PAPER

Room- and service-temperature tensile properties of repeatedly welded austenitic stainless steel by Gleeble simulation of heat-affected zone

Y H Guo, Li Lin, X M Pan, Donghui Zhang, Lili Liu and M K Lei

1 School of Materials Science and Engineering, Dalian University of Technology, Dalian 116024, People’s Republic of China
2 China Nuclear Industry 23 Construction Co., Ltd, Beijing 101300, People’s Republic of China

E-mail: surfeng@dlut.edu.cn

Keywords: austenitic stainless steel, repeated welding, heat-affected zone, gleeble weld simulator, tensile properties

Abstract

The effects of repeated welding on the microstructure and tensile properties of the heat-affected zone (HAZ) of AISI 304N austenitic stainless steel at room and service temperatures were investigated using a Gleeble welding simulator. Thermal cycle parameters were obtained by in situ measurements of the HAZ of seven sequential weld beads. By thermal cycling in the Gleeble simulator, specimens were prepared to simulate the HAZs of steels in the as-welded condition and after one to five repeated welding processes, designated as AW and RW1–RW5, respectively. The HAZ microstructures were characterised and tensile tests were conducted at room temperature and the equipment service temperature of 260 °C. The repeatedly welded specimens showed microstructures of austenitic matrices with δ-ferrite inclusions. The average austenite grain size was 41.4–47.3 μm and the δ-ferrite content was 0.69–3.13 vol%. RW2 exhibited the maximum percentages of 31.8% and 15.9% high-angle boundary (HAB) and coincidence site lattice, respectively, while RW4 exhibited the minimum values of 22.3% and 9.0%, respectively. At room temperature, the ultimate tensile strength (UTS) and yield strength (YS) were 525–670 MPa and 190–245 MPa, respectively; at 260 °C, they were 460–475 MPa and 200–215 MPa, respectively. The room-temperature UTS and YS mainly depended on the contents of δ-ferrite and HABs. As the δ-ferrite content decreased and the frequency of HABs increased, the UTS and YS were increased. However, the boundary activation energy and dislocation resistance decreased at the service temperature, where the effects of the δ-ferrite and HAB contents on the UTS and YS were weaker.

1. Introduction

Repeated welding is the most widely used technology for manufacture, construction, and repair throughout the chemical, petroleum, and nuclear industries [1]. Under repeated welding, the heat-affected zone (HAZ) of the welded joint undergoes repeated heating and cooling and its strength and ductility can be upset by this thermal cycling. The main cause of the degradation in tensile properties in repeatedly welded components is the variation in grain size and precipitation amount [2–9]. The degradation in mechanical properties can affect the service lifetime and safety of welded equipment. Therefore, the tensile properties of HAZs are very important factors in safety assessments of repeated welding.

AghaAli et al [9] reported that, with one to four repeated welds by shielded metal arc welding (SMAW) on 316L austenitic stainless steel, the tensile strength and elongation of the repeatedly welded joint varied because of the grain size variation. Schino et al [10] reported an investigation on the effect of the grain size on the tensile properties, showing a Hall–Petch relationship for low-Ni austenitic stainless steel. Watanabe et al [11] reported the grain-boundary distribution and texture of repeatedly welded materials by electron back-scatter diffraction (EBSD). It was proposed that cracks could easily propagate along random grain boundaries during tensile
Table 1. Chemical composition mechanical properties of AISI 304N austenitic stainless steel.

| C  | Cr | Ni   | Mn | Si  | P   | S   | N   | UTS | YS  | elongation |
|----|----|------|----|-----|-----|-----|-----|-----|-----|------------|
| 0.070 | 18.545 | 8.211 | 1.600 | 0.500 | 0.034 | 0.001 | 0.120 | 550 MPa | 240 MPa | 25%         |

fracture; however, fracture ductility could be increased by increasing the proportion of low-angle grain boundaries. Kumar et al [12] studied the effects of grain-boundary character distribution (GBCD) on the tensile properties of 304L austenitic stainless steel at room temperature; the observed increase in percent elongation and decrease in yield strength (YS) were correlated with the special coincidence site lattice (CSL) structure. Saha et al [13] studied the effects of texture and misorientation on AISI 304 austenitic stainless steel in the HAZs of welding joints prepared by gas metal arc welding (GMAW). These studies indicated that when the frequency of high-angle boundaries (HABs) increased, the tensile strength and elongation of the welding joint were increased. The tensile properties are important in evaluating welding performance; those of the HAZ are particularly crucial in determining the suitability of repeated welding for a given application. Therefore, it is necessary to study the evolution and mechanisms of the HAZ tensile properties.

Most investigations on the tensile properties at room and service temperature of AISI 304N, which is widely used in the nuclear industry, have focused on the bulk material [14, 15]. Available data are scarce regarding the effects of repeated welding on the tensile properties of AISI 304N austenitic stainless steel at room and service temperatures. Because equipment service life is affected by repeated welding, the effect of repeated welding on the tensile properties of AISI 304N at room and service temperatures requires detailed understanding, not only for the design of repeated welding processes, but also for evaluating the effects of repeated welding on equipment safety. However, accurately locating and testing the HAZ for tensile properties by general methods is difficult because of the narrowness of the HAZ in actual welded joints [16, 17]. Furthermore, manual welding methods do not offer precise control of the welding heat input, thus hindering the controlled study of repeated welding on the HAZ. The Gleeble welding simulator, a physical simulation apparatus used in studying the hot-working of metals, may be effective in investigating the HAZ of repeatedly welded joints with good repeatability and reliability, because it can generate a controllable heat input over an area sufficient for tensile testing.

In this study, the Gleeble welding simulator was used to simulate one to five cycles of repeated welding to study the tensile properties of the HAZ formed on AISI 304N austenitic stainless steel at both room and service temperatures. Thermal cycles in the welding of seven repeated beads were measured experimentally by in situ thermocouples. The collected multi-bead welding thermal cycle data were used to prepare the as-welded and repeatedly welded HAZ specimens. All specimens were examined to systematically investigate the mechanistic effects of repeated welding on the microstructure, GBCD, and tensile properties of AISI 304N austenitic stainless steel.

2. Experimental

An AISI 304N austenitic stainless steel welding plate with a size of 240 × 150 × 20 mm was used in the multi-bead welding process. The chemical composition of AISI 304N austenitic stainless steel is listed in table 1. Welding was conducted using automatic gas tungsten arc welding (auto-GTAW) with ER308L filler wire of 2.0 mm in diameter. A 60° bevelled groove with the depth of 10 mm was machined in the centre of the welding plate. Figure 1 shows the V-groove joint, welding schedule and arrangement of the thermocouples. Thermocouples were located 5 mm away from the groove centre, where it was nearest to the fusion line. The welding process was similar to that performed in commercial welding and consisted of seven beads, designated as L1–L7. The welding process parameters are listed in table 2. Welding was conducted by a Liburdi GT VI machine with automatic-tracking of the arc voltage. The cycles of the HAZ were measured by K-type thermocouples with thermocouple wires of 0.1 mm in diameter. A temperature logger by Fluke NetDAQ was used to record the temperature field with a sampling frequency of 60 Hz. A DSI Gleeble3800 weld-simulator, a dynamic thermomechanical testing device used for widespread applications, including materials processes and physical simulation, with the maximum heating speed $10^4 ^\degree C/s$, was used to simulate the welding and repeated welding processes. The Gleeble weld-simulator was advantageous in generating HAZs with large volumes.

Figure 2 shows the temperature field measured in situ in the HAZ and the Gleeble input and output temperature curves. Figure 2(a) shows the temperature field for beads L1–L7 over time for temperatures of $\geq 300 ^\degree C$. The peak temperatures of the L1 and L2 beads are 1314 ^\degree C and 1290 ^\degree C, respectively; significant decreases of peak temperature are observed for beads L3–L7 with a gradual slope. For L6 and L7, the peak temperature is $< 700 ^\degree C$. Below this temperature, according to the 70 wt% Fe pseudo-binary phase diagram of
the Fe–Cr–Ni alloy with different Cr/Ni ratios [18], the phase transformation does not occur. Therefore, the welding thermal cycles of beads L1–L5 were selected as input of the Gleeble simulator because of the negligible effect of L6 and L7 beads on the solid-state transformation of the HAZ for the Fe–Cr–Ni austenitic stainless steel. Figure 2(b) shows the heat input of the simulator based on the temperature fields of L1–L5. This heat input was applied once to AISI 304N austenitic stainless steel, yielding the as-welded (AW) specimen; additional heat input of one to five times were applied to obtain the repeated welding (RW) specimens RW1 to RW5, respectively. Figure 2(c) shows once the real-time temperature fields of the Gleeble simulation process, obtained from a thermocouple welded to the centre of the specimen surface. Two geometric types of simulated specimens were used: rectangular-section bars of 10.5 mm × 10.5 mm × 60 mm and round-section bars of φ10 mm × 80 mm. The rectangular-section bars were prepared for microstructural and EBSD analysis. The round-section bars were used in tensile testing at room temperature and 260 °C (equipment service temperature). Figure 3 shows an original macrograph of the Gleeble weld-simulator specimen before machining for sampling, showing a simulated HAZ of approximately 10 mm in width at the centre.
The specimens with square section for metallographic observation were prepared through using #200, #320, #400, #800, and #1200 grit emery sheets, followed by final polishing with 0.5 μm diamond paste slurry. Etching of the specimen was performed electrolytically in a 10% oxalic acid solution at 10 V for 60 s. The polished and etched specimens were examined for microstructure by Leica MEF4 optical microscope. Etching of the specimen was performed electrolytically in a 10% oxalic acid solution at 10 V for 60 s. The x-ray diffractometer measurement was performed in a PANalytical Empyrean x-ray diffractometer using monochromatic CuKα radiation operated at 40 kV tube voltages, with a scan rate of 4° per minute and 2θ angle range 30° to 90°. The assessment of the grain size was performed according to ASTM E-112. Each specimen was analysed in ten fields of view at 200×. The amount of δ-ferrite was measured using three methods, including the quantitative metallography method, magnetic method, and eddy current method [19]. In the quantitative metallographic method, the δ-ferrite contents were measured with a 200× metallographic image covering five fields of view for each specimen in the centre area. Each specimen was measured at ten points by the magnetic method using an F-IA device. The eddy current method, using an Olympus NORTEC600 defectoscope and contact probes with the frequency of 0.5 MHz was also applied to measure the trajectory of signal variation of the eddy current phase angle.

In order to analyse the grain-boundary characteristics using EBSD, all the specimens were mechanically polished as optical microscopy (OM) specimens. Etching was performed electrolytically in a 10% oxalic acid solution at 10 V for 30 s; the etched specimens were then finely polished with a 0.05-μm colloidal silica solution for 3 h using a Buehler Vibromet2 vibratory polisher. EBSD testing was performed in an integrated scanning electron microscopy—EBSD—energy-dispersive x-ray spectroscopy (SEM—EBSD—EDS) system (Zeiss Supra 55). The EBSD patterns were acquired using an acceleration voltage of 20 kV and a specimen tilt of 70°. The typical area analysed was 958 × 548 μm with the scanning step of 4 μm. The sample coordinate system CS0 (rolling direction—transverse direction—normal direction; RD–TD–ND) was coordinated with the EBSD coordinate system CSn (x_n–y_n–z_n).

According to ASTM E8 and E21, tensile testing was conducted at 2 mm min⁻¹ at room temperature (25 °C) and the service temperature of 260 °C, which is the working temperature of some essential equipments under normal operating conditions in the advanced pressurised-water reactor nuclear power plants [15]. Figure 4 shows the sampling positions, orientations, and dimensions of the tensile specimens. The gauge section was 4 mm in length and 6 mm in diameter. For tensile testing at 260 °C, an electric furnace was used to heat the specimen; thermocouples were attached to the specimens at the gauge sections to monitor the temperature during testing. The tensile fracture surfaces after testing were analysed using SEM and EDS.

3. Results

3.1. Microstructure of the repeated welding specimens

Figure 5 shows the microstructure of the AW and RW1–RW5 specimens obtained by Gleeble weld simulation. The XRD patterns show γ-austenite and δ-ferrite in all specimens. The austenitic matrix mostly comprises equiaxed crystals and twins with a few secondary twins. Longer δ-ferrite laths and dispersed δ-ferrite precipitates are also observed in the grain boundaries. Grain size assessment of the AW and RW1–RW5 specimens was performed using the linear intercept method according to ASTM E-112. A slight variation in the austenitic grain size of 41.4–47.3 μm was observed: 41.4 μm, 47.3 μm, 41.6 μm, 45.5 μm, 46.5 μm, and 43.6 μm for the AW, RW1, RW2, RW3, RW4, RW5 specimens. The smallest grain size of 41.4 μm was obtained in the AW and the largest of 47.3 μm in RW1.

Figure 6 shows the δ-ferrite contents measured in the AW and RW1–RW5 AISI 304N austenitic stainless steel samples. As the number of repeated welding cycles increases, the δ-ferrite content first decreases to 0.69 vol% for the RW2 specimen from 3.13 vol% for the AW specimen, increases to 1.50 vol% for RW4, and reaches 1.18 vol% for RW5 as calculated using the quantitative metallography method. The largest and smallest contents of δ-ferrite are obtained for the AW and RW2 specimens, respectively. A similar trend in the δ-ferrite content is observed by the magnetic method, as 1.10 FN (Ferrite Number), 0.75 FN, 0.63 FN, 0.88 FN, 0.95 FN, and 0.64...
FN with the AW and RW1–RW5 specimens, respectively. While quantitative $\delta$-ferrite contents are not attained by the eddy method, a similar trend in the $\delta$-ferrite contents is observed via the impedance phase angle.

### 3.2. GBCD of the repeatedly welded specimens

Figure 7 shows the misorientation angle distribution and the networks of misorientation angles of $>15^\circ$ in the austenitic matrices for the repeated-welding specimens. All misorientation angle distributions deviate from the standard distribution curve of austenitic stainless steel; most misorientation angles are $<5^\circ$ and $>55^\circ$. Grain boundaries can be characterised crystallographically and geometrically by the relative orientation relationship between two grains, and classified by the misorientation angle into two categories: low-angle boundaries (LABs) and high-angle boundaries (HABs) \[20\]. LABs are defined as boundaries with misorientation angles of $<15^\circ$; HABs are those with misorientation angles $>15^\circ$. As the number of repeated welds is increased, a variation in the LAB fraction of 68.2%–77.7% is observed, whereas the HAB fractions of the AW and RW1–RW5 specimens
show large variations of 30.1%, 27.9%, 31.8%, 25.7%, 22.3%, and 23.7%, respectively. Networks of HABs in the AW and RW1–RW5 specimens are observed in the misorientation maps. The networks of HABs are distributed continuously in AW, RW1, and RW2 with small degrees of clustering. By comparison, those in RW3, RW4, and RW5 have sparse distributions and poor continuity.

Figure 8 shows the CSL fraction as a function of $\Sigma$ for the repeated welding specimens. The boundary type of CSL is $\Sigma^m$ ($m = 1, 2, 3$), which is mainly $\Sigma^3$ boundary in the repeated welding specimens. The CSL boundaries with $\Sigma$ values $< 29$ are defined as special boundaries; those with values $> 29$ are random boundaries [11]. In austenitic stainless steel, the grain boundary of $\Sigma^3$ is a twin boundary with the boundary angle of 60° [21]. With increasing repeated welding, the special boundary fractions of AW, RW1, and RW2 are 18.7%, 15.0%, and 18.5%, respectively. Among them, the fractions of $\Sigma^3$ special boundaries are 11.2%, 9.0%, and 10.8%, respectively. The fractions of special boundaries in RW3, RW4, and RW5 are 11.2%, 9.0%, and 10.8%, all of which are lower than that in RW1. The fraction of $\Sigma^{27}$ boundaries is low to be negligible.

Figure 9 shows the angular deviation from the ideal Kurdjumov–Sachs (K-S) orientation of $\{111\}_{\gamma}/\{110\}_{\delta}, \{110\}_{\gamma}/\{111\}_{\delta}$ [22] for the $\delta$-ferrite/austenite interface in the repeated welding specimens. The angular deviation exhibits two peaks, one at $\sim 3.5^\circ$ for AW, RW1, RW2, and RW5 and another at $\sim 4.5^\circ$ for RW3 and RW4. Because the K-S relationship represents the best atomic match within the $\delta$-ferrite/austenite interfaces, these interfaces have the lowest energy [23, 24]. With repeated welding, the K-S relationship of RW3 is broken down because of grain refinement during the second repeated welding process. With increased numbers of repeated welding, the $\delta$-ferrite/austenite interface becomes a lower-energy interface; therefore, the fraction of angular deviation of 4.5° for RW4 is decreased. The frequency of angular deviation from 3.5° of RW5 reaches a second peak value from that of RW4.

3.3. Tensile properties and fracture surfaces at room temperature

Figure 10 shows the engineering stress–strain curves for the repeated welding specimens. The failure zones are all located at the centres of the gauge sections. The room-temperature engineering stress–strain curves show typical
low YS values and strain-hardening behaviours [25]; with increasing strain, the tensile strength is increased to the maximum value. With repeated welding, the ultimate tensile strength (UTS) increases first from 555 MPa for AW to the maximum value of 670 MPa for RW2, decreases to 525 MPa for RW4, and reaches 530 MPa for RW5. In all repeated welding specimens, the specimens with the highest and lowest UTS are RW2 and RW4, respectively. The trend in YS is similar to that of UTS. The elongation of the repeated-welding specimens varies from 14.5% to 17.5%, with the minimum for RW1 and the maximum for RW2.

Figure 11 shows details of the fracture surface morphologies in the macroscale fracture surface of the entire section and the microscale fracture surface in the middle of the section. Figures 11 (a)–(f) shows that the macro fracture surface is divided into the crack initiation zone (fibrous zone) and the final rupture zone (shear lip zone). In the fibrous zone, many cracks are parallel to each other. The cracks of RW4 in figure 11(e) extend to the shear lip zone. Figures 11(a1)–(f1) shows the microscale fracture surface in the fibrous zone; it is coved by dimples and microcracks, as typically observed for ductile fracture [26]. AW shows a mixed mode of fracture with both microcracks and ductile dimples. In RW1, the number of cracks is decreased comparing to that in AW. However, many tear ridges are observed with very fine dimples. After two repeated welding cycles, RW2 shows ductile fracture with many fine ductile dimples and few microcracks. The fracture surface of RW3 contains very fine dimples and cleavage facets. The fracture morphology of RW4 is similar to that of AW, but the microcracks
are widely distributed. After five repeated welds, RW5 shows a mixed fracture mode with many ductile dimples as well as microcracks. Observing the fracture surfaces in the shear slip zone of all the repeated welding specimens, the fracture surfaces show few cracks and smooth areas, typical of slip–plane separation fracture, as shown in figures 11(a2)–(f2).

Figure 10. Engineering stress–strain curves of the repeated welding specimens at room temperature.

Figure 11. Macroscale and microscale fracture surface morphologies and EDS analysis of the repeated welding specimens: (a), (a1), (a2) AW; (b), (b1), (b2) RW1; (c), (c1), (c2) RW2; (d), (d1), (d2) RW3; (e), (e1), (e2) RW4; (f), (f1), (f2) RW5.
EDS is used to analyse elements at the microcrack positions P1 and P2 as shown in figures 11(a1)–(f1); the results are listed in table 3. The Ni contents of the measurement points are all lower than that in the base metal; therefore, microcracks may be located in δ-ferrite.

3.4. Tensile properties and fracture surfaces at service temperature

Figure 12 shows the engineering stress–strain curves for the repeated welding specimens at 260 °C. The failure zones are also located at the centres of the gauge sections. For the repeated welding specimens, the UTS and YS are in the ranges 460–475 MPa and 200–215 MPa, respectively. The elongation varies from 16.2% to 17.7% for AW and RW1–RW5. At the service temperature, the engineering stress–strain curves also show lower YS and strain-hardening behaviours. The difference of the engineering stress–strain curve is in the change in curve shape at service temperature compared to that at room temperature. Remarkable necking occurs at the first turn of the stress–strain curve [27]. The engineering stresses of all repeated welding specimens are increased consistently with engineering strain until necking occurs. The strain-hardening effects in the repeated welding specimens differ slightly thereafter. The highest UTS of 475 MPa is obtained for RW1 with the maximum elongation of 17.7%; the lowest value of 460 MPa is obtained for RW3 with the minimum elongation of 16.2%. At the service temperature, the strain-hardening significantly affects the UTS [27].

Figure 13 shows details of the fracture surface morphologies of the repeated welding tensile specimens at the service temperature. Figures 13(a)–(f) shows the macroscale fracture surfaces, which lack obvious fibrous and shear lip zones. In a three-dimensional (3D) view, the fracture appears as a spiral initiating approximately at the centre of the fracture surface and proceeding along the [111] direction. Meanwhile, the microscale fracture features of AW and RW1–RW5 are shown in figures 13(a1)–(f1). The microscale fracture surfaces of AW and RW2–RW5 show ductile dimples in different sizes with voids located at the bottoms of the dimples, generated by the rupture of δ-ferrite laths under normal stress. Moreover, the ductile dimples of AW and RW2 are elliptical, while the fracture surface of RW1 contains equiaxial dimples and some shell-shaped cleavage planes. The fracture surfaces of RW3–RW5 also show equiaxial dimples.

4. Discussion

Figure 14 shows the UTS, YS, and elongation of AW and RW1–RW5. The stress–strain curves show typical lower YS and strain hardening behaviours of austenitic stainless steel at both the room and service temperatures. At room temperature, with repeated welding, similar trends in UTS and YS are observed. The highest UTS and YS are 670 MPa and 245 MPa, respectively, for RW2; the lowest values of 525 MPa and 190 MPa, respectively, are

|   | AW  | RW1 | RW2 | RW3 | RW4 | RW5 |
|---|-----|-----|-----|-----|-----|-----|
| Cr | 18.1| 16.9| 22.0| 30.2| 24.4| 18.5|
| Ni | 5.9 | 4.1 | 2.8 | 1.0 | 2.4 | 5.1 |

Table 3. Contents of Cr and Ni in fracture surface on AW, RW1–RW5 specimens measured using EDS wt%.

Figure 12. Engineering stress-strain curves for the repeated welding specimens at 260 °C.
obtained for RW4. The maximum and minimum elongations of 17.5% and 14.5% are measured for RW2 and RW1, respectively. Considering the strength and ductility, RW2 has the best mechanical properties and RW4 the worst. At 260 °C, the highest UTS of 475 MPa and maximum elongation of 17.7% are obtained for RW1. The lowest UTS of 460 MPa and minimum elongation of 16.2% are obtained for RW3. Compared to the room-temperature UTSs, those at service temperature are sharply decreased with slight variations; however, the YS and elongation of all specimens at 260 °C are not significantly decreased, but the variation is decreased.

4.1. Effect of δ−ferrite and grain boundary character on room−temperature tensile properties
At room temperature, the difference in grain size of 5.9 μm between 41.4 and 47.3 μm may not strongly influence the tensile properties according to the Hall–Petch relationship. The Hall-Petch relationship is described by the follow equation [12]:

![Figure 13](image URL) Macroscale and microscale fracture surface morphologies of the repeated welding specimens: (a), (a1) AW; (b), (b1) RW1; (c), (c1) RW2; (d), (d1) RW3; (e), (e1) RW4; (f), (f1) RW5.

![Figure 14](image URL) Tensile properties of the repeated-welding specimens at room and service temperature.
where \( \sigma \) stands for the tensile strength, \( \sigma_0 \) and \( K \) (the magnitude of \( K \) is generally \( 10^2 \) [28]) are constants, and \( d \) is the grain size. In the present study, according to the equation, the difference of stress between the maximum and minimum grain size was just 0.01K. As shown in figures 6 and 14, the YS of the repeated welding specimens are increased with decreased \( \delta \)-ferrite contents. Tensile fracture occurs in two stages: crack initiation and crack propagation.

For the crack initiation stage, the tensile fracture of austenitic stainless steel with good ductility is strongly affected by secondary phases [20, 29] such as \( \delta \)-ferrite. The number of tensile cracks is directly related to the \( \delta \)-ferrite content of the repeated welding specimens, as shown by the fracture morphologies and EDS measurements from figure 11. During the plastic deformation of the repeated welding specimens, \( \delta \)-ferrite may hinder dislocation slip because the austenite matrix and \( \delta \)-ferrite precipitation have different crystal structures. Therefore, many dislocation pile-ups occur at the interface between the \( \delta \)-ferrite and austenite matrix. With increasing strain, a head dislocation from the pile-up migrates to the \( \delta \)-ferrite/austenite interface under shear stress, thus causing microcrack generation along the interface [30]. Furthermore, the interface orientation of \( \delta \)-ferrite and the austenite matrix also affects the crack initiation. As shown in figure 9, the angular deviations from the K-S orientation for RW3 and R4 are greater than those of AW, RW1, RW2, and RW5; therefore, the \( \delta \)-ferrite/austenite boundaries of RW3 and RW4 have weak coherence compared to the other specimens [31]. These weakly coherent interfaces are prone to breakage under lower shear stresses [24]; therefore, microcracks are more likely to generate at these interfaces first, because of the dislocation pile-ups. Figures 11(d1) and (e1) show that the microcrack fracture morphologies are slip fractures indicating low coherence in RW3 and RW4, while the microcracks morphologies of the AW, RW1, RW2, and RW5 specimens indicate brittle fracture via tearing. Throughout the above analysis, the initiation of microcracks during the tensile test is dependent on the \( \delta \)-ferrite content and the interface orientation of the \( \delta \)-ferrite and austenitic matrix.

The microcracks continue to propagate after initiation. Crack propagation in austenite stainless steel mainly occurs by multiple slips of the grain boundaries [27]. Therefore, the grain boundaries have an important role in resisting the migration of dislocation pileups during crack growth. The crack propagation path is much longer for HABs than for LABs. In addition, the \( \Sigma \)3 CSL boundary is characterised as a low-energy boundary [11, 20]. When cracks encounter such grain boundaries, they require more energy to induce boundary slip, thus preventing crack propagation [23]. Figure 15 shows the frequency of HABs and the UTS as a function of repeated welding numbers at room temperature. The UTS is increased with decreasing ferrite content and increasing HAB fraction. Therefore, the UTS and YS are affected by the combined effects of \( \delta \)-ferrite and low-energy grain boundaries during the crack initiation and propagation stages. However, for the AW specimen with the highest \( \delta \)-ferrite content, the UTS and YS are not the lowest because of the many HABs. In figures 14 and 15, the trend in the elongation of repeated welding specimens is similar to that of HABs; the elongation of the repeated welding specimens is increased with increasing HAB fraction [23].

**4.2. Effect of \( \delta \)-ferrite and grain boundary character on service — temperature tensile properties**

At 260 °C, the deformation trends of the UTS, YS, and elongation are similar to those at room temperature, as shown in figure 14. However, the dislocation migration and microcrack morphologies show some differences between room and service temperatures because the elevated testing temperature decreases the interfacial activation energy of dislocation movement [27, 32]. During crack initiation, dislocation pile-ups are generated at the interface of \( \delta \)-ferrite and austenite by grain sliding. Because the interface motivation energy is decreased with the decrease in elasticity modulus at the elevated testing temperature [24], the interface of the \( \delta \)-ferrite and austenite shows reduced resistance to dislocation movement, causing easy crack propagation at the interface. As shown in figures 6 and 14, the YS of the repeated welding specimens is increased with decreasing \( \delta \)-ferrite contents. The number of tensile cracks is directly related to the \( \delta \)-ferrite content of the repeated welding specimens, as shown by the fracture morphologies and EDS measurements from figure 11. During the plastic deformation of the repeated welding specimens, \( \delta \)-ferrite may hinder dislocation slip because the austenite matrix and \( \delta \)-ferrite precipitation have different crystal structures. Therefore, many dislocation pile-ups occur at the interface between the \( \delta \)-ferrite and austenite matrix. With increasing strain, a head dislocation from the pile-up migrates to the \( \delta \)-ferrite/austenite interface under shear stress, thus causing microcrack generation along the interface [30]. Furthermore, the interface orientation of \( \delta \)-ferrite and the austenite matrix also affects the crack initiation. As shown in figure 9, the angular deviations from the K-S orientation for RW3 and R4 are greater than those of AW, RW1, RW2, and RW5; therefore, the \( \delta \)-ferrite/austenite boundaries of RW3 and RW4 have weak coherence compared to the other specimens [31]. These weakly coherent interfaces are prone to breakage under lower shear stresses [24]; therefore, microcracks are more likely to generate at these interfaces first, because of the dislocation pile-ups. Figures 11(d1) and (e1) show that the microcrack fracture morphologies are slip fractures indicating low coherence in RW3 and RW4, while the microcracks morphologies of the AW, RW1, RW2, and RW5 specimens indicate brittle fracture via tearing. Throughout the above analysis, the initiation of microcracks during the tensile test is dependent on the \( \delta \)-ferrite content and the interface orientation of the \( \delta \)-ferrite and austenitic matrix.

The microcracks continue to propagate after initiation. Crack propagation in austenite stainless steel mainly occurs by multiple slips of the grain boundaries [27]. Therefore, the grain boundaries have an important role in resisting the migration of dislocation pileups during crack growth. The crack propagation path is much longer for HABs than for LABs. In addition, the \( \Sigma \)3 CSL boundary is characterised as a low-energy boundary [11, 20]. When cracks encounter such grain boundaries, they require more energy to induce boundary slip, thus preventing crack propagation [23]. Figure 15 shows the frequency of HABs and the UTS as a function of repeated welding numbers at room temperature. The UTS is increased with decreasing ferrite content and increasing HAB fraction. Therefore, the UTS and YS are affected by the combined effects of \( \delta \)-ferrite and low-energy grain boundaries during the crack initiation and propagation stages. However, for the AW specimen with the highest \( \delta \)-ferrite content, the UTS and YS are not the lowest because of the many HABs. In figures 14 and 15, the trend in the elongation of repeated welding specimens is similar to that of HABs; the elongation of the repeated welding specimens is increased with increasing HAB fraction [23].

**Figure 15.** Frequency of HABs and the UTS as a function of repeated welding numbers at room temperature.

\[
\sigma = \sigma_0 + Kd^{-1/2}
\]
content at the service temperature. However, a slight variation in the YS of 200–215 MPa is observed for the RW1–RW5 specimens because of the decreasing effect of δ-ferrite on the YS. As shown in figure 13, few large-scale cracks are present on the fracture surface compared to the room-temperature fracture surface.

The microcracks continue to propagate after initiation. Deformation predominantly occurs by grain-boundary sliding, which easily generates edge dislocations. Edge dislocations climb to nearby sliding planes with the assistance of the thermal activation energy at the service temperature, thus decreasing strain hardening because of the decreased stress concentration. Therefore, the UTs at the service temperature are lower than those at room temperature. Furthermore, grain boundaries show viscosity; adjacent grain boundaries are prone to slip, which reduces the effect of the grain-boundary character. Therefore, slight UT fluctuations are observed at the service temperature from figure 14. Dislocation movement occurs by collaborative deformation via dislocation climbing; almost no local stress concentration forms, yielding fracture instability.

AISI 304N is an austenitic stainless steel with a face-centred cubic (FCC) structure. The slip face is {111} and the slip direction is {110}. Under normal stress, the sliding surface (111) moves along the [110] direction, possibly causing the spiral fracture morphologies shown in figures 13(a)–(f). However, at room temperature, the dislocation pile-ups are first generated at the δ-ferrite/austenite interface, causing instability and fracture and eventually forming fractures with the ‘cup and cone’ geometry.

5. Conclusions

(1) With one to five cycles of repeated welding, the δ-ferrite contents of all specimens were 0.69–3.13 vol% and the grain sizes were 41.4–47.3 μm. The maximum and minimum fractions of Σ3 were 15.9% and 9.0% for RW2 and RW4, respectively. Meanwhile, the maximum and minimum HAB fractions of 31.8% and 22.3% were observed for RW2 and RW4, respectively. The angular deviation of RW3 and RW4 from the ideal K-S orientation was ~4.5°.

(2) At room temperature, the UTs, YS, and elongation of all repeated-welding specimens had the ranges 525–670 MPa, 190–245 MPa, and 14.5%–17.5%, respectively. The grain size of 41.4–47.3 μm might not strongly affect the tensile properties. The tensile properties depended instead on the δ-ferrite and HAB contents. As the δ-ferrite content decreased and the HAB fraction increased, the UTs and YS were increased.

(3) At the service temperature of 260 °C, the UTs and YS are 460–475 MPa and 200–215 MPa, respectively, increased with decreasing δ-ferrite contents and increasing HABs. The effects of the δ-ferrite content and HAB fraction were weaker than at room temperature. The UTs at the service temperature was lower than that at room temperature. The fluctuation of YS was slight because of the elevated test temperature.

Acknowledgments

The authors are grateful to Mr J Y Gao and Ms. N Ding for their helpful discussion and technical assistance. This work was supported by National Basic Research Program of China (973 Program) under Grant No.2015CB057306.

ORCID iDs

Y H Guo 0 https://orcid.org/0000-0002-8313-6514

References

[1] Song S and Dong P 2014 Residual stresses in weld repairs and mitigation by design Proc. of the ASME 2014 33rd Int. Conf. on Ocean, Offshore and Arctic Engineering (San Francisco, California, USA) (ASME) p 24547
[2] Daei S A H and Vakili-Tahami F 2012 Experimental study of the creep behavior of parent, simulated HAZ and weld materials for cold-drawn 304L stainless steel Eng. Fail. Anal. 21 78–90
[3] Yi H J, Lee Y J and Lee K O 2015 Influences of the welding heat input and the repeated repair welding on Ti-3Al-2.5 V titanium alloy Acta Metall. Sin. (English Lett.) 28 684–91
[4] Varghese P, Prasad M S, Joseph F, Varkey M J, Antony K and Sreekanth A 2015 The effect of repeated repair welding on the corrosion behaviour of austenitic stainless steel and mild steel dissimilar weldment Proc. of Int. Conf. on Advances in Materials, Manufacturing and Applications pp 864–9
[5] Nascimento M P, Voorwald H J C and Payão Filho J D A C 2012 Effects of several TIG weld repairs on the axial fatigue strength of AISI 4130 aeronautical steel-welded joints Fatigue Fract. Eng. Mater. Struct. 35 191–204
[6] Hia S W 2008 The effect of repeated weld-repairs on the microstructure and texture of 304 stainless steel weld National Taiwan University of Science and Technology M9503808 http://ir.lib.ntust.edu.tw/handle/987654321/6377
Lin C M, Tsai H L, Cheng C D and Yang C 2012 Effect of repeated weld-repairs on microstructure, texture, impact properties and corrosion properties of AISI 304L stainless steel Eng. Fail. Anal. 21 9–20

Vega O E, Hallen J M, Villagomez A and Contreras A 2008 Effect of multiple repairs in girth welds of pipelines on the mechanical properties Mater. Charact. 59 1498–507

AghaAli I, Farzam M, Golozar M A and Danacee I 2014 The effect of repeated repair welding on mechanical and corrosion properties of stainless steel 316L. Mater. Des. 54 331–41

Schino A D, Barteri M and Kenny J M 2003 Effects of grain size on the properties of a low nickel austenitic stainless steel J. Mater. Sci. 38 4725–33

Watanabe T 1993 Grain boundary design for advanced materials on the basis of the relationship between texture and grain boundary character distribution (GBCD) Texture Microstruct 20 195–216

Kumar B R, Chowdhury S G, Narasaih N, Mahato B and Das S K 2007 Role of grain boundary character distribution on tensile properties of 304L stainless steel Metall. Mater. Trans. A 38A 1136–43

Saha S, Mukherjee M and Pal T K 2015 Microstructure, texture, and mechanical property analysis of gas metal arc welded AISI 304 austenitic stainless steel J. Mater. Eng. Perform. 24 1125–39

Lee W-S and Lin C-F 2001 Impact properties and microstructure evolution of 304L stainless steel Mater. Sci. Eng. A 308 124–35

Blandford R K, Morton D K, Snow D S and Rahi T E 2007 Tensile stress-strain results for 304L and 316L stainless steel plate at temperature 2007 ASME pressure vessels and piping division conference (San Antonio, Texas, USA) p 26906

Silvat B, Li L, Decuester A and Griffiths B 2013 Effect of postweld heat treatment on the toughness of heat-affected zone for grade 91 steel Weld. J. 92 80s–87s

Liu W, Lu F, Yang R, Tang X and Cui H 2015 Gleeble simulation of the HAZ in Inconel 617 welding J. Mater. Process. Technol. 225 221–8

Lippold J C and Savage W F 1979 Solidification of austenitic stainless steel weldments: part I-a proposed mechanism AWS 60th Annual Meeting (Detroit, USA) pp 362s–374s

Kukla D, Grzywna P, Kopeć M and Kowalewski Z 2016 Eddy current method for thickness assessment of carburized layers Physical Metalurgy and Material Science Conf.-Advanced Materials and Technologies (Poland) pp 2–4

Gourgues A-F 2002 Electron backscatter diffraction and cracking Mater. Sci. Technol. 18 119–33

Caul M, Fiedler J and Randle V 1996 Grain-boundary plane crystallography and energy in austenitic steel Scr. Mater. 35 831–6

Tomida T, Wakita M, Yasuyama M, Sugaya S, Tomota Y and Vogel S C 2015 Memory effects of transformation textures in steel and its prediction by the double Kurdjumov–Sachs relation Acta Mater. 61 2828–39

Sinha S, Kim D-I, Fleury E and Suwas S 2015 Effect of grain boundary engineering on the microstructure and mechanical properties of copper containing austenitic stainless steel Mater. Sci. Eng. A 628 175–85

Lee D N and Han H N 2012 Orientation relationships between precipitates and their parent phases in steels at low transformation temperatures J. Solid Mech. Mater. Eng. 6 323–38

Kim M-Y, Kwak S-C, Choi I-S, Lee Y-K, Suh J-Y, Fleury E, Jung W-S and Son T-H 2014 High-temperature tensile and creep deformation of cross-weld specimens of weld joint between T92 martensitic and Super304H austenitic steels Mater. Charact. 97 161–8

Mills W J 1997 Fracture toughness of type 304 and 316 stainless steels and their welds Int. Mater. Rev. 42 45–82

Bryan T S, Hashimoto N and Farrell K 2004 Temperature dependence of strain hardening and plastic instability behaviors in austenitic stainless steels Acta Mater. 52 3899–90

Ulvan E and Koursaris A 1988 The effect of grain size on the bulk formability and tensile properties of austenitic stainless steel types 304 and 316 Metall. Trans. A 19A 2287–98

Bujat S, Besson J, Gourgues A-F, N’guyen F and Pineau A 2001 Microstructure and damage initiation in duplex stainless steels Mater. Sci. Eng. A 317 32–6

Yang C, Huang H, Thorogood G J, Jiang L, Ye X, Li Z and Zhou X 2016 The effect of grain size and dislocation density on the tensile properties of Ni–SiC NP composites during annealing J. Mater. Eng. Perform. 25 726–33

Kamiya O, Kumagai K and Kikuchi Y 1990 Effects of δ ferrite morphology on low temperature fracture toughness of SUS304L steel weld metal Trans. Japan Weld. Soc. 21 57–62 http://ci.nii.ac.jp/naid/110003380318/en/

Pei H X, Zhang H L, Wang L X, Li S L, Li D Z and Wang X T 2014 Tensile behaviour of 316LN stainless steel at elevated temperatures Mater. High Temp. 31 198–203