Microstructural Characterization of Dissimilar Metal Welded Joints in Nozzle Safe-Ends of Third-Generation Nuclear Power Plants (Postprint)

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Abstract

Using microscopic analysis techniques including OM, TEM, SEM, microhardness tester, AFM, magnetic force microscopy (MFM), and scanning Kelvin probe (SKPFM), the microstructure, microhardness, distribution of major alloying elements, grain boundary types, and residual strain of the dissimilar metal welding joint of low-alloy steel A508/nickel-based weld metal 52M/austenitic stainless steel 316L at the safe end of the reactor pressure vessel in advanced pressurized water reactor nuclear power plants were analyzed, and the microstructure and properties across different thicknesses of the entire welding joint were compared. The results indicate that there is no significant difference in microstructure and hardness along the weld thickness direction. A layer of fine equiaxed grains formed by unmelted weld metal appears at the root weld position. The residual strain within the 316L base metal heat-affected zone (HAZ) is higher than that in other locations of the welded component. The vicinity of the fusion line exhibits complex distributions of microstructure, microhardness, grain boundary types, elemental composition, and residual strain. TEM and MFM analyses reveal that there are granular precipitates rich in Cr and Mo elements within the 316L base metal matrix. SKPFM results demonstrate that this precipitate phase has a more negative Volta potential than the matrix, and is therefore less corrosion-resistant.

Full Text

Micro-Characterization of Dissimilar Metal Weld Joint for Connecting Pipe-Nozzle to Safe-End in Generation III Nuclear Power Plant

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Abstract

The dissimilar metal weld joint (DMWJ) in the primary water system of pressurized water reactors (PWRs) has proven to be a vulnerable component owing to its proneness to various types of flaws. Thus, maintaining the integrity of such joints in the presence of defects is of great importance to the design and safe management of nuclear power plants (NPPs). For reliable integrity analysis of DMWJs, it is essential to understand the microscopic characteristics in all regions of the joint. In this work, optical microscopy (OM), transmission electron microscopy (TEM), scanning electron microscopy (SEM), durometer, atomic force microscopy (AFM), magnetic force microscopy (MFM), and scanning Kelvin probe force microscopy (SKPFM) were utilized to investigate the microstructure, microhardness, distribution of main elements, grain boundary characteristics, and residual strain in the A508/52M/316L DMWJ used for connecting the pipe safe-end and the nozzle of the reactor pressure vessel in PWRs. A comparative analysis of the microstructure and properties along the radial direction of the DMWJ was also performed. The results showed that no region differed dramatically from other parts of the weldment in terms of microstructure and microhardness. A layer of fine grains resulting from unmelted filler metal was found in the backing weld part of the joint. The residual strain in the heat-affected zone (HAZ) of 316L was higher than that in other regions.

Meanwhile, drastic variations in microstructure, chemical composition distribution, and grain boundary character distribution (GBCD) were observed in both the 316L/52Mw and 52Mb/A508 interface regions. TEM and MFM analyses showed that numerous chromium- and molybdenum-rich precipitate particles were distributed both along grain boundaries and inside grains in the 316L base metal, which were identified as precipitates with complex elemental composition rather than normal string delta ferrite in 316L austenitic stainless steel. SKPFM results indicated that these precipitates were more prone to corrosion than the base metal. Therefore, further investigation into the cause of deformation and its impacts on corrosion resistance, particularly the stress corrosion cracking (SCC) sensitivity of the precipitates, needs to be carried out.

Keywords: dissimilar metal welding, microstructure, microhardness, chemical composition distribution, grain boundary character, residual strain

Introduction

Dissimilar metal weld joints (DMWJs) in PWR nuclear power plants are susceptible components in the primary coolant system due to their unique material characteristics. The main damage and failure modes include stress corrosion cracking (SCC), fatigue cracking, corrosion cracking, hot cracking, and lack-of-fusion defects. The formation of these defects is associated with factors such as mechanical property variations along the weld joint, differences in thermal expansion coefficients and alloy elements between materials, welding residual stress, carbon migration, and service environment [1-10]. Over the past several decades, numerous cracking and leakage incidents in reactor primary coolant pipe safe-ends have been reported at nuclear power plants worldwide [1,2], making the integrity of these joints a critical prerequisite for safe NPP operation. Previous studies [3-7] have shown that alloy element diffusion near the fusion line, differences in crystal structure between weld and base metals, and welding heat flow lead to gradient changes in microstructure, microhardness, grain boundary type, and residual stress distribution in this region, which subsequently affect the mechanical properties and corrosion resistance of the entire weld joint, particularly its SCC susceptibility. The narrow region adjacent to the fusion line exhibits higher SCC susceptibility than both the base metal and weld metal. Therefore, detailed micro-characterization of these DMWJs, especially at the fusion line, is essential for understanding their safety and stability in PWR primary coolant environments.

The selected reactor type for this study uses 52M nickel-based alloy as the filler metal for safe-end welding in the primary system. This filler metal belongs to the 690 series of nickel-based alloys and offers superior SCC resistance compared to 600 series alloys due to its higher Cr content [8-10]. Extensive research has been conducted on this type of DMWJ both domestically and internationally. Wang et al. [11] characterized the microstructure, microhardness, and compositional variations at the fusion line interface of A508/52M/316L DMWJs, finding significant differences in fracture toughness and crack propagation behavior at the weld/base metal interface compared to other joint regions. Choi et al. [12] investigated the effects of aging treatment on microhardness, composition, and intergranular Cr carbide precipitation at the fusion line interface of low-alloy steel/nickel-based filler metal DMWJs using TEM and three-dimensional atom probe tomography (3D APT). Hou et al. [13] studied the influence of transition zone microstructure on SCC resistance in Inconel 182/A533B low-alloy steel DMWJs. However, these studies lacked systematic investigation of all microstructural features across the entire joint.

Achieving localization of nuclear power plant design, construction, and raw materials is crucial for China' s nuclear energy industry to break free from international constraints and promote sustainable energy development. Concurrently, establishing a technically sound methodology for integrity design and assessment of metal weld joints is vital for the safe operation of Generation III PWRs, which mainly include AP1000 and EPR reactor types. China has developed CAP1000 and CAP1400 reactors based on the AP1000 design, which was developed by Westinghouse and is used in most Chinese NPPs under construction or planned. Only with reliable safety assessment can scientific decisions be made regarding continued service, repair, or replacement of components when joint defects or cracks are detected. However, previous micro-characterizations of these joints have primarily focused on the weld width direction, neglecting potential structural and property variations along the joint thickness direction. This limitation hinders comprehensive understanding of the overall microstructure, composition, and properties of the weldment, potentially creating discrepancies between research results on mechanical, electrochemical, and SCC behavior and actual performance.

Therefore, this work analyzed the microstructure, microhardness, distribution of main alloy elements, grain boundary types, and residual strain in the A508/52M/316L DMWJ of an advanced PWR reactor pressure vessel safe-end, and compared the microstructure and properties at different thicknesses of the entire weld joint to provide a basis for predicting and evaluating welding process and in-service performance of the joint.

1 Experimental Methods

The research object in this work was a domestically produced DMWJ for a PWR primary coolant pipe safe-end. The reactor nozzle and safe-end were made of low-alloy steel A508 and 316L stainless steel (SS), respectively. The welding procedure consisted of: (1) buttering approximately 27 mm of nickel-based alloy 52M onto the A508 nozzle as a buffer layer, followed by stress relief heat treatment; (2) subsequently welding the nozzle and safe-end together using nickel-based alloy 52M filler metal through multi-pass welding. The buttering layer and butt weld used the same filler metal type, but due to different batch numbers and welding processes, the deposited metals had slight differences in

Cr and Ni content, designated as 52Mb and 52Mw for distinction. To prevent direct contact between the corrosion-prone A508 and the PWR primary coolant environment, a nickel-based cladding approximately 7 mm thick was typically deposited on the inner wall surface near the port of A508, while the remaining inner wall was clad with 308 stainless steel. The main alloy element contents of each joint component are listed in . The macroscopic morphology of the joint cross-section is shown in [Figure 1: see original paper]a. Due to the large sample size, it was cut into 8 pieces along the wall thickness direction, as illustrated in [Figure 1: see original paper]b, where position 1 corresponds to the inner pipe wall and position 8 corresponds to the outer pipe wall.

Different etching solutions were selected for different regions. Two methods were used for etching the weld and cladding layers: (1) a solution of 16 g $FeCl_3$ + 80 mL HCl + 2 mL HNO_3 + 11 mL H_2O, and (2) electrolytic etching in 10% chromic acid aqueous solution (mass fraction) at 3 V for 20 s. For 316L SS, etching was performed using either a solution of 5 g $CuSO_4$ + 25 mL HCl + 25 mL H_2O or electrolytic etching in 10% ammonium persulfate aqueous solution for 2-3 min at 4 V. A508 was etched with 4% nital. The microstructure of the entire joint was examined using an Observer.Z1m optical microscope (OM). A JEM-2100F field-emission transmission electron microscope (TEM) was used for local region analysis. TEM samples were prepared as 3 mm diameter discs, mechanically ground to 50 m, dimpled, and ion-milled. Microhardness measurements were conducted on an MHVD-1000AP microhardness tester with a load of 200 g and dwell time of 15 s. To reflect microhardness trends near the fusion line, the spacing between adjacent hardness points was 150 m near the fusion line and 500 m in regions more than 3 mm away from the fusion line. The A508 HAZ and fusion line interface represent the most SCC-susceptible part of the entire weldment [11], requiring measurements at smaller step sizes. Therefore, this region was measured with a load of 50 g, dwell time of 15 s, and spacing of 50 m between adjacent points. To determine the diffusion of main alloy elements across the fusion line interface, an XL30-FEG environmental scanning electron microscope (ESEM) equipped with energy-dispersive spectroscopy (EDS) was used with a scanning step of 0.42 m and dwell time of 2.0 s per point. Magnetic force microscopy (MFM) and scanning Kelvin probe force microscopy (SKPFM) tests were performed on a MultiMode IIID scanning probe microscopy system.

The EBSD attachment of the ESEM was used to measure grain boundary types and residual strain distribution near the fusion line between the weld and base metal. Samples were pre-ground with 2000-grit sandpaper, polished, and ultrasonically cleaned in alcohol for 30 min before measurement. An accelerating voltage of 15 kV was used, and data were analyzed using OIM software. The threshold for calculating misorientation between adjacent points was set at 5°, meaning points with misorientation below this value were considered to be within the same grain.

2 Results and Discussion

2.1 Microstructure

2.1.1 52M Nickel-Based Alloy Weld The microstructure of the 52M nickel-based alloy weld is shown in [Figure 2: see original paper]. As seen in [Figure 2: see original paper]a, the butt weld microstructure consists of fully austenitic columnar grains that tend to grow vertically. The buttering layer 52Mb differs from the butt weld 52Mw in that its grain structure grows horizontally, as shown in [Figure 2: see original paper]b. The grain growth direction in the weld is determined by the cooling direction of the molten pool during welding, with obvious epitaxial growth between successive weld passes.

Observation of the entire weld pool microstructure revealed relatively uniform distribution along the weld depth direction without significant variation. [Figure 3: see original paper] shows the macroscopic microstructure of the weld. [Figure 3: see original paper]a presents the macroscopic morphology of the sample at the middle position 4, clearly showing the vertical grain growth in the butt weld and horizontal grain growth in the buttering layer. In the middle region of the weld joint, a layer of fine equiaxed grains appeared in the backing weld portion (located at position 3, approximately 45 mm from the inner weld wall), as indicated by the arrow in [Figure 3: see original paper]b. Researchers [14] have suggested that this equiaxed grain layer is formed by the accumulation of unmelted filler metal.

2.1.2 316L Stainless Steel Base Metal The microstructure of the 316L SS base metal is shown in [Figure 5: see original paper]a, consisting of relatively uniform equiaxed annealed austenite grains with some twin structures. Numerous granular precipitates were observed both in the HAZ near the weld and in the base metal far from the weld, with some large precipitate particles forming clusters and distributed sporadically in the matrix, as shown in [Figure 5: see original paper]b. Two types of precipitates were identified: (1) larger particles several micrometers in size with irregular shapes, and (2) smaller particles only a few hundred nanometers in size, present in significantly lower quantities than the large particles. EDS analysis revealed that the Cr and Mo contents in the large and small precipitates were 27.23% and 21.91%, and 11.02% and 19.44%, respectively—both higher than those in the 316L matrix, with correspondingly lower Fe and Ni contents.

The morphology of large precipitates after etching with $CuSO_4$ and HCl solution is shown in [Figure 6: see original paper]a, with the corresponding MFM image in [Figure 6: see original paper]b. The precipitate regions appear lighter than the matrix, indicating these precipitates are non-magnetic. The normal microstructure of 316L SS consists of austenite grains with δ-ferrite precipitated along grain boundaries. Since ferrite is ferromagnetic at room temperature and would appear darker than non-magnetic austenite grains in MFM images, these precipitates can be identified as not being normal δ-ferrite in 316L austenitic stainless steel. SKPFM results in [Figure 6: see original paper]c show that the Volta potential of the large precipitates is lower than that of the surrounding austenite matrix. The potential profile along the dashed line in [Figure 6: see original paper]d indicates a potential difference of up to 30 mV between the precipitates and matrix, suggesting that these precipitates act as anodes in a micro-galvanic cell with the matrix under corrosive electrolyte conditions and are therefore more susceptible to corrosion.

Studies [15–18] have shown that when austenitic stainless steel is tempered at 600–1050 °C, the ferrite undergoes a eutectoid reaction ($\delta \rightarrow \sigma + \gamma$_{new}) to form tetragonal σ-phase, which can proceed until all ferrite is transformed. The chemical composition of σ-phase varies with heat treatment time and temperature but is generally richer in Cr and Mo and poorer in Ni compared to δ-ferrite. In austenitic stainless steels, σ-phase is a common intermetallic phase, and ferrite-stabilizing elements such as Cr, Mo, and Si promote its formation. Research [16] has also shown that σ-phase has lower Volta potential than the γ-phase. Based on its composition, morphology, SKPFM data, and previous research findings, the large precipitate particles are likely σ-phase. As a hard and brittle intermetallic phase, σ-phase adversely affects fracture toughness. Considering that 316L is in direct contact with the harsh corrosive environment of PWR primary coolant, the extensive distribution of σ-phase in 316L SS will inevitably impact the corrosion resistance of the safe-end.

SEM and TEM images of the small precipitate particles are shown in [Figure 7: see original paper]a and b. Their particle size is significantly smaller than that of the large precipitates. EDS analysis results in [Figure 7: see original paper]c show no obvious Cr- or Mo-depleted zones around the small precipitates. Electron diffraction pattern analysis revealed that these precipitates have a bcc structure. Studies [18–22] have shown that Mo-containing austenitic stainless steels, ferritic stainless steels, and duplex steels may form a bcc-structured -phase when tempered at 600–900 °C, with space group I$\bar{4}$3m and lattice constant of 0.892 nm. -phase has a wide stoichiometric range, with a typical composition of $Fe_{36}Cr_{12}Mo_{10}$, though it usually contains other alloying elements such as Ti and Ni under actual conditions. Based on electron diffraction analysis and elemental composition, the small spherical precipitates are identified as -phase with a molecular composition of $(Fe, Ni)_{18}Cr_6Mo_5$. -phase is also a brittle phase that adversely affects both the corrosion resistance and mechanical properties of the safe-end.

The OM image of the HAZ near the 316L/52Mw fusion line is shown in [Figure 8: see original paper]. The grains adjacent to the fusion line show some coarsening, as indicated by arrows, and twin structures are reduced compared to the base metal. Otherwise, the microstructure shows no obvious difference from the base metal, and precipitate distribution and morphology remain unchanged despite welding heat input [23].

2.1.3 A508 Low-Alloy Steel Base Metal The microstructure of A508 low-alloy steel is shown in [Figure 9: see original paper]. The matrix consists of tempered bainite, containing a small amount of granular bainite within the feathery upper bainite matrix. Both low-carbon regions ([Figure 9: see original paper]a) and high-carbon regions ([Figure 9: see original paper]b) exist, with significantly increased granular bainite content in the high-carbon regions.

The microstructure of the A508 HAZ is shown in [Figure 10: see original paper]. The HAZ width is approximately 2.0–2.5 mm. [Figure 10: see original paper]a illustrates the microstructural evolution from the 52Mb/A508 fusion line interface toward the A508 base metal, with the solid line marking the boundary between the HAZ and base metal. The HAZ can be subdivided into four regions (I–IV) separated by dashed lines, corresponding to the fusion zone, coarse-grain zone, fine-grain zone, and tempered zone. Region I is bright ferrite formed by decarburization adjacent to the fusion line. Region II corresponds to temperatures above the Ac_3 transformation temperature, where original base metal grains grew significantly, forming quenched martensite upon cooling. Depending on cooling rate and welding heat input, bainite may also appear, creating a mixed martensite-bainite structure, as indicated by arrows in [Figure 10: see original paper]b. Region III corresponds to the Ac_1–Ac_3 temperature range where ferrite dissolved minimally into austenite while pearlite, bainite, and sorbite fully transformed into austenite. During cooling, the austenite transformed into fine martensite while the original ferrite remained unchanged but grew somewhat, forming a mixed ferrite-martensite structure ([Figure 10: see original paper]c). Region IV exhibits obvious tempered microstructural characteristics ([Figure 10: see original paper]d).

The HAZ microstructure on the A508 side adjacent to the inner wall cladding interface is similar to that near the buttering layer fusion line, though some regions show coarse lath martensite rather than bright ferrite immediately adjacent to the fusion line ([Figure 11: see original paper]). The regions farther from the fusion line show microstructures consistent with the fine-grain and tempered zones of the HAZ.

2.1.4 316L/52Mw Fusion Line Interface The microstructure near the 316L/52Mw fusion line interface is shown in [Figure 12: see original paper]. A special region composed of fine columnar crystals appears between the 316L SS and 52Mw weld. Studies [24] have shown that heat flow and element diffusion during welding lead to the formation of complex microstructures at this interface that differ from both base metal and weld metal. Due to differences in melting temperatures between base and weld metals, an unmixed zone (UZ) commonly appears near the fusion line interface. This UZ is not continuously distributed along the entire fusion line, and its width varies at different thicknesses of the weld. The microstructure on the 52Mw weld side consists of transition-oriented austenitic columnar grains that grow epitaxially from the 316L SS grains.

2.1.5 52Mb/A508 Fusion Line Interface The microstructure near the 52Mb/A508 fusion line is shown in [Figure 13: see original paper]. A bright transition zone (TZ) caused by compositional transition exists between 52Mb and A508 ([Figure 13: see original paper]a), indicating higher etching resistance in this zone compared to the adjacent base metal and weld under this etching system. Type-I and Type-II grain boundaries characteristic of dissimilar metal welding were also observed in some regions ([Figure 13: see original paper]b). Type-II boundaries are parallel to the fusion line interface, while Type-I boundaries connect the fusion line interface to Type-II boundaries. Studies [25-28] have shown that such boundaries form due to differences in crystal structure between weld and base metals and sharp compositional changes across the fusion line. Type-I boundaries grow epitaxially from base metal grain boundaries [25,29], and both Type-I and Type-II boundaries are high-angle grain boundaries with higher SCC susceptibility than the fusion line, readily becoming paths for crack initiation and propagation [13,30,31]. Research on 182-LAS DMWJs [32] has shown that this region exhibits the highest residual strain values. The presence of these special boundaries is a primary reason for the SCC susceptibility of DMWJs. However, observation of the interface between the inner wall cladding and A508 ([Figure 11: see original paper]) revealed no Type-I or Type-II boundaries, only a bright transition zone.

2.2 Microhardness

The microhardness distribution across the weldment is shown in [Figure 14: see original paper]. The maximum microhardness appears in the A508 HAZ at approximately 320 HV. Three regions of microhardness jumps correspond to the 316L/52Mw, 52Mw/52Mb, and 52Mb/A508 interfaces, with the most dramatic hardness change occurring at the 52Mb/A508 interface. The hardness of 316L SS ranges from 165 to 185 HV. From inner to outer wall, the HAZ hardness first increases then decreases, with maximum values exceeding 255 HV at approximately 3 mm from the fusion line. Austenitic stainless steel has a high coefficient of thermal expansion and low thermal conductivity, and its thermal property differences from 52M nickel-based alloy can lead to significant residual stress and deformation in the welding HAZ, causing increased microhardness and crack susceptibility.

The hardness variation in the weld region is relatively gradual, with the butt weld showing slightly higher overall hardness than the buttering layer, likely due to post-weld stress relief treatment and unconstrained welding conditions. The weld metal adjacent to the 316L/52Mw and 52Mb/A508 interfaces represents the lowest hardness region in the entire weld, possibly caused by elemental migration near the interface reducing Cr and Ni content and weakening solid solution strengthening effects [33].

The microhardness distribution in the A508 HAZ is shown in [Figure 14: see original paper]d. From the fusion line toward the base metal, the A508 microhardness increases sharply then gradually decreases, with the highest hardness occurring approximately 600 m from the fusion line, corresponding to the mixed martensite-bainite structure region in the coarse-grain zone ([Figure 10: see original paper]b). In the transition region from HAZ to base metal, the microhardness is lower than in both the HAZ and base metal, creating a local softening zone. This softening results from the welding heat flow tempering this region at high temperature, promoting migration of C and other elements and softening the tempered bainite structure of the base metal [11]. However, the microhardness of the A508 HAZ adjacent to the inner wall cladding was lower than that near the 52Mb buttering layer, possibly due to different welding processes for the cladding and buttering layers.

Comparison of microhardness data from inner wall, outer wall, and middle regions shows similar distributions along the weld thickness direction. The average microhardness of the backing weld sample at position 5 was higher than that of samples at inner wall position 1 and outer wall position 8, likely because regions closer to the backing weld experience more thermal cycles, resulting in greater hardness effects.

2.3 Distribution of Main Alloy Elements Near Fusion Line

2.3.1 316L/52Mw Fusion Line The distribution of main alloy elements Fe, Cr, Ni, and Mo near the 316L/52Mw fusion line at the middle weld position 4 is shown in [Figure 15: see original paper], with [Figure 15: see original paper]a and b corresponding to regions without and with an unmixed zone (UZ), respectively. The results show significant concentration changes near the interface caused by element migration driven by concentration gradients between base metal and weld metal during welding. However, the elemental mass fractions in the UZ differ from those in both base metal and filler metal, likely due to melting temperature differences causing partially melted base metal to be inadequately diluted [34]. The element composition in the transition zone (TZ) on the 52Mw side shows gradient variation, with TZ width being larger when UZ is present. Studies [28,29] have shown that elemental migration at the interface promotes formation of complex microstructures in the 316L/52Mw interface region, affecting the mechanical properties and SCC resistance of the DMWJ. Comparison of inner wall, middle, and outer wall data shows similar compositional transition trends near the 316L/51Mw interface at different thickness positions.

2.3.2 52Mb/A508 Fusion Line The distribution of main alloy elements Fe, Cr, and Ni near the 52Mb/A508 fusion line at middle position 4 is shown in [Figure 16: see original paper], with [Figure 16: see original paper]a and b corresponding to regions without and with Type-I and Type-II boundaries, respectively. As shown in [Figure 16: see original paper]a, Fe content gradually increases while Ni and Cr contents gradually decrease in the TZ on the 52Mb side; compositional changes on the A508 side are not obvious. In regions with Type-I and Type-II boundaries, Fe content decreases slightly while Cr and

Ni contents increase somewhat between the Type-II boundary and fusion line. The element distribution near the fusion line interface between the inner wall cladding and A508 is similar to that in regions without Type-I and Type-II boundaries. Likewise, comparison of inner wall, middle, and outer wall data shows similar compositional transition trends near the 52Mb/A508 interface at different thickness positions.

2.4 Grain Boundary Character Distribution Near Fusion Line

2.4.1 316L/52Mw Fusion Line EBSD images of grain boundary character distribution on both sides of the 316L/52Mw fusion line are shown in [Figure 17: see original paper]. The grain boundary character distribution (GBCD) is presented in [Figure 18: see original paper]. The results indicate that 316L SS grain boundaries are dominated by high-angle grain boundaries and coincidence site lattice (CSL) boundaries, with low-angle grain boundaries comprising a small proportion. From the fusion line toward the base metal, the proportion of CSL boundaries gradually increases, exceeding that of high-angle grain boundaries at 7 mm from the fusion line, with most twin boundaries identified as CSL boundaries. Statistical results show that CSL boundaries have excellent SCC resistance [35-38]. The proportion of CSL boundaries in the 316L HAZ is significantly lower than in the base metal, indicating that the welding process adversely affects SCC resistance. The 52Mw weld region is dominated by high-angle grain boundaries with relatively low contents of low-angle and CSL boundaries. Numerous small "grains" appear in some weld regions ([Figure 17: see original paper]a6, b7, c5), with most of their boundaries identified as CSL boundaries. The distribution trend is consistent along the weld thickness direction, with similar proportions of the three boundary types.

2.4.2 52Mb/A508 Fusion Line The grain boundary character distribution on both sides of the 52Mb/A508 fusion line is represented by the outer wall (position 7) in [Figure 19: see original paper]. The GBCD across the 52Mb/A508 fusion line at inner and outer walls is shown in [Figure 20: see original paper]. Both sides of this fusion line are dominated by high-angle grain boundaries with small amounts of low-angle grain boundaries and almost no CSL boundaries. Grain size first decreases then increases from the fusion line toward the A508 matrix. Comparison of inner and outer wall data shows consistent trends in grain boundary character distribution near this interface along the weld thickness direction.

2.5 Residual Strain Distribution Near Fusion Line

Kernel average misorientation (KAM) values can quantitatively reflect the degree of strain hardening in materials. The residual strain distribution near the fusion line interface of the inner wall (position 1) is shown in [Figure 21: see original paper], with KAM value statistics presented in [Figure 22: see original paper]. Near the 316L/52Mw fusion line, KAM values increase then decrease from the fusion line toward the 316L base metal. This occurs because welding heat flow creates localized high temperatures and rapid heating/cooling rates near the fusion line, generating maximum residual strain. As distance from the fusion line increases, peak temperature and heating/cooling rates decrease, gradually reducing residual strain [27,39]. Residual strain values in the backing weld region are higher than in the inner and outer walls, likely due to more thermal cycles experienced at the backing weld position [40]. Weld KAM values are lower than those in the base metal HAZ, but inner and outer wall KAM values first decrease, then increase, then decrease again from the fusion line toward the weld metal. The backing weld does not show this trend, possibly because the scanned region at 7 mm from the fusion line in the backing weld is near the 52Mw/52Mb fusion line, while the corresponding regions in the inner and outer walls are located in the middle of the weld pool.

Near the 52Mb/A508 fusion line, KAM values show smaller fluctuations and much lower average values than near the 316L/52Mw fusion line, because the buttering layer welding was performed without constraint, resulting in less welding deformation. Inner and outer wall data show consistent trends, with slightly higher KAM values at the outer wall.

Conclusions

1. Both the butt weld and buttering layer consist of coarse columnar grain structures. A transition zone (TZ) of certain width exists near the fusion line. A discontinuous unmixed zone (UZ) composed of small columnar crystals was observed at the 316L/52Mw interface, while discontinuous Type-I and Type-II grain boundaries appeared on the A508 side. Significant compositional fluctuations occurred in the UZ, Type-I, and Type-II boundary regions.

2. The A508 base metal of this DMWJ has a tempered bainite microstructure, with a complex HAZ microstructure including fusion zone, coarse-grain zone, fine-grain zone, and tempered zone. The 316L base metal consists of austenite with numerous granular precipitates and some twin structures, showing no obvious HAZ microstructure.

3. The maximum microhardness of the DMWJ appears in the A508 HAZ. Three microhardness jumps occur near the fusion lines. Both the 316L SS and A508 HAZ show microhardness that first increases then decreases with distance from the fusion line.

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Note: Figure translations are in progress. See original paper for figures.