

Coexistence of two distinct fatigue failure mechanisms in Super304H welded joint at elevated temperatures

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ABSTRACT

Super304H steel is a promising candidate for boiler tube materials in thermal power plants operating under ultra-supercritical conditions, and the fatigue properties of its welded joint play a key role in the structural reliability of the thermal power plants. Herein, we report that there is a transition in the fatigue failure location (i.e., fatigue failure mechanism) from the base metal at a high strain amplitude ($\geq \sim 0.4\%$) to the weld metal at a low strain amplitude ($\leq \sim 0.3\%$) at a constant temperature in the operating range of 500–700 °C, and this leads to a significant reduction in the fatigue resistance of the welded joint as compared to that of the base metal. We found that a longer fatigue test time at a low strain amplitude induces a thermal aging effect that promotes different microstructural evolutions in the base metal and weld metal, deepening material inhomogeneity in the welded joint, and thereby triggering strain localization in the weld metal. With increasing fatigue test time (i.e., thermal aging time), Cr₂₃C₆ carbides precipitated in the interdendritic region of the weld metal and at the austenitic grain boundary of the base metal, and Nb (C, N) phases precipitated in the interior of austenitic grain in the base metal, which increased local hardness, whereas there was no significant microstructural change in the dendrite core of the weld metal, retaining its initial hardness. The intensified local hardness inhomogeneity caused the softest zone in the welded joint, wherein strain localization occurred, serving as a crack nucleation site and propagation path, to shift from the base metal to the weld metal (dendrite core). This induced a shift in the fatigue failure location from the base metal to the weld metal with decreasing strain amplitude.

1. Introduction

Owing to its superior mechanical properties and corrosion resistance, Super304H steel is a promising candidate for heat exchanger tube materials in the boiler system of thermal power plants operating under ultra-supercritical (USC) conditions of 30 MPa and 600 °C [1–4]. Heat exchanger tubes are used in various complex shapes to meet design requirements and improve thermal efficiency, and hence, the welding of Super304H steel to other components such as the boiler head and other boiler tubes is imperative [5–7]. Welded parts, however, are known to be the most vulnerable sections, impairing strength, ductility, and creep and corrosion resistance [8–10]. In this regard, the mechanical properties of the Super304H welded joint, in particular its fatigue properties,

are a key factor in ensuring the structural reliability of the boiler system in USC thermal power plants [11,12].

According to our previous study, wherein the low-cycle fatigue properties of Super304H welded joints were investigated in the temperature range of 500–700 °C [13], the welded joint has a fatigue resistance comparable to that of the base metal at high strain amplitudes of $\geq \sim 0.4\%$. However, there is a significant reduction in the fatigue resistance of the welded joint at low strain amplitudes of $\leq \sim 0.3\%$, as compared with that of the base metal. This reduction in fatigue resistance is associated with the shift of the fatigue failure location from the base metal at high strain amplitudes to the weld metal at low strain amplitudes. To the best of our knowledge, the underlying mechanisms causing such a shift in the fatigue failure location have not been

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elucidated yet. Some studies have reported the fatigue failure location shift from the base metal to the weld metal for Cr–Ni–Mo–V steel welded joint [14,15] and dissimilar welded joints in 9% Cr and CrMoV steels [16]. However, these steels are subjected to high-cycle fatigue loading, and the fatigue failure location shifts under a very high-cycle fatigue regime, whereupon cracks originate and grow from the internal defects in the weld metal, eventually leading to failure. Although a recent study reported the fatigue failure location shift in modified 9Cr/CrMoV dissimilar welded joints under low-cycle fatigue loading, the shift occurs from the CrMoV over-tempered zone (heat-affected zone) at high strain amplitudes of $>0.4\%$ to CrMoV base metal at low strain amplitudes of $\leq 0.4\%$, i.e., showing opposite locations of fatigue failure as compared to our study, and the responsible mechanisms have not been properly studied. In general, the fatigue properties of metallic materials at elevated temperatures are affected by oxidation, creep, and creep-fatigue-oxidation interaction [17–20]. A longer fatigue test time at a low strain amplitude increases the possibility of these three factors affecting the fatigue properties and degrading the fatigue resistance. However, our previous study shows that the effect of the three factors on the fatigue properties of Super304H welded joints are insignificant and can be neglected [13]. Another factor that becomes important at low strain amplitudes appears to be associated with the fatigue failure location shift in the Super304H welded joint. It is known that thermal aging of type 310 stainless steel welded joints at $700\text{ }^{\circ}\text{C}$ leads to precipitation of chromium carbides in the interdendritic region of the weld metal, which increases hardness and decreases impact toughness [21]. In addition, the thermal aging of Super304H steel at $650\text{ }^{\circ}\text{C}$ induces chromium carbide precipitation in the austenitic grain boundary and secondary Nb (C, N) phase and Cu-rich phase precipitation in the interior of the austenitic grain, which increases hardness and strength while decreasing ductility and impact toughness [1,22]. Therefore, we speculate that a longer fatigue test time at the low strain amplitude, which allows a longer high-temperature exposure to a test specimen, will have a thermal aging effect that induces different microstructural evolutions in the base and weld metals, promoting the inhomogeneity of the local mechanical properties and consequently affecting the fatigue failure mechanism of the Super304H welded joint.

In this study, we explored the microstructural evolution of the base metal and weld metal of Super304H welded joints during fatigue deformation (i.e., thermal aging) at $600\text{ }^{\circ}\text{C}$ and their effect on the local mechanical properties of both weld regions. We combined microstructural analyses using optical microscopy, scanning electron microscopy (SEM), energy-dispersive X-ray spectroscopy (EDS), and X-ray diffraction (XRD) with local hardness measurements using nanoindentation. The detailed fatigue failure mechanisms and the reason for the fatigue failure location shift with decreasing strain amplitude were explained based on the relationship between the local hardness inhomogeneity and strain localization in the welded joint.

2. Experimental details

2.1. Materials and welding process

The base metal and filler metal used in this study were Super304H tubes with an outer diameter of 42.2 mm and thickness of 6.6 mm and T-304H rods with a diameter of 2.4 mm, respectively. The chemical composition of the base metal was C–0.09, P–0.06, S–0.01, Ni–8.39, Cr–17.8, Cu–2.87, Nb–0.42, N–0.2, Mn–0.76, Si–0.21, and Fe–balance (wt%) and that of the filler metal was C–0.13, P–0.03, S–0.03, Ni–16.0, Cr–18.0, Cu–3.0, Mo–0.9, Nb–0.6, N–0.15, Mn–3.2, Si–0.3, and Fe–balance (wt%). The Super304H tubes were cut into smaller tubes of lengths 50 mm and 70 mm and then welded using a total of five passes of direct current straight polarity manual gas tungsten arc welding with a 60° V-shaped groove and 3 mm root gap. During the welding process, the formation of weld defects was carefully checked through visual inspection at the end of each welding pass. After the

welding process, internal defects were further examined by conducting radiographic and penetrant testing. The details of the welding process are given in our previous study [13].

2.2. Microstructural characterization

Cuboidal specimens with dimensions of 26 mm (length) $\times$ 6 mm (width) $\times$ 2 mm (thickness), including the base metal, heat-affected zone, and weld metal, were machined from the Super304H welded tubes, and some of them were thermally annealed at $600\text{ }^{\circ}\text{C}$ for 2 h and 8 h to examine the effect of thermal aging on microstructural evolutions in the welded joint. The machined specimens were hot-mounted with epoxy resin, and the surface of the specimens was wet-grinded with SiC papers from #320 to #4000 and then polished with 9, 6, 1, and $0.04\text{ }\mu\text{m}$ diamond pastes, in that order. Next, the surfaces of the specimens were electrochemically etched at an applied voltage of 6 V for 20 s in 2 wt% oxalic acid solution [13,23]. After the etching process, the specimen surfaces were observed and analyzed using an optical microscope (BX53 M, Olympus, Japan), SEM (S-4800, Hitachi, Japan), and EDS (EMAX ENERGY EX-350 x-act, Horiba, Japan) to characterize the microstructural features of the welded joint and their variation with thermal aging. In addition, XRD analysis of the welded joint with the same specimens was conducted using an X-ray diffractometer (D/Max-2500, Rigaku, Japan) with Cu-K α radiation. The XRD measurements were run for $2\theta = 20^{\circ}$ – 100° , and the lattice parameters were determined using JADE software (version 9.0, Material Data, Inc., USA).

Fatigue failure mechanisms were examined by observing the failure location and longitudinal section of the test specimens fatigue-failed in the base metal (at a strain amplitude of 0.4%) and weld metal (at a strain amplitude of 0.3%) using an optical microscope and SEM; the same polishing and etching processes, mentioned above, were used to prepare the observation specimens.

2.3. Hardness tests

Vickers hardness tests (HM-124, Akashi, Japan) were conducted on the cross-section of the welded joints, including the base metal, heat-affected zone, and weld metal, with the same specimens used for microstructural characterization by following the ASTM E92-82 standard [24]. Indentations were made with a load of 500 gf for 10 s, and the distance between the indentations was kept at 0.3 mm to exclude the effect from nearby indentations.

Nanoindentation tests (Nano indenter XP, MTS, USA) were also conducted on the base metal and weld metal of the welded joint using the same specimens employed for microstructural characterization by following ISO 14577-1 [25] to characterize the local hardness. Indentation arrays (4×15) were set on the base metal and weld metal to make indentations to 200 nm depth, while a distance of 20 μm was kept between indentations to exclude the effect from nearby indentations. After the nanoindentation tests, the indentation locations were carefully examined to determine the correlation between the microstructure and local hardness.

2.4. Low-cycle fatigue tests

Low-cycle fatigue test specimens, with a gauge section of 16 mm length and 6 mm diameter, were machined from Super304H welded tubes along the direction perpendicular to the welding direction [26]. The gauge section of the specimens included the base metal, heat-affected zone, and weld metal, where the weld metal was placed at the center. Fatigue tests were conducted using a hydraulic material testing machine with a capacity of 100 kN (Landmark, MTS, USA). A three-zone resistance-type furnace (653, MTS, USA) was used to control the test temperature within a variation of $\pm 1\text{ }^{\circ}\text{C}$ at steady state. To measure and control the strain during the fatigue test, a high-temperature uniaxial extensometer with a gauge length of 12 mm

(632.53F-14, MTS, USA) was used.

Low-cycle fatigue tests were conducted under a fully reversed strain control mode with a triangular waveform at 500, 600, and 700 °C. The strain rate was fixed at $5 \times 10^{-3} \text{ s}^{-1}$, and total strain amplitudes of 0.25%–1.0% were used. The reproducibility of the test results was confirmed by repeating the test at each strain amplitude.

3. Results and discussion

3.1. Fatigue resistance

The total strain amplitude–fatigue life data of the welded joint at 600 °C are presented in Fig. 1a, and the data of the base metal are inserted for comparison. The fatigue life data of the welded joint at 500 and 700 °C are also given in Table 1. The welded joint shows a general tendency of metallic materials, i.e., the fatigue life decreases with increasing strain amplitude. The fatigue resistance of the welded joint is comparable to that of the base metal at high strain amplitudes of $\geq \sim 0.4\%$, but there is a drastic reduction in the fatigue resistance of the welded joint as compared to that of the base metal as the strain amplitude reduces to less than 0.4% (Fig. 1a). The observation on the fatigue failure location, given in Table 1, indicates and this is associated with the shift of the fatigue failure location from the base metal at high strain amplitudes of $\geq \sim 0.4\%$ (Fig. 1b) to the weld metal at low strain amplitudes of $< \sim 0.4\%$ (Fig. 1c). It should be noted that similar results are observed at 500 and 700 °C, although the transition strain amplitude varies slightly with temperature (Table 1).

As described earlier, the effects of oxidation, creep, and creep-fatigue-oxidation interactions on the fatigue resistance of the welded joint are insignificant [27], and the fatigue failure location shift can be attributed to the microstructural evolution during fatigue deformation. A longer fatigue test time at low strain amplitude implies that the test specimen is exposed to high temperature for a longer time, which enhances the thermal aging effect. The test times required to complete fatigue test at each strain amplitude are summarized in Table 1. The fatigue test time that causes the shift of the fatigue failure location to the weld metal is $\sim 4 \text{ h}$ at 500 °C, less than $\sim 5 \text{ h}$ but more than $\sim 3 \text{ h}$ at 600 °C (not accurately determined because of insufficient data), and $\sim 3 \text{ h}$ at 700 °C, indicating that it decreases with increasing temperature. This temperature dependence indirectly supports the idea that the fatigue failure location shift is associated with the thermal aging effect.

3.2. Microstructure and its evolution during fatigue deformation

To examine the microstructural evolutions in the base metal and weld metal of the welded joint during the fatigue test, we conducted thermal aging experiments at 600 °C for aging times of 2 and 8 h; the two thermal aging times were determined considering that the time required for inducing fatigue failure location shift at 600 °C was less than $\sim 5 \text{ h}$ but more than $\sim 3 \text{ h}$ (Table 1). In this regard, the microstructure of the as-welded material and the material aged for 2 h corresponds to that of the welded joint fatigue-tested at high strain amplitudes ($\geq \sim 0.4\%$), whereas the microstructure of the material aged for 8 h represents that of the welded joint fatigue-tested at low strain amplitudes ($\leq \sim 0.3\%$).

Fig. 2 shows the microstructure of the base metal in the welded joint and its evolution with thermal aging treatment. As shown in Fig. 2a, the as-welded material has an equiaxed homogeneous austenitic grain structure with an average grain size of $\sim 22 \mu\text{m}$, and some precipitates were observed in both the grain interiors and grain boundaries, which were identified as the primary Nb (C, N) phase by EDS analysis (inset in the left panel of Fig. 2a). The right panel of Fig. 2a presents a magnified image of region A in the left panel, where EDS analysis results of the grain interior and grain boundary are given. The similar chemical compositions of the grain interior and boundary indicate that there was no precipitate at the grain boundary other than the Nb (C, N) phase. With thermal aging of 2 h, no change in the microstructure of the grain interior was observed, and the primary Nb (C, N) phase remained stable without growth, whereas new precipitates started to form at some grain boundaries (Fig. 2b). The EDS analysis results for the grain interior and grain boundary in the right panel of Fig. 2b (a magnified image of region B in the left panel) reveal that these new precipitates at the grain boundary had a higher Cr content compared to that of the grain interior and were found to be Cr_{23}C_6 carbides [1,22]. As the thermal aging time increased to 8 h, there was still no change observed in the microstructure of the grain interior and the primary Nb (C, N) phase, but the precipitation of Cr_{23}C_6 carbides had advanced at most grain boundaries along with coarsening of the carbides (Fig. 2c). These results indicate that during the thermal aging of the base metal in the welded joint, Cr_{23}C_6 carbides precipitated at the grain boundaries, which was promoted with increasing thermal aging time, while the primary Nb (C, N) phase and austenitic grain structure remained stable without growth.

The microstructure of the weld metal in the welded joint and its evolution during thermal aging are shown in Fig. 3. The as-welded

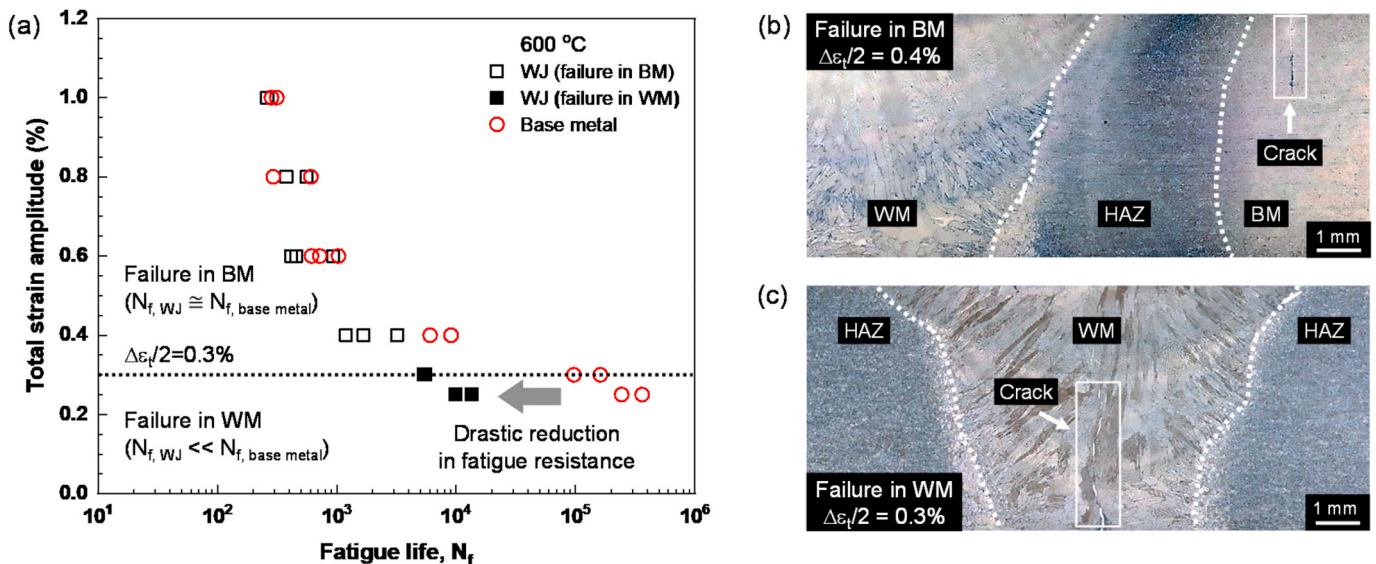
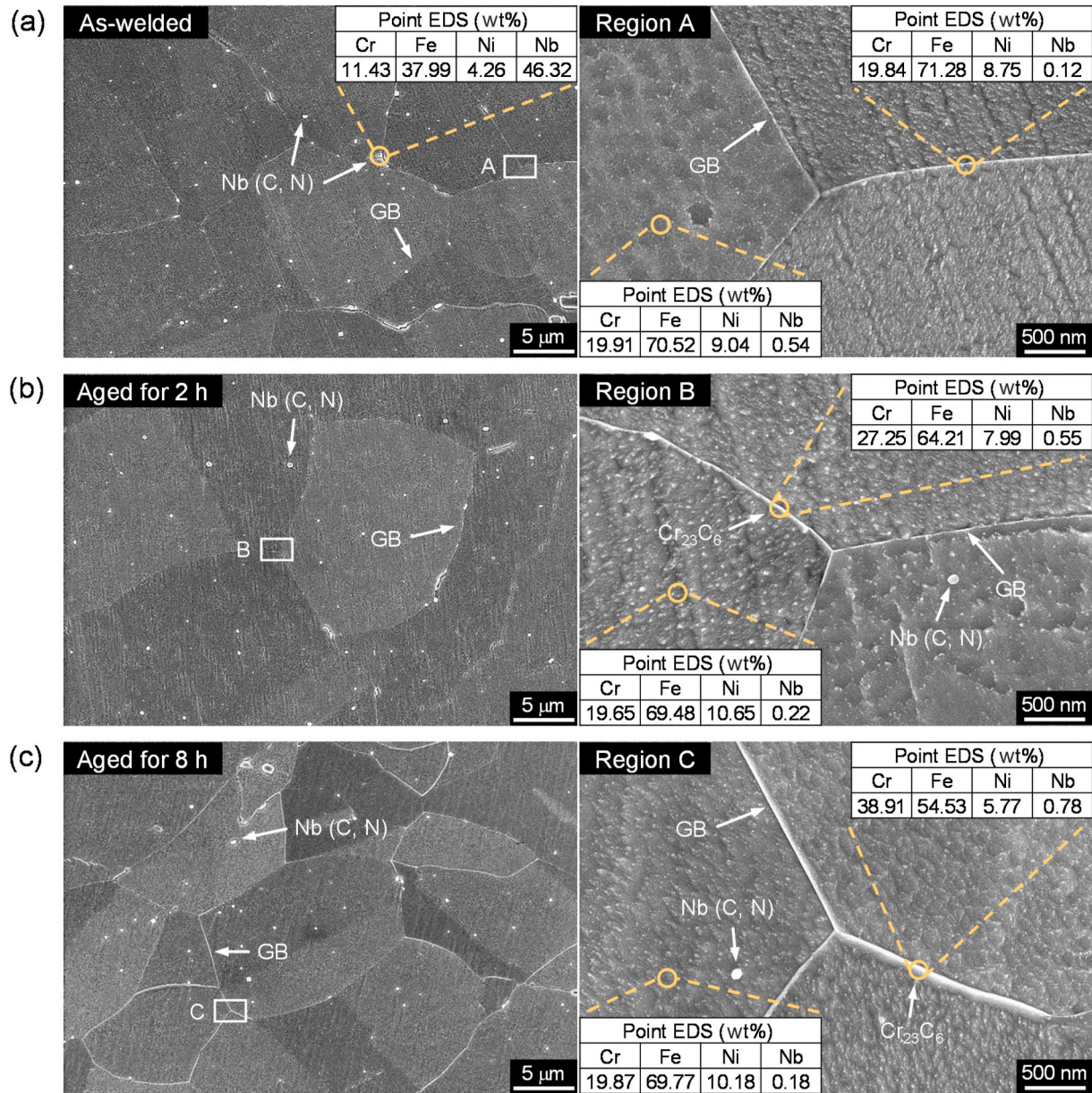


Fig. 1. Low-cycle fatigue characteristics of Super304H welded joint at 600 °C. (a) Total strain amplitude-fatigue life data where the data of the base metal are inserted for comparison. Optical micrographs showing the longitudinal section of the specimens fatigue-failed at (b) $\Delta\epsilon_t/2 = 0.4\%$ and (c) $\Delta\epsilon_t/2 = 0.3\%$. Welded joint (WJ), base metal (BM), weld metal (WM), and heat-affected zone (HAZ).

Table 1

Low-cycle fatigue life of Super304H welded joint as a function of total strain amplitude and temperature.

$\Delta\epsilon_t/2$ (%)	500 °C			600 °C			700 °C		
	N_f (cycle)	Test time (h)	Failure location	N_f (cycle)	Test time (h)	Failure location	N_f (cycle)	Test time (h)	Failure location
0.25	9,628	6.85	WM	13,473	8.99	WM	13,765	9.15	WM
0.25	12,304	8.34	WM	9,943	7.02	WM	3,264	3.31	WM
0.3	4,662	4.61	WM	5,353	5.07	WM	4,578	4.55	BM
0.3	5,111	4.91	WM	5,493	5.16	WM	2,211	2.97	BM
0.4	2,662	3.87	WM	1,678	2.99	BM	1,088	2.47	BM
0.4	2,171	3.43	BM	1,192	2.56	BM	824	2.23	BM
0.4	–	–	–	3,210	4.35	BM	–	–	–
0.6	785	2.55	BM	417	2.06	BM	373	2.00	BM
0.6	812	2.58	BM	934	2.75	BM	323	1.93	BM
0.6	–	–	–	457	2.11	BM	–	–	–
0.8	402	2.21	BM	378	2.17	BM	194	1.84	BM
0.8	637	2.63	BM	559	2.49	BM	192	1.84	BM
1.0	362	2.30	BM	261	2.08	BM	120	1.77	BM

**Fig. 2.** Microstructural evolution in the base metal of the welded joint during thermal aging at 600 °C. (a) As-welded material, (b) aged for 2 h, and (c) aged for 8 h. Insets indicate the results of EDS point analysis. Grain boundary (GB).

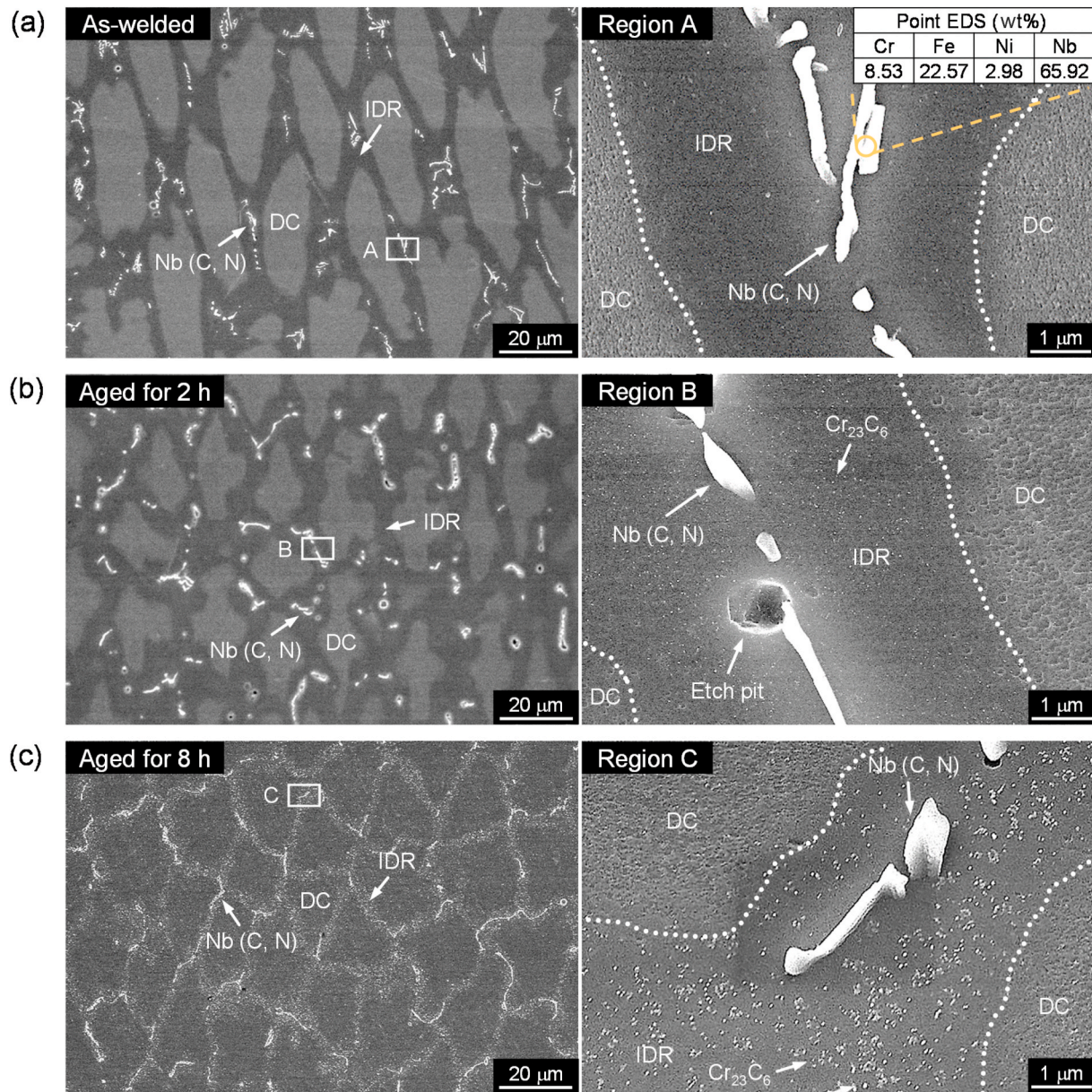


Fig. 3. Microstructural evolution in the weld metal of the welded joint during thermal aging at 600 °C. (a) As-welded material, (b) aged for 2 h, and (c) aged for 8 h. The inset in the right panel of (a) indicates the result of EDS point analysis. Dendrite core (DC) and interdendritic region (IDR).

material showed a full austenite matrix with an average grain size of $\sim 202 \mu\text{m}$, where a dendrite structure developed, forming a subgrain structure (Fig. 3a). Rod-like precipitates observed in the interdendritic regions (between the dendrite cores) were identified as Nb (C, N) phase by EDS analysis, as shown in the inset of the right panel in Fig. 3a (a magnified image of region A in the left panel). After thermal aging for 2 h, no change was observed in the dendrite core and the Nb (C, N) phase (Fig. 3b). However, small precipitates were formed in the interdendritic regions, away from the Nb (C, N) phase (right panel of Fig. 3b). As the thermal aging time increased to 8 h, the number and size of precipitates formed in the interdendritic regions increased significantly, whereas the dendrite core and Nb (C, N) phase remained unchanged. Because the spatial resolution of EDS analysis was not sufficient to identify the small precipitates, XRD was used for analysis. Fig. 4 presents the XRD spectrum of the weld metal in the welded joint, and its variation with increasing thermal aging time; the XRD measurement was conducted over an area of $3 \text{ mm} \times 3 \text{ mm}$. The XRD spectrum of the as-welded material showed strong diffraction peaks of the austenite phase

together with relatively weak diffraction peaks of the Nb (C, N) phase. On thermal aging for 2 h, there was no notable change in the XRD spectrum, indicating that the formation of small precipitates in the interdendritic regions, as shown in Fig. 3, is still insignificant. After thermal aging for 8 h, however, new peaks appeared in the XRD spectrum, which were identified as the Cr_{23}C_6 carbide precipitates formed in the interdendritic regions of the weld metal in the welded joint. In addition, as described earlier, the precipitation of Cr_{23}C_6 carbides was suppressed near the Nb (C, N) phase (Fig. 3b and c) because the formation of this phase consumes a large amount of carbon and creates a surrounding carbon depletion zone [21]. Based on these results, it was concluded that thermal aging of the weld metal of the welded joint induced precipitation of Cr_{23}C_6 carbides only in the interdendritic regions, and the number and size of the carbides increased with increasing thermal aging time. However, there was no change in the dendrite core, thus preserving the initial microstructure.

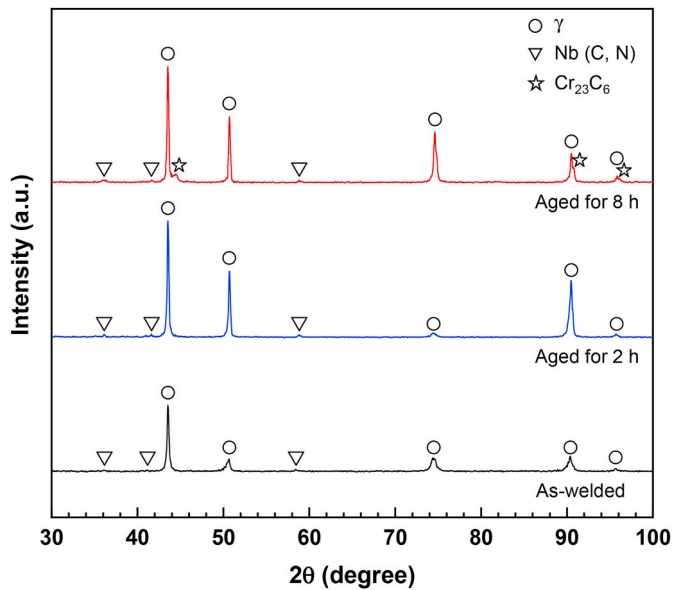


Fig. 4. X-ray diffraction spectrum of the weld metal in the as-welded material, and its variation with aging for 2 and 8 h at 600 °C.

3.3. Local mechanical property and non-uniformity of deformation

To examine the effect of the different thermal aging induced microstructures in the base metal and weld metal on the mechanical properties of the welded joint, we conducted Vickers hardness tests. Fig. 5 shows the Vickers hardness distribution over the three regions of the welded joint, the weld metal, heat-affected zone, and base metal, and its variation with the duration of thermal aging treatment; the indentation locations are marked with a white arrow in the inset. In the as-welded material, the hardness was the highest in the weld metal (~ 212 HV(0.5)), and while it decreased sharply in the heat-affected zone (~ 200 HV(0.5)), an almost similar value was maintained in the base metal (~ 197 HV(0.5)). On thermal aging for 2 h, the hardness increased slightly in all three regions of the welded joint. As the thermal aging time increased to 8 h, a marked increase in hardness was observed in all three regions, and was the highest in the weld metal; ~ 230 HV

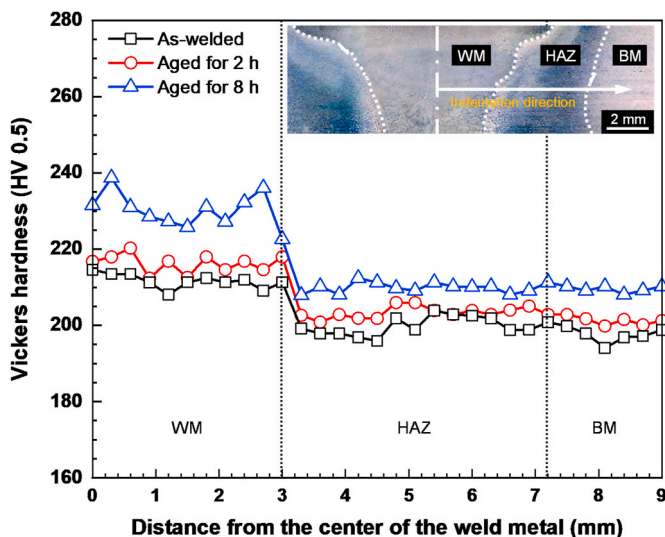


Fig. 5. Vickers hardness distribution in the weld metal, heat-affected zone, and base metal of the as-welded material, and its variation with different aging times of 2 h and 8 h at 600 °C. The inset shows indentation locations in the welded joint.

(0.5) in the weld metal and ~ 210 HV(0.5) in the heat-affected zone and base metal. These results reveal that thermal aging increases hardness in all three regions of the welded joint, and the aging effect is more pronounced in the weld metal, increasing the inhomogeneity of local hardness in the welded joint. In addition, it is noted that even with the thermal aging treatment, the tendency of hardness distribution in the welded joint remains unchanged, that is, the highest in the weld metal, and lower and comparable in the heat-affected zone and base metal.

The increase in the hardness of the welded joint during thermal aging arises from the change in the microstructure of each region. As described earlier, during thermal aging, Cr_{23}C_6 carbides precipitate at the austenitic grain boundary of the base metal and in the interdendritic regions of the weld metal, which induces a precipitation hardening effect and consequently increases the hardness of the base metal and weld metal. The Vickers hardness measurement results show that the base metal is the softest region in the welded joint, irrespective of the duration of thermal aging treatment. Considering the principle of the low-cycle fatigue test, in which a constant strain amplitude is applied to the gauge section of a test specimen, a larger strain would develop in the softest region (base metal) compared to the hard region (weld metal), causing strain localization in the base metal. This implies that the base metal serves as a crack nucleation site and provides a path for crack propagation, consequently leading to failure during fatigue deformation. This is true for the welded joints fatigue-tested at high strain amplitudes of $\geq \sim 0.4\%$, which correspond to the as-welded material and the material aged for 2 h. However, the failure location of the welded joints fatigue-tested at low strain amplitudes of $\leq \sim 0.3\%$ (corresponding to the material aged for 8 h) shifts to the weld metal, indicating that the weld metal is the softest region; thus, Vickers hardness is not a suitable parameter to correlate the microstructure with the local mechanical property. In fact, considering that Vickers indentation is made over an area of a few thousand square micrometers, it is impossible to accurately characterize the hardness of the microstructural features with micrometer size, such as dendrite core and interdendritic region, in the base metal and weld metal.

To measure the local hardness of each microstructure in the base metal and weld metal, we conducted nanoindentation tests. Fig. 6 presents the nanoindentation hardness of the three important microstructural features, i.e., the grain interior of the base metal, and the dendrite

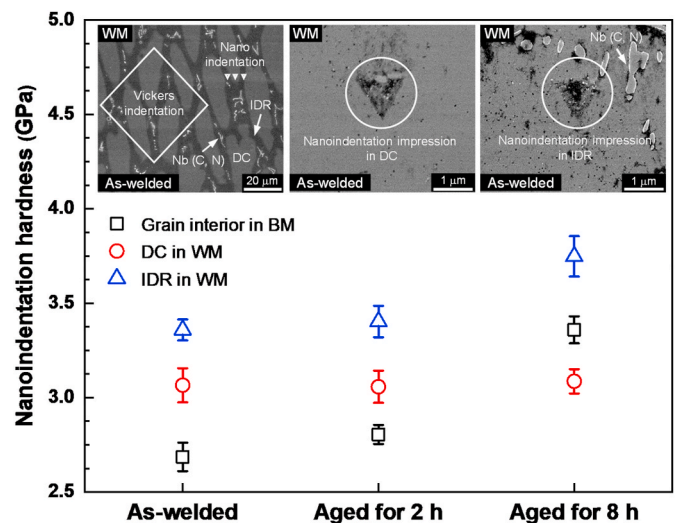


Fig. 6. Comparison of nanoindentation hardness between the dendrite core and interdendritic region in the weld metal and grain interior in the base metal for the as-welded material and the materials aged for 2 h and 8 h at 600 °C. The left panel of the inset compares impression size between Vickers indentation and nanoindentation, and the center and right panels show nanoindentation impressions on the dendrite core and interdendritic region in the weld metal of the as-welded material, respectively.

core and interdendritic region of the weld metal, and their evolution with increasing thermal aging time. The left panel of the inset in Fig. 6 compares the size of the indentation impression between Vickers indentation and nanoindentation; the center and right panels of the inset show the nanoindentation impressions in the dendrite core and interdendritic region of the weld metal, respectively, demonstrating the suitability of nanoindentation to measure the local hardness of each microstructure. The three microstructural features exhibit a distinct difference in the variation of hardness with thermal aging time. In the weld metal, there was no change in the hardness of the dendrite core on thermal aging, as it retained the initial hardness of 3.07 GPa, whereas the hardness in the interdendritic region increased significantly from 3.36 GPa (as-welded material) to 3.77 GPa (material aged for 8 h). This difference in hardness with thermal aging matches well with the microstructural evolutions in both regions, i.e., with thermal aging, Cr_{23}C_6 carbides precipitate in the interdendritic region, but there is no change in the microstructure of the dendrite core. As for the grain

interior of the base metal, the hardness markedly increases from 2.68 GPa (as-welded material) to 3.36 GPa (material aged for 8 h) with thermal aging, even though there is no change in the microstructure (Fig. 2). It is known that the secondary Nb (C, N) phase, which is nanometer-sized and beyond the spatial resolution of the SEM used in this study, precipitates in the grain interior of Super304H steel during thermal aging at 625 and 650 °C [22,28,29] and induces a hardening effect. This explains the increase in the hardness of the grain interior in the base metal with thermal aging. These results imply that the different microstructural evolutions in the base metal and weld metal during fatigue deformation (i.e., thermal aging) intensify the inhomogeneity of the local hardness in the welded joint and deepen the non-uniformity of deformation, leading to strain localization in the softest region. According to the nanoindentation results, the softest region in the welded joint is the base metal (grain interior) for the as-welded material; however, it shifts to the weld metal (dendrite core) for the material aged for 8 h. This change in the softest region in the welded joint, caused by

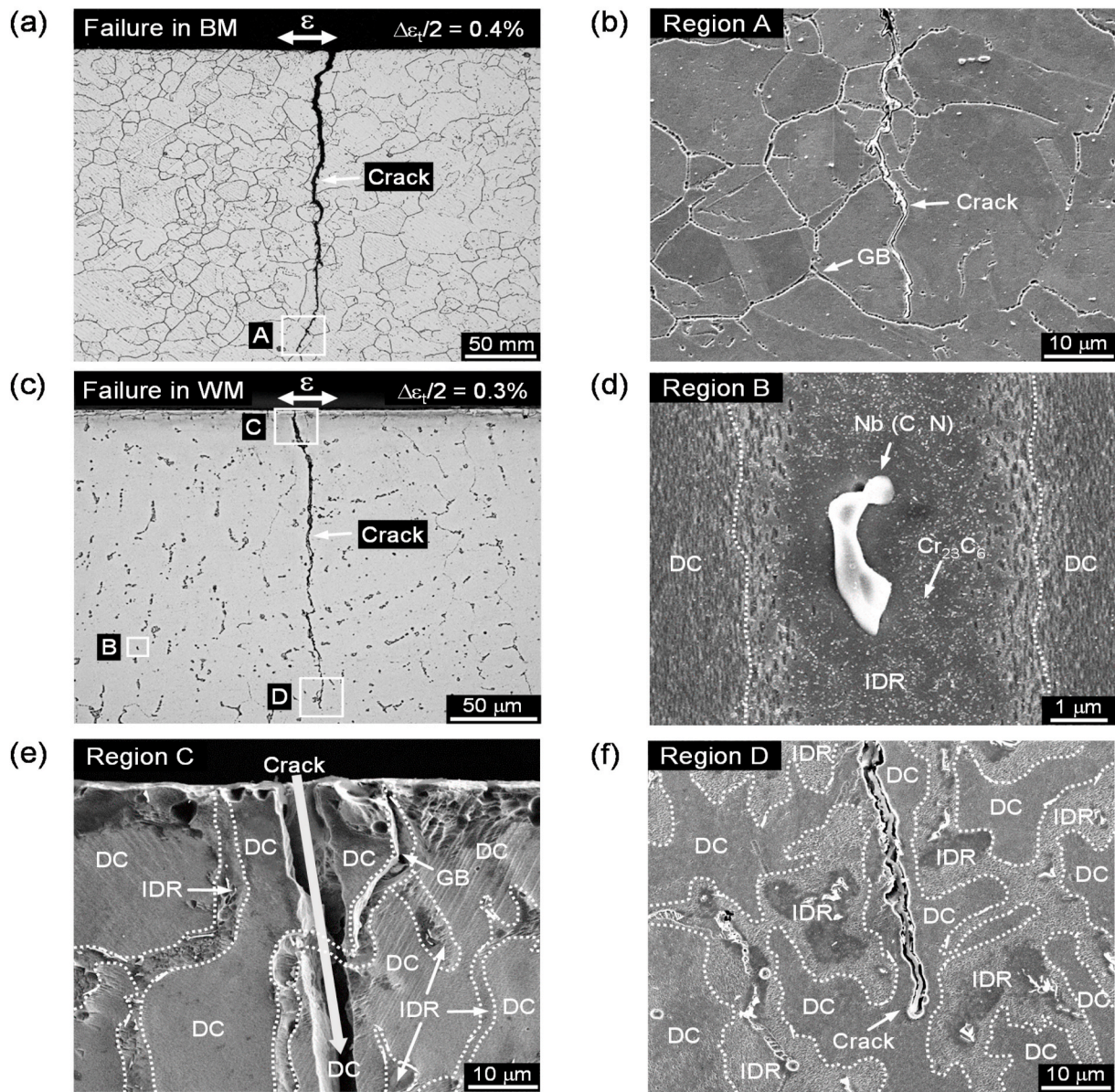


Fig. 7. Optical and SEM micrographs showing the two different fatigue failure mechanisms of the welded joint at 600 °C. (a) Optical micrograph showing the longitudinal section of the fatigue-failed specimen at $\Delta\epsilon_f/2 = 0.4\%$ (failure in the base metal) and (b) magnified SEM image of region A in (a). (c) Optical micrograph showing the longitudinal section of the fatigue-failed specimen at $\Delta\epsilon_f/2 = 0.3\%$ (failure in the weld metal) and magnified SEM images of regions (d) B, (e) C, and (f) D in (c).

the thermal aging effect, appears to be responsible for the shift in the fatigue failure location from the base metal at a high strain amplitude of $\geq \sim 0.4\%$ to the weld metal at a low strain amplitude of $\leq \sim 0.3\%$, as observed in this study.

3.4. Fatigue failure mechanisms

Fig. 7 presents the optical and SEM micrographs of the longitudinal sections of the specimens that fatigue-failed at 600°C . At a high strain amplitude of 0.4% , fatigue failure occurred in the base metal (Fig. 7a and b). A main crack initiated on the specimen surface and propagated to the interior along the direction perpendicular to the loading axis in the transgranular mode. As described earlier, the base metal is the softest region in the welded joint fatigue-tested at a high strain amplitude of $\geq \sim 0.4\%$ (corresponding to the as-welded material and the material aged for 2 h). Thus, the applied strain is concentrated in the base metal, causing strain localization and eventually failure. As Cr_{23}C_6 carbides precipitate at the grain boundary and enhance the hardness, the weaker grain interior serves as a crack nucleation site and propagation path, leading to transgranular fracture. At a low strain amplitude of 0.3% , the fatigue failure location shifts to the weld metal (Fig. 7c). The magnified SEM image of region B in Fig. 7c shows the precipitation of Cr_{23}C_6 carbides in the interdendritic region and absence of any microstructural change in the dendrite core, which is identical to the microstructures developed in the material aged for 8 h (Fig. 7d). Fig. 7e and f shows the magnified SEM images of the crack initiation region C and crack tip region D in Fig. 7c. It is evident that a main crack initiates in the dendrite core on the specimen surface and propagates to the interior along the dendrite core. As described earlier, the dendrite core in the weld metal is the softest region in the welded joint fatigue-tested at a low strain amplitude of $\leq \sim 0.3\%$; thus, the strain is localized in the dendrite core during fatigue deformation, serving as a crack nucleation site and propagation path. This corresponds well with the experimental results shown in Fig. 7. Based on these results, we can hypothesize the mechanism causing the shift of the fatigue failure location from the base metal

to the weld metal with decreasing strain amplitude. For the fatigue test at a high strain amplitude of $\geq \sim 0.4\%$, the test time was not sufficient to produce the thermal aging effect, and the base metal continued to be the softest region, where the applied strain was localized and led to failure. However, the fatigue test at the low strain amplitude of $\leq \sim 0.3\%$ provides a longer test time, which is sufficient to produce the thermal aging effect. Different microstructural evolutions occur in each region of the welded joint, promoting inhomogeneity of local hardness and thereby altering the softest region to the dendrite core in the weld metal. This change in the softest region of the welded joint leads to a shift in the fatigue failure location from the base metal to the weld metal. The microstructural evolution in the weld metal during fatigue deformation at elevated temperatures and the coexistence of two different fatigue failure mechanisms are schematically illustrated in Fig. 8.

4. Conclusions

In this study, we explored microstructural evolutions in the Super304H welded joint during fatigue deformation in the operating temperature range of $500\text{--}700^\circ\text{C}$ and their effect on the local mechanical properties and fatigue failure mechanism. Our results showed that depending on the time required to complete the fatigue test, thermal aging affects the microstructure in each region of the welded joint differently, causing nonuniformity in local hardness and deformation, and eventually altering the fatigue failure mechanism. At a high strain amplitude of $\geq \sim 0.4\%$, the fatigue test time was not sufficient to produce the thermal aging effect and the initial microstructures remained unchanged. Under these conditions, the base metal continued to be the softest region in the welded joint where the applied strain was localized, offering a crack nucleation site and propagation path, and consequently leading to the fatigue failure in the base metal. At a low strain amplitude of $\leq \sim 0.3\%$, however, the longer test time offered a sufficient high-temperature exposure to produce the thermal aging effect, which introduced a different microstructural change in each region of the welded joint: the precipitation of Cr_{23}C_6 carbides in the interdendritic

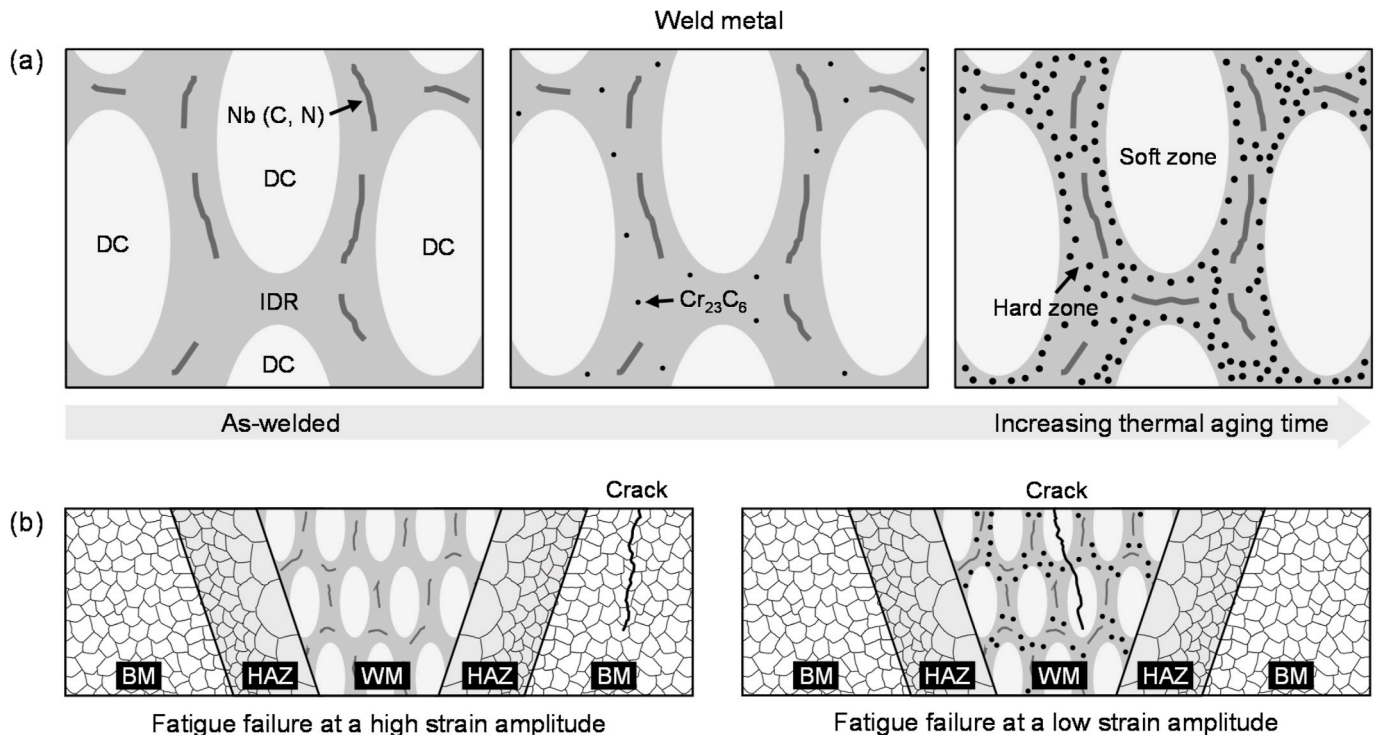


Fig. 8. Schematics illustrating the (a) microstructural evolution in the weld metal of Super304H welded joint during fatigue deformation (i.e., thermal aging) and (b) two different fatigue failure mechanisms coexisting at elevated temperatures. The left and right panels in (b) depict the fatigue failure mechanisms in the base metal at a high strain amplitude ($\geq \sim 0.4\%$) and in the weld metal at a low strain amplitude ($\leq \sim 0.3\%$), respectively.

region of the weld metal and at the austenitic grain boundary of the base metal and the precipitation of nanometer-sized Nb (C, N) phase in the austenitic grain interior of the base metal. The formation of these precipitates intensified the inhomogeneity of the local hardness in each region of the welded joint, which changed the softest region from the base metal to the weld metal (dendrite core). Therefore, strain localization occurred in the dendrite core of the weld metal during fatigue deformation, which served as a crack nucleation site and propagation path, and consequently led to fatigue failure in the weld metal.

Data availability

The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.

CRediT authorship contribution statement

Geun Dong Song: Conceptualization, Methodology, Formal analysis, Writing – original draft. Jeong Ho Hwang: Conceptualization, Investigation, Formal analysis, Writing – review & editing. Dae-Woong Kim: Investigation, Writing – review & editing. Byong Chon Park: Formal analysis, Writing – review & editing. Junhee Hahn: Project administration, Formal analysis, Writing – review & editing. Seong-Gu Hong: Conceptualization, Supervision, Writing – review & editing.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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