

Results and Discussion of High-temperature Tensile Test

This chapter provides details of the high-temperature tensile test carried out on the P92/304L dissimilar weld joint (DWJ). The detailed discussion regarding the strength value, microstructure properties, fracture surface analysis and micro-hardness value obtained after high-temperature tensile test in temperature range of 450-850 °C is provided.

7.1 High-temperature tensile test for P92/304L dissimilar weld joint

The dissimilar weld joint (DWJ) of P92 creep strength enhanced ferritic (CSEF) steel and 304L austenitic stainless steel (ASS) are expected to operate at a temperature range of 570-650 °C and pressure of 25-30 MPa. Hence, it is important to investigate the stability of these weld joints at elevated temperatures (Lee et al., 2016). The high-temperature tensile strength test was conducted at different temperatures (450 °C, 550 °C, 650 °C, 750 °C, and 850 °C) to examine the ultimate tensile strength (UTS) and yield strength (YS) values of the P92-ERNiCrMo-3-304L DWJ at elevated temperature. The UTS of the 588 MPa was observed at room temperature, and decreased to 47 MPa at 850 °C. The UTS and YS value decreased with the increase of the test temperature. During the room temperature tensile test, the specimen failed from the 304L BM due to the low UTS value of the 304L steel than P92 steel and ERNiCrMo-3 filler weld fusion zone at room temperature. The P92/304L DWJ failed from the 304L base metal (BM) up to the temperature of 550 °C. The P92 steel exhibits a higher UTS value than the 304L up to 650 °C due to the precipitation strengthening and solid solution strengthening effect. However, due to the coarsening of the precipitates and new secondary phase formations, the precipitation and solid solution strengthening effect becomes weak at high-temperature. Thus, it can be concluded that at high temperature, P92 steel is the most susceptible to failure among the other region. The fracture location shifted from the 304L BM at 450 °C and 550 °C to the P92 steel BM at 650 °C, 750 °C, and 850 °C. Thus, the mechanical and microstructural properties of the P92 steel degrade quickly after the maximum operating temperature range. Although the test was carried out up to 850 °C, the weld fusion zone did not crack. This indicates the excellent stability of the ERNiCrMo-3 filler metal at elevated temperature service. At high temperatures, because of some detrimental metallurgical changes in the P92 steel, the microstructure and mechanical properties of the P92 steel degrade quickly after the maximum operating temperature range of 560-600 °C. Figure 7.1(a-e) shows the appearance of the ruptured high-temperature tensile specimen. It becomes difficult to determine whether a fracture originated from the HAZ or BM if it occurs close to the weld interface. However, as in present case the fracture location was far away from the weld interface, the exact fracture location was easy to identify.

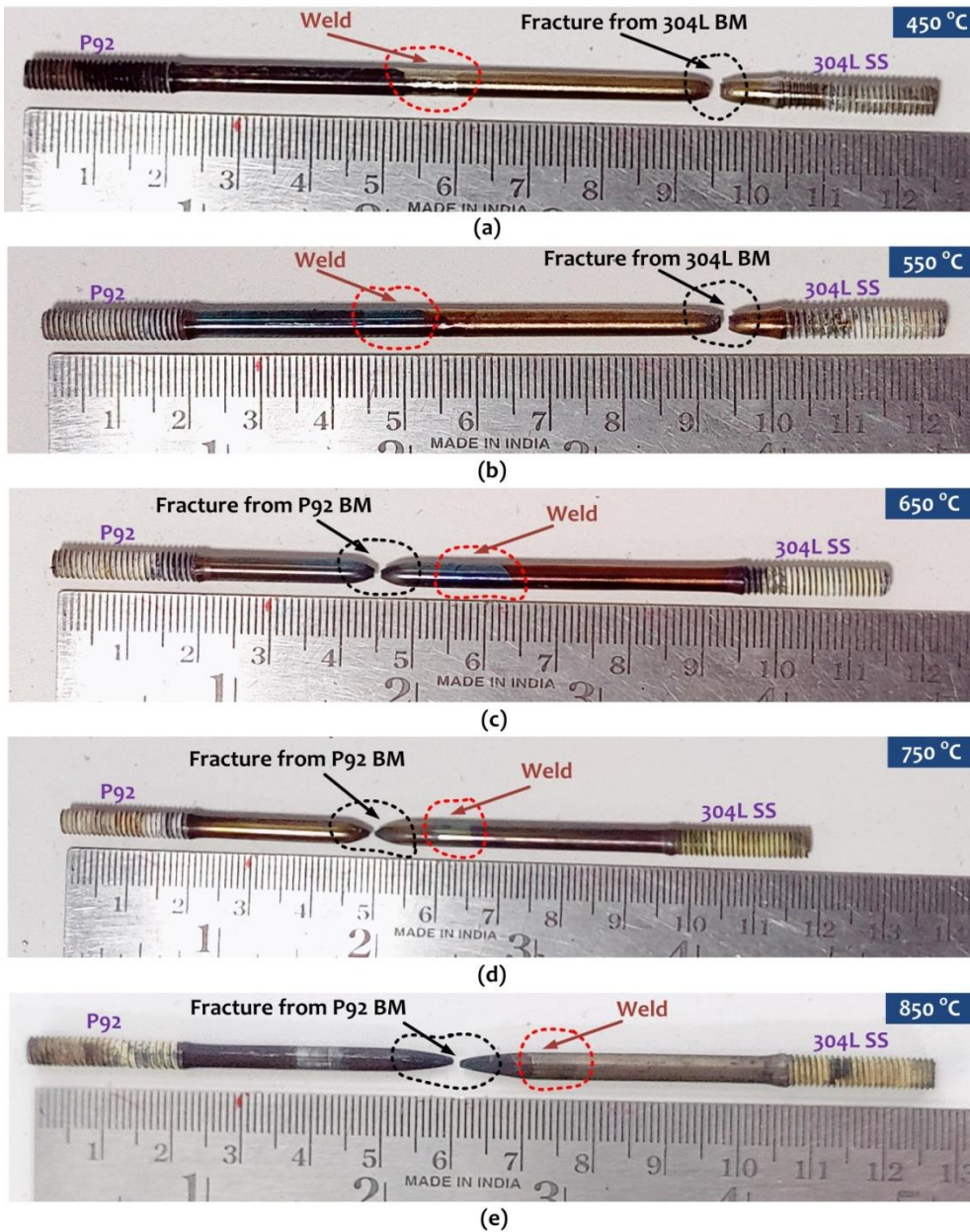


Figure 7.1: Photograph of the ruptured high-temperature tensile specimen (P92-ERNiCrMo-3-304L DWJ), showing the fracture location at different test temperature (450 °C-850 °C)

Figure 7.2(a-e) shows the high-temperature tensile specimens (P92-ERNiFeCr-2-304L DWJ) ruptured at different test temperature. The specimen tested at a room temperature and at elevated temperatures (450 °C and 550 °C) were failed from the 304L BM region, as shown in Figure 7.2(a, b), respectively. However, the crack formation was observed at the weld interface and P92 steel HAZ for the specimen tested at a temperature of 450 °C and 550 °C as shown in the inset image. This indicates the poor bonding between the 304L and ERNiFeCr-2 filler at high-temperature. The specimen tested at 650 °C, 750 °C, and 850 °C failed from the P92 steel BM region, as shown in Figure 7.2(c-e). The UTS of the 630 MPa was observed at room temperature, and it was decreased to 50 MPa at 850 °C. The UTS was decreased with the increase of the temperature.

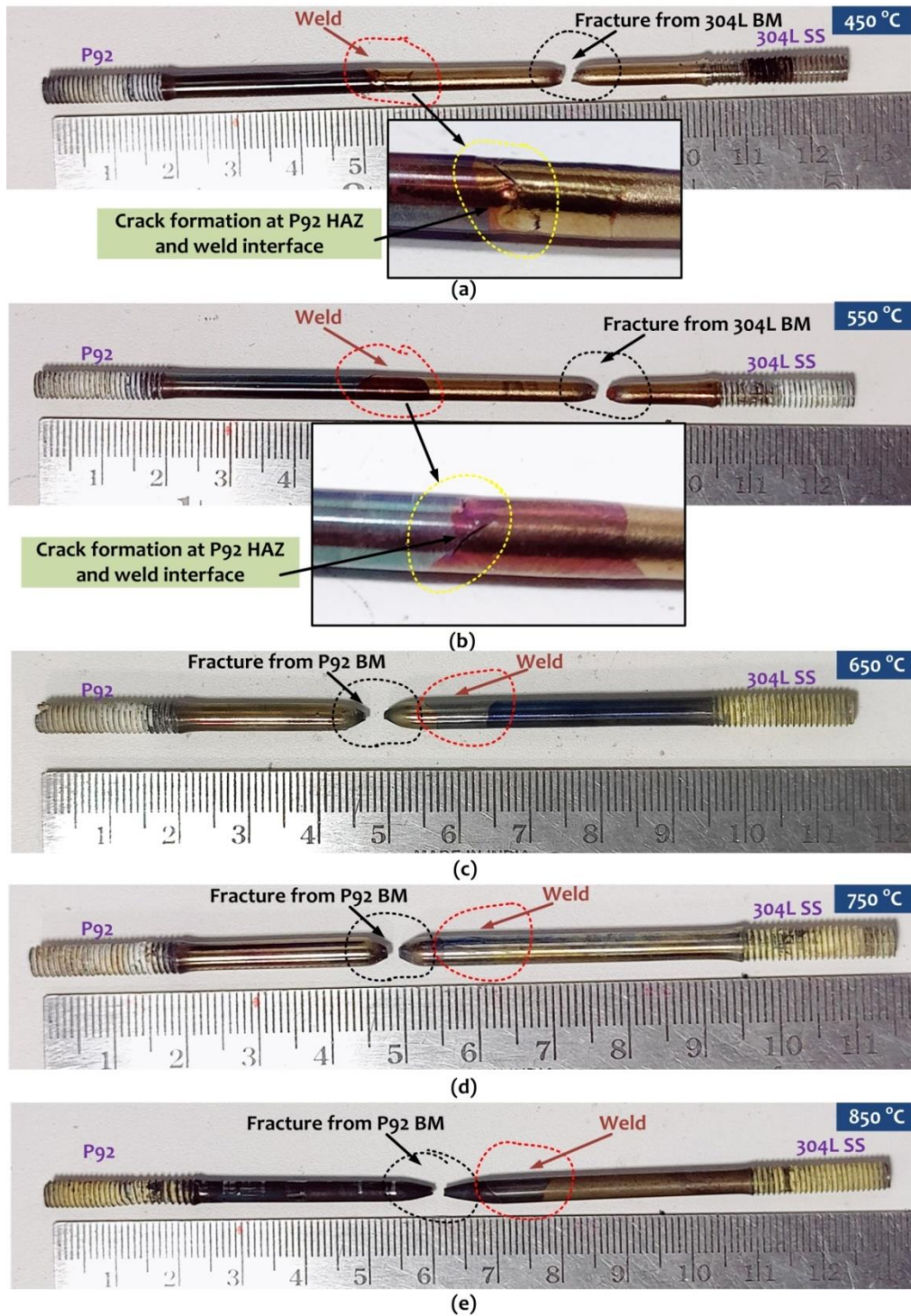


Figure 7.2: Photograph of the ruptured high-temperature tensile specimen (P92-ERNiFeCr-2-304L DWJ), showing the fracture location at different test temperature (450 °C-850 °C)

The effect of test temperature on the strength of the P92-Thermanit MTS 616-304L DWJ was also investigated. The appearance of the fractured high-temperature tensile test specimens is shown in Figure 7.3(a-e). From the test, it was observed that the specimen tested at a temperature of 450 °C and 550 °C failed from the 304L BM. Also, the P92 HAZ region was safe during the room temperature tensile test. The specimen tested at a temperature of 650 °C, 750 °C, and 850 °C failed from the P92 BM. Although the high-temperature tensile test was conducted at a wide range of temperatures, the weld fusion zone did not undergo any significant deformation.

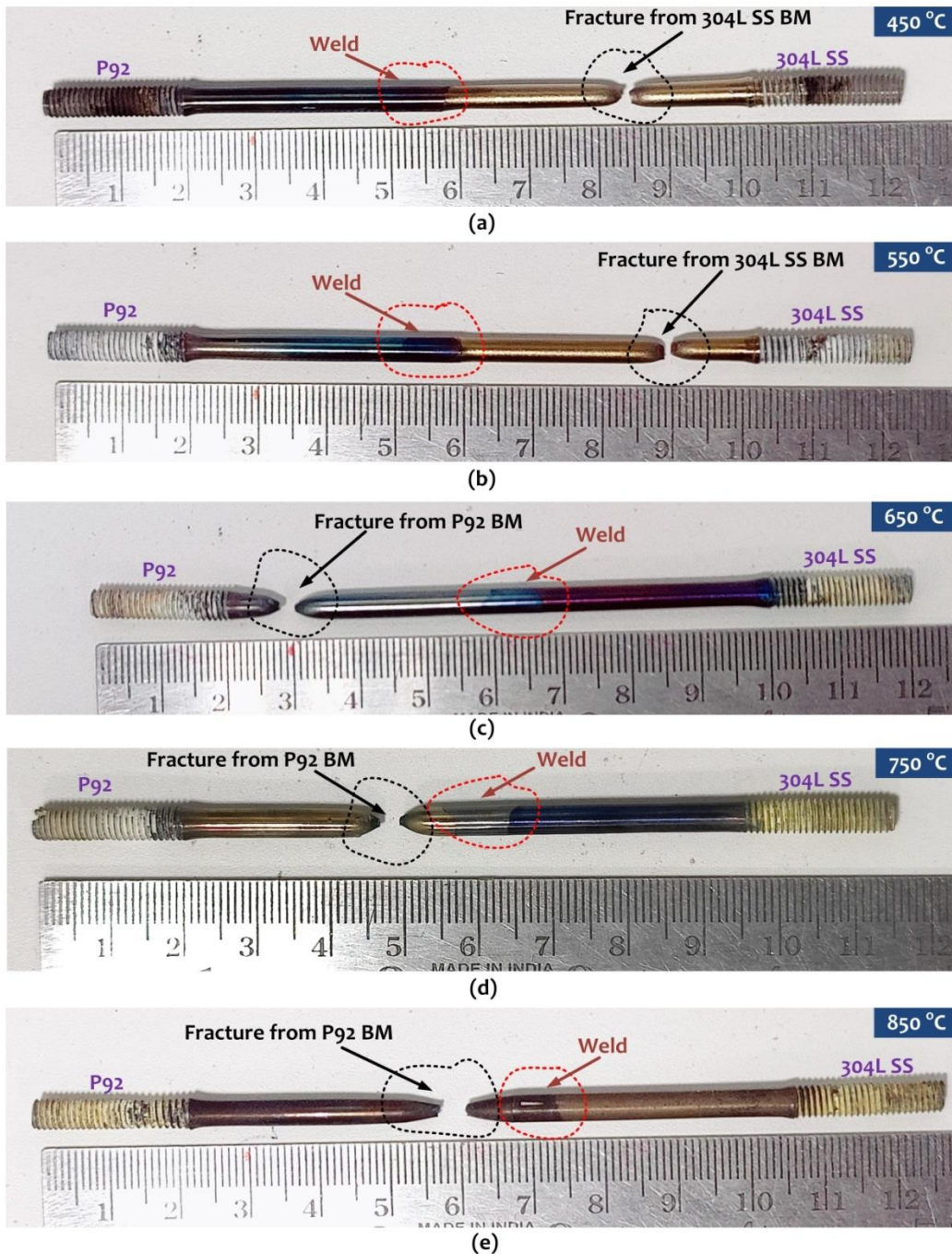


Figure 7.3: Photograph of the ruptured high-temperature tensile specimen (P92-Thermanit MTS 616-304L DWJ), showing the fracture location at different test temperature (450 °C-850 °C)

The stress-strain curve obtained at room and elevated temperature for P92-ERNiCrMo-3-304L DWJ is shown in Figure 7.4. The stress-strain curve indicated the serrated flow/jerky flow behavior during the plastic deformation for the high-temperature test. The strain aging process hardens the material. Thus, the stress required to deform the material increases after strain/work hardening. Strain aging can also occur during plastic deformation; in that case, it is called dynamic strain aging (DSA). DSA occurs at a particular temperature and strain rate. DSA is an unwanted phenomenon as it decreases the ductility and toughness of the material. DSA occurs due to the pinning of the dislocations by solute atoms. This phenomenon is also called serrated flow behavior, as the stress-strain curve exhibits serration during the plastic deformation. DSA results from the interaction between dislocations and solute atoms during plastic deformation. When dislocations encounter impediments during plastic deformation, they are momentarily halted. As the solute atoms diffuse around the pinned dislocations during this period, the

hold of the barrier on the dislocations is further strengthened. With enough tension, these dislocations eventually pass the obstacles and proceed to the next one where they are stopped, allowing the process to continue. This interaction will result in the inhomogeneous deformation behavior, called serration in the stress-strain curve, as seen from the stress-strain curve at 450 °C and 550 °C. The serrated flow occurs due to the locking and releasing of the dislocations by the solute atom, which results in the jerky flow during the high-temperature tension test (Rahman et al., 2009). The Portevin and Le-chatelier developed this theory of serration, hence also called the Portevin Le-chatelier (PLC) effect. There are three types of serration behavior depending on the profile of the stress-strain curve: Type A, Type B, and Type C. The serration type varies with the temperature and strain rate (Sainath et al., 2014). According to results in the literature, DSA mostly happens in 304L (Antoun et al., 2019; Zergani et al., 2020). Also, the DSA happens at a specific strain rate and temperature (Sainath et al., 2014). DSA primarily happens in alloys with interstitial solute atoms because these atoms can quickly and forcefully segregate to dislocations and lock them during straining. At elevated temperatures and low strain rates, the mobility of solute atoms in this steel increases to the point that they can follow a dislocation during its motion and segregate to its core while it has to wait in front of an obstacle. DSA happens when the repeated interaction of interstitial solute atoms prevents dislocation motion. Serrations or wavy patterns resembling saw teeth in the stress-strain plot are seen in this situation. Most studies have shown that the serrated flow is suppressed with increasing strain rate and decreasing temperature (Gupta et al., 2013). The room temperature and elevated temperature UTS value, YS value, % elongation, and % reduction in cross-sectional area of the P92/304L DWJ prepared using ERNiCrMo-3 filler are listed in Table 7.1. From the high-temperature tensile test results, it was noticed that as the test temperature increases, the time to failure of the P92/304L DWJ decreases. The metallurgical changes toward the specimen failed from the 304L and P92 side were examined by taking the cross-section of the fracture surface.

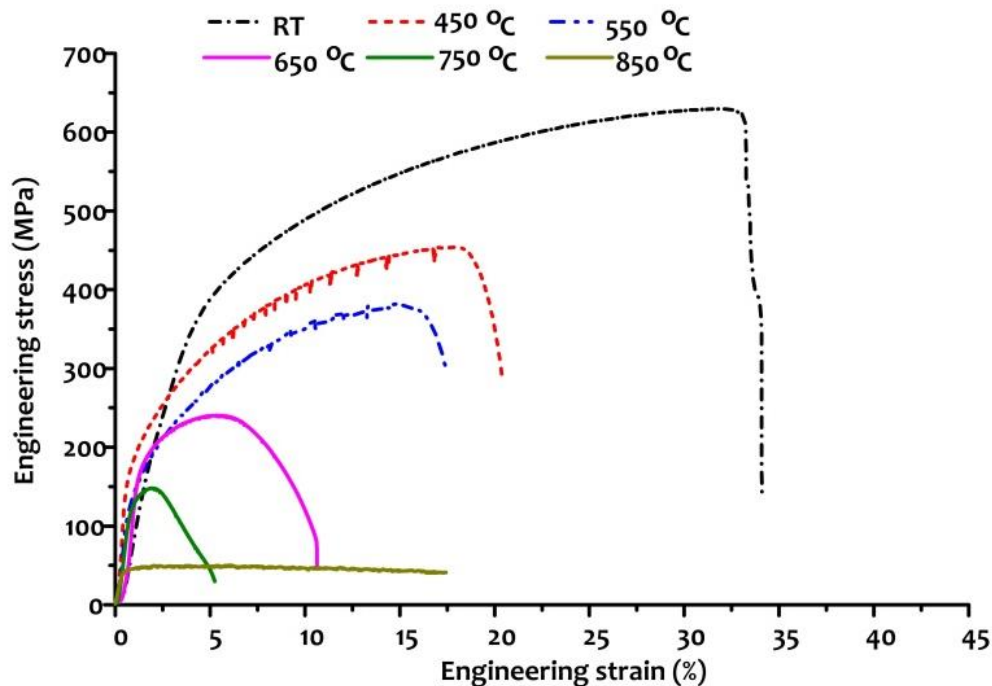
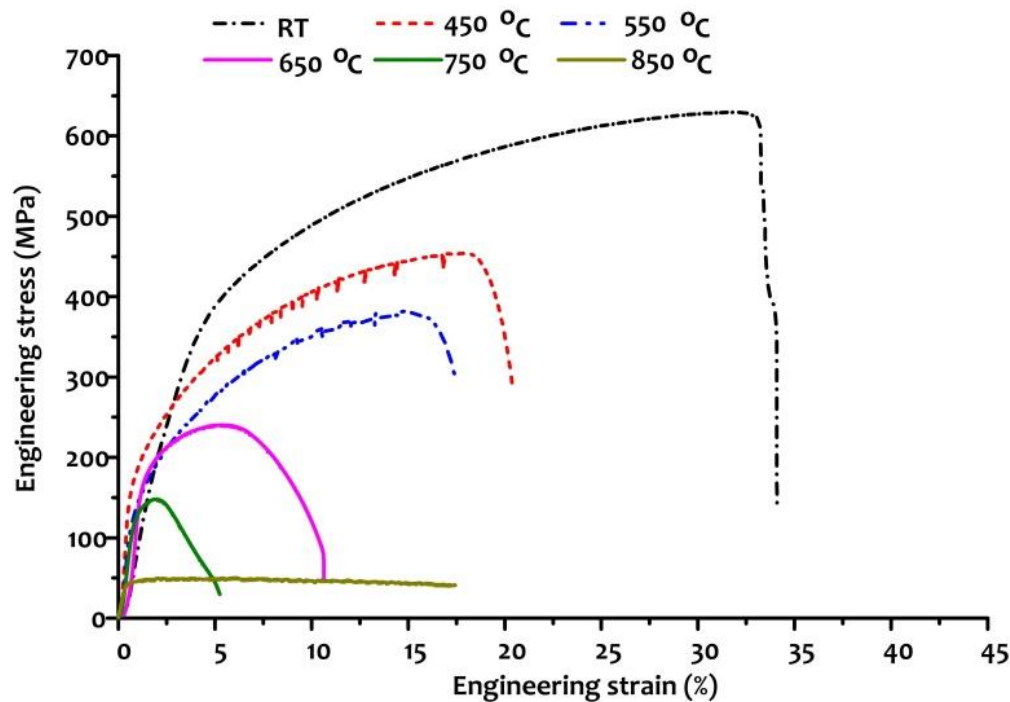


Figure 7.4: Engineering stress-strain curve at room and elevated temperature (450 °C-850 °C), serrated flow behavior can be observed from the curve at 450 °C and 550 °C temperature (P92-ERNiCrMo-3-304L DWJ)

Table 7.1 Room temperature and high-temperature tensile strength of the P92-ERNiCrMo-3-304L DWJ

Temperature (°C)	Ultimate tensile strength (MPa)	Yield strength (MPa)	% Elongation	% Reduction in cross-section area	Fracture location
25 (Room temperature)	588	375	31	-	304L BM
450	423	168	13.7	93	304L BM
550	396	152	12.33	88	304L BM
650	290	130	4.11	87	P92 BM
750	107	90	5.48	94	P92 BM
850	47	35	9.59	91	P92 BM

Figure 7.5 shows the stress-strain curve obtained at room temperature and at the temperature range of 450 °C-850 °C for P92-ERNiFeCr-3-304L DWJ. The obtained UTS and YS values and tensile ductility (% elongation and % reduction in cross-sectional area) for specimens tested at different temperatures is given in Table 7.2. The serrated flow behavior at elevated temperature test can be observed from the stress-strain curve. The P92-ERNiFeCr-3-304L DWJ exhibited serration flow behavior when tested at 450 °C and 550 °C.

**Figure 7.5:** Engineering stress-strain curve at room and elevated temperature (450 °C-850 °C), serrated flow behavior can be observed from the curve at 450 °C and 550 °C temperature (P92-ERNiFeCr-2-304L DWJ)**Table 7.2** Room temperature and high-temperature tensile strength test results of the P92-ERNiFeCr-2-304L DWJ

Test Temperature (°C)	Ultimate tensile strength (σ_u) (MPa)	Yield Strength ($\sigma_{y(0.2)}$) (MPa)	% Elongation (%)	% Reduction in cross-section area (%)	Fracture location
25 (Room temperature)	630	228	34	41	304L BM
450	454	150	14	87	304L BM

550	382	135	15	85	304L BM
650	240	120	8.5	83	P92 BM
750	148	105	9.5	85	P92 BM
850	50	38	20	86	P92 BM

Figure 7.6 shows the stress-strain curve obtained at room temperature and elevated temperature for P92-Thermanit MTS 616-304L DWJ. The serrated/jerky flow behavior of the P92/304L DWJ at 450 °C and 550 °C test temperature can be observed from the stress-strain curve. The tensile strength of the 660 MPa was observed at room temperature, and it was decreased to 58 MPa at 850 °C. The details of the UTS value, YS value, % elongation, and % reduction in area for specimens tested at different temperatures is listed in Table 7.3.

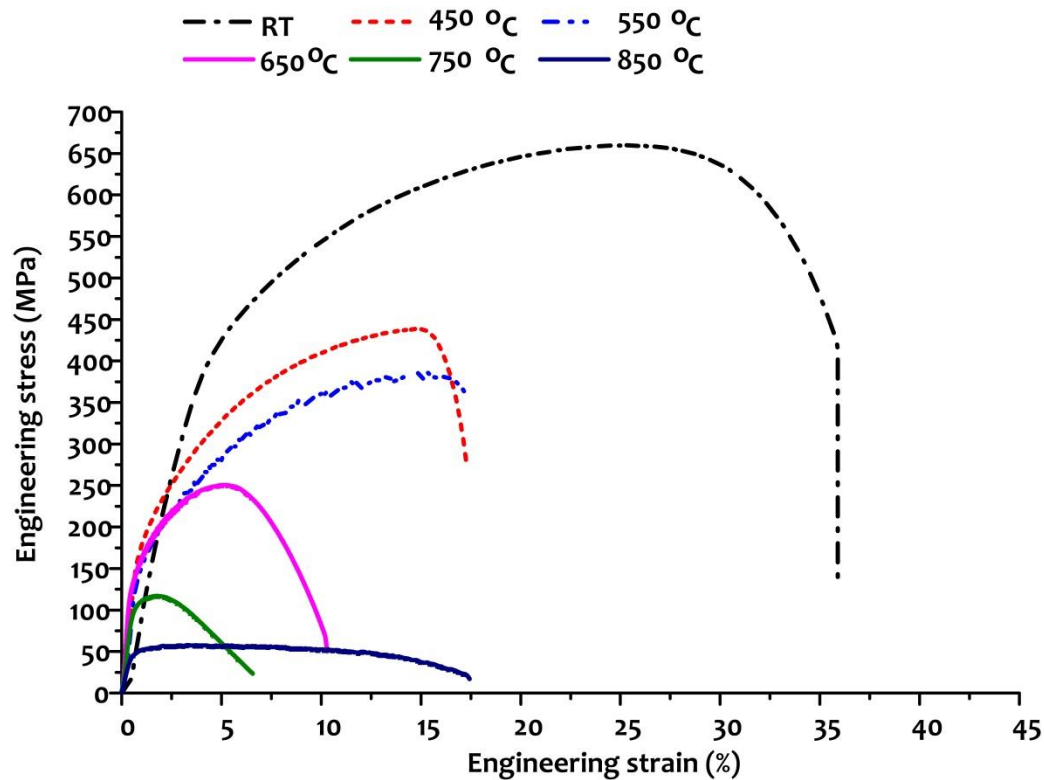


Figure 7.6: Engineering stress-strain curve at room and elevated temperature (450 °C-850 °C) (P92-Thermanit MTS 616-304L DWJ)

Table 7.3 Room temperature and high-temperature tensile test results of the P92-Thermanit MTS 616-304L DWJ

Test Temperature (°C)	Ultimate tensile strength (σ_u) (MPa)	Yield Strength ($\sigma_{y(0.2)}$) (MPa)	Percentage Elongation (%)	Percentage reduction in cross-section area (%)	Fracture location
25 (Room temperature)	660	291	35	-	304L BM
450	439	155	14	93	304L BM
550	387	125	13	88	304L BM
650	251	110	16	90	P92 BM
750	117	90	17	91	P92 BM
850	58	41	20	94	P92 BM

7.2 Macrostructure of the fractured high-temperature tensile test specimens

Figure 7.7(a-e) shows the macrograph of the cross-section of the fractured high-temperature tensile test specimen for P92-ERNiCrMo-3-304L DWJ. The macrograph shows the fracture specimen consisting of P92 BM, 304L BM, ERNiCrMo-3 weld fusion zone and P92 HAZ in gauge length. The macrographs showed no significant deformation in the weld fusion zone after the high-temperature tensile test. The weld fusion zone was free from cracking and deformation for all the test temperature. This indicates the reliability of the ERNiCrMo-3 filler metal for elevated temperature service.

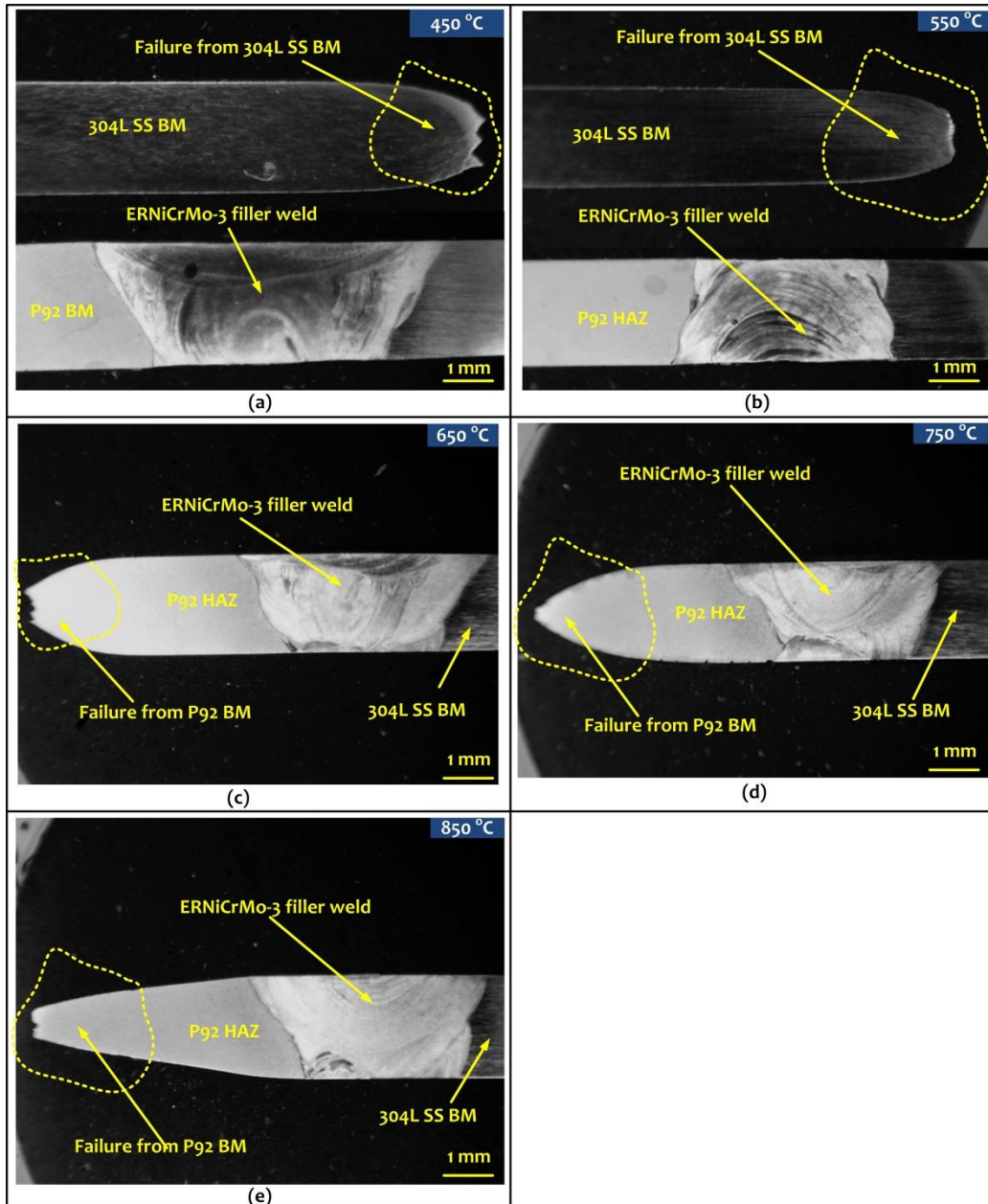


Figure 7.7: (a, b) showing macrostructure of the tensile tested sample at 450 °C and 550 °C, which was taken from the localized portion of the 304L BM sample and (c-e) showing macrostructure of the tensile tested sample at 650 °C, 750 °C and 850 °C, which was taken from the localized portion of the P92 BM sample (P92-ERNiCrMo-3-304L DWJ)

The macrograph of the cross-section of the fractured high-temperature tensile test specimens for P92-ERNiFeCr-2-304L DWJ is displayed in Figure 7.8(a-e). The macrograph showed that the cracking had occurred at the weld fusion zone and P92 HAZ interface. The crack formation generally originates from the weaker section of the interface (unmixed

zone). The crack formation occurs during tension and it indicates the weak bonding of the ERNiFeCr-2 filler with 304L BM at elevated temperature.

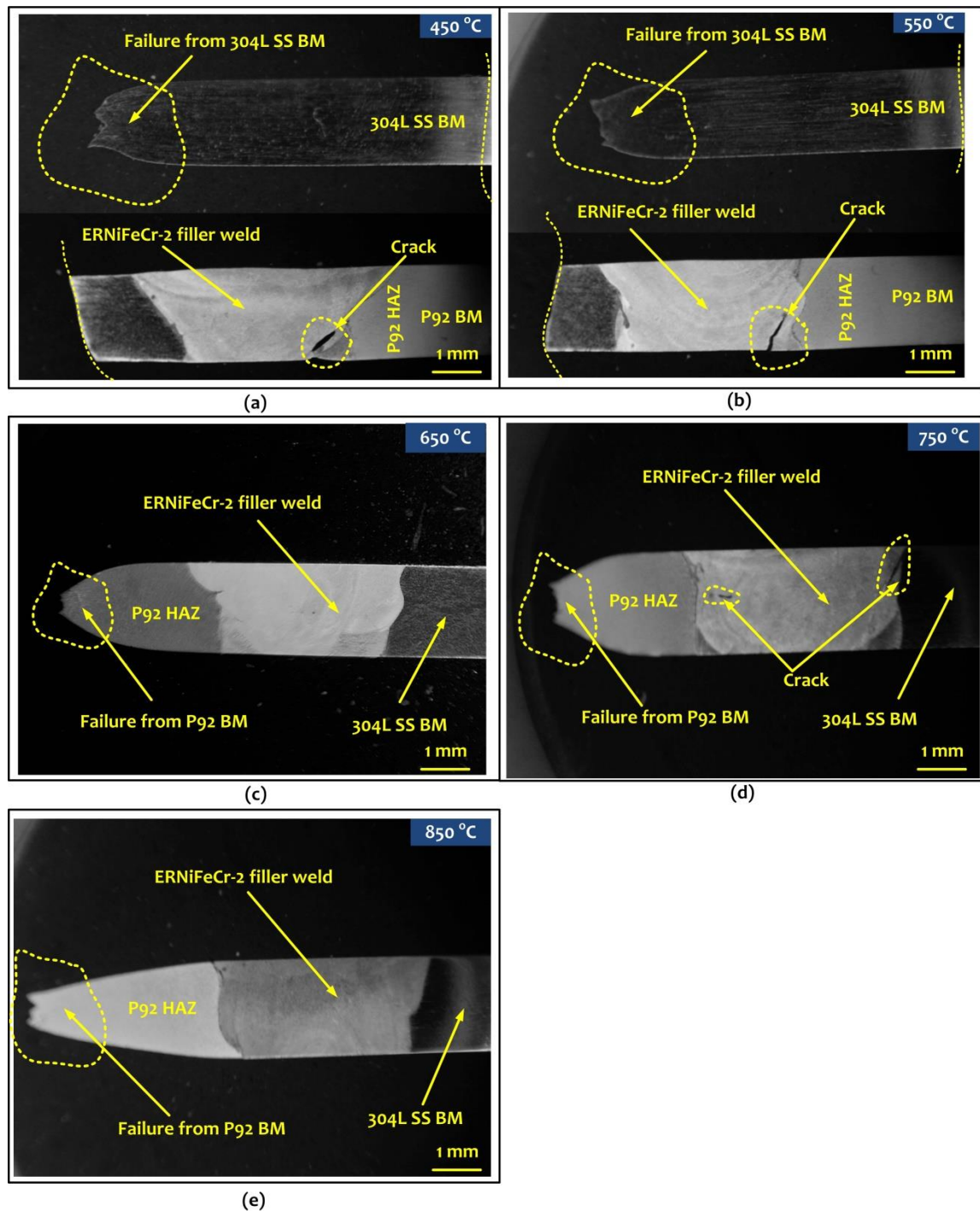


Figure 7.8: (a, b) shows the macrostructure of the tensile tested sample at 450 °C and 550 °C, which was taken from the localized portion of the 304L BM and (c-e) shows the macrostructure of the tensile tested sample at 650 °C, 750 °C and 850 °C, which was taken from the localized portion of the P92 BM (P92-ERNiFeCr-2-304L DWJ)

The macrograph of the cross-section of the fractured high-temperature tensile specimens (P92-Thermanit MTS 616-304L DWJ) tested at a temperature of 450 °C, 550 °C, 650 °C, 750 °C, and 850 °C is shown in Figure 7.9(a-e). From the macrograph, no cracking and voids were observed in the weld fusion zone. The flat fracture surface appearance was observed for the specimen tested at 450 °C and 550 °C and failed from the 304L, indicating the possible brittle kind of fracture. It is due to the increase in the brittleness of the 304L due to the DSA phenomenon. The obvious necking was observed for all specimens failed

from the P92 steel. The further microstructure characterization near the necking region was carried out using field emission scanning electron microscope (FESEM) and optical microscope.

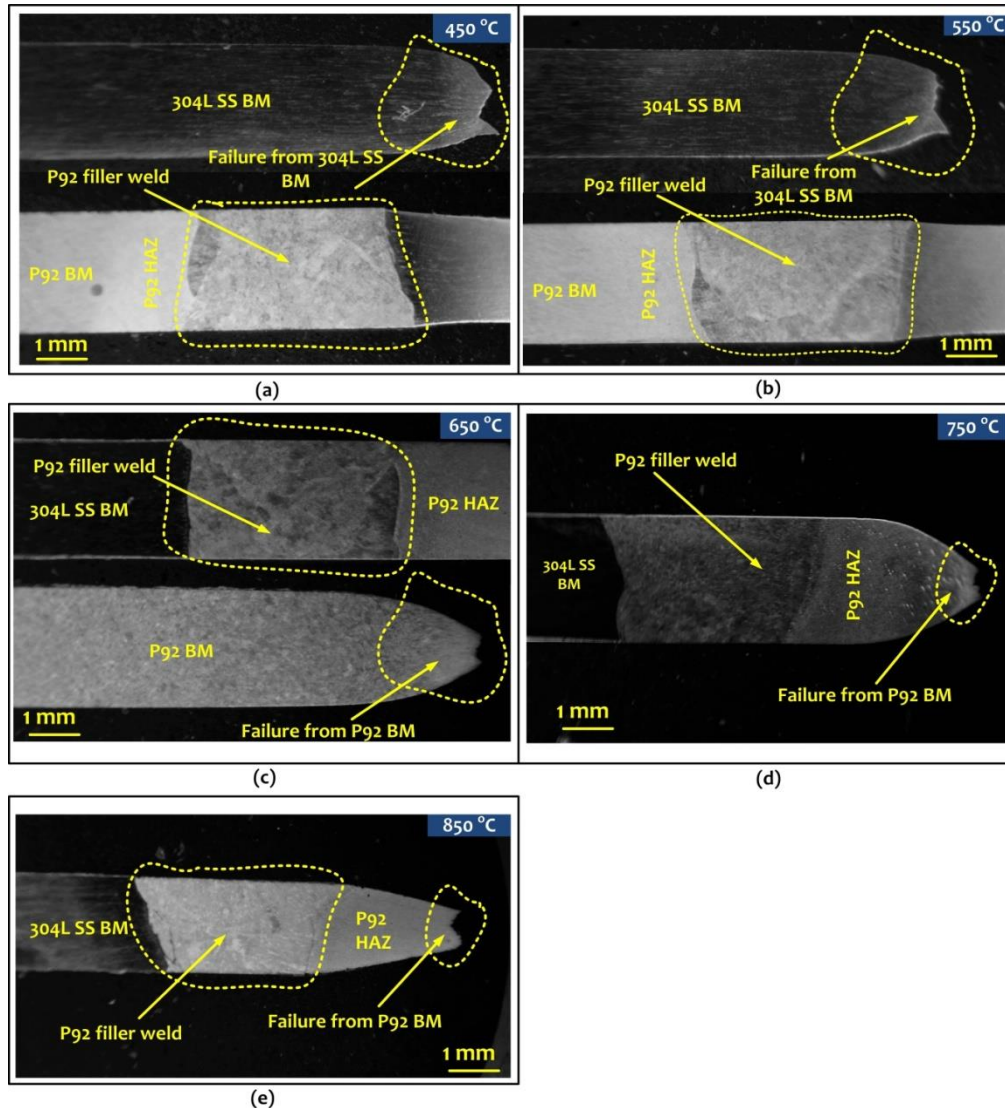


Figure 7.9: (a-e) shows the macrostructure of the cross-section of the fracture high-temperature tensile sample, tested at 450 °C-850 °C temperature (P92-Thermanit MTS 616-304L DWJ)

7.3 Microstructure changes after high-temperature tensile test

The major microstructural variation was observed near the fracture tip surface. The specimen tested at 450 °C and 550 °C failed from the 304L BM. The failure from the 304L BM during the test at 450 °C and 550 °C was easily identifiable, as HAZ formation was insignificant toward the 304L side. The microstructure near the fracture region of these high-temperature tensile specimens is shown in Figure 7.10(a-f). The as-received 304L austenitic stainless steel primarily consists of austenite grains, twins, and ferrite stringers, as already discussed in previous section. The 304L is a non-heat treatable alloy. Thus, it does not undergo any phase transformation or microstructural evolution at elevated temperature service. It undergoes only strain/work hardening during plastic deformation at elevated temperature. The austenite grains and twins were stretched during the elevated temperature tensile test. Thus, the specimen tested at 450 °C and 550 °C and fractured from the 304L does not contain twins and austenite grains in its microstructure, as shown in Figure 7.10(a-f). The elongated ferrite stringers were only observed after the high-temperature tensile test. (Song et al., 2014) also observed similar microstructure characteristics for the HR3C advanced austenitic steel after the high-temperature tensile

test. It should be noted that the specimen fractured from the 304L at 450 °C and 550 °C showed the similar kinds of microstructural characteristics for all the filler metal.

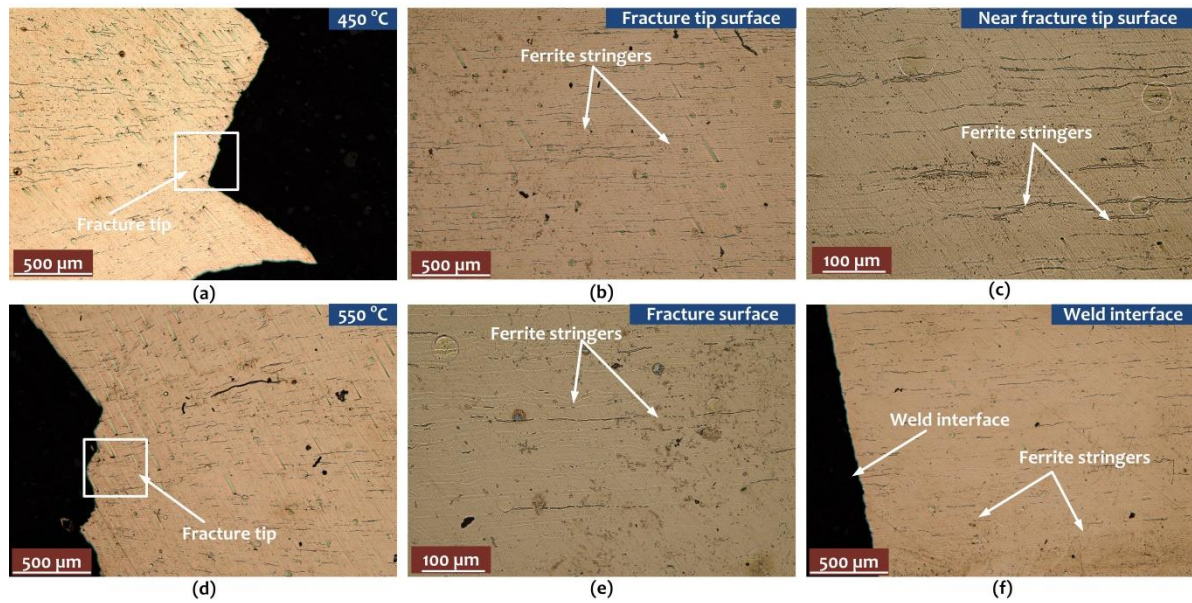


Figure 7.10: (a) Optical microstructure of the tensile specimen tested at 450 °C: fracture region (b) microstructure of the fracture region (c) microstructure near the fracture tip surface. (d) Optical microstructure of the tensile specimen tested at 550 °C: fracture surface (e) microstructure near the zone of failure and (f) microstructure at the weld interface

The microstructure of the specimen tested at 650 °C, 750 °C, and 850 °C, and fractured from the P92 BM is shown in Figure 7.11(a-i). The microstructural observation near the fracture surface showed the absence of the lath martensite structure and prior austenite grains (PAGs). The precipitate (possibly $M_{23}C_6$ and MX type) of different sizes was observed near the fracture surface. Also, very few prior austenite grain boundaries (PAGBs) were observed near the fracture tip surface. Many cavity formations near the necking region indicate that the specimen has undergone a significant plastic deformation before fracture. The plastic deformation causes the softening of the matrix. Thus, this region becomes weaker during the tensile test and decreases the tensile strength. Small cracks and voids were formed near the necking surface. Crack formation and crack propagation are the two main steps of fracture. The slow crack propagation leads to ductile fracture, while the rapid crack propagation leads to brittle fracture. The tension load will create the necking at the low-strength region. As time passes, small micro-void formation occurs at the necking region. In the later stage, the coalescence of the micro-voids takes place. In this stage, the micro-voids become sufficiently large to form a crack. The number of cavities near the fracture tip surface increased with the increase of the test temperature, as shown in Figure 7.11(a, d and g). After the high-temperature tension test, the microstructure near the weld interface, coarse grain heat affected zone (CGHAZ) and fine grain heat affected zone (FGHAZ) region was also investigated. It should be noted that fewer voids were observed in the CGHAZ and FGHAZ regions, indicating the less effect of the applied load and temperature on the microstructural properties of the P92 HAZ region. The large number of voids formation near the fracture tip surface indicates that the specimen tested at high temperature undergoes significant plastic deformation before fracture. The extensive void formation also indicates the ductile mode of the failure at a temperature of 650 °C, 750 °C, and 850 °C. The tensile test was performed at 650 °C, 750 °C, and 850 °C, which is approximately the tempering temperature of the P92 steel. At this temperature, new carbide precipitates and martensite lath degradation (formation of the sub-grain from the tempered martensite lath structure) occurs. The microstructure degradation toward the P92 side is the main reason behind the failure of the P92/304L

DWJ at high-temperature from the P92 BM region. The P92 steel loses its properties after 650 °C due to the dissolution of the precipitates and weakening of the grain boundaries.

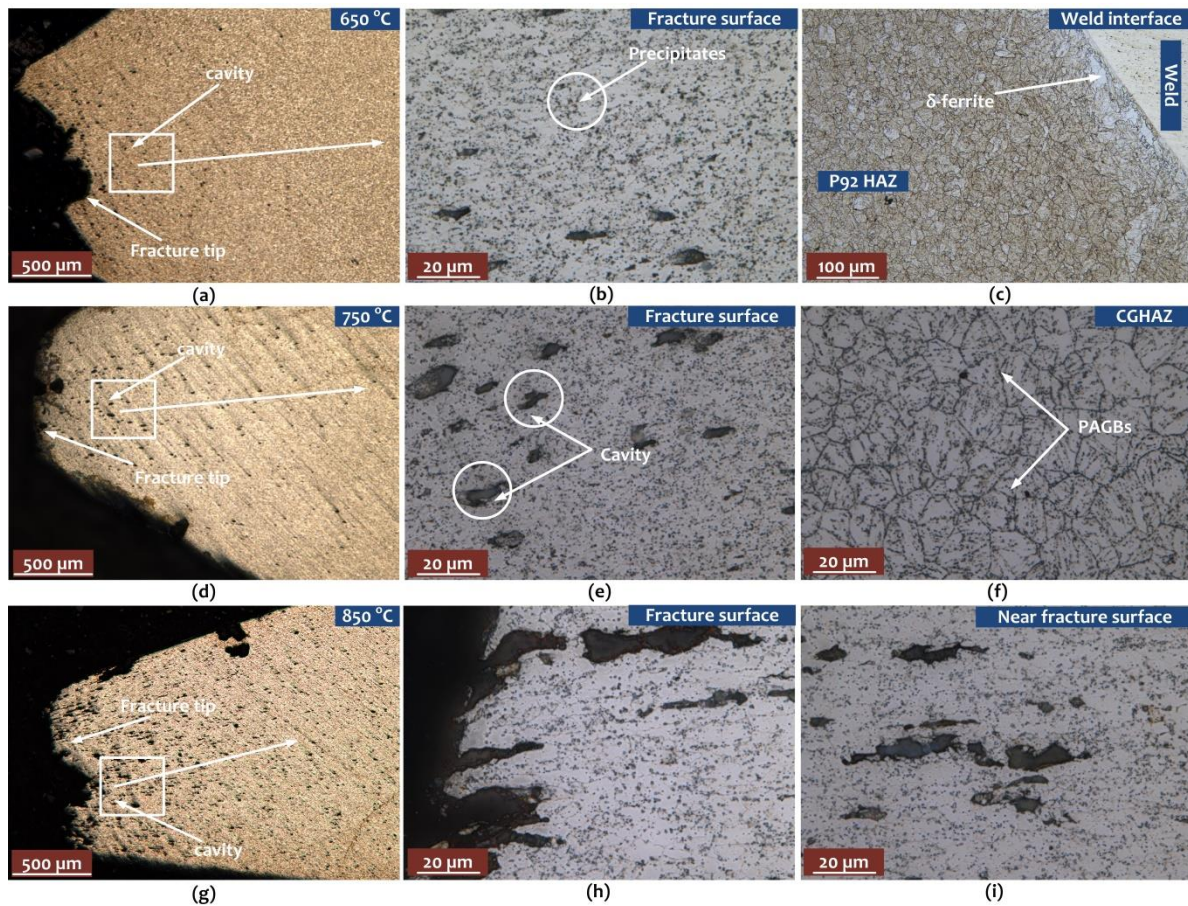


Figure 7.11: Optical microstructure of the tensile tested specimen till fracture; consisting of fracture region, micro and macro voids can be observed (a) Optical microstructure of the fracture tensile specimen tested at 650 °C (b) microstructure of the fracture region (c) microstructure at the weld and P92 HAZ interface. (d) Optical microstructure of the fracture tensile specimen tested at 750 °C (e) microstructure near the zone of failure (f) microstructure of the P92 CGHAZ region. (g) Optical microstructure of the fracture tensile specimen at 850 °C (h) microstructures of the fracture region and (i) microstructure near the fracture region

The mechanical properties are linked with the microstructure of the material. The normalizing heat treatment for the P92 steel produces the lath martensite structure with high dislocation density. The tempering heat treatment after normalizing produces the tempered martensite structure with PAGs and sub-grains boundaries. The small needle shape lath martensite structure transform into the large equiaxed grains. The microstructure of the P92 steel mainly consists of PAGs, PAGBs, lath/sub-grain boundaries, and precipitates ($M_{23}C_6$ and MX). The PAG is further subdivided into the lath martensite packets, which consist of lath martensite blocks. The lath martensite block is composed of sub-grains with high dislocation density. The strength of the P92 steel is due to its tempered martensite lath structure and precipitates. This martensite lath structure and precipitates pin the movement of the dislocation and maintain the strength of the P92 steel. The FESEM microstructure of the specimen prepared from the cross-section of the fractured high-temperature tensile specimen near the fracture tip surface is shown in Figure 7.12(a-i). The softening of the lath martensite structure near the necking region and the formation of voids soften the necking region. The $M_{23}C_6$ precipitates are decorated along the PAGBs. The FESEM micrograph shows that the cavities formations occur from the PAGBs and at triple points. From the literature, it was found that the $M_{23}C_6$ are comparatively less stable at higher temperature than the MX precipitates. Due to the

difference in the co-efficient of thermal expansion between the precipitates and the base metal, these precipitates coarsen during the high-temperature tensile test. Also, the bonding strength between the precipitates and the base metal reduces. The coarse precipitates generally separate from the base metal, and thus high-stress concentration is created. Thus, during the high-temperature tensile test, the crack is initiated from the coarse $M_{23}C_6$ precipitates due to the high-stress concentration in this region. The crack grows faster with the increase of stress, which leads to the failure of the welded joint (Lei et al., 2015). The maximum working temperature of the P92 steel is 620-625 °C (Xue et al., 2012). Under elevated temperature tensile test, elongated grain boundaries were observed near the fracture tip surface. The grain boundaries are elongated due to the high tensile load applied at elevated temperature. The numerous voids that occurred around the PAGBs after the high-temperature tensile test are shown in Figure 7.12(a-i). The cavity formation occurs mainly along the triple point of the PAGBs (Meng et al., 2014), as shown in Figure 7.12(b, e and h). The large numbers of voids that occur near the fracture surface coalesce together and form cracks. From the elevated temperature tensile test results, it is clear that the P92 steel softens more rapidly than the 304L steel with increasing temperature. During the high-temperature tensile test, the growth of the $M_{23}C_6$ precipitates in the P92 steel increases, resulting in the loss of the precipitation strengthening effect due to the segregation of the important Cr and Ni elements from the base metal matrix. The Energy-dispersive-spectroscopy (EDS) point analysis was conducted near the fracture tip surface containing coarse and fine secondary phase particles around the cavity. The EDS spectrum result was C: 7.82%, Cr: 8.98%, Fe: 75.86%, Nb: 3.72%, W: 2.28%, Mo: 1.08%, Ni: 0.33%. The result indicated that the $M_{23}C_6$ precipitates near the cavity. The W element presents in the $M_{23}C_6$ precipitates retard the coarsening rate of these particles. At high temperature, carbide-forming elements such as Cr and C from the matrix diffuse into the $M_{23}C_6$ precipitates, hence increasing the coarsening of these particles (Godec and Balantič, 2016). The grain boundaries decorated with the $M_{23}C_6$ precipitates mainly obstruct the dislocation movement under applied tension load. At high temperature, because of the coarsening of the $M_{23}C_6$ precipitates, the pinning effect provided by these precipitates gets suppressed. Thus, at high-temperature, failure takes place from the P92 steel.

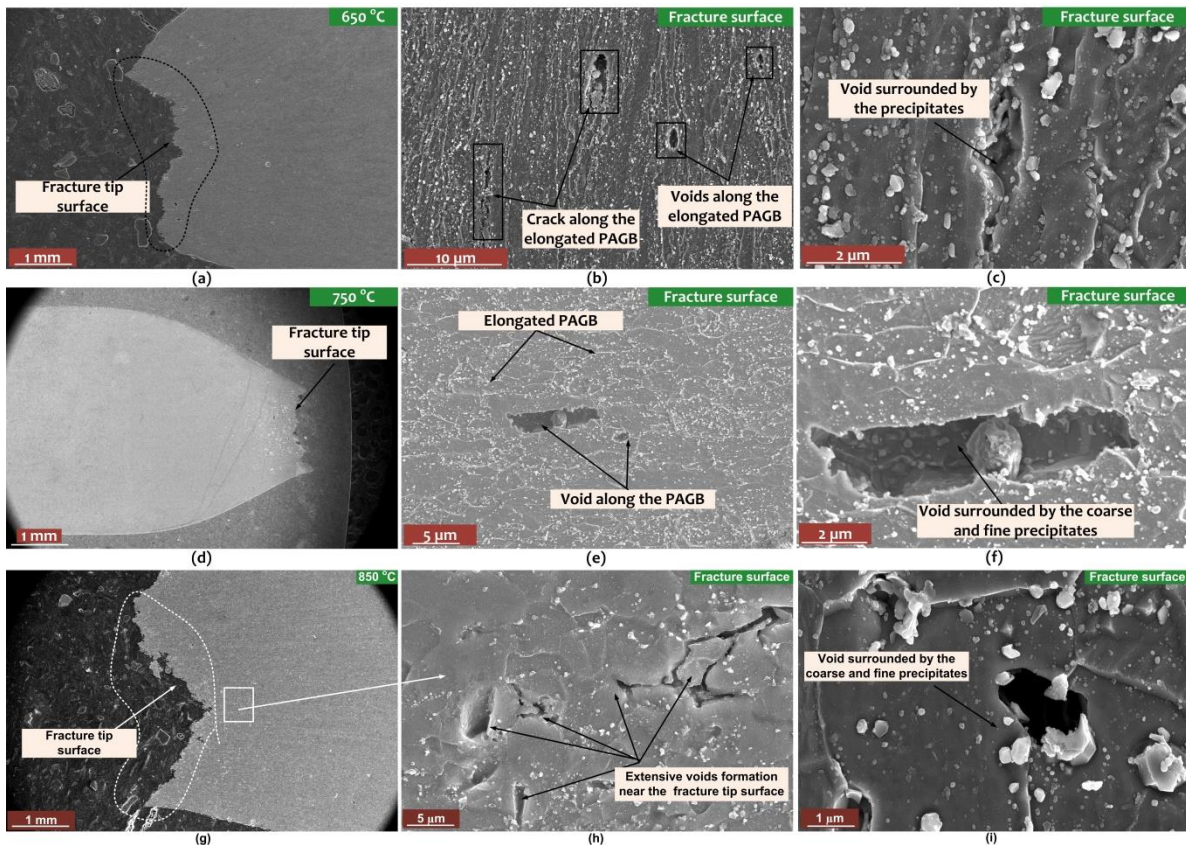


Figure 7.12: (a) Fracture tip surface of the specimen tested at 650°C (b) microstructure near the fracture tip surface, cavities along the elongated grain boundaries can be seen (c) microstructure taken at higher magnification to reveal cavities and precipitates (d) cross-sectional view of the fracture surface for the specimen tested at 750 °C (e) microstructure near the fracture tip surface (f) microstructure taken at higher magnification to reveal cavities and precipitates (g) cross-sectional view of the fracture surface for the specimen tested at 850°C (h) extensive voids formation near the fracture tip surface due to the applied tension load at high-temperature and (i) cavities and coarsening of the precipitates can be seen at the fracture surface

7.4 Micro-hardness distribution after high-temperature tensile test

Based on the micro-hardness distribution, the precise fracture location can be estimated. The micro-hardness variation at each point also shows the extensive microstructural evolution and microstructural inhomogeneity after tensile testing at different temperatures. The Vickers micro-hardness test was conducted over the same specimen used for microstructural characterization consisting of the necking region, P92 HAZ, P92 BM, weld fusion zone, 304L BM, and 304L HAZ. The indentations were taken at a regular interval of 0.25 mm by applying a load of 500 gf for 10 s. The Vickers microhardness distribution across the cross-section of the fracture high-temperature tensile test specimen is presented in Figure 7.13(a-e). The micro-hardness across the P92/304L DWJ after welding is presented in previous section. The microhardness value obtained after welding was also taken into consideration to approximate the fracture location. For all the specimens, the fracture point was selected as the starting point for the hardness measurement ($X = 0$). The cross-section of the fractured tensile test specimen was divided into three parts: fracture point and corresponding BM, weld fusion zone, and un-affected BM. The specimen tested at 450 °C and 550 °C failed from the 304L BM. From the micro-hardness data across the 304L region after the high-temperature tensile test, it was pointed out that the hardness of the 304L was higher than that of the as-received 304L BM (~190 HV). The maximum micro-hardness of the 304L increased to 310 HV near the fracture point after the high-temperature tensile test. The increase in the micro-hardness value of the 304L after the high-temperature tensile test is due to the work hardening or strain

hardening effect (Desu et al., 2016). The DSA phenomenon discussed in the previous section is mainly responsible for the increase in the hardness value of the 304L after the high-temperature tensile test. DSA reduces the ductility and toughness, and increases the brittleness of material. The DSA is a consequence of strain hardening during the plastic deformation. The 304L is a non-heat treatable and ductile metal. Thus, during the plastic deformation, because of the increase in the number density of the dislocation, the hardness of the 304L steel increases. It can be pointed out from the micro-hardness distribution that the micro-hardness near the fracture tip surface was changed dramatically for all the test temperature due to the microstructure changes. From the micro-hardness across the necking surface, it can be confirmed that failure took place from the P92 BM for the specimen tested at 650 °C-850 °C. The low micro-hardness value was observed in the rupture location of the P92 steel (as low as 110-90 HV for the specimen tested at 850 °C). The extremely low micro-hardness value at the fracture tip surface was due to the extensive plastic deformation that occurred in the P92 steel and thus large numbers of voids formation. It was observed that as the tensile test temperature increases, the hardness value of the P92 steel decreases at the fracture location. It should be noted that the hardness of the P92 steel near the fracture region (for the specimen tested at 650 °C-850 °C, and failed from the P92 BM) decreased noticeably after the high-temperature tensile test. The reason why P92 steel shows a low micro-hardness value after a high-temperature tensile test is because of the softening of the lath martensite structure and the coarsening of the precipitates at elevated temperatures. Also, the degradation of the martensite lath structure and the decrease in the dislocation density are mainly responsible for a drop in the micro-hardness value of the P92 steel after the high-temperature tensile test. The hardness value across the P92 HAZ decreases considerably with an increase in the test temperature due to the tempering effect once the high-temperature tensile test is completed. The hardness value of the CGHAZ and ICHAZ region vary significantly for each test specimen after a high-temperature tensile test at different temperature due to the heating at various temperatures (450-850 °C) and then cooling after a fracture. The hardness gradient for the nickel-based weld fusion zone was smaller for the different high-temperature tensile test temperature. It shows the good stability of the nickel-based filler metal at elevated temperature. Figure 7.14 (a-e) shows the variation in micro-hardness across the ruptured high-temperature tensile specimens (P92-ERNiFeCr-2-304L DWJ) after testing from 450-850 °C. Figure 7.15 (a-e) shows the variation in micro-hardness across the cross-section of the ruptured high-temperature tensile specimens (P92-Thermanit MTS 616-304L DWJ) after testing from 450-850 °C.

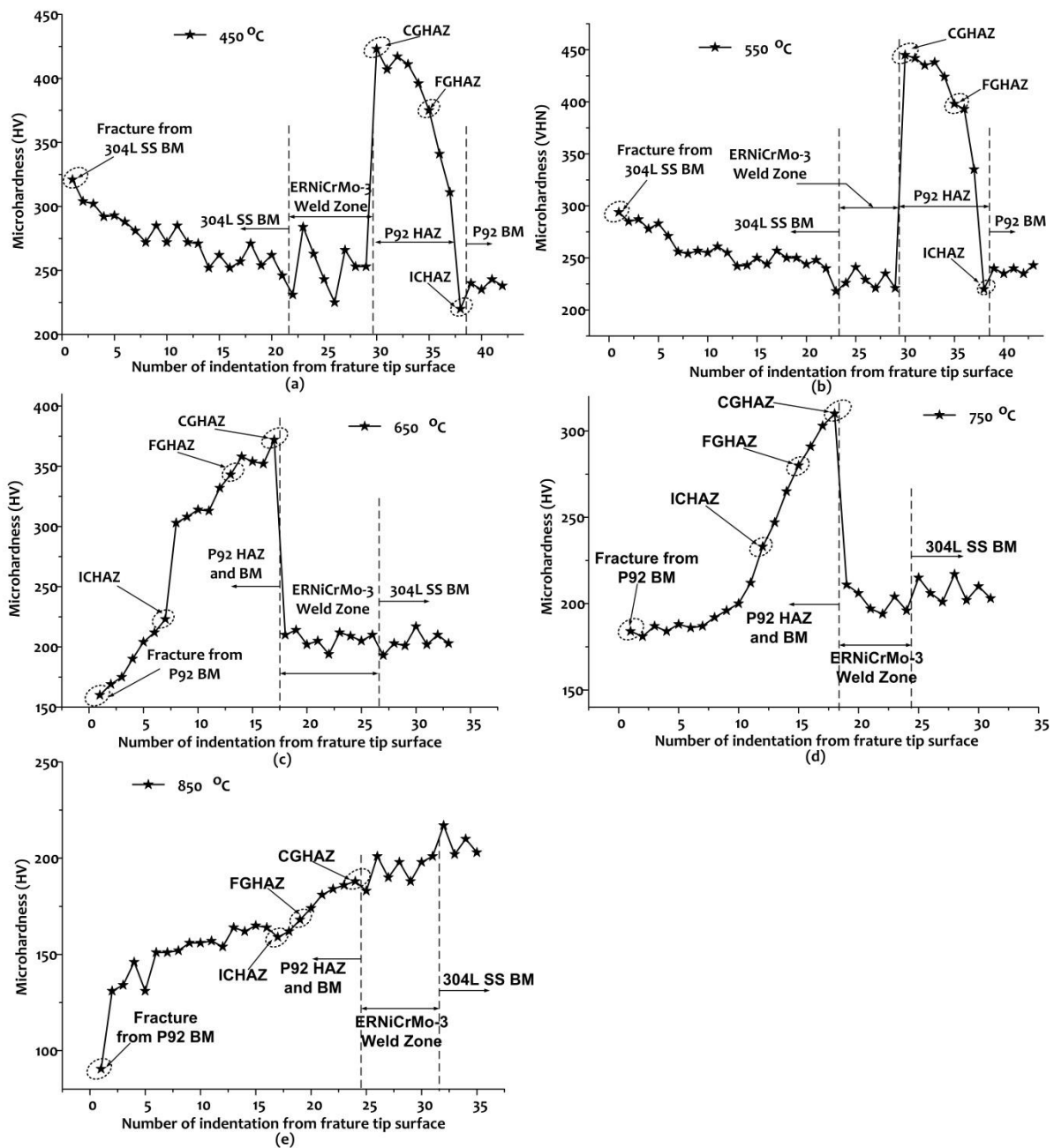


Figure 7.13: (a-e) Micro-hardness profile across the tensile fractured specimens tested at various temperatures in the 450-850 °C range (P92-ERNiCrMo-3-304L DWJ)

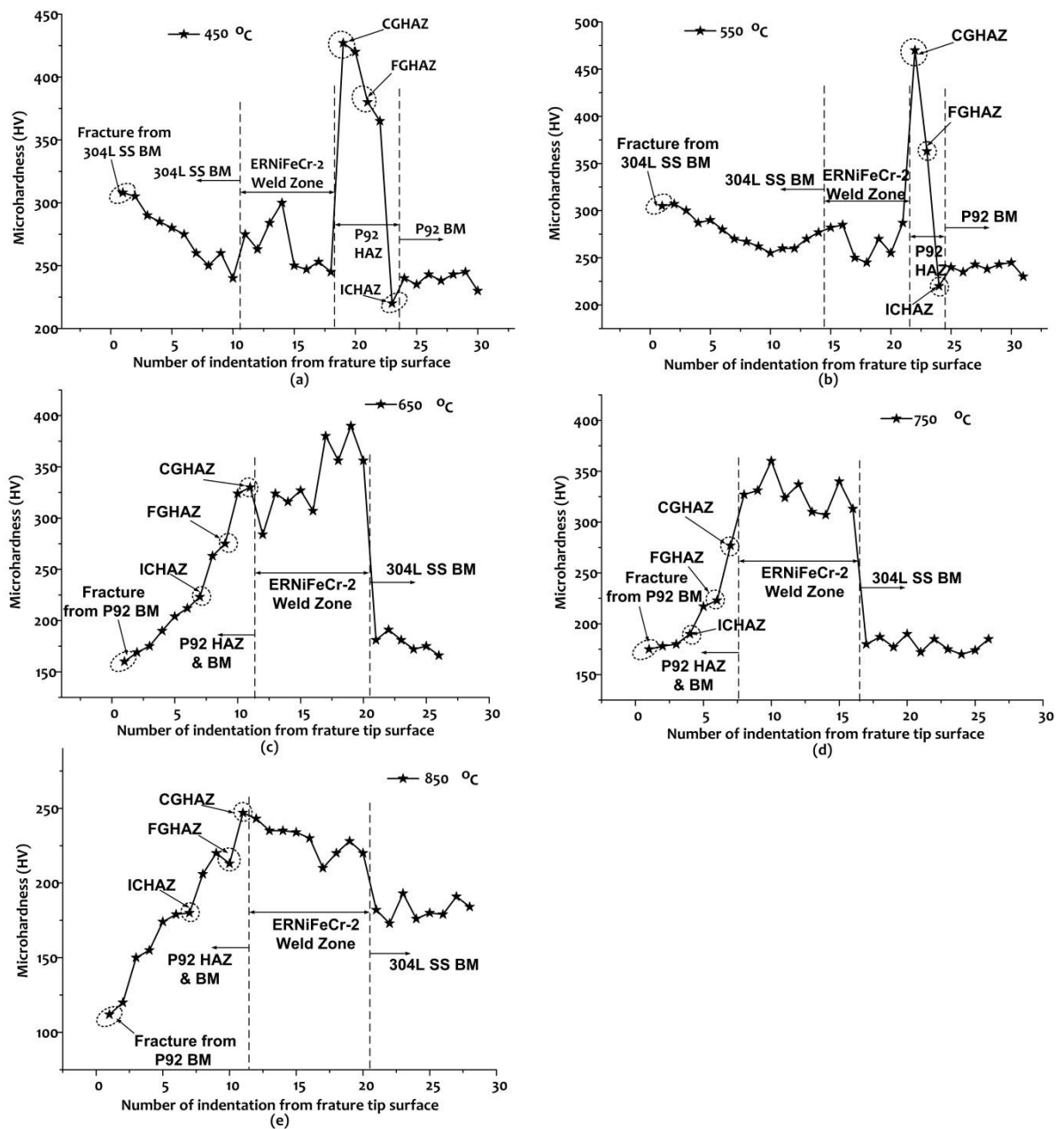


Figure 7.14: Micro-hardness profile across the cross-section of the fractured tensile specimens tested at different temperatures: (a) 450°C (b) 550°C (c) 650°C (d) 750°C and (e) 850°C (P92-ERNiFeCr-2-304L DWJ)

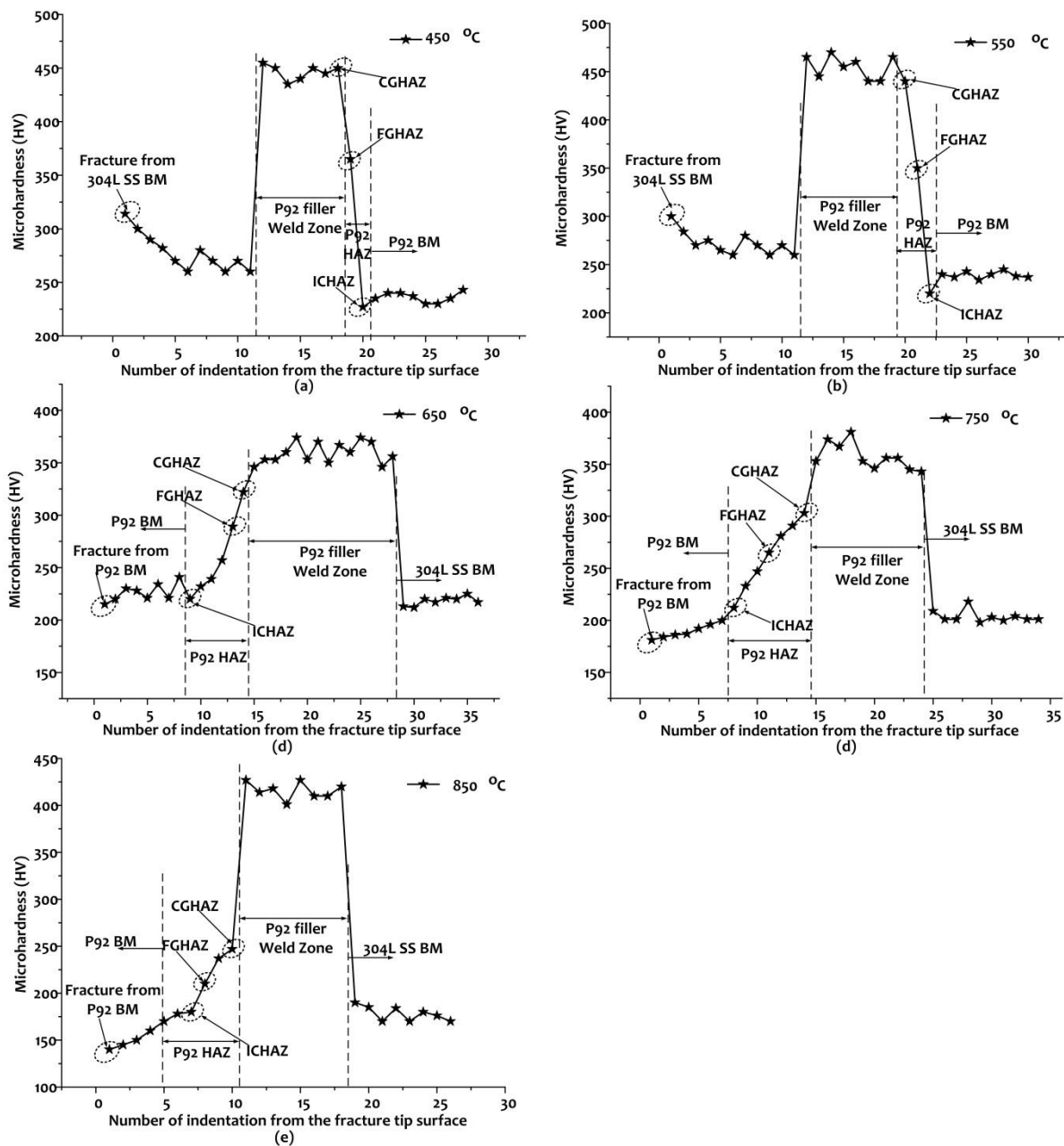


Figure 7.15: (a-e) Micro-hardness profile across the tensile fractured specimens tested in the range of 450-850 °C temperatures (P92-Thermanit MTS 616-304L DWJ)

7.5 Fracture surface analysis of high-temperature tensile specimens

After microstructure and micro-hardness examination across the fracture tip surface, the fracture behavior of the ruptured specimen after tensile test was investigated with the help of SEM. The SEM microstructure of the fracture surface of the specimen at low (50X) and high magnification (1500X and 2500X) is shown in Figure 7.16(a-e). The low magnification image, as shown in Figure 7.16(a-e) was used to measure the reduced diameter of the fractured high-temperature tensile specimen with the help of the SEM measurement tool. The low-magnification image also shows the overall appearance of the fracture surface. The initial diameter of the specimen was 3 mm and the final diameter (reduced diameter) due to the necking during high-temperature tensile test was later measured with the help of low magnification SEM image. The corresponding microstructure of the center region at high magnifications (1500X and 2500X) is shown in the subsequent image (towards right side). It should be noted that the P92/304L dissimilar weld joint fractured from the 304L BM during the tensile test at 450 °C and 550 °C. The fracture surface of the 304L BM is shown in Figure 7.16(a-b). From the low magnification

image of the specimen tested at 450 °C and 550 °C, it can be observed that the fracture surface contains few voids and dimples. The very small size micro-voids were observed from the high-magnification microstructure image. The flat fracture surface appearance was also observed near the voids and dimples. Thus, the ductile-brittle mode of the fracture occurred for the specimen that failed from the 304L BM. Also, due to the DSA phenomenon occurring in the 304L steel, the brittleness of the 304L material increases. This phenomenon is responsible for the brittle mode of the fracture from the 304L. The specimen was fractured from the P92 BM when tested at 650 °C, 750 °C, and 850 °C. Specimens that failed from the P92 steel at temperatures range between 650°C-850°C showed a typical cup and cone fracture. A typical cup and cone fracture reveals the ductile fracture mode. According to the fractograph shown in Figure 7.16(c-e), P92 steel experiences substantial plastic deformation before the fracture. The necking phenomenon of the P92 steel before fracture is indicated by the drop in diameter and the corresponding decrease in cross-sectional area. The fracture surface of the P92 steel is shown in Figure 7.16(c-e). The fracture surface of the elevated temperature tensile tested specimen mainly consists of dimples and voids of different sizes and shapes. Large voids were also observed at the surface of the fractured specimen with higher test temperature, i.e., 650-850 °C. From the microstructure of the fractured surface, it was observed that the microstructure at the center of the fractured surface and the edge was different. The comparatively larger voids were observed at the center region compared to the edge of the fractured surface. The large number of voids at the fracture surface indicates the high plastic deformation due to the applied stress. The small and large voids at the fracture surface indicate the ductile fracture. (Chen et al., 2012) also reported similar fracture modes. The authors reported small, shallow, and equiaxial dimples at low test temperature and large, deep, and completely parabolic dimples at elevated temperature. The type and size of dimples change with the change of the test temperature. From the fractography of the failed surface at different temperatures, it was observed that the number of dimples and voids increases as the test temperature increases. The fracture surface of the specimen tested at 850 °C, as shown in Figure 7.16(e). The fracture surface of the specimen tested at 850 °C showed that the oxidation occurred at such a high temperature on the surface of the P92 steel. It is due to the poor oxidation resistance ability of the P92 steel at such a high temperature.

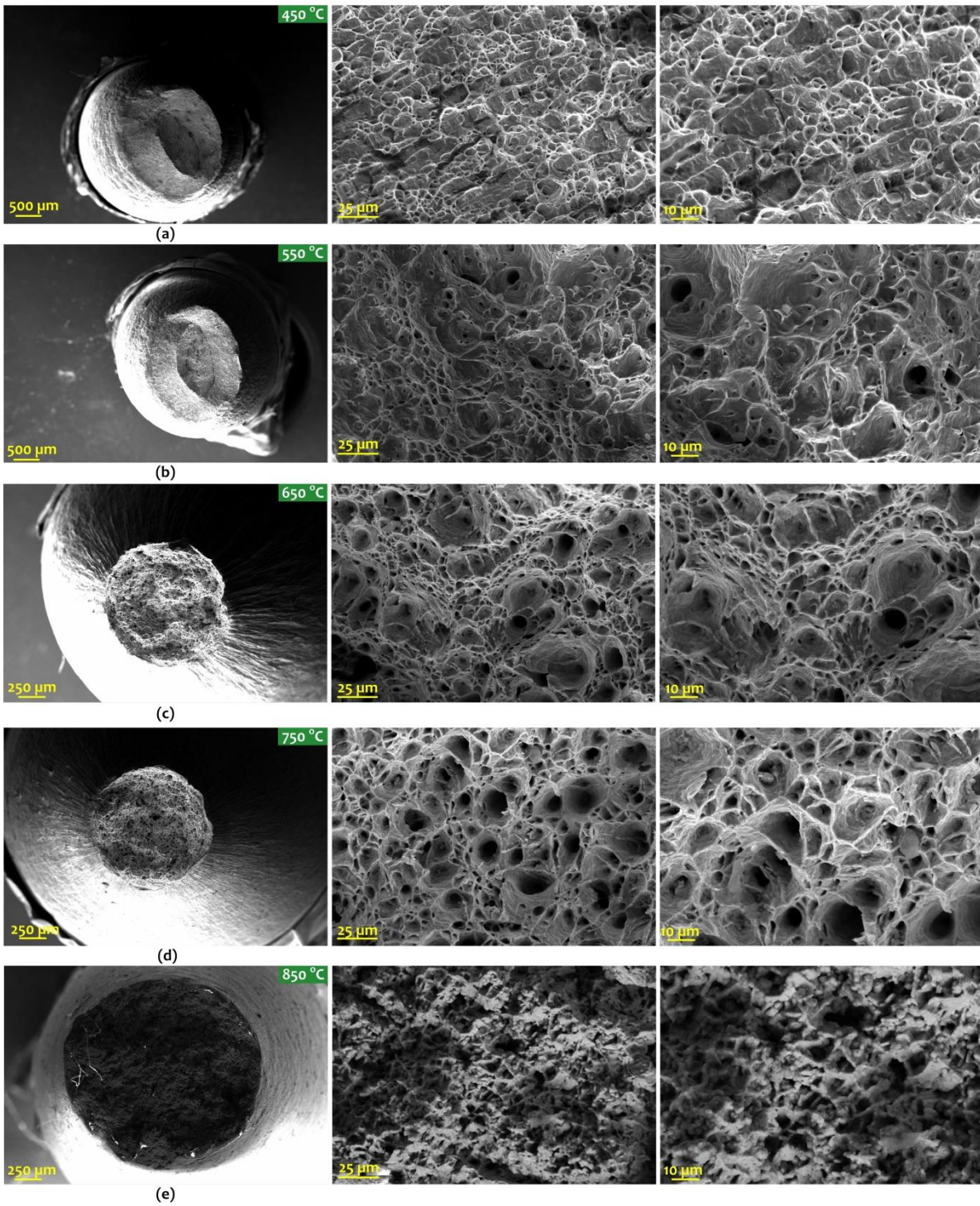


Figure 7.16: SEM micrograph of the fracture surface of the high-temperature tensile specimen tested at (a) 450 °C (b) 550 °C (c) 650 °C (d) 750 °C and (e) 850 °C (P92-ERNiCrMo-3-304L DWJ)

...