS of a few selective regions of microstructures were taken to reveal the chemical composition of different precipitates observed in these welds which were also subjected to XRD analysis.

2.3. Mechanical testing of welds

2.3.1. Microhardness studies

Polished specimens were subjected to microhardness evaluation using a microhardness tester of two kg capacity using testing conditions of 500 gm load with a dwell time of 15 s. Different zones of the weld joint were examined.

2.3.2. Tensile testing of welds

The specimens were extracted from the welded plates as shown in Figure 1(b), such that tensile load could be applied transversally to the welding direction, thus revealing information about the transverse tensile properties.
### Table 1. Chemical composition of the base material and the filler material.

| Material | C%  | Si% | Mn% | P%  | Cr% | Ni% | Mo% | Cu% | Co% | Ti% | Nb% | Al% | V%  | N%  | Fe% |
|----------|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|
| 2205     | 0.024 | 0.562 | 0.762 | 0.003 | 0.028 | 22.213 | 5.323 | 3.110 | 0.120 | 0.054 | 0.010 | 0.00 | 0.00 | 0.128 |
| E2209    | 0.026 | 0.90 | 0.90 | 0.002 | 0.025 | 22.10 | 10.00 | 2.84 | —    | —    | —    | —    | —    | —    | 0.18 |
| ER2209   | 0.016 | 0.42 | 0.42 | —    | —    | 23.07 | 8.60  | 3.20  | 0.16  | —    | —    | —    | —    | —    | 0.160 |
of the welded joints. The size, shape and orientation of transverse tensile specimens were selected in accordance with ASTM E08/E08M-11 standard [17] and testing was carried out on a digitally controlled computerized Universal testing machine of 50 kN capacity (Model: HK 50, Make: Tinius Olsen, UK).

2.3.3. Impact toughness testing of welds
Charpy V-notch impact specimens of full size were extracted from the welded joints and prepared in accordance with ASTM E-23 standard with V-notch lying at the center of the weld metal [18]. An impact tester of 300 J capacity was used for testing these specimens at room temperature. The specimens were extracted and prepared in triplicate and the average of each was recorded.

2.3.4. Fatigue crack growth rate test (FCGR) of welds
The fatigue crack growth behaviour of welds was studied as per ASTM E-647 standard. Samples referred to as CT (compact tension) specimens, were extracted from the welded joints as shown in figure 1 (b) (specimens 1 and 1’ represent the base metal with crack orientation along and across the rolling direction respectively, whereas 1” represents the crack orientation along the weld centreline), using a wire cut EDM, and were prepared with a pre-cracked length of 5.5 mm, with a width (W) of 25 mm and a/W ratio of 0.416, and were subjected to cyclic loading using loading and unloading rate of 0.025 mm s$^{-1}$ and 0.008 mm s$^{-1}$ respectively. The geometrical details of the CT specimens used in the present work are shown in figure 2.

Constant loading and unloading conditions were maintained on a computerized and servo hydraulically controlled low cycle fatigue testing machine of 25 kN capacity. The plots were generated between crack propagation and the number of cycles based upon the data generated which was acquired with the dedicated software. For each specimen the same testing conditions were used for FCGR testing and crack extension from 5.5 mm to 11.25 mm was analyzed for studying the fatigue crack propagation behaviour of different specimens comprising of the base metal and the weld metal, both in the untreated as well as the aged conditions. Further for facilitating characterization of the nature of the fracture modes, all the specimens were pulled apart up to the point of fracture on the same machine. The fractured surfaces were further examined with SEM fractography and EDS of few selected regions from the fractographs were also recorded.
| Weld pass sequence no. | Welding process | Current (I) (A) | Voltage (V) (V) | Gas flow rate (l min⁻¹) | Welding speed (S) (mm min⁻¹) | Heat input per weld pass = η(VI)/S, where η = 0.7 is the heat transfer efficiency (kJ mm⁻¹) |
|------------------------|-----------------|----------------|----------------|--------------------------|-------------------------------|----------------------------------------------------------------------------------|
| 1 (Root pass)          | GTAW            | 120            | 15             | 12                       | 62.5                          | 1.296                                                                             |
| 2 (Middle pass-top side)| SMAW            | 115            | 24             | —                        | 93                            | 1.335                                                                             |
| 3 (Middle pass-bottom side)| SMAW        | 115            | 24             | —                        | 96                            | 1.293                                                                             |
| 4 (Cover pass-top side)| SMAW            | 120            | 24             | —                        | 82                            | 1.58                                                                              |
| 5 (Cover pass-bottom side)| SMAW       | 120            | 24             | —                        | 82                            | 1.58                                                                              |
2.3.5. Pitting corrosion testing of welds
Specimens were extracted from the welded joints for carrying out pitting corrosion measurements using PAP (Potentiodynamic anodic polarization) technique in accordance with ASTM G5-13 standard [19]. The surface preparation of the samples was done using emery paper up to 1000 grit size. Testing conditions as reported in the literature [20, 21] were used for this test where electrolyte having acid solution with chlorides containing 0.5 M H₂SO₄ (sulfuric acid) + 0.5 M NaCl (sodium chloride) at room temperature was used. This testing was carried out on a Potentiostat/Galvanostat (Make: Gamry Instruments, USA; Model: Reference 600) which was controlled by a dedicated software DC 105. The experimental procedure used for testing these specimens comprised of firstly a 5 min delay at open circuit potential (Voc) followed by anodic potentiodynamic scan which started at 50 mV below Voc to 1500 mV (SCE) at a potential scan rate of 100 mV min⁻¹. For each specimen two readings of pitting potential (Epitt) were taken and each individual reading was taken using a fresh test solution.

3. Results and discussion
3.1. Ferrite studies
Ferrite measurements carried out on different specimens at different locations showed a significant variation in different zones of the weld, the results of which are presented in table 3.

In the as welded condition, the root region of the weld zone showed the highest ferrite content (65%) followed by the middle pass (61.8%) and then the cover weld pass (47.6%). The ferrite content in the HAZ and the unaffected base metal was found to be 54.9% and 52.4% respectively. Ferrite content in all the zones reduced moderately and significantly, when subjected to thermal aging conditions T₁ (350 °C/30 min) and T₂ (850 °C/30 min) respectively. The ferrite values of shielded metal arc DSS welds as found in the present case were on the higher side as compared to gas metal arc DSS welds [5], thus indicating that welding processes variation owing to their specific parametric selection as well as shielding medium can have a strong influence on the metallurgical balance of γ/δ (austenite/ferrite).

3.2. Microstructural results
Optical as well as scanning electron micrographs of different zones of the weldments were taken to identify different phase formations that occurred under aging conditions which are shown in figures 3 and 4 respectively. Different microstructural phases were also subjected to XRD examination to study different compound

Figure 2. Geometrical details of the CT specimens used in the present work (all dimensions in mm).

Table 3. Ferrite (%) results of different specimens.

| Region/pas          | Untreated | Aging condition T₁ (350 °C/30 min) | Aging condition T₂ (850 °C/30 min) |
|---------------------|-----------|-----------------------------------|-----------------------------------|
| Weld metal Root pass| 65        | 52.5                              | 33.8                              |
| Middle pass         | 61.8      | 54.2                              | 28.6                              |
| Cover pass          | 47.6      | 43.7                              | 18.6                              |
| HAZ (near the root pass) | 54.9  | 48.2                              | 41.3                              |
| HAZ (near the root pass) | 55.9  | 52.3                              | 39.6                              |
| Unaffected base metal | 57.5  | 54.7                              | 43.4                              |
formations. The precipitation mechanism of DSS welds showed different morphologies of austenite viz. primary (γ₁), secondary austenite (γ₂) and tertiary austenite (γ₃) [4]. Primary austenite (γ₁) can directly form from liquid to solid transformation i.e. \( L \rightarrow \gamma_1 \) and \( L \rightarrow \delta + \gamma_3 \), whereas subsequent solid state transformation of ferrite into austenite can also occur [5]. Such mechanism of austenite formation from ferrite can give rise to different morphologies of austenite like GBA (grain boundary austenite), WA (Widmanstätten austenite), IGA (intragranular austenite) and PTA (partially transformed austenite), which in the present case was seen to form in the specimen WMT₂ as shown in figure 3(d). Figure 4(a) shows the SEM micrograph of the weld metal of specimen WMT₂ corresponding to region 2 (middle weld pass as shown in figure 1) which lies at a distance of around 3.5 mm from the top of the base plate surface. σ-phase, which is seen as white reflective phase (encircled by a square in figure 3(a)) was found to form on the interfacial regions of δ/γ. Sigma phase usually starts after carbide formation where δ/γ interfaces act as preferential sites for heterogeneous nucleation [22]. In the process of sigma phase formation, Cr gets absorbed and Ni gets rejected into adjacent regions, thus favoring the tendency of γ₂ formation near sigma phase. Secondary austenite (γ₂) was also observed as indicated on this micrograph which had lower levels of Cr and Mo as compared to the primary austenite (γ). It is reported that ferrite changes into sigma phase and secondary austenite [1], and this reaction continues till all the ferrite gets consumed. Secondary austenite (γ₂) formed in the interior of ferrite and at the δ/γ₁ interfacial regions when the middle weld pass was subjected to high temperature heat cycles. Table 4 shows EDS results of the aged welds taken at different locations so as to indicate the chemical composition associated with different regions. Further two EDS spectra viz. Figures 6(f) and (l) as mentioned in table 4, were also taken corresponding to the fractured

![Figure 3. Optical micrographs of base and weld metal samples at different aged conditions.](image-url)
ends of the tensile tested and Charpy V-notch tested specimens respectively, so as to find out and attribute the loss of ductility as well as impact toughness associated with the aged welds to changes in the chemical composition that would indicate towards some kind of intermetallic compound formation.

Figure 4. SEM micrographs of the specimen taken from different zones of the weldment.

Table 4. EDS results of the aged welds.

| Micrograph | Region       | Cr  | Ni  | Mo  | Si  | Fe  | C   | O   | Co  | V   |
|------------|--------------|-----|-----|-----|-----|-----|-----|-----|-----|-----|
| figure 4(a) | 1            | 17.53 | 9.25 | 0.79 | 0.71 | 61.23 | 3.32 | 6.71 |
| figure 4(a) | 2            | 16.76 | 12.40 | 1.64 | 0.34 | 63.55 | 2.80 | 1.93 |
| figure 4(b) | 3            | 19.90 | 11.59 | 1.66 | 0.52 | 60.43 | 2.83 | 1.67 | 1.27 |
| figure 4(b) | 4            | 17.77 | 12.45 | 2.06 | 0.51 | 60.29 | 3.03 | 2.62 |
| figure 4(c) | 5            | 20.26 | 8.27 | 0.95 | 2.17 | 45.88 | 7.32 | 9.75 |
| figure 4(c) | 7            | 21.86 | 5.08 | 1.95 | 0.22 | 54.49 | 11.28 | 3.46 | 1.47 |
| figure 4(d) | 6            | 20.41 | 9.11 | 2.06 | 0.49 | 47.82 | 11.05 | 7.08 |
| figure 4(d) | 7            | 15.01 | 5.58 | 0.88 | 2.60 | 27.75 | 15.13 | 24.73 |
| figure 4(f) | Spectrum 1 (σ-phase) | 25.66 | 5.22 | 6.19 | 1.82 | 58.39 | —   | —   | —   |
| figure 4(f) | Spectrum 2    | 18.74 | 3.52 | 1.42 | 0.36 | 54.21 | 16.71 | —   | 2.42 | 0.61 |
Table 5. Microhardness, Tensile, Impact toughness, fatigue cycles count and Pitting potential results.

| Property                        | Base metal | Weld metal |
|---------------------------------|------------|------------|
| Microhardness (HV<sub>1.5gms</sub>) | 262.2 307 332 | 339 350 360 |
| Tensile Yield stress (MPa)      | 621 614 604 | 602 582 583 |
| UTS (MPa)                       | 767 732 722 | 746 742 734 |
| Elongation (%)                  | 47.8 43.7 40.5 | 39.7 31.8 14.9 |
| Impact toughness (CVN value)    | 276 J 264 J 202 J | 217 J 198 J 8 J |
| Cycles count (across the grains)| 123 307 — 88 227 | 232 240 — 120 205 |
| Cycles count (along the grains) | 152 183 — 99 217 | — — — |
| Pitting potential (mV)          | 335.6 309 253.6 | 625.7 535.9 439.2 |

Figure 4(b) shows the SEM micrograph of the weld metal comprising of the cover pass. Here sigma phase formation was observed at δ/γ interfacial regions (region 3), besides which it also occurred as having coral morphology (shown by region 3′) which further is seen dispersed as fine regions as well as coarse regions.

Further, austenite grain coarsening was also observed here which may be attributed to high heat input used for this particular weld pass, since relatively larger groove volume necessitated the use of higher metal deposition rate, thus leading to slower cooling rate which resulted into austenite grain coarsening. Spectrum taken on region 3 showed the presence of oxide inclusions mainly comprising of Si and Co (viz SiO<sub>2</sub> and CoCo<sub>2</sub>O<sub>4</sub>) as indicated by the XRD results), whereas spectrum taken at region 4 showed the formation of γ<sub>2</sub>, thus indicating a region deficient in Ni, besides the formation of different oxides which as per the XRD examination were identified as NiCr<sub>2</sub>O<sub>4</sub>, NiO<sub>2</sub>, TiO<sub>2</sub> and Ti<sub>3</sub>O<sub>5</sub> compounds. Figure 4(c) shows the SEM micrograph of the root region of fusion zone where PTA as well as WA is observed besides large GBA. Such an observation may be attributed to the fact that some portion of the root pass got remelted while middle weld pass was laid and thus there formed a HAZ within the root weld pass which experienced a temperature range where the formation of PTA occurred. Point spectrums taken at region shown by 5 in this micrograph indicate the presence of oxides (comprising of Ti, Si and Co) which when analyzed through XRD examination were found to be TiO<sub>2</sub>, Ti<sub>3</sub>O<sub>5</sub>, SiO<sub>2</sub> and CoCo<sub>2</sub>O<sub>4</sub>. Spectrum at region 6 shows ferrite containing carbides as well as oxides. Figure 4(d) shows the SEM micrograph of HAZ of the weld joint where the formation of carbides as well as oxides in austenite region was observed as shown by spectrum at region 7. The formation of carbides like SiC, Cr<sub>7</sub>C<sub>3</sub> and oxide NiO<sub>2</sub> resulted into lowering of Cr and Mo which further lowered the level of Fe to 27.75% as indicated by the EDS results. Figures 4(e) and (f) show the SEM micrograph of the grain coarsened HAZ and the composite zone of the weld, where again different morphologies of austenite were observed.

3.3. Microhardness results

Microhardness possessed by the cover pass on both the sides of the weld (as-welded condition) was in the range of 268–287 HV as compared to the middle weld pass which showed a value of 277 HV, whereas HAZ and the base metal showed microhardness of 242 and 254 HV respectively. Higher microhardness associated with the top and the bottom weld pass could be attributed to the higher cooling rate as compared to the root/middle region which received heat during laying down of the subsequent weld passes and hence underwent relatively slower cooling rate and thus possessed relatively low hardness. HAZ of the weld joint, owing to experiencing microstructural grain coarsening, showed a lower hardness level. When subjected to thermal aging conditions, all the zones showed a significant increase in their respective microhardness values which could be attributed to intermetallic precipitation formation in these joints.

3.4. Tensile testing results

The results of the transverse tensile testing conducted on different specimens are presented in table 5 and the stress strain curves for the transverse tensile tested welds are shown in figure 5.

The base metal as well as the welded joint did not show much variation in the UTS value, although the base metal exhibited better ductility (percentage elongation) as compared to the weld metal under all conditions. However, the base metal showed a degradation of ductility under aging conditions T<sub>1</sub>(350 °C/30 min) and T<sub>2</sub> (850 °C/30 min) such that the total elongation reduced from 49.7% to 40.5%. In comparison, the weld metal showed a drastic loss of ductility from 39.7% to 14.9%, thus recording a loss of 62.46%. Such a drastic loss of ductility in the weld may be attributed to the presence of sigma phase precipitation as well as intermetallic carbides and oxide inclusions which formed during thermal aging.

Fractographs as shown in figure 6, indicate different fracture morphologies which explain the mechanism followed by these specimens during the fracturing process under tensile loading conditions. Figure 6(a) shows...
that the specimen BMT₀ (untreated base metal) showed a classical dimpled morphology (indicative of high ductility) with consistent tearing ridges on the entire fracture surface, which continued to dominate for the aged specimens BMT₁ as well as BMT₂, with the only difference that the presence of intermetallic precipitation and oxides in these specimens led to void formation which resulted into their reduced ductility. For the welds, specimen WMT₀ (as-welded) showed the presence of large sized voids which tend to coalesce and thus form large, deep and continuous cracks. The extent of precipitation being higher in case of specimen WMT₁, these voids (shown by arrows) became larger and thus showed the signs of reduced ductility. The most detrimental loss of ductility as found in specimen WMT₂ (figure 6) was accompanied by the change of tensile fracture from ductile to quasi-cleavage as indicated by shallow and flat featureless regions with absence of deep dimples and tearing ridges due to precipitation of brittle 𝜎-phase whose presence was evidenced from its EDS analysis as shown in table 4.

3.5. Impact testing results of the welds

The CVN (Charpy V-notch) testing results of different specimens tested at room temperature are presented in table 5. The base metal possessed higher CVN value than the weld metal which may be attributed to its typically metallurgical balanced duplex structure mainly comprising of ferrite and austenite in favorable proportions due to which it could, under all conditions, retain its CVN value in the range of 202–276 J, although corresponding to aging condition T₂ (850 °C/30 min) a loss of CVN value of around 74 J was recorded. However, in comparison, the CVN value corresponding to the specimen WMT₂ (as-welded) showed a catastrophic deterioration of impact toughness which showed a CVN of only 8 J as compared to 217 J possessed by the specimen WMT₀, thus showing a drastic loss of 96.31% due to aging. Such a serious degradation of impact toughness may be attributed to the detrimental effect induced by different precipitating phases when the joint was subjected to high temperature aging condition. As suggested by the literature, sigma phase formation, and secondary austenite precipitation, beside carbides as well as oxide formation in the DSS alloy has been reported to adversely affect its impact toughness properties [12]. In general, it was observed that higher austenite content improved ductility and hence the impact toughness of the weld. When these welds were subjected to high temperature aging, multiple factors came into play like grain coarsening of ferrite as well as austenite, formation of hard and brittle sigma phase along the δ/γ interfaces, besides the formation of intermetallic carbides as well as oxide inclusions, which tend to weaken the loading capacity of the weld to such an extent that the weld metal experienced a significant degradation in its impact toughness. From figure 6(g), it is seen from the SEM micrographs that dimple fracture prevailed in the base metal as well as the weld metal in the untreated condition. The presence of γ₂ was responsible for preventing crack growth through brittle ferrite by absorbing more impact energy. The presence of spherical inclusions (indicated by arrows) acted as nucleation sites for dimple formation. Discontinuous fracture of the dimples resulted due to these inclusions which resulted into reduced loading area and hence low impact toughness. Figure 6(l) shows that quasi cleavage fracture mode dominated in
the specimen WMT2 which was identified from the deep, wide and continuous cracks accompanied by flat featureless regions, which could be attributed to σ-phase formation as well as carbide formation as evident from its EDS result (table 4).

3.6. Fatigue crack growth rate (FCGR) test results

FCGR test carried out on different specimens resulted into variable response in terms of number of cycles taken to propagate a fatigue pre-crack size from 5.5 mm to 11.25 mm and the results of the same are presented in table 5 and shown in figure 7(a). For the base metal, fatigue crack orientation kept along the rolling direction (specimen 1 in figure 1(b)), resulted into greater number of fatigue cycles i.e. 152 183 as compared to when the crack was oriented across the rolling direction (specimen 1' in figure 1(b)), which withstood 123307 number of cycles. Aging condition T2 deteriorated the FCGR performance of the base metal. The weld metal exhibited (specimen 1' in figure 1(b)), better fatigue performance as compared to the base metal, but showed a loss of fatigue performance in terms of number of fatigue cycles withstood during testing from 232 240 cycles (as welded condition) to 120 205 cycles (under T2 aging condition), thus resulting into a loss of 48% in its fatigue performance. This degradation in the fatigue performance of the weld joint could be attributed to different

Figure 6. SEM fractographs of tensile and impact toughness tested specimens; Tensile tested specimens [Base metal: (a) to (c) and Weld metal: (d) to (f)] and Charpy impact tested specimens [Base metal: (g) to (i) and Weld metal: (j) to (l)].
precipitating phases like sigma as well as carbides and oxides inclusions which as already discussed previously, tend to induce embrittling tendencies into the joint, thus adversely affecting its load carrying ability under fatigue loading conditions. Further, the influence of such a combination of different precipitates on the fracturing mechanism of the weld joint was studied via SEM fractography carried out on fatigue tested specimens and few of these fractographs are shown in figures 7(b) to (d). The fracture morphology shows that intermetallic precipitation resulted into deep, wide and continuous cracks (figure 7(d)), thus indicating the reduced fatigue performance.

3.7. Pitting corrosion test results

The results of the pitting corrosion evaluation of the welds using PAP technique are presented in table 5 and the anodic polarization curves so generated are shown in figure 8. All the anodic polarization curves exhibited an active-passive-transpassive polarization behavior. It can be seen that degradation of pitting resistance is found both in the base metal as well as the weld metal under aging conditions. This loss of pitting potential may be attributed to precipitation which led to segregation effects thus making the matrix deficient in Cr, Ni, Mo etc which made the surface vulnerable to pitting attack. In the present case, sigma phase precipitates which were enriched in Cr and Mo tend to enhance segregation, thus leaving the adjoining areas prone to pitting attack. Further, pitting tendencies are also found to be promoted due to the presence of oxide inclusions which act as preferential regions and source of nuclei for the pit formation figures 9(a) to (c) show evidence of pits formation in welds at different aged conditions. It was observed that the weld with higher pitting potential (WMT ε) exhibited smaller pits formation tendencies and when subjected to thermal aging the size of the pits increased. Owing to the higher pitting potential possessed by WMT ε weld metal, this weld also had lower transpassive dissolution region under the same ending potential of 1500 mV for all the welds, due to which pits formed were smaller in size (figure 9(a)). Besides this, another important point worth noting for WMT ε was that it possesses more noble corrosion potential value relative to other welds which indicates higher thermodynamic stability of electrolyte solution and sample electrode interface.
4. Conclusions

1. Austenite to ferrite ratio showed a significant variation among different zones of the weld. Weld metal possessed least ferrite followed by the HAZ and the base metal. Thermal aging resulted into dissolution of ferrite which lowered the ferrite content significantly across all the weld passes. The order of ferrite reduction that followed was, the weld cover pass (47.6% to 18.6%), middle weld pass (61.8% to 28.6%), root region (65% to 33.8%), HAZ (55.9% to 39.6%) and the base metal (52.4% to 43.4%).

2. Aging treatment of 850 °C/30 min resulted into intermetallic precipitation comprising of sigma phase, carbides and oxides inclusions in the weld joint due to which an overall increase in microhardness was observed across different zones of the weld.

3. Yield and ultimate tensile strength, of the base metal and the weld metal did not show any significant variation due to precipitation, but a significant loss of ductility (percentage elongation) of around 62.46%, as compared to the as-welded condition, was observed in the weld metal corresponding to 850 °C/30 min aging condition.

4. The base metal possessed higher impact toughness as compared to the weld metal and both of them nearly retained it even after precipitation, except in the case of 850 °C/30 min aged weld metal which suffered a catastrophic degradation (CVN reduced from 217 J to 8 J) due to the combined effect of sigma phase, intermetallic carbides and oxide inclusions.

Figure 8. Pitting curves for the weld metal under different conditions.

Figure 9. Evidence of pits formation in welds: (a) WMT<sub>0</sub>, (b) WMT<sub>1</sub>, (c) WMT<sub>2</sub> at different aged conditions.
5. The weld metal showed a superior FCGR performance as compared to the base metal. However, precipitation that occurred due to 850 °C/30 min aging condition led to a loss of fatigue performance of the weld metal by 48.24% and the base metal by 28.44%, when measured in terms of total number of cycles taken to propagate the fatigue crack under the same testing conditions.

6. The weld metal exhibited superior pitting corrosion performance as compared to the base metal and both these zones showed a loss in their respective pitting potentials due to precipitation.

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