

Results and Discussion of Creep Test

This chapter provides details of the creep test carried out on the P92/304L dissimilar weld joint (DWJ). The creep behavior and high-temperature tensile strength of DWJ between P92 and 304L steel prepared with nickel-based and Thermanit MTS 616 filler metal, however, have not been thoroughly studied. Further study and research into the creep tendencies of these particular DWJ is required to fill this literature gap. The detailed discussion regarding the creep service life, microstructure properties, fracture surface analysis, and micro-hardness value obtained after creep test in stress range of 80-200 MPa and at a temperature of 650 °C is provided.

8.1 Creep test for P92-ERNiCrMo-3-304L dissimilar weld joint

The creep is an essential property for any material used at high temperature and stress condition. The creep properties of the P92/304L DWJ were investigated at 650 °C with applied stress in the 100-200 MPa range. The rapid microstructural changes occur in the P92 steel after 650 °C. Thus, the operating temperature was selected as 650 °C to to accelerate the creep test and reduce the experiment time duration. Figure 8.1(a-e) shows the creep ruptured specimens of P92/304L DWJ at 650 °C and applied stress of 100 MPa, 130 MPa, 150 MPa, 180 MPa, and 200 MPa. The fracture occurred from the P92 steel base metal (BM) region. The ERNiCrMo-3 filler weld fusion zone and 304L BM was safe after the creep test. The creep test results showed that the 304L and ERNiCrMo-3 filer have better creep strength and thermal stability than P92 steel.

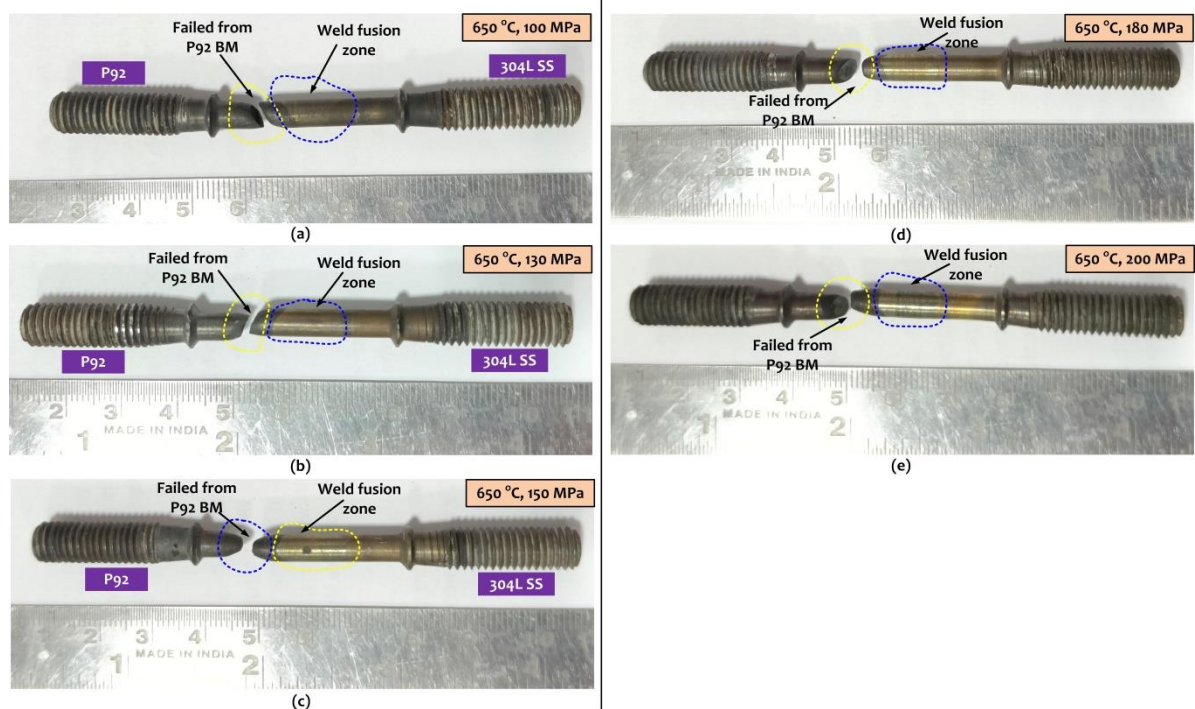


Figure 8.1: P92/304L DWJ specimens tested at different creep tests condition: (a) 650 °C/100 MPa (b) 650 °C/130 MPa (c) 650 °C/150 MPa (d) 650 °C/180 MPa and (e) 650 °C/200 MPa

The creep curve of the various specimens tested under constant load at 650 °C is shown in Figure 8.2(a, b). The maximum creep rupture time (706 h) was obtained for the specimen tested at 650 °C under 100 MPa. At the same time, the minimum rupture time (11 h) was obtained for the specimen tested at 650 °C under 200 MPa. The different stage of the creep curve, including the primary (transient), secondary (steady-state), and tertiary stage, was obtained for each test condition, as shown in Figure 8.2(a, b). Before the primary stage begins, there is instantaneous deformation due to the applied load on the specimen, which is insignificant. In the primary stage, the creep/strain rate decreases with time due to the work-hardening effect. In this stage, the material gets hardened due to the restriction in the dislocation movement. In the second stage, the creep rate remains constant over time, mainly due to the balance between the work hardening and softening behavior of the material. Work hardening and recovery are two opposite mechanisms. In the third stage, a process occurs quickly due to the necking or crack formation with effective cross-sectional area decreasing till specimen failure. The summary of the creep test results is listed in Table 8.1. Furthermore, the generated data have been plotted in the form of log-log scale in terms of creep rate against time, as shown in Figure 8.2(c, d) to measure the minimum creep rate (MCR) at different applied stresses. The measured creep strain is insignificant at 150 MPa and 130 MPa. Similarly, we have measured MCRs for 80 MPa and 100 MPa stresses also. The representation of creep behavior can be seen in the form of three different stages of creep. One can be noticed from Figure 8.2(c, d), the obtained MCR is found to be decreased with decrease in the applied stress.

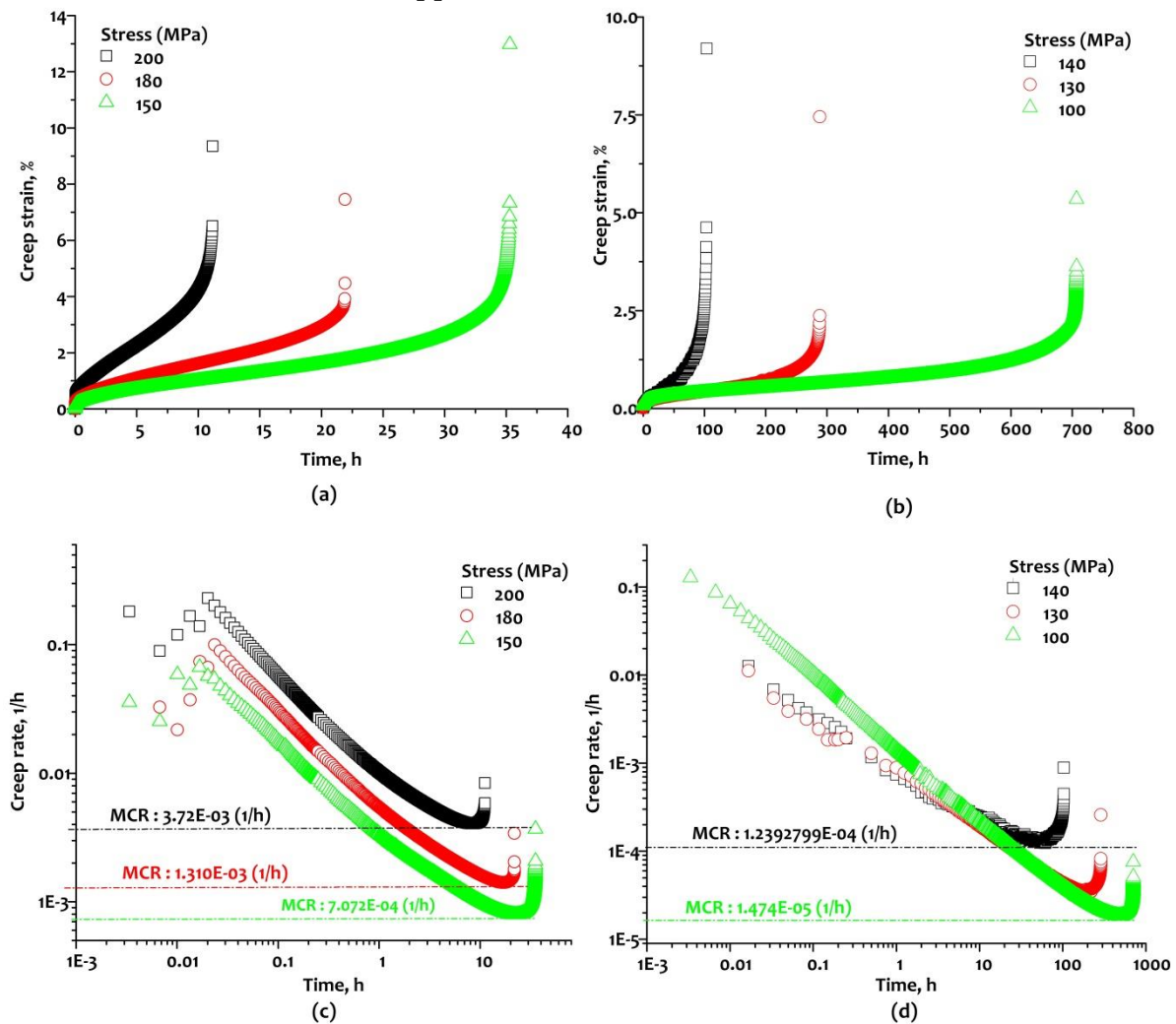


Figure 8.2: Plot of creep strain as a function of time for P92/304L DWJ have been conducted at 650 °C under the stress range of 150-200 MPa and 100-140 MPa, as shown in (a) and (b), respectively. The

corresponding creep rate vs. time plot is drawn in (c) and (d) to distinguish MCR with respect to the applied stress

Table 8.1 Creep test results of P92/304L DWJ (ERNiCrMo-3 filler metal) at 650 °C under the various stress levels

Temperature (°C)	Stress (MPa)	Creep Strain (%)	Rupture time (h)	Elongation (%)	Reduction Area (%)
650	100	5.35	706.9	5.35	43.11
650	130	16.94	288.3	7.45	42.31
650	140	18.02	103.5	18.02	72.17
650	150	12.97	35.3	12.97	68.06
650	180	7.46	21.9	7.46	47.12
650	200	9.35	11.2	9.35	48.11

The obtained creep data from the ERNiCrMo-3 weld joints specimens have been plotted by using the well-known power law equation ($\dot{\epsilon}_m = A \sigma^n$), where A is the material constant. The stress dependence of minimum creep rate ($\dot{\epsilon}_m$) of ERNiCrMo-3 weld joints are linearly fitted and resulted with the best fit. The slope of the fitted line is referred as stress exponent (n) and found to be 5.8, as shown in Figure 8.3. The creep test is a time-consuming process. Hence, it is not feasible to carry out the creep test for the entire creep life of the material. A mathematical relationship must be developed between the applied stresses (σ) and ruptured life (t_R). On the X-axis, the independent variable (stress in the present case) and the dependent variable (rupture time) were chosen on the Y-axis. From the creep experimental data, it was pointed out that the rupture time varies quickly with the change in the applied stress (Table 8.1). Thus, log-log plot have been plotted to obtain the best fit line of the data using linear fit analysis is shown in Figure 8.3. The obtained R-squared value (co-efficient of the determination) was close to 0.979, which indicates that the excellent fitting of the experimental data within the trend line. (Wu et al., 2021) also used a similar approach to develop the linear regression model between the applied stresses and rupture times. The obtained equation to predict the creep life at applied stress could be given by Eq. (8.1):

$$\log(t_r) = \log(14.39695) + (-5.79039) \log(\sigma) \quad (8.1)$$

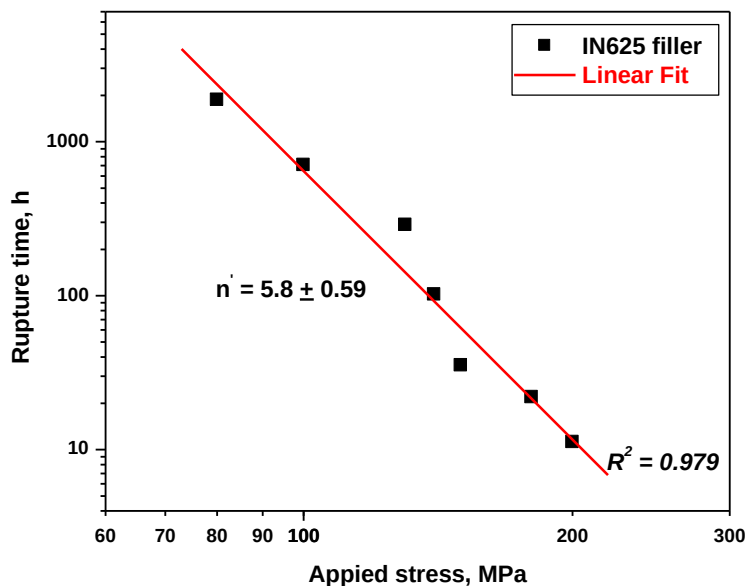


Figure 8.3: Plot of rupture time against the applied stress at 650°C has been linearly fitted to find the stress exponent (n). The value of 'n' is found to be 5.8, and the creep mechanism falls under the dislocation climb-controlled process in the stress range of 80-200 MPa

The other approach most frequently used to predict the creep life of the material is the Larson-Miller parameter (LMP) equation. The creep life of the material depends upon the applied stress and temperature. The LMP equation considers applied stress and temperature to predict the component's creep life. The LMP equation can be given by Eq. (8.2) as follows (Swaminathan et al., 2008):

$$LMP = T(C + \log t_r) \quad (8.2)$$

Where LMP is the Larson-Miller parameter, C is the Larson-Miller constant, t_r (h) is the rupture time, and T (K) is the test temperature. The value of C depends on the material; generally, it is in the range of 20-30 (Pavan et al., 2015; Shin et al., 2018). In the present case, the value of C is chosen 20 (Gong et al., 2010). Based on the reported data in Table 8.2, the plot of applied stress v/s LMP has been applied to the second-order polynomial fit, as shown in Figure 8.4. From the trend line, the value of LMP can be obtained at a given applied stress. Therefore, the corresponding time to rupture can be calculated at 650°C. Various researchers have also used this approach to predict the creep service life (Y. Wang et al., 2021). The obtained equation to predict the creep life at the applied stress could be given by Eq. (8.3):

$$Y = (-595673) + (9.70368E-4)X + (-2.81172E-8)X^2 \quad (8.3)$$

Table 8.2 Rupture time and LMP values of P92/304L DWJ (ERNiCrMo-3 filler metal) at 650°C under various stress levels

Temperature (°C / K)	Stress (MPa)	Rupture time (h)	Larson-Miller parameter (LMP)
650 / 923	100	706.9	21090
650 / 923	130	288.3	20730
650/923	140	103.5	20320
650 / 923	150	35.3	19889
650 / 923	180	21.9	19698
650 / 923	200	11.2	19428

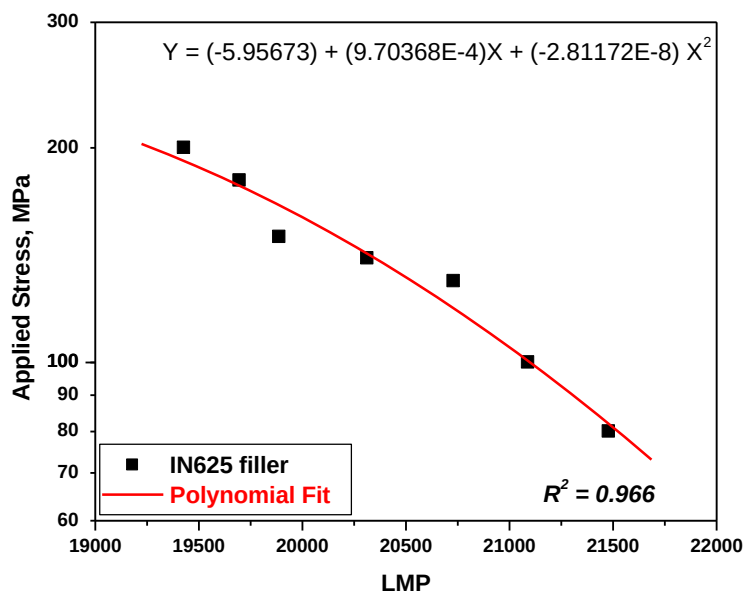


Figure 8.4: Plot of applied stress against LMP at 650°C has been fitted by applying the second order polynomial fit in the stress range of 80-200MPa. The constant value of LMP was taken as 20

8.1.1 Macrostructure after creep test

The creep behavior of the P92/304L DWJ was investigated at 650 °C under the stress range of 100-200 MPa. The creep test specimen failed from the P92 steel BM for all the test conditions. The different creep service life was obtained for the specimen tested at different stress. The macrograph of the cross-section of the fractured creep specimen is shown in Figure 8.5(a-e). The macrograph consists of a fracture surface, BM, and weld fusion zone. The macrographs of the cross-section of the fractured creep specimen also confirm the excellent creep strength of the ERNiCrMo-3 filler and 304L BM.

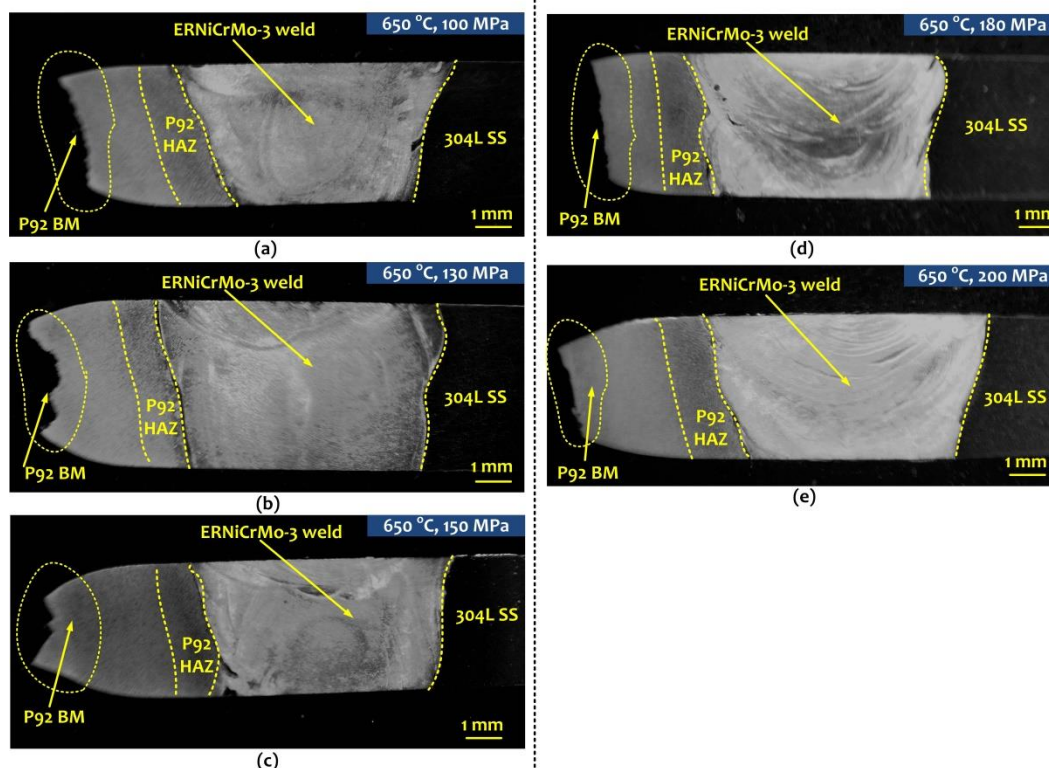


Figure 8.5: Cross-sectional macrograph of the creep ruptured specimen tested at 650 °C under the applied stresses of (a) 100 MPa (b) 130 MPa (c) 150 MPa (d) 180 MPa and (e) 200 MPa

8.1.2 Microstructure changes after the creep test

The microstructure evolution near the fracture tip surface after the creep test is shown in Figure 8.6(a-f) and Figure 8.7(a-i). The large number of voids formed near the fracture tip surface indicates the extensive plastic deformation at the necking region. Comparatively higher numbers of the micro and macro-voids were observed in the necking region compared to the coarse grain heat affected zone (CGHAZ) and fine grain heat affected zone (FGHAZ), as seen from the optical microscope image. The ERNiCrMo-3 weld fusion zone showed good stability during the creep test at 650 °C with an applied stress range of 100-200 MPa. The creep fracture was observed from the P92 BM. The microstructural instability of the P92 steel at elevated temperature is the main limiting factor for its long-term service. The strength of the P92 steel is due to its tempered martensite lath structure with large numbers of dislocations. The dislocations are introduced into the metal during its manufacturing process (rolling process). The dislocations present in the metal help in plastic deformation. The movement of the dislocations causes the plastic deformation of the metal. However, if dislocations can't move, plastic deformation does not occur. Solid solution strengthening and precipitation strengthening are the two basic mechanisms by which the motion of the dislocations is hindered. The movement of the dislocations is restricted by the precipitates present in the metal. Also, the movement of the dislocations is hindered at the grain boundaries due to the presence of the precipitates. $M_{23}C_6$ and MX precipitates are important for the creep

strength of the P92 steel. The $M_{23}C_6$ precipitates are decorated along the prior austenite grain boundaries (PAGBs). These precipitates maintain the creep strength of P92 steel by hindering the motion of the dislocations and grain boundaries. However, the combined strengthening effect by the above discussed mechanism may change at elevated temperature service. The coarsening of the $M_{23}C_6$ precipitates during the high-temperature service influences the creep strength of the steel. The finer precipitates effectively pin the boundaries and dislocations compared to the coarse precipitates (Sakthivel et al., 2019). The formation of the creep cavity starts along the PAGBs. It is because grain boundaries become weak at high temperature due to the precipitate coarsening and loss of the solid solution strengthening elements. Also, formation of the Laves phase $[Fe_2 (Mo, W)]$ after long-term exposure provides precipitation strengthening. However, this effect lowers after the coarsening of the Laves phase (Wang et al., 2017). The Laves phase provides the possible nucleation sites for the creep cavities. The high-stress concentration at this place converts the cavities into micro-cracks. These micro-cracks later transformed into macro-cracks and caused the fracture from the P92 steel (Wang et al., 2017). The formation of the Laves phase also results in the loss of the solid solution strengthening W and Mo elements from the P92 steel matrix. At low-stress condition, when the rupture time is high, the fracture from the P92 steel is due to the formation of the Laves phase. The number density of the $M_{23}C_6$ carbide precipitates in the P92 BM matrix is higher compared to the MX precipitates due to the increased weight percentage of the Cr element in the alloy composition. Thus, the stability of the $M_{23}C_6$ carbides is more critical to maintain the creep strength of the P92 steel at elevated temperature service. The absence of a pinning effect by these precipitates is mainly responsible for the void formations along the grain boundaries due to the lack of boundary strengthening. The MX carbonitride precipitates have a small size and low coarsening rate at elevated temperature service. These precipitates provide stable precipitation hardening to the P92 steel. However, $M_{23}C_6$ carbide precipitates have a big size and high coarsening rate at elevated temperature service. Due to this, the pinning effect of these particles on lath boundaries disappears once these precipitates coarsen. Also, the dissolution of the $M_{23}C_6$ carbides during the welding leads to the high concentration of the C and Cr elements in the matrix. During creep, the C and Cr elements re-precipitate on the remaining particles, resulting in coarse precipitates. This phenomenon mainly causes the tri-axial stress concentration between the coarse precipitates and the BM matrix. This ultimately leads to the softening of this region and void generation near the coarse precipitates (M. Y. Kim et al., 2014b).

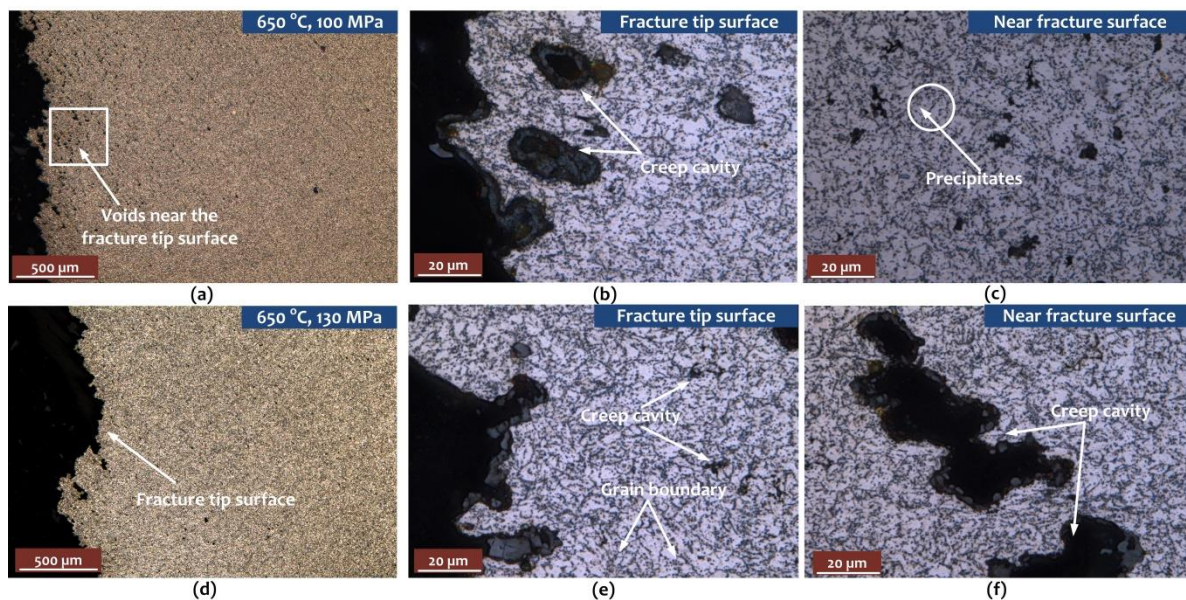


Figure 8.6: (a) Optical microstructure of the creep rupture specimen tested at 650 °C/100 MPa: fracture surface, voids and cavities can be observed (b) microstructure of the fracture tip surface (c)

microstructure near the fracture surface (d) microstructure of the creep rupture specimen tested at 650 °C/130 MPa: fracture surface (e) microstructure of the fracture tip surface and (f) microstructure near the fracture surface

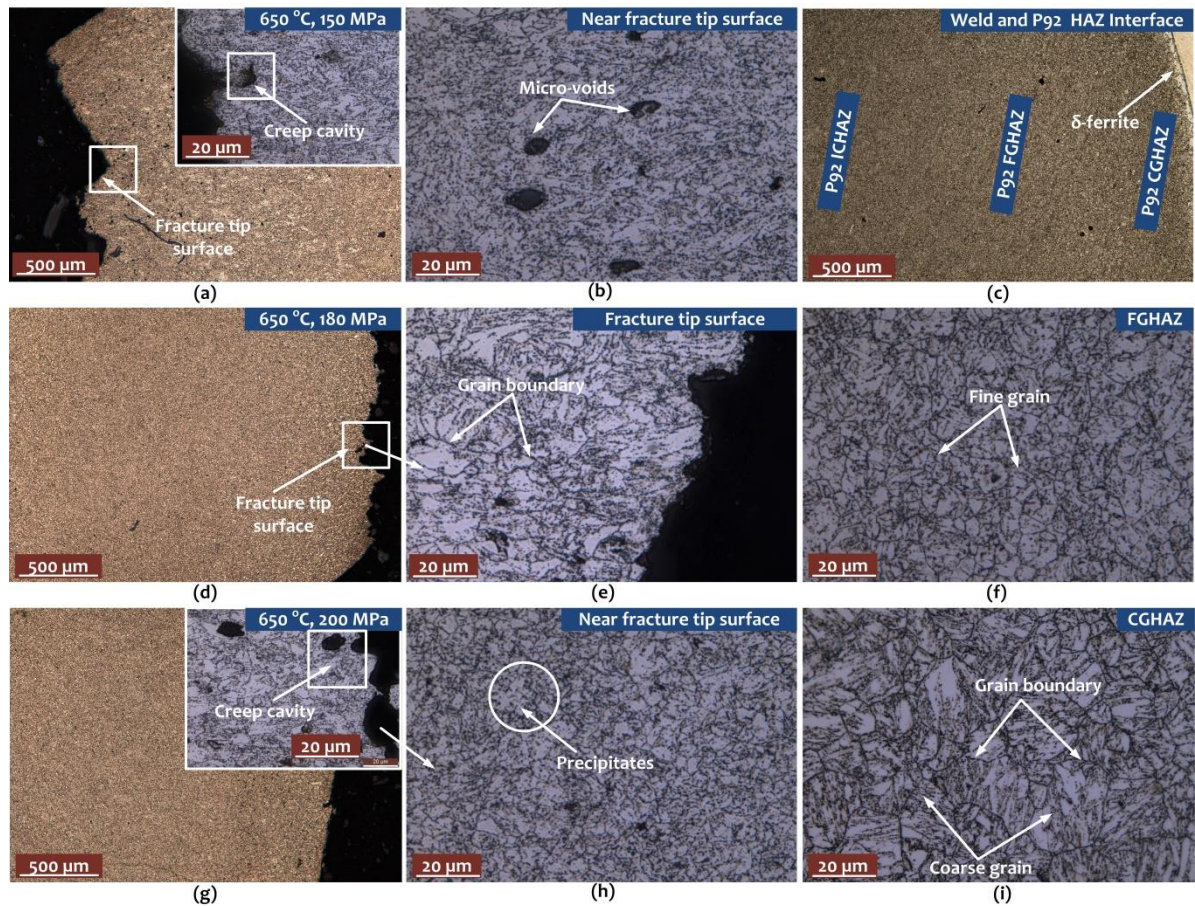


Figure 8.7: (a) Optical microstructure of the creep rupture specimen tested at 650 °C/150 MPa: fracture surface (b) microstructure near the fracture tip surface (c) microstructure at the weld and P92 HAZ interface (d) microstructure of the creep rupture specimen tested at 650 °C/180 MPa: fracture surface, voids and cavities can be observed (e) microstructure of the fracture tip surface (f) microstructure at the P92 FGHAZ region (g) microstructure of the creep rupture specimen tested at 650 °C/200 MPa: fracture surface, voids and cavities can be observed (h) microstructure near the fracture tip surface and (i) microstructure at the P92 CGHAZ region

The FESEM microstructural investigation at the fracture tip surface, weld interface, and weld fusion zone was also carried out for the creep ruptured specimen. The FESEM image of the microstructure evolution toward the P92 steel after the creep test at 650 °C under 100 MPa is shown in Figure 8.8(a-f). The creep cavities formed near the fracture surface are shown in Figure 8.8(b, c). The FESEM microstructure of the ERNiCrMo-3 weld fusion zone and BM interface is shown in Figure 8.8(d-f). The EDS line scan was performed at the BM and weld fusion zone interface to identify the element distribution in gradient microstructure, as shown in Figure 8.9(a-d). The steep variation of the Cr, Ni, and Fe elements was observed at the weld fusion line, which was in accordance with the chemical composition of the BM and weld fusion zone. The solidification of the weld pool after welding leads to the segregation of the various elements. This segregation causes the formation of the secondary phase in the weld fusion zone. The EDS map of the secondary phase particle developed in the weld fusion zone due to the rapid cooling after welding is shown in Figure 8.9(e). The segregation of the Nb, Mo, and Ti can be observed from the map. The segregation of Mo and Nb takes place in the inter-dendritic region due to their distribution coefficient value of less than 1. The depletion of the Fe and Cr elements can be observed from the area map. The composition of these particles determined by EDS point analysis is listed in Table 8.3. Based on the published results, the possible secondary phase

formation can be (Nb,Ti)C (Silva et al., 2013). As the specimen was fractured from the P92 steel for all the creep test condition, the major aim of this study was to perform a detailed microstructural investigation of the P92 steel after creep service.

The FESEM image of the microstructure evolution toward the P92 steel after the creep test at 650 °C under 130 MPa and 150 MPa is shown in Figure 8.10(a-f). The microstructure of the P92 CGHAZ region with voids formed along the PAGBs after the creep test is shown in Figure 8.9(d). From the FESEM microstructure, it can be seen that the precipitation of the $M_{23}C_6$ carbides increased significantly after the creep test. The undissolved precipitates tend to coarsen during the creep service. The creep strain causes the coarsening of the existing $M_{23}C_6$ precipitates. The coarsening of the $M_{23}C_6$ precipitates results in a decrease in the number density of the precipitates. This ultimately degrades the precipitation-strengthening mechanism. The formation of the Laves phase takes place during the elevated temperature creep exposure. A significant amount of the W and Mo is consumed during the formation of this phase, which ultimately reduces the solid solution strengthening. During elevated temperature service, the $M_{23}C_6$ precipitates and Laves phases grow rapidly along the PAGBs. The FESEM-EDS point and area map analysis were conducted near the fracture tip surface containing coarse and fine secondary phase particles around the cavity to identify the composition of the secondary phase. The EDS results showed a strong peak of the W elements. The dominating particles near the creep cavities are rich in Cr and W elements. Such particle was identified as the Laves phase and carbide precipitates. The possible Laves phase can be $(Fe, Cr)_2(W, Mo)$ type Laves phase and $(Fe, Cr)_{23}C_6$ secondary precipitates (Nandha Kumar et al., 2020). The FESEM-EDS area mapping analysis revealed the spot of the W element. Thus, the possible Laves phase can be $(Fe, Cr)_2(W)$ type. Also, P92 steel (W:2 wt.% max., Mo:0.6 wt.% max.) has a higher amount of the W and less amount of the Mo element in its chemical composition compared to the P91 steel (W: nil., Mo:1.05 wt.% max.) (Zhao et al., 2016). Thus, the majority of the Laves phase formation is of $(Fe, Cr)_2(W)$ type instead of $(Fe, Cr)_2(Mo)$ type in the P92 steel. The $(Fe, Cr)_2(Mo)$ type Laves phase mostly forms in P91 steel (Bong Yoon et al., 2020). The EDS elemental map of the creep cavity present near the fracture tip surface is shown in Figure 8.11. The spot of the Cr and W elements was observed from the area map analysis. The chemical composition of the various particles determined by EDS is listed in Table 8.3. The FESEM microstructure evolution toward the P92 steel after the creep test at 650 °C under the stress range of 150-200 MPa is shown in Figs. 22(a-i). The formation of the Laves phase was not observed for this test condition. The coarse carbide precipitates were only observed around the cavity. The grain boundary sliding and diffusion are the two main plastic deformation mechanisms during creep tests. During the creep service, a cavity is formed along the PAGBs due to the grain boundary sliding. The glide (motion of dislocation in its slip plane) and climb (motion of dislocation in adjacent parallel slip plane) of the dislocation in the lattice cause creep. When the applied stress is high, the Laves phase does not form due to the low service time. In this condition, failure occurs due to the plastic deformation caused by the grain boundary sliding in the base metal. This large plastic deformation forms dimples of various shapes and sizes along the PAGBs, resulting in the transgranular fracture (Zhang et al., 2019).

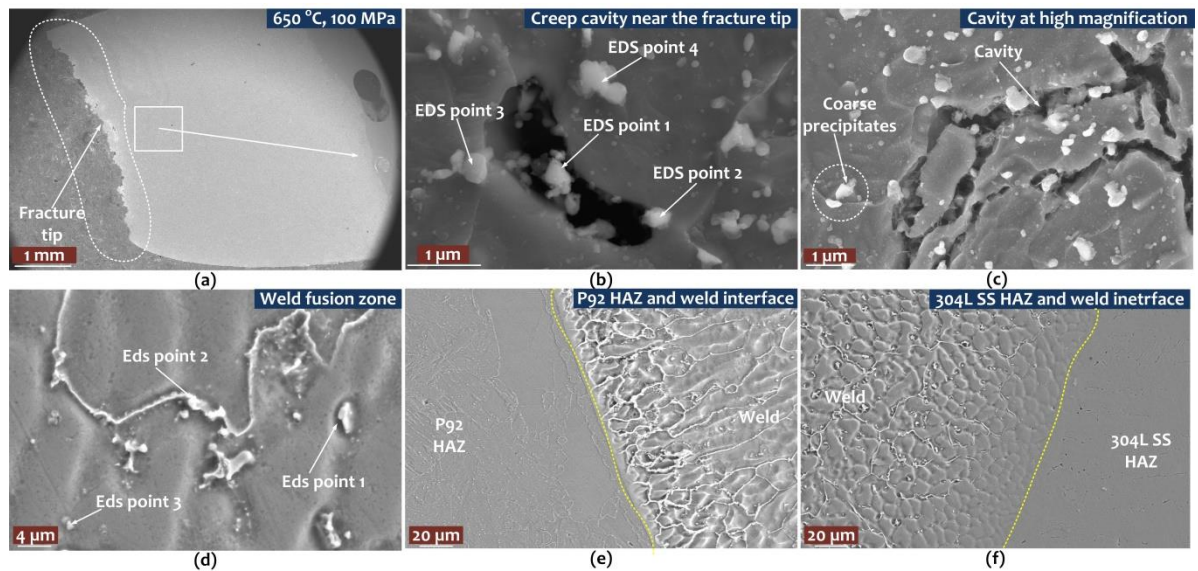


Figure 8.8: (a) FESEM microstructure of the creep rupture specimen at 650 °C/100 MPa: fracture tip surface (b, c) creep cavity near the fracture tip surface at high magnification (d) microstructure of the ERNiCrMo-3 weld fusion zone (e) P92 HAZ and weld interface and (f) 304L HAZ and weld interface

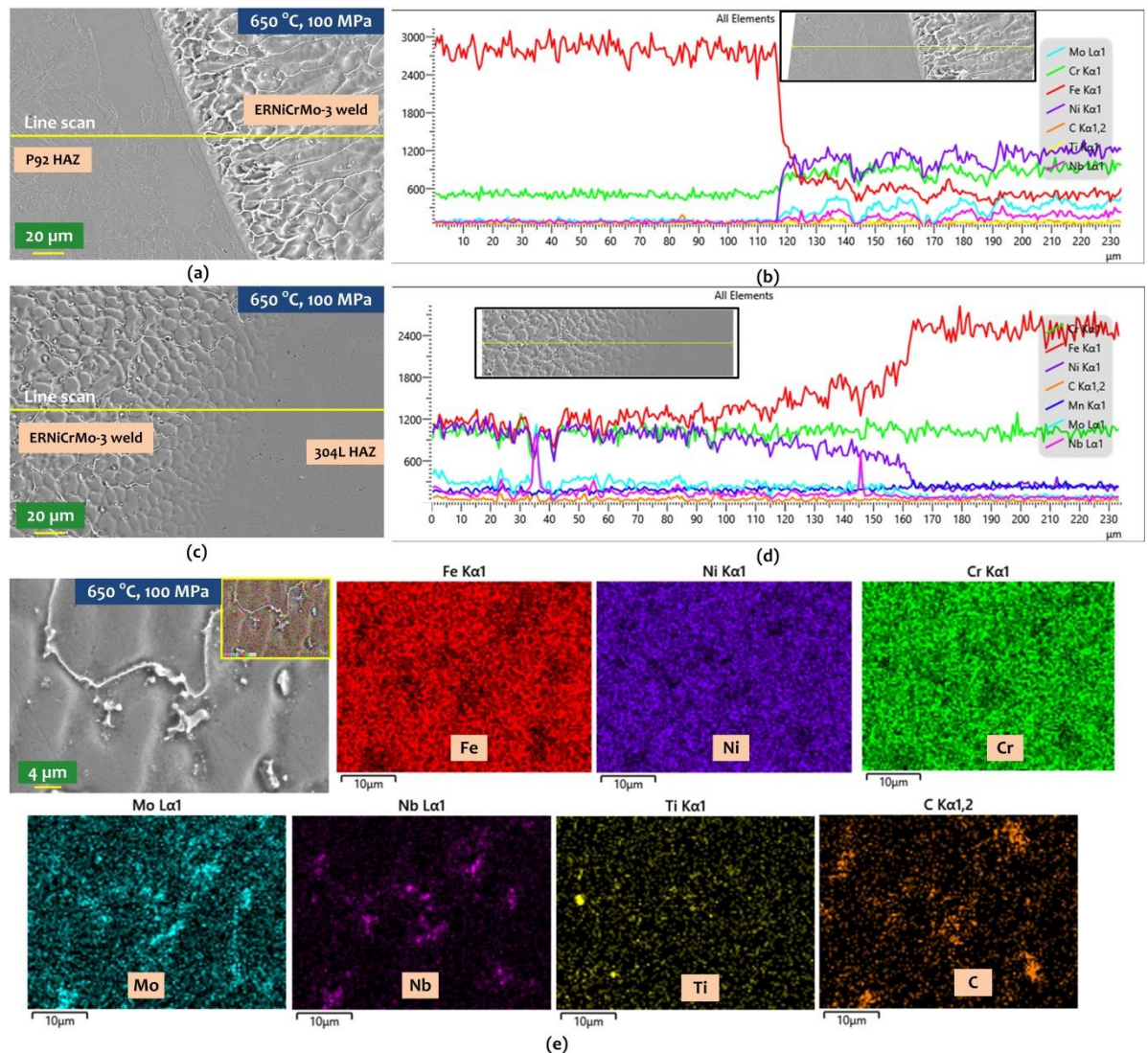


Figure 8.9: For the specimen creep tested at 650 °C, 100 MPa (a, b) EDS line scan results across P92 HAZ and ERNiCrMo-3 weld interface (c, d) EDS line scan results across 304L HAZ and ERNiCrMo-3 weld interface and (e) EDS elemental map of ERNiCrMo-3 weld fusion zone

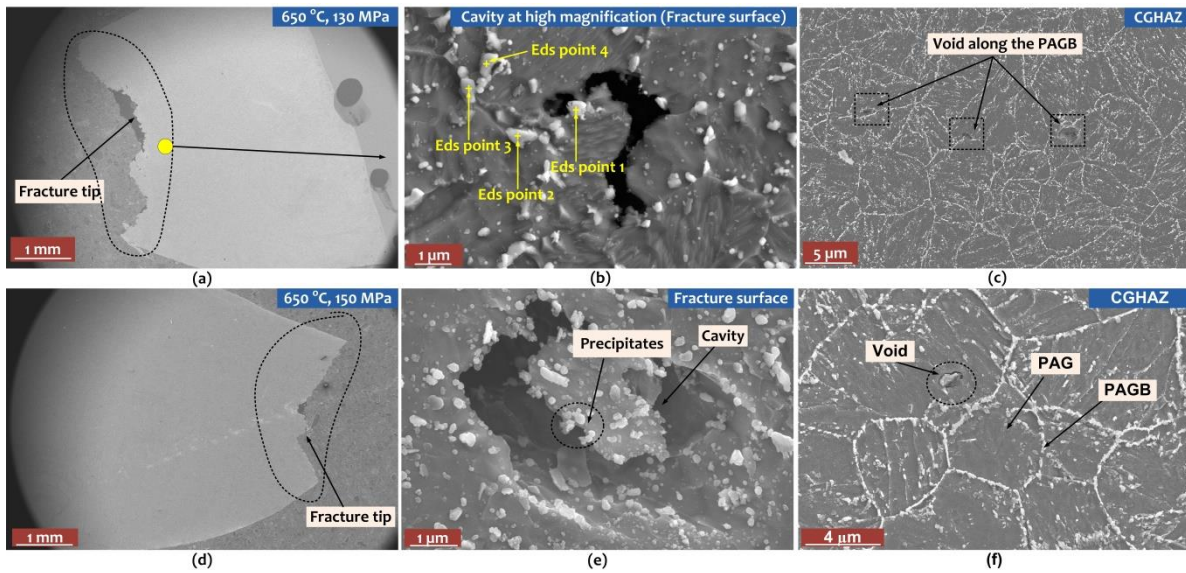


Figure 8.10: (a) FESEM microstructure of the creep rupture specimen at 650 °C/130 MPa: fracture tip surface (b) microstructure near the fracture tip surface, voids can be seen (c) creep cavity near the fracture tip surface at high magnification (d) microstructure of the CGHAZ region, voids can be seen along the PAGBs (e) FESEM microstructure of the creep rupture specimen at 650 °C/150 MPa: fracture surface (f) microstructure of the CGHAZ region, voids can be seen along the PAGBs

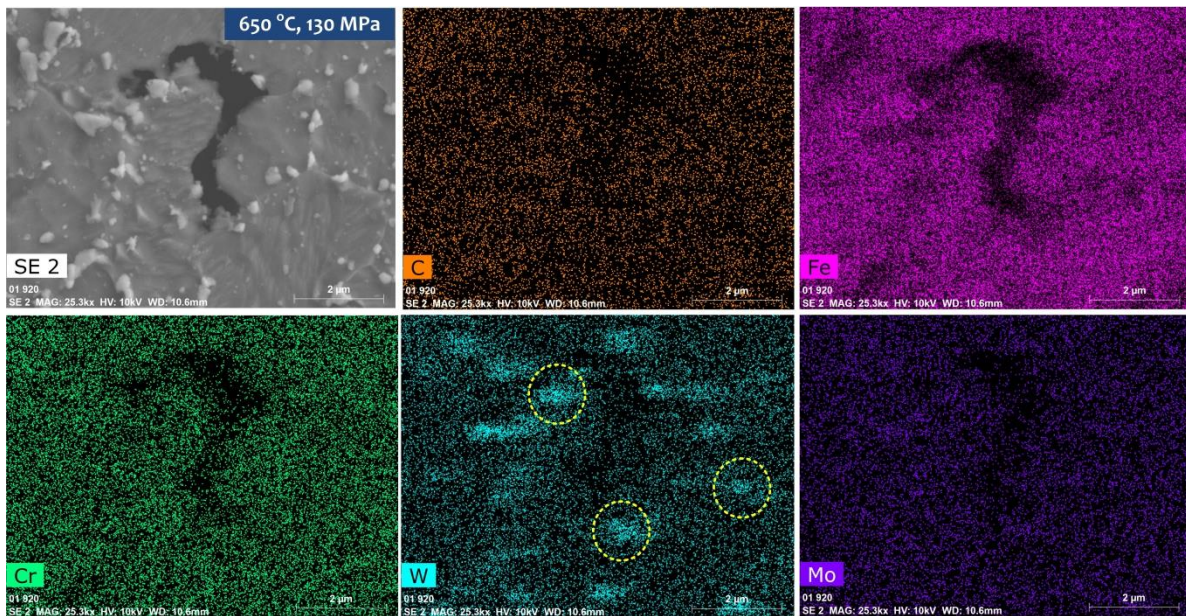


Figure 8.11: EDS elemental map of creep cavity present near the fracture tip surface for the specimen tested at 650 °C, 130 MPa, Figure 8.10(b) (few W element spots are marked by a circle)

Table 8.3 EDS results of the precipitate particles near the fracture surface

	Location	C	Fe	Cr	Mo	Nb	Ni	W	Ti
Figure 8.8(b)	1	22.97	62.61	6.55	0.0	1.91	0.29	5.63	-
	2	19.17	69.87	6.23	0.0	1.63	0.24	2.36	-
	3	17.92	65.28	6.05	0.02	1.82	0.10	5.37	-
	4	15.19	72.53	7.14	0.0	1.51	0.20	2.22	-
Figure 8.8(d)	1	6.5	27.6	18.2	4.7	2.4	40.5	-	0.2
	2	11.9	17.7	14.3	11.7	13.2	30.9	-	0.2
	3	23.8	15.6	11.4	8.7	16.5	22.7	-	1.2
Figure 8.10(b)	1	9.12	60.39	8.50	1.28	4.48	1.07	15.88	-
	2	17.61	57.51	5.74	-	2.54	0.59	4.27	-

	3	15.22	41.57	20.88	0.27	3.75	0.45	8.58	-
	4	22.06	40.50	15.06	-	4.04	0.79	2.66	-

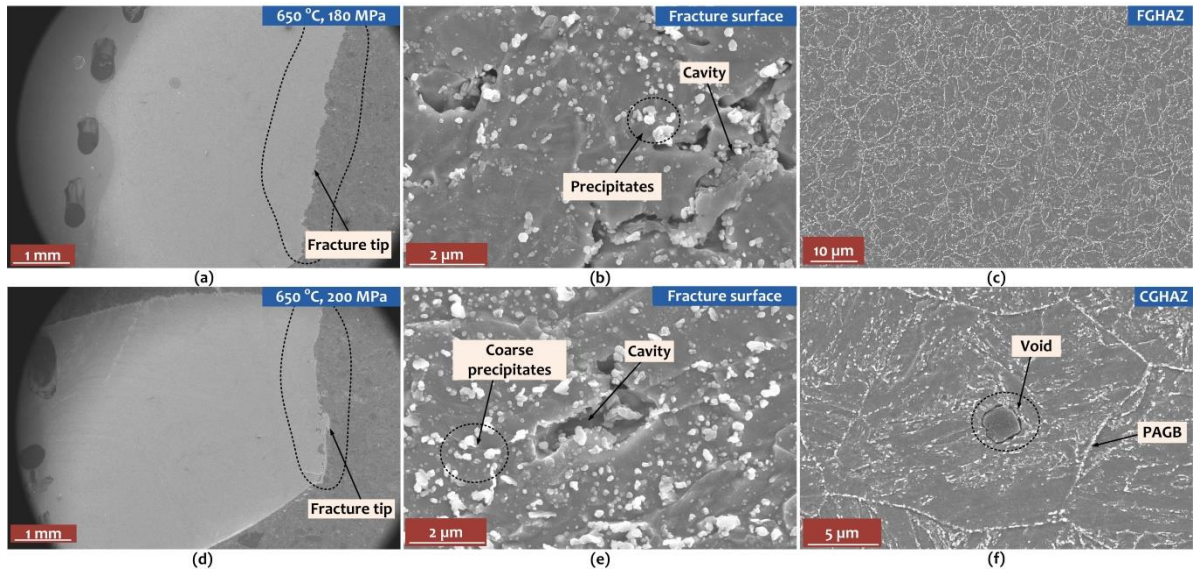


Figure 8.12: (a) FESEM microstructure image of the creep rupture specimen tested 650 °C/ 180 MPa: fracture surface (b) microstructure of the fracture surface, voids can be seen (c) microstructure of the FGHAZ region (d) FESEM microstructure of the creep rupture specimen tested at 650 °C/200 MPa: fracture surface (e) microstructure of the fracture surface, voids can be seen and (f) microstructure of the CGHAZ region, void can be seen

The schematic diagram of the failure mechanism from the P92 steel after elevated temperature and stress condition is shown in Figure 8.13. The microstructure of the P92 steel is composed of tempered martensite lath structure, $M_{23}C_6$ carbide, and MX carbonitride precipitates, as discussed in the introduction section. The formation of the new $M_{23}C_6$ precipitates and the coarsening of the existing $M_{23}C_6$ precipitates increases with the increase of the creep strain. The grain boundaries mainly provide strengthening against creep failure. The $M_{23}C_6$ carbide precipitates mainly consist of Cr (M stands for Cr) and C in their composition. The dissolved Cr and C re-precipitate during the high-temperature working environment. This results in the coarsening of the $M_{23}C_6$ carbide precipitates. The formation of the Laves phase starts after 100 h of service at the elevated temperature of 650 °C. The mechanism of the Laves phase formation is as follows. The Laves phase develops near the chromium-depleted region adjacent to the $M_{23}C_6$ carbide. Thus, before the formation of the Laves phase, the formation of the $M_{23}C_6$ carbide is required. The region adjacent to the $M_{23}C_6$ carbide has a high concentration of the W and Mo elements, which act as a possible site for the Laves phase formation (Nandha Kumar et al., 2020). The Laves phase forms at the PAGBs at the cost of consumption of the important solid solution strengthening W and Mo elements from the P92 BM matrix. Thus, the solid solution strengthening provided by the W and Mo elements weakened. The formation of the Laves phase results in the softening of the P92 BM matrix, providing a possible nucleation of the creep cavities. Thus, a micro-void is created around these coarse secondary phase particles during the applied tension load. These micro-voids later converge together and form the crack. Thus, failure of the DWJs takes place from the P92 BM before the actual service life.

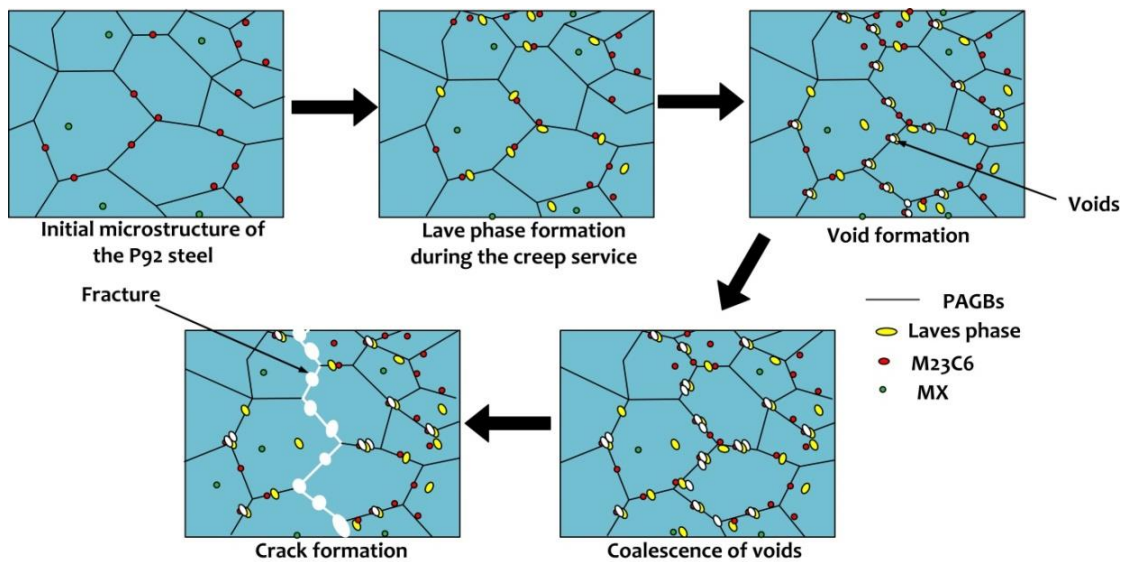


Figure 8.13: Failure mechanism during creep service from P92 steel (Wu et al., 2021)

8.2 Creep test for P92-ERNiFeCr-2-304L dissimilar weld joint

Figure 8.14(a-d) shows the creep ruptured specimens tested at 650 °C under the applied stress in the range of 80-180 MPa. The creep test was performed on the P92/304L DWJ to determine the rupture time and fracture locations at different temperatures and stresses. The fracture occurred from the P92 BM for all the test condition. The creep test results showed that the 304L BM and ERNiFeCr-2 nickel-based filler metal has better creep strength and excellent thermal stability at high-temperature and stress than the P92 steel.

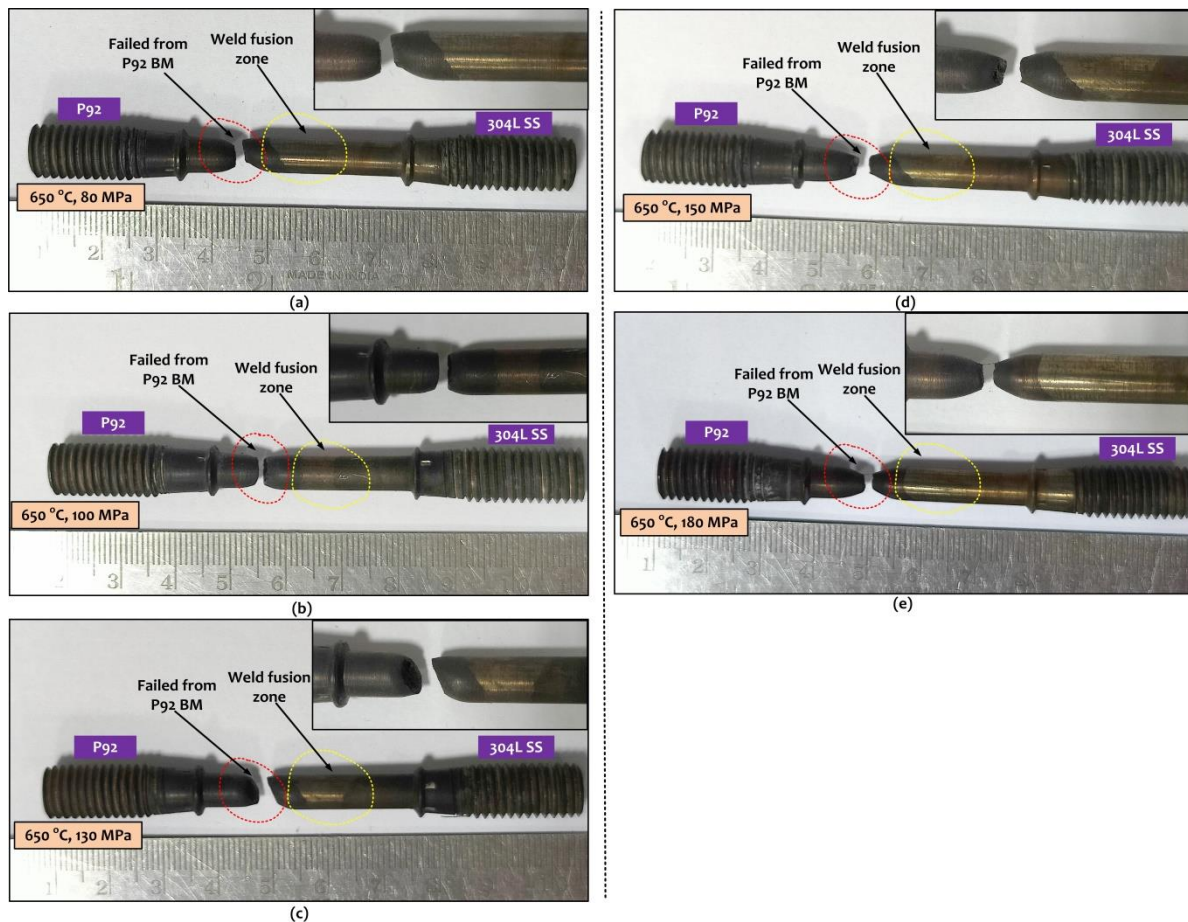
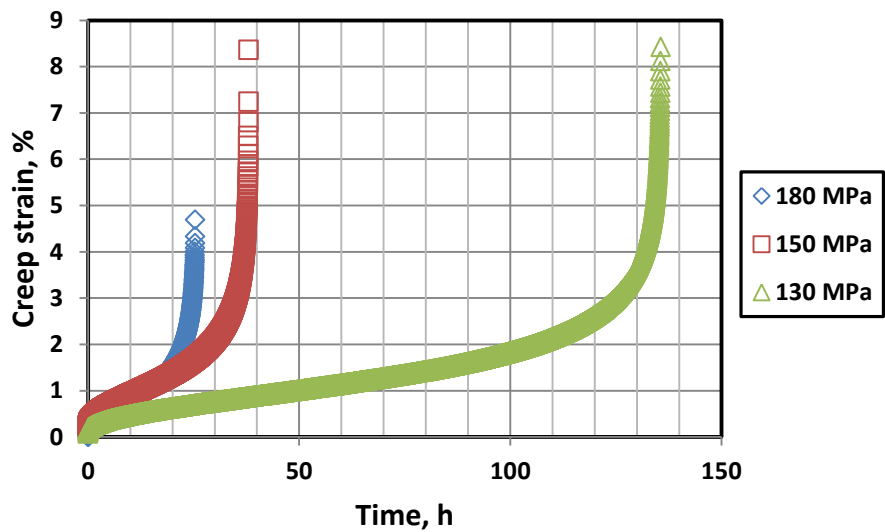
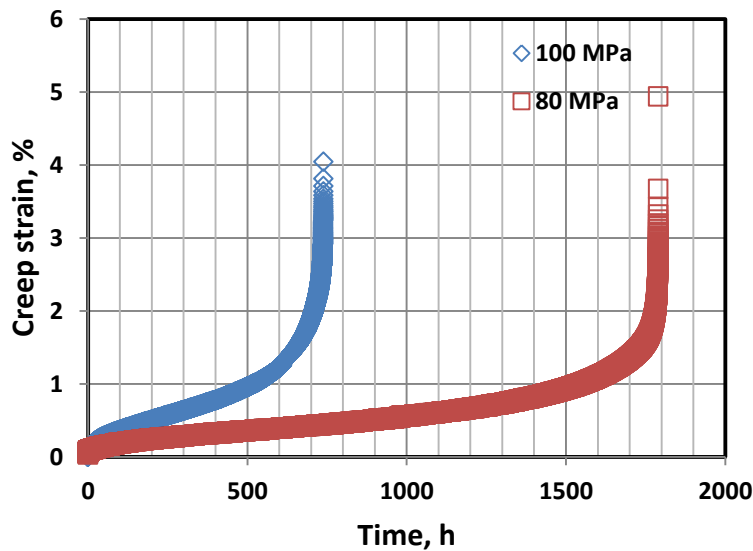


Figure 8.14: P92/304L DWJ specimens after creep test at 650 °C under different stresses: (a) 80 MPa (b) 100 MPa (c) 130 MPa (d) 150 MPa and (d) 180 MPa

Figure 8.15(a, b) shows the creep strain against time plot of ERNiFeCr-2 weld joints, which have been performed till rupture. The maximum creep rupture time of (1789 h) was recorded for the specimen tested at 650 °C under 80 MPa for longer duration. At the same time, the specimen tested at 650 °C under 180 MPa and the value of minimum rupture time for (25 h) was noticed. The summary of all the creep test results is listed in Table 8.4. Rupture elongation and reduction in the cross-sectional area were also measured from the fractured specimens, as reported in Table 8.4.



(a)



(b)

Figure 8.15: Creep curves (creep strain vs. time) of ERNiFeCr-2 weld joints performed at 650°C under (a) the higher stress range of 130-180MPa and (b) at the lower stresses (100MPa and 80MPa)

Table 8.4 Creep tests results of ERNiFeCr-2 weld joints

Temperature (°C)	Stress (MPa)	Minimum Creep Rate (1/h)	Rupture time (h)	Elongation (%)	Reduction Area (%)
650	180	7.43E-04	25.4	11.88	60.89
650	150	6.63E-04	38.0	9.58	71.60
650	130	1.79E-04	135.6	14.34	69.75
650	100	1.91E-05	738.9	4.05	43.15
650	80	5.50E-06	1789.8	4.94	39.59

Furthermore, the generated data have been plotted in the form of log-log scale in terms of creep rate against time, as shown in Figure 8.16 to measure the MCR at different applied stresses. The measured creep strain is insignificant at 150 MPa and 130 MPa. Similarly, we have measured MCRs for 80 MPa and 100 MPa stresses also. The representation of creep behavior can be seen in the form of three different stages of creep. One can be noticed from Figure 8.16, the obtained MCR is found to be decreased with decrease in the applied stress. The obtained creep data from the ERNiFeCr-2 weld joints specimens have been plotted by using the well-known power law equation ($\dot{\epsilon}_m = A \sigma^n$), where A is the material constant. The stress dependence of minimum creep rate ($\dot{\epsilon}_m$) of ERNiFeCr-2 weld joints are linearly fitted and resulted with the best fit. The slope of the fitted line is referred as stress exponent (n) and found to be 6.65 along with the value of A is obtained as $1\text{E-}18 \text{ MPa}^{-n} \text{ h}^{-1}$, as shown in Figure 8.17.

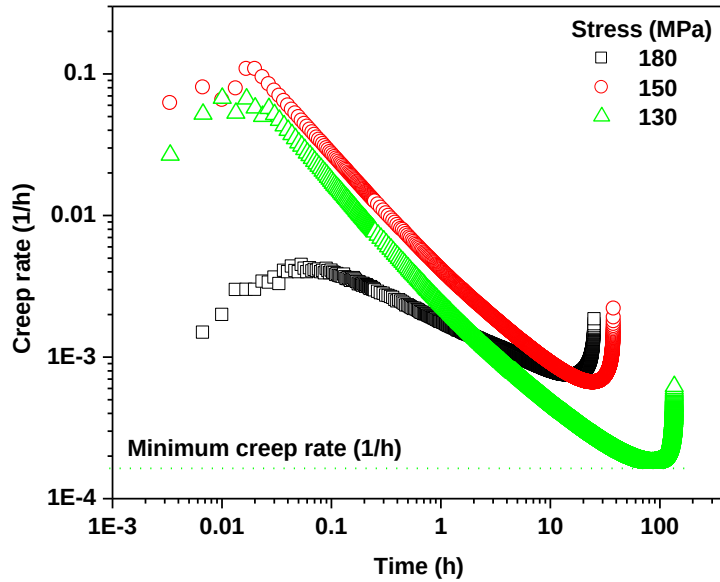


Figure 8.16: The minimum creep rate has been measured by plotting the data in terms of log-log scale

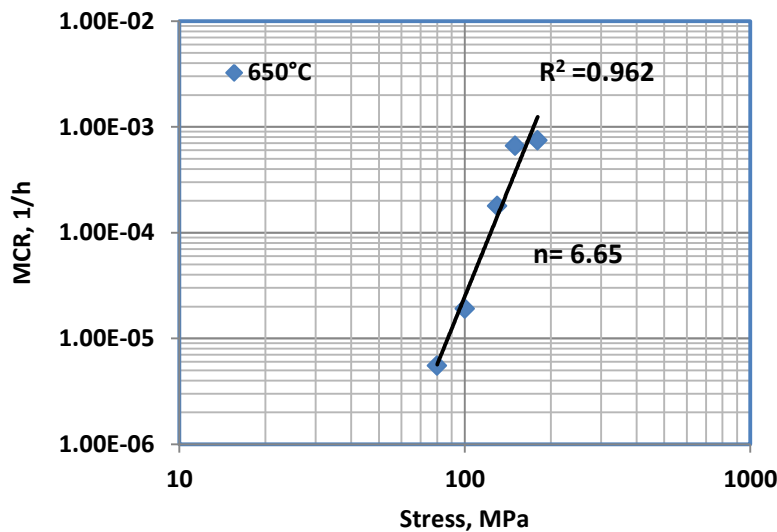


Figure 8.17: log-log plot of minimum creep rate against the applied stress of ERNiFeCr-2 weld joints. The obtained experimental data are linearly fitted to determine stress exponent

The relationship between the applied stress (σ) and rupture time (t_R) was determined by using the linear regression analysis. From the creep experimental data,

it was observed that the rupture time varies with the applied stress as shown in Table 8.4. Thus, log-log plot of rupture times and applied stresses have been drawn and fitted using the linear fit analysis. The obtained R-squared value (co-efficient of the determination) was close to 0.989, which is an indication of the best fit of the experimental data and is closer to trend line under the stress range of 80-180 MPa. The obtained stress exponent is approximately 5.7 from the Figure 8.17. Therefore, one can understand that the deformation mechanism of creep is confirmed as dislocation climb controlled mechanism of creep, as shown in Figure 8.17 and Figure 8.18. The obtained equation to predict the creep life at applied stress could be given by Eq. (8.4):

$$\log(t_R) = \log(14.07356) + (-5.6592) \log(\sigma) \quad (8.4)$$

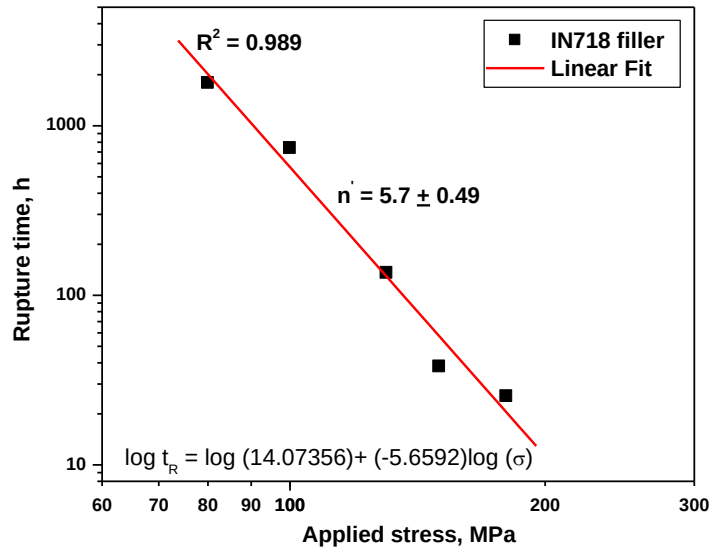


Figure 8.18: Plots of rupture time vs. applied stress at 650°C

Based on the data obtained in Table 8.5, the plot of applied stress v/s LMP on the log scale was obtained, as shown in Figure 8.19. From the trend line, the value of LMP corresponding to any applied stress can be obtained from the graph. The obtained equation to predict the creep life at the applied stress could be given by Eq. (8.5):

$$Y = 1.08167 + (2.84401E-4)X + (-1.14474E-8)X^2 \quad (8.5)$$

Table 8.5 LMP value corresponds to the applied stress and rupture time

Temperature (°C/K)	Applied Stress (MPa)	Rupture time (h)	Larson-Miller parameter (LMP)
650 °C /923 K	80	1789.8	21462
650 °C /923 K	100	738.9	21108
650 °C /923 K	130	135.6	20428
650 °C /923 K	150	38.0	19918
650 °C /923 K	180	25.4	19757

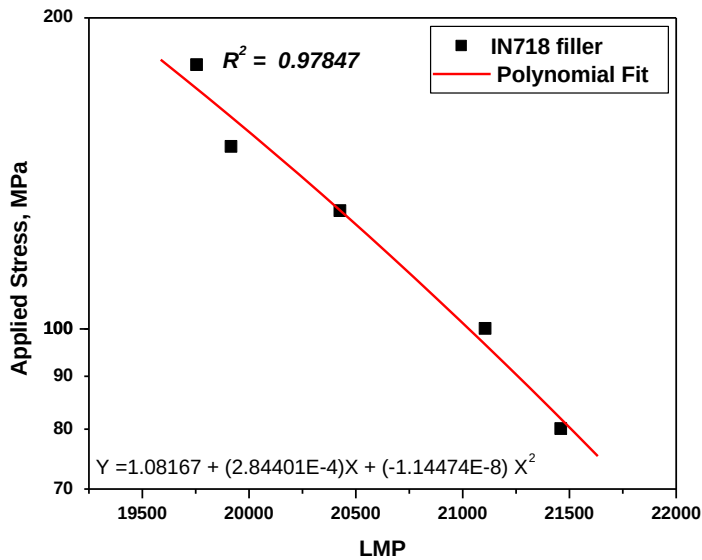


Figure 8.19: Plots of applied stress vs. LMP at 650°C

8.2.1 Macrostructure after creep test

The macrograph of the creep ruptured specimen is shown in Figure 8.20(a-e). The macrograph shows the cross-section of the fracture surface of the creep specimen consisting of P92 BM, 304L BM, ERNiFeCr-2 filler weld fusion zone and P92 HAZ. From the macrograph, few cracks were observed in the weld fusion zone. This is due to the creep service at elevated temperature and stress. The further microstructure characterization near the necking region, weld fusion zone, HAZ, and BM was carried out using FESEM and optical microscope.

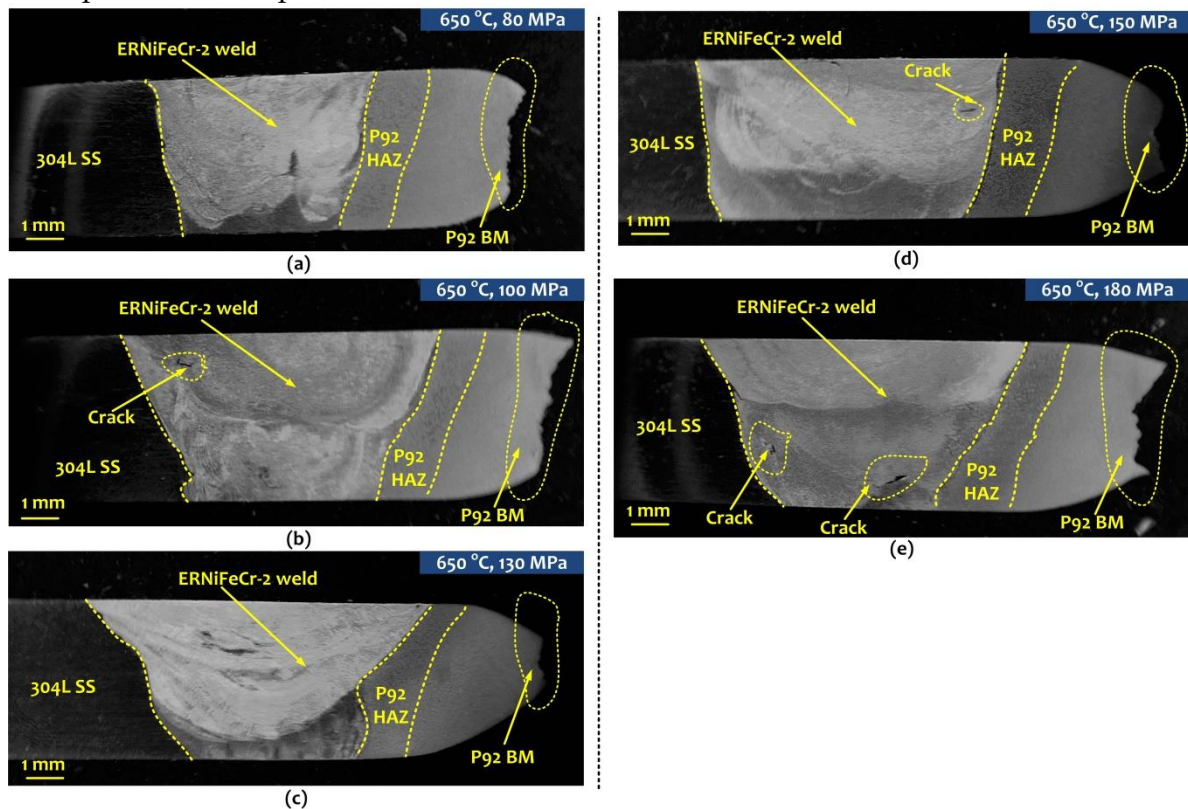


Figure 8.20: Macrostructure of the longitudinal section of the fracture creep specimens. The macrostructure of the samples clearly indicates the creep fracture from the P92 BM and cracks within the ERNiFeCr-2 weld region, as shown with arrows

8.2.2 Microstructure changes after creep test

The optical microstructure near the necking region and other locations of the P92/304L DWJ after creep test at 650 °C under 80 MPa, 100 MPa and 130 MPa is shown in Figure 8.21(a-i). The fracture was observed in the P92 BM, close to the inter-critical heat affected zone (ICHAZ). The microstructure near the P92 CGHAZ and FGHAZ region after creep test is also shown in Figure 8.21(f, i). The microstructure features in these regions were unaffected during the creep test. The grain boundaries decorated with the $M_{23}C_6$ precipitates mainly obstruct the dislocation movement under the applied load. The dissolution temperature of the $M_{23}C_6$ and MX precipitate is 830 °C and 1200 °C, respectively. Thus, during the elevated temperature test, $M_{23}C_6$ precipitates easily dissolved and coarsened compared to the MX precipitates (Zhou et al., 2021). At high-temperature because of the coarsening of the $M_{23}C_6$ precipitates, the pinning effect provided by these precipitates gets suppressed. Also, due to the coarsening of the precipitates, ductility decreases. Thus, the strength of the P92 steel falls rapidly at high-temperature due to the coarsening of the $M_{23}C_6$ precipitates (Lei et al., 2015). Hence, at high-temperature, failure takes place from the P92 BM. The optical microstructure near the necking region and other locations of the weld joints after the creep test at 650 °C under 150 MPa and 180 MPa is shown in Figure 8.22(a-f). The large number of voids formed near the necking region is due to the plastic deformation caused by the applied stress and softens the matrix. The plastic deformation softens the matrix. The microstructure stability of the P92 steel is crucial for its long-term service at high-temperature and pressure condition. The creep resistance materials are developed based on the principle of obstructing the movement of the dislocations. The appropriate selection of the alloying elements leads to the formation of precipitates ($M_{23}C_6$ and MX) in the matrix, which obstruct the movement of the dislocations (Magnusson and Sandström, 2007). The precipitation strengthening and solid solution strengthening are the two main mechanisms that provide the high-temperature strength to the P92 steel. The finely dispersed $M_{23}C_6$ and MX precipitate prevent the sliding of the grain boundaries and dislocation movement. Thus, enhancing the creep strength of the P92 steel. The other method used to obstruct the movement of the dislocations is by increasing the overall dislocation density in the material. This can be achieved by strain hardening or work hardening process. The high creep strength of the P92 steel is due to its tempered martensite lath structure and high dislocation density in its microstructure. The $M_{23}C_6$ precipitates contribute more to the creep strength of the P92 steel compared to the MX precipitates due to their high number density. Although, the stability of the $M_{23}C_6$ precipitates is lower than the MX precipitates at high temperature. The high stability of the tempered martensite lath structure is due to the high Zener drag force (retarding force of grain boundary motion caused by the presence of fine particles, which leads to a finer grain structure) exerted by particles ($M_{23}C_6$ and MX) of different regions. These particles strengthen the boundary by exerting the Zener force. The Zener drag force starts decreasing under creep condition and transforms the grain boundaries to sub-grain boundaries. The formation of the sub-grain boundaries and coarsening of the precipitates reduce the creep strength of the P92 steel.

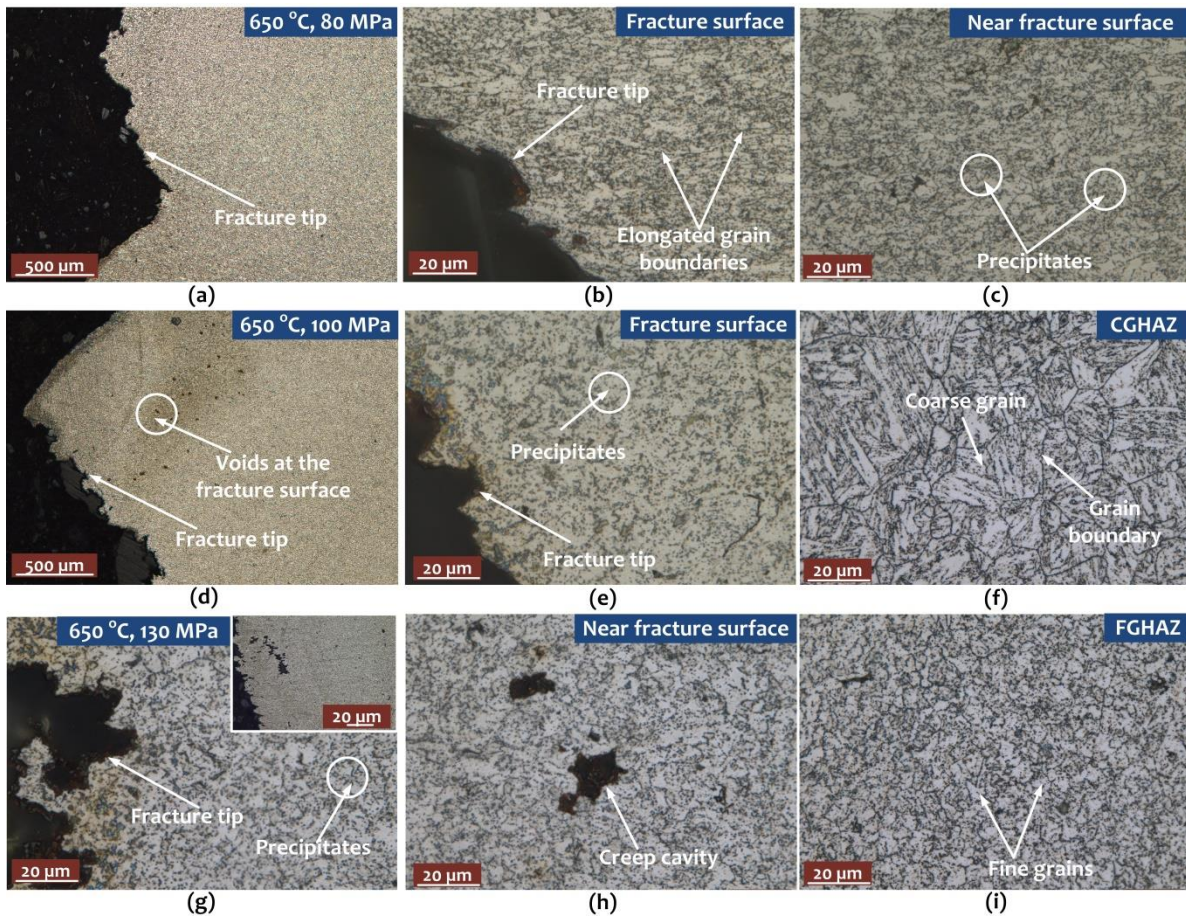


Figure 8.21: Optical microstructure of the creep fracture specimen at 650°C under 80 MPa: (a) fracture surface (b) microstructure at the fracture tip surface (c) microstructure near the fracture tip surface. Optical microstructure of the creep fracture specimen at 650°C under 100 MPa: (d) fracture surface, voids, and cavities can be observed (e) microstructure at the fracture tip surface (f) microstructure at the P92 CGHAZ region. Optical microstructure image of the creep fracture specimen at 650°C under 130 MPa: (g) fracture surface (h) microstructure near the fracture surface and (i) microstructure at the P92 FGHAZ region

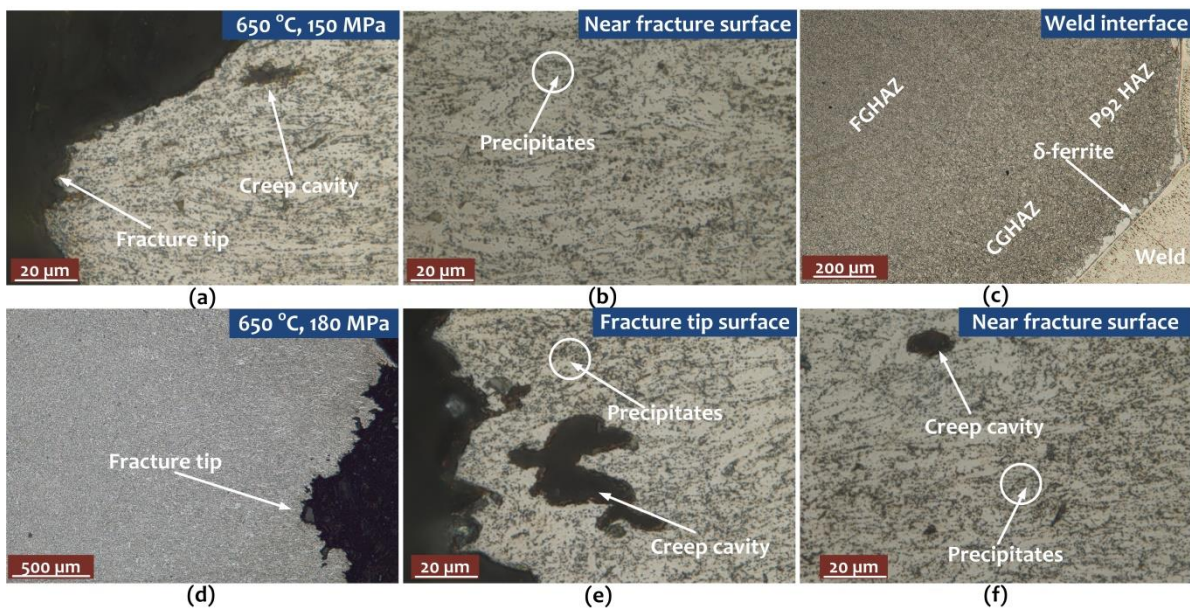


Figure 8.22: Optical microstructure of the creep fracture specimen tested at 650°C under 150 MPa: (a) fracture tip surface (b) microstructure near the fracture surface (c) microstructure at the weld and P92 HAZ interface. Optical microstructure image of the creep fracture specimen tested at 650°C under 180 MPa: (d) fracture tip surface (e) microstructure of the fracture tip surface and (f) voids can be seen near the fracture surface

The FESEM image of the microstructure evolution toward the P92 steel after the creep test at 650 °C under 80 MPa is shown in Figure 8.23(a-f). The FESEM microstructure image showed the absence of the tempered martensite lath structure near the fracture surface. The presence of coarse and fine precipitates near the fracture tip surface can be observed from the microstructure. The $M_{23}C_6$ carbides precipitates, which were partially present in the steel after tempering and further precipitate and coarsen during ageing, play a substantial role in the creep strength of P92 steel. To determine the segregation of the key elements near the fracture region, FESEM EDS and area map analysis was conducted, as shown in Figure 8.23(c) and Figure 8.24. The FESEM EDS analysis showed that, $M_{23}C_6$ precipitates and Laves phase were present near the cavity. The Laves phase form at the cost of consumption of the important W and Mo solid solution strengthening elements from the P92 BM matrix. Thus, the solid solution strengthening effect provided by the W and Mo elements get weaken. The detailed of the failure mechanism due to the formation of the Laves phase after elevated temperature and stress condition is discussed in previous section. The EDS line scan was performed at the BM and weld fusion zone interface for the specimen ruptured after maximum creep service life (650 °C, 80 MPa) to identify the element distribution, as shown in Figure 8.25(a-d). The steep variation of the Cr, Ni, and Fe elements was observed at the weld fusion line, which was in accordance with the chemical composition of the base and weld metal. The EDS map of the secondary phase particle developed in the weld fusion zone for the specimen tested at 650 °C, 80 MPa is shown in Figure 8.25(e). The segregation of the Nb, Mo, and Ti can be observed from the map. The composition of these particles determined by EDS point analysis is listed in Table 8.6.

The FESEM image of the microstructure evolution toward the P92 steel side after the creep test at 650 °C under applied stresses of 100 MPa and 130 MPa is shown in Figure 8.26(a-f). The grain boundaries become obscure as the creep service time increases. The FESEM EDS point and area map analysis was carried out near the fracture tip surface containing coarse and fine secondary phase particles around the cavity to identify the composition of the secondary phase (Figure 8.26(c, f) and Figure 8.27). The area mapping result showed the spot of the W and Cr elements. The EDS point analysis results also showed a strong peak of the W and Cr elements. The dominating particles near the creep cavities are rich in Cr and W elements. Such particle was identified as the Laves phase and carbide precipitates. The possible Laves phase can be $(Fe, Cr)_2(W, Mo)$ type Laves phase and $(Fe, Cr)_{23}C_6$ secondary precipitates. The FESEM-EDS area mapping analysis revealed the spot of the W element. Thus it can be said that the possible Laves phase can be $(Fe, Cr)_2(W)$ type. (Hu et al., 2009) investigated the effect of Laves phase precipitation on the creep resistance of heat-resistant steel. The Laves phase precipitation can be divided into three stages: nucleation stage, growth stage, and coarsening stage. During the nucleation stage, the Laves phase forms at the cost of consumption of the solid solution strengthening W and precipitation strengthening Mo element. In the growth stage, the volume percentage of the Laves phase increases. Once the nucleation and growth stage is completed, the coarsening of the Laves phase occurs. (Brittan et al., 2020) reported that during the initial stage, Laves phase gives beneficial precipitation strengthening effect by reducing the creep rate, but once it coarsens, the creep rate increases. The size of the Laves phase greatly influences the creep properties of heat-resistant steel. Also, as the creep service time increases, the Laves phase loses its coherent relationship with the adjacent laths resulting in the loss of the precipitation strengthening. The solid solution W and Mo elements improve the ductility of the P92 steel by offering a solid solution strengthening. The formation and coalescence of the Laves phase are responsible for the ductile to brittle mode fracture. The chemical composition of the various particles determined by EDS is listed in Table 8.6. The FESEM image of the microstructure evolution toward the P92 steel after the creep test at 650 °C, 150 MPa, and 180 MPa is shown in Figure 8.28(a-f). The EDS

analysis was carried out on the particles around the micro-voids. The FESEM-EDS point analysis (Figure 8.28(f)) revealed a strong peak of Fe and Cr and the low peak of W. Thus, it can be said that particles around the micro-void are $(Fe, Cr)_{23}C_6$. The formation of the Laves phase was not observed for this test condition. At the higher stresses, the creep failure is mainly due to the plastic deformation caused by the grain boundary sliding in the base metal and microstructure examination suggests that without formation of the Laves phase. This large plastic deformation forms in the presence of dimples having various shape and size along the PAGBs and resulting the transgranular fracture (Zhang et al., 2019).

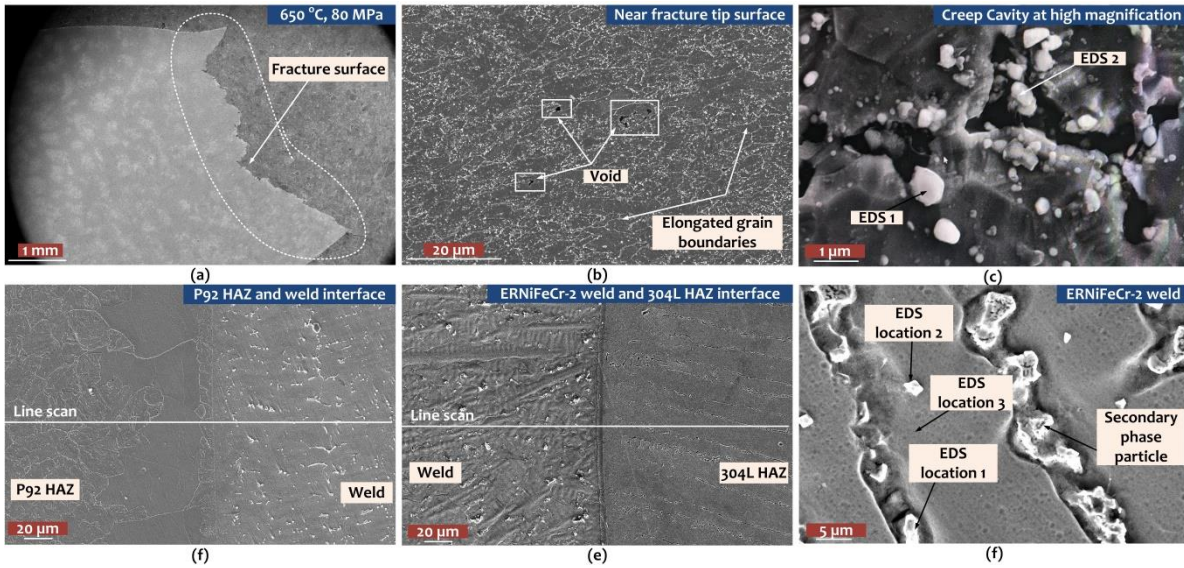


Figure 8.23: FESEM microstructure taken at different region of the creep fracture specimen tested at 650°C under 80 MPa: (a) fracture surface (b) microstructure of the fracture surface (c) creep cavity near the fracture surface at high magnification (d) microstructure at the weld and P92 HAZ interface (e) microstructure at the weld and 304L HAZ interface and (f) microstructure of the ERNiFeCr-2 weld fusion zone

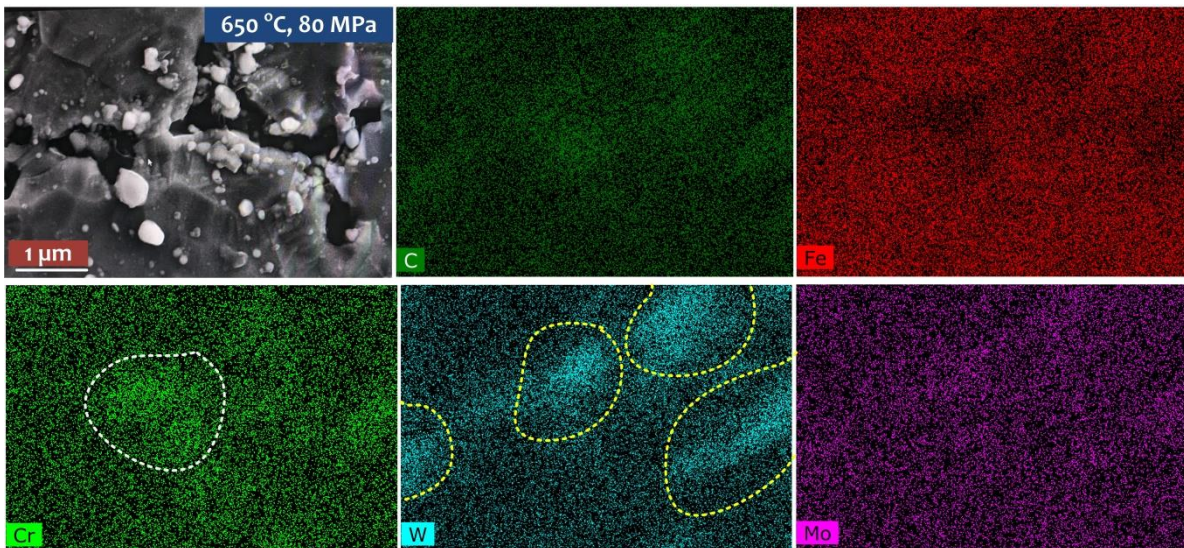


Figure 8.24: FESEM EDS elemental map of creep cavity present near the fracture tip surface for the specimen tested at 650 °C, 80 MPa, Figure 8.23(c) (W and Cr element spots are marked by circle)

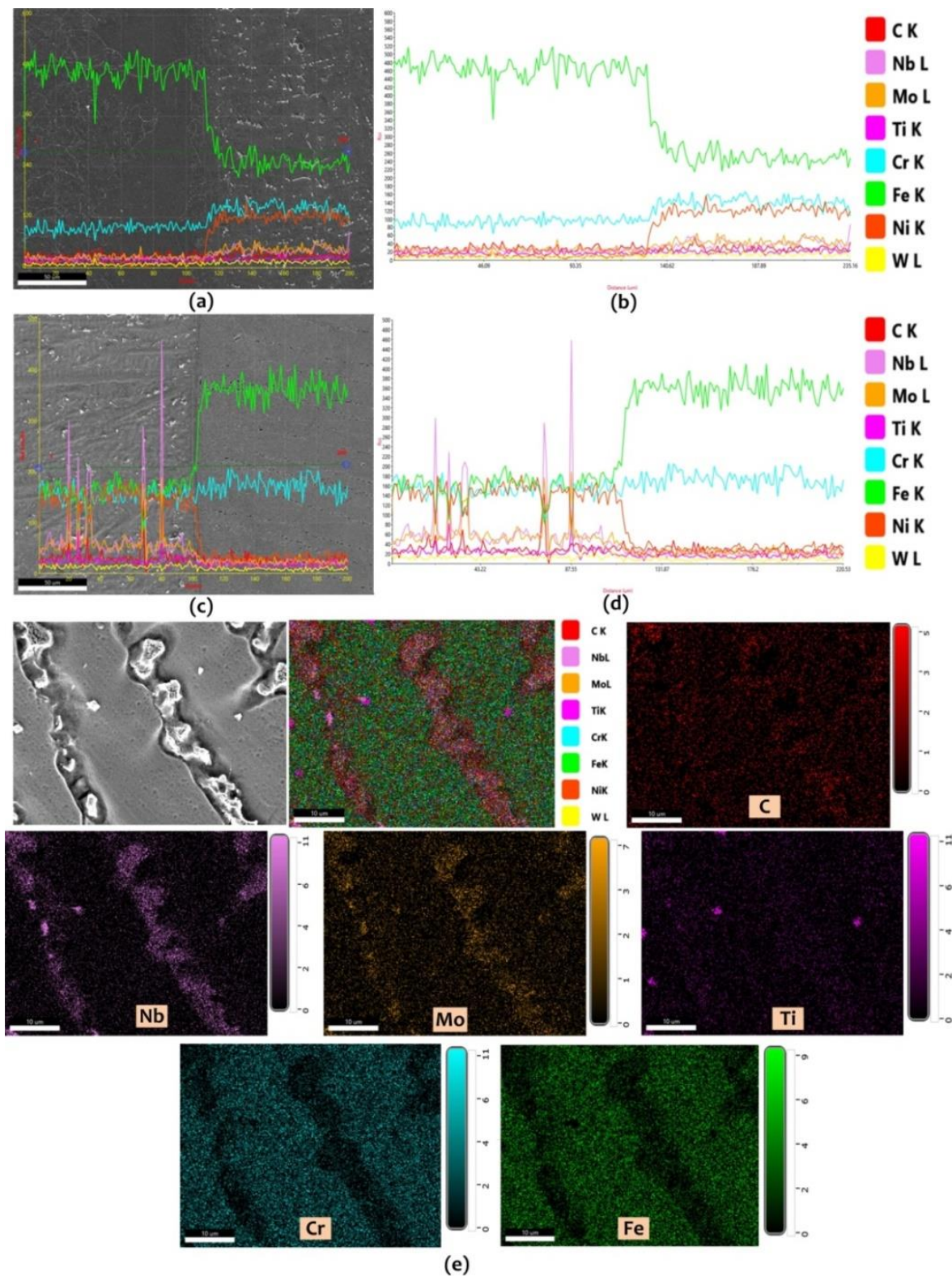


Figure 8.25: (a, b) EDS line scan results across P92 HAZ and weld interface (c, d) EDS line scan results across 304L HAZ and weld interface and (e) EDS elemental map of ERNiFeCr-2 weld fusion zone

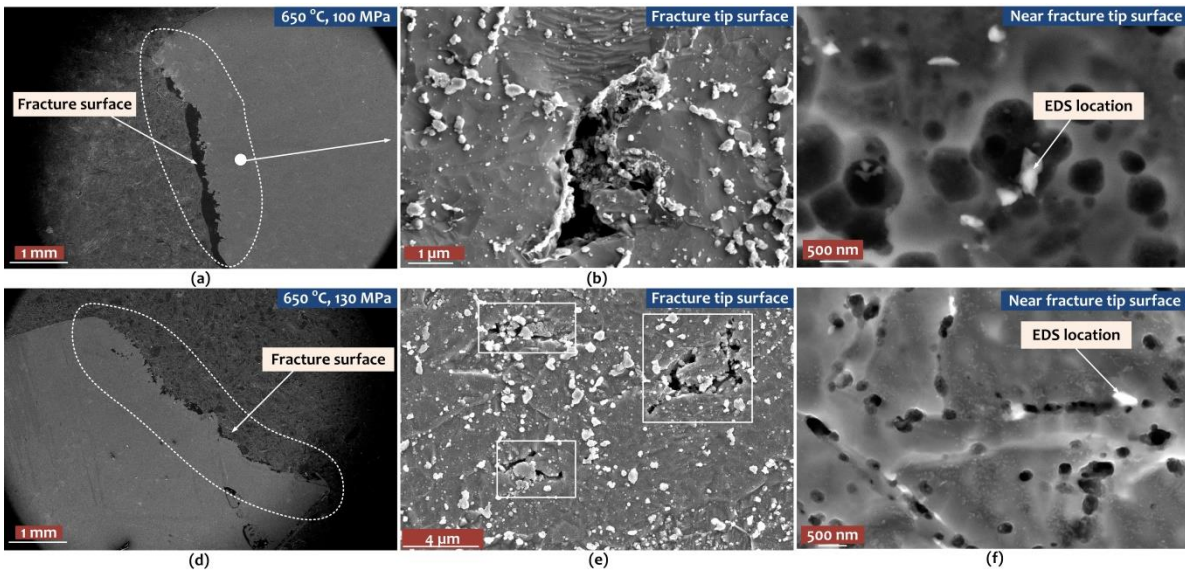


Figure 8.26: FESEM microstructure taken at different regions of the fracture creep specimen tested at 650°C under 100 MPa: (a) fracture surface (b) creep cavity near the fracture tip (c) creep cavity near the fracture tip at high magnification. FESEM microstructure taken at different regions of the fracture creep specimen tested at 650°C under 100 MPa: (d) fracture surface (e) microstructure of the fracture surface, voids can be seen and (f) microstructure near the fracture surface at high magnification

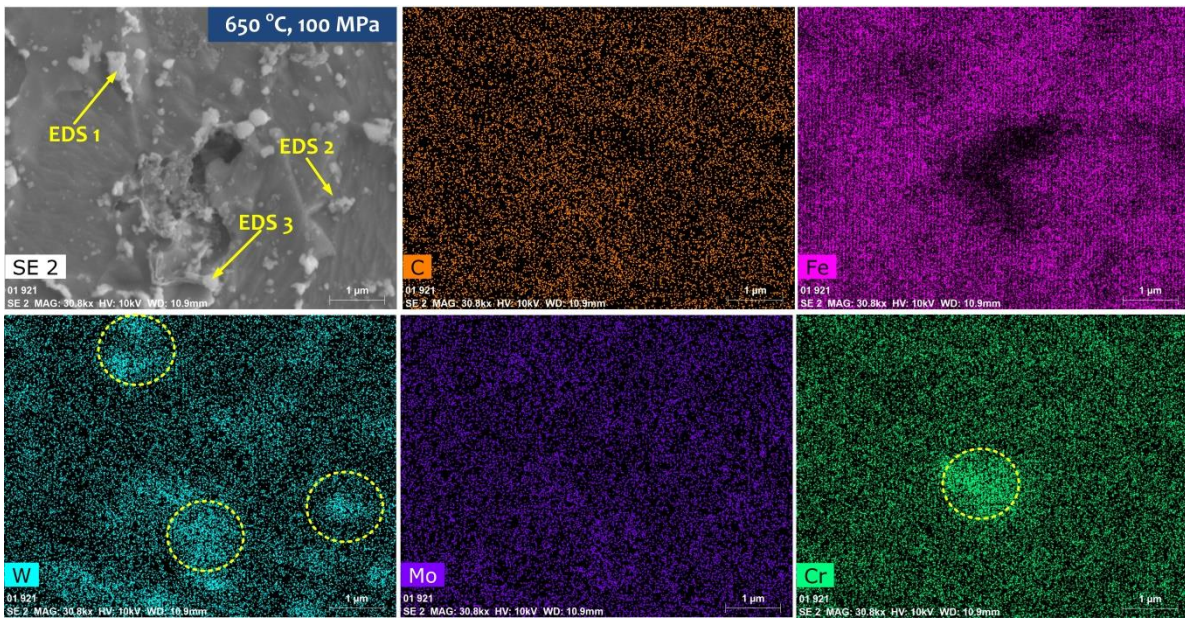


Figure 8.27: FESEM EDS elemental map of creep cavity present near the fracture tip surface for the specimen tested at 650 °C, 100 MPa, Figure 8.26(b) (W and Cr element spots are marked by circle)

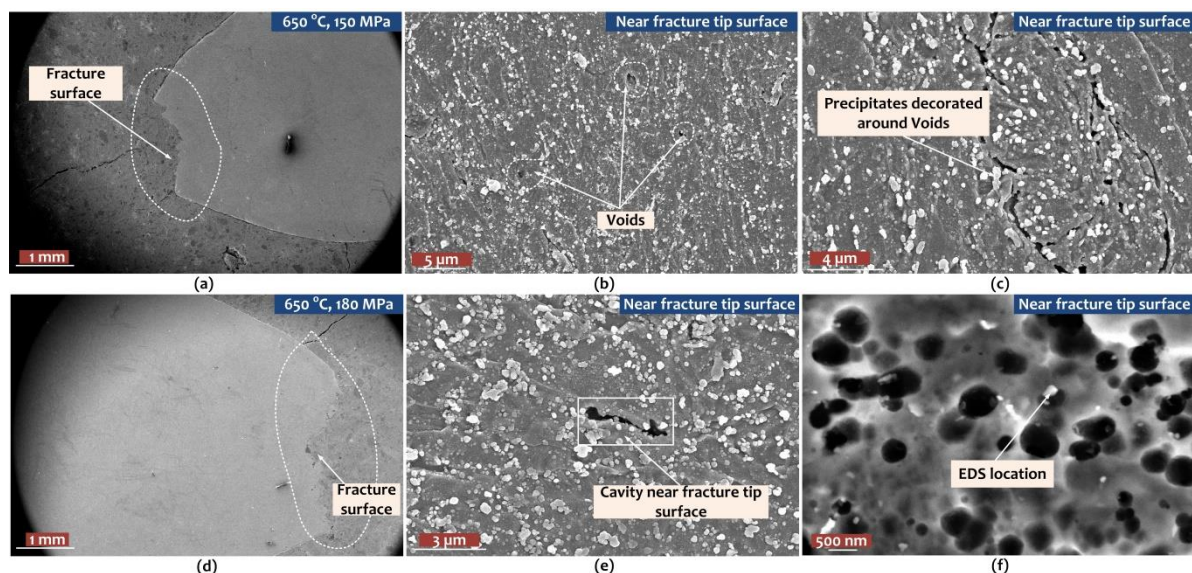


Figure 8.28: FESEM microstructure image of the creep fracture specimen tested at 650°C under 150MPa: (a) fracture surface (b) voids can be seen near the fracture tip surface (c) coarse particles can be seen near the fracture surface at high magnification. FESEM microstructure image of the creep fracture specimen tested at 650°C under 180MPa: (d) fracture surface (e) coalescence of cavities can be seen near the fracture surface and (f) microstructure of the fracture surface at high magnification

Table 8.6 EDS results of the precipitate particles near the fracture surface

	Location	C	Fe	Cr	Mo	Nb	Ni	W	Ti
Figure 8.23(c) (650 °C, 80 MPa)	1	17.11	64.86	6.13	0.17	1.36	0.17	6.15	0.1
	2	11.04	54.09	16.35	0.31	5.41	0.90	2.08	0.1
Figure 8.23(f) (650 °C, 80 MPa)	1	19.0	10.4	5.3	4.2	23.6	34.3	1.4	1.8
	2	27.0	2.4	2.2	1.1	39.1	4.2	0.6	23.4
	3	12.1	24.2	15.1	2.1	2.1	43.6	0.1	0.6
Figure 8.26(c) (650 °C, 100 MPa)	1	28.6	45.5	6.2	2.8	0.3	1.6	14.9	0.1
Figure 8.26(f) (650 °C, 100 MPa)	1	28.8	46.7	6.0	3.1	0.2	0.3	14.9	0.0
Figure 8.27 (650 °C, 130 MPa)	1	9.12	60.39	8.50	1.28	4.48	1.07	15.88	0.1
	2	17.61	57.51	5.74	0.0	2.54	0.59	4.27	0.0
	3	15.22	41.57	20.88	0.27	3.75	0.45	8.58	0.0
Figure 8.28(f) (650 °C, 180 MPa)	1	15.6	76.1	6.7	0.2	0.0	0.4	1.0	0.2

8.3 Creep test for P92-Thermanit MTS 616-304L dissimilar weld joint

The appearance of the fractured creep test specimens is shown in Figure 8.29(a-d). The creep rupture tests were conducted at 650 °C with applied stress ranges of 130-200 MPa. The fracture occurred from the P92 BM for all the test condition. The creep test showed that the Thermanit MTS 616 filler weld fusion zone and 304L have higher creep strength than the P92 steel.

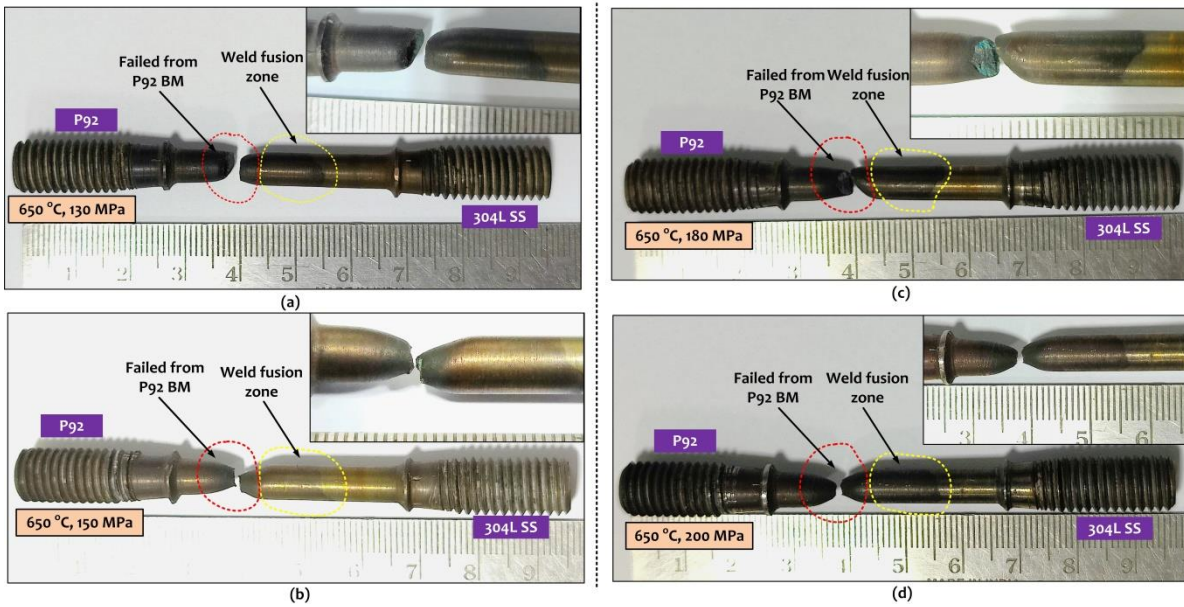
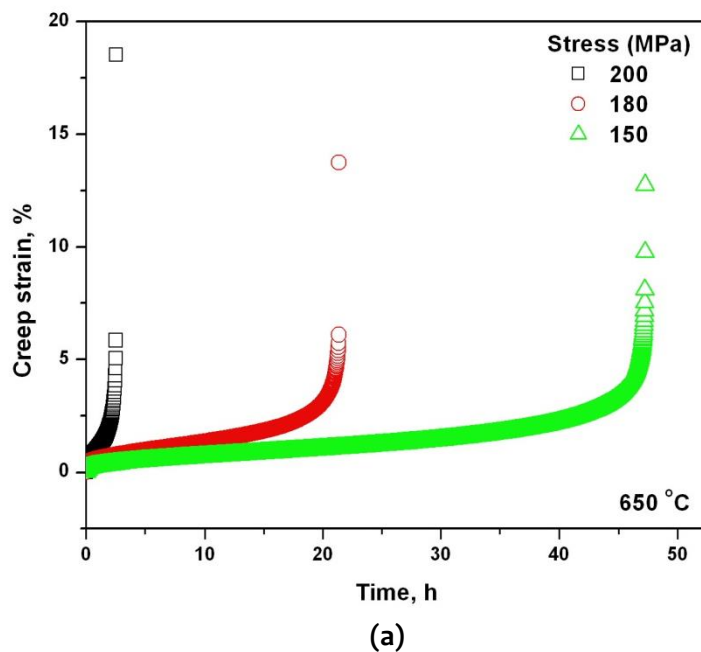


Figure 8.29: Creep failed specimen at different test condition (a) 650 °C, 130 MPa (b) 650 °C, 150 MPa (c) 650 °C, 180 MPa and (d) 650 °C, 200 MPa

The creep curve is the plot of the plastic strain v/s time. In the creep test, the deformation strain was measured as function of time under the constant load and temperature. Figure 8.30(a, b) shows the creep strain against the time plot of P92/304L DWJ, which have been performed until failure of the specimen. The maximum creep rupture time (242.1 h) was obtained at 650 °C, 130 MPa. At the same time, minimum rupture time (2.53 h) was obtained at 650 °C, 200 MPa. The different stages of the creep curve, including the primary creep stage, secondary stage, and tertiary stage, can be observed from the creep curve. The summary of the creep test results is given in Table 8.7. The percentage elongation and reduction in the cross-sectional area were also measured from the fractured specimens, as reported in Table 8.7. Furthermore, the measured data have been plotted in the form of a log-log scale between creep rate and time in Figure 8.31 to measure the MCR at each applied stress. One can observe from Figure 8.31, that MCR is found to be decreased with a decrease in the applied stress.



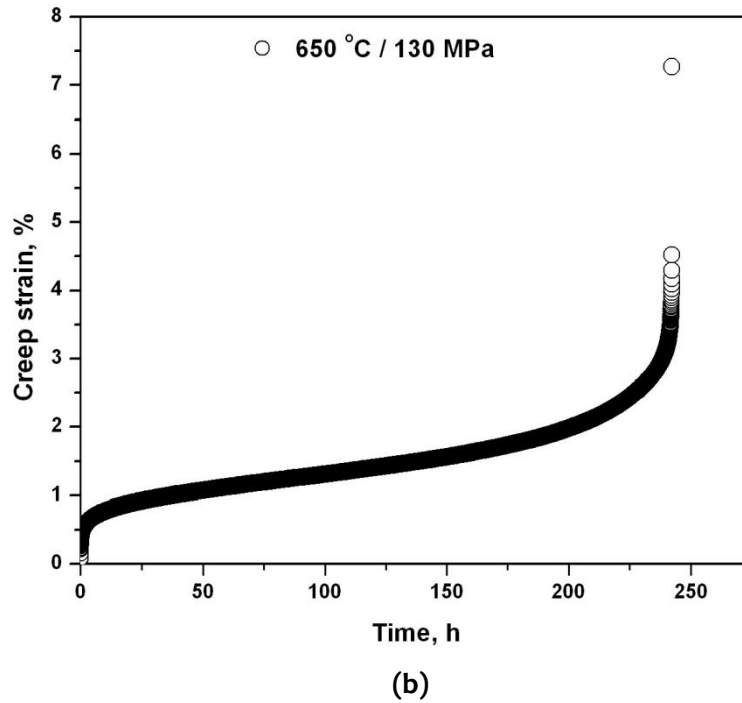


Figure 8.30: Creep curves (creep strain vs. time) of P92 weld joints performed at 650°C under the higher stress range of 150-200 MPa and 130 MPa as shown in (a) and (b), respectively

Table 8.7 Summary of the creep tests results of Thermanit MTS 616 filler weld joint

Temperature (°C)	Stress (MPa)	Rupture time (h)	Elongation (%)	Reduction Area (%)
650	130	242.10	7.27	50.71
650	150	47.25	12.76	78.57
650	180	21.34	13.75	62.31
650	200	2.53	18.51	84.9

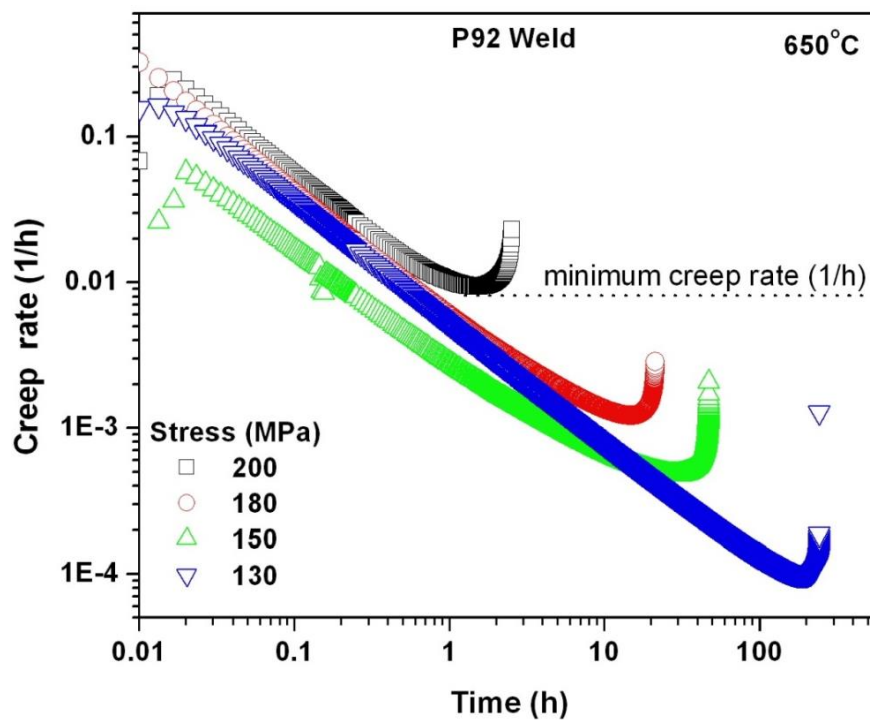


Figure 8.31: Creep rate vs. time plot of P92 weld joint shown to find minimum creep rate by plotting the data in the form of log-log scale

The obtained creep data from the P92 weld joints specimens have also been plotted in the form of log-log scale using the power law equation ($\dot{\epsilon}_m = A \sigma^n$), where A is the material constant. The stress dependence of minimum creep rate ($\dot{\epsilon}_m$) data is linearly fitted and results in the best fit. The slope of the fitted line is known as stress exponent (n) and found to be 9.70, along with the value of A ($2.963\text{E-}25 \text{ MPa}^{-n} \text{ h}^{-1}$), as shown in Figure 8.32. The obtained A and n are comparable with the literature data of the P92 steel (Guguloth and Roy, 2017). Furthermore, the creep rupture time and stress data have been plotted in a log-log scale, as shown in Figure 8.33. The data were linearly fitted and the obtained stress exponent is approximately 9.6. One can notice that the values of stress exponents from Figure 8.32 and Figure 8.33 are insignificant and it has been found for P92 weld joints (Guguloth and Roy, 2017). The creep test cannot be conducted for the whole creep service life of the material. Thus, the linear regression analysis was used to determine the correlation between the applied stress (σ , MPa) and rupture time (t_R , hr). The obtained trend line using linear fit is shown in the graph (Figure 8.33). The obtained R-squared value (co-efficient of the determination) is determined as ≈ 0.953 , which indicates the excellent fitting of the data in the trend line. The obtained equation to predict the creep life at applied stress could be given by Eq. (8.6):

$$\log(t_R) = 22.57566 + (-9.55294) \log(\sigma) \quad (8.6)$$

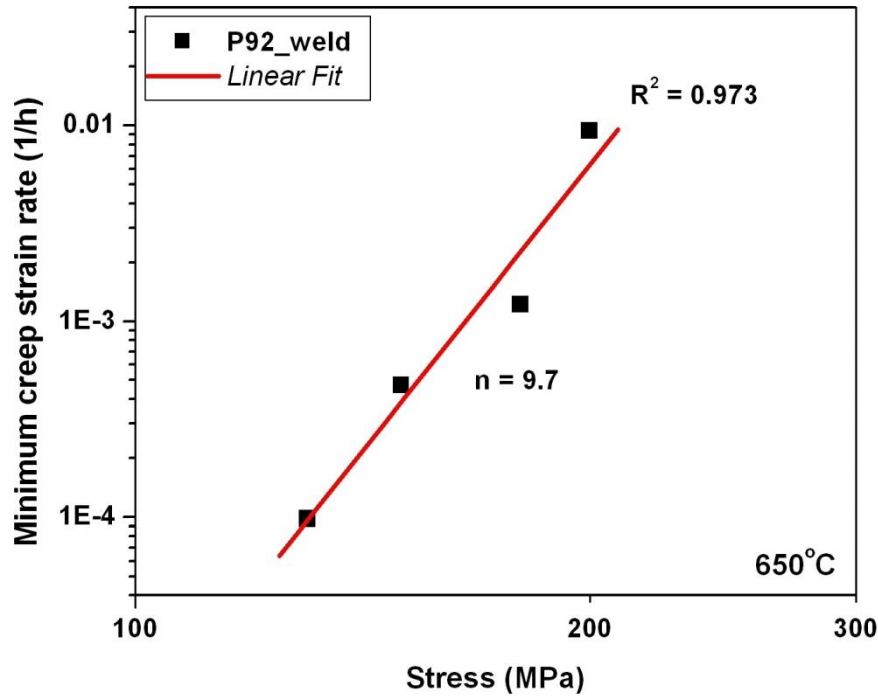


Figure 8.32: Log-log plot of minimum creep rate against the applied stress of P92 weld joints. The data are linearly fitted with the best-fit line to determine stress exponent ($n \approx 9.7$)

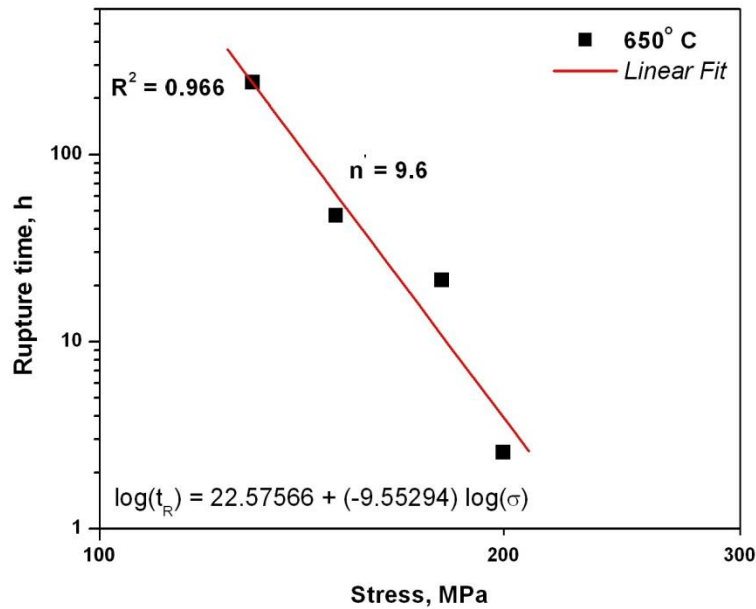


Figure 8.33: Log-log plot of rupture time vs. stress for P92/304L DWJ at 650°C

Figure 8.34 illustrates the plot of stress v/s LMP on the log scale that was created using the data from Table 8.8. From the trend line, the value of LMP corresponding to the practical stress can be obtained. Based on the obtained LMP value, the corresponding rupture life can be estimated. The obtained equation to predict the creep life at the applied stress could be given by Eq. (8.7):

$$Y = (-10.35158) + (0.00114)X + (-2.55763E-8)X^2 \quad (8.7)$$

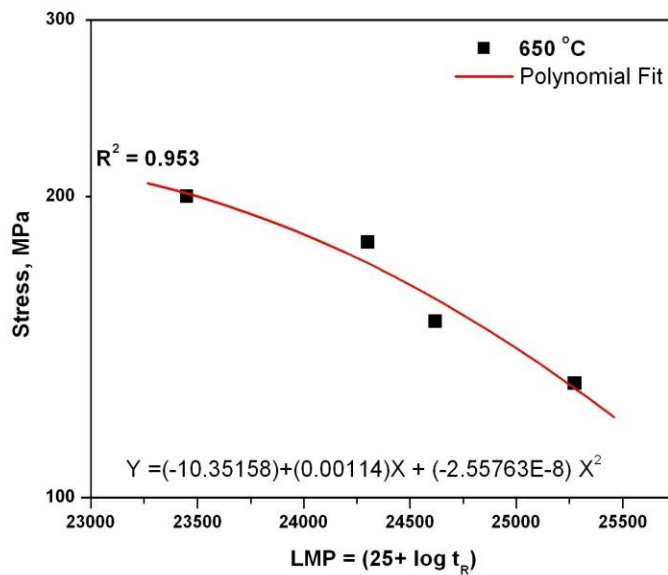


Figure 8.34: Plot of stress vs. LMP data at 650°C have been fitted using second order polynomial fit

Table 8.8 LMP value and rupture time at a temperature of 650 °C

Temperature (°C/K)	Stress (MPa)	Rupture time (h)	Larson-Miller parameter (LMP)
650 °C /923 K	130	242.10	25275
650 °C /923 K	150	47.25	24620
650 °C /923 K	180	21.34	24301
650 °C /923 K	200	2.53	23447

8.3.1 Macrostructure after creep test

Figure 8.35(a-d) shows the cross-section macrostructure of the creep-tested specimens consisting of the fracture surface, both the BM and weld fusion zone. For all the test conditions, the fracture occurred from the P92 BM. The crack formation near the weld and 304L BM interface and at the P92 filler weld fusion zone was observed. However, the failure occurred from the region away from the 304L BM and P92 filler weld fusion zone. Additionally, it was seen from the macrograph that the specimen that failed during the creep test lacked sufficient necking. The obvious necking for the specimen tested at 200 MPa indicated the ductile fracture. For all the creep test conditions, the specimen failed from the P92 BM, indicating the sound quality of the weld at elevated temperature and stress conditions. The creep test results also showed that the 304L has better creep strength and thermal stability at high-temperature and stress than P92 steel.

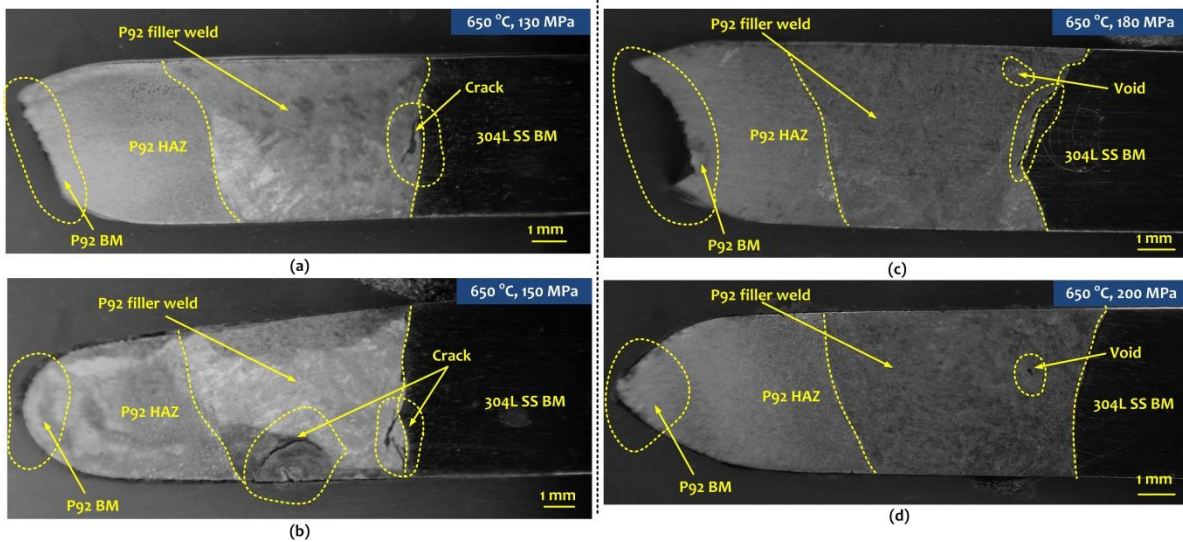


Figure 8.35: (a-d) Overall macrostructure of the creep ruptured specimen consisting of necking region, P92 HAZ, weld fusion zone and 304L BM

8.3.2 Microstructure changes after creep test

After the creep test, the microstructure near the necking region, weld fusion zone, and P92 HAZ region are shown in Figure 8.36(a-f) and Figure 8.37(a-f). From the optical microscope image, precipitates and elongated PAGBs were observed near the fracture surface. The creep cavities nucleate along the PAGBs. The formation of the creep cavities at the intersection of the grain boundaries was observed during the microstructure study. The creep strength of the P92 steel depends upon the thermal stability of the $M_{23}C_6$ and MX precipitates at high-temperature. Also, the thermal stability of the $M_{23}C_6$ and MX precipitates is different. The $M_{23}C_6$ precipitates dissolve at a temperature of approximately 830 °C, while the MX precipitates dissolve at a temperature beyond 1200 °C. The coarsening and dissolution of the precipitates are the two main phenomena occurring at high-temperature. The small particles dissolve while the coarse particles coarsen (Zhou et al., 2021). The $M_{23}C_6$ precipitates are generally bigger in size compared to the MX precipitates. These $M_{23}C_6$ precipitates coarsen rapidly at higher temperatures and cause the stress concentration along the PAGBs. Due to this, the ability of these precipitates to hinder the motion of the dislocation decreases. The $M_{23}C_6$ particles are mainly enriched with Cr and Mo. Also, Cr and Mo are the main solid-solution strengthening elements in the P92 base metal matrix. The coarsening of the $M_{23}C_6$ particles will decrease the solubility of the solid solution strengthening elements in the matrix. This would lead to the softening of the matrix. The coarsening of the $M_{23}C_6$ particles will also result in a higher stress concentration between the grain boundaries and the precipitates. This will result in the formation of voids at the grain boundaries (Liu et al., 2015). After long-term creep service, the formation of the lath phase occurs. The formation of the lath phase [such as $Fe_2(Mo)$

and $\text{Fe}_2(\text{W})$] takes place due to the depletion of the W from the solid solution. The formation of the lath phase takes place at the grain and sub-grain boundaries. When the size of the precipitated lath phase is below the critical limit, it provides precipitation strengthening. Although, when the size of the lath phase reaches beyond the critical limit, it provides a possible cavity site for the creep rupture. The cavities formed near the lath phase cause the brittle fracture. The formation of the Z-phase takes place after a long service. The formation of the Z-phase $[(\text{Cr}, \text{Nb})\text{N}]$ at the cost of consumption of the MX carbonitrides $[\text{Nb}(\text{C}, \text{N})$ and $\text{V}(\text{C}, \text{N})]$ takes place after the long-term creep test of about 10,000 h (Fedoseeva et al., 2016).

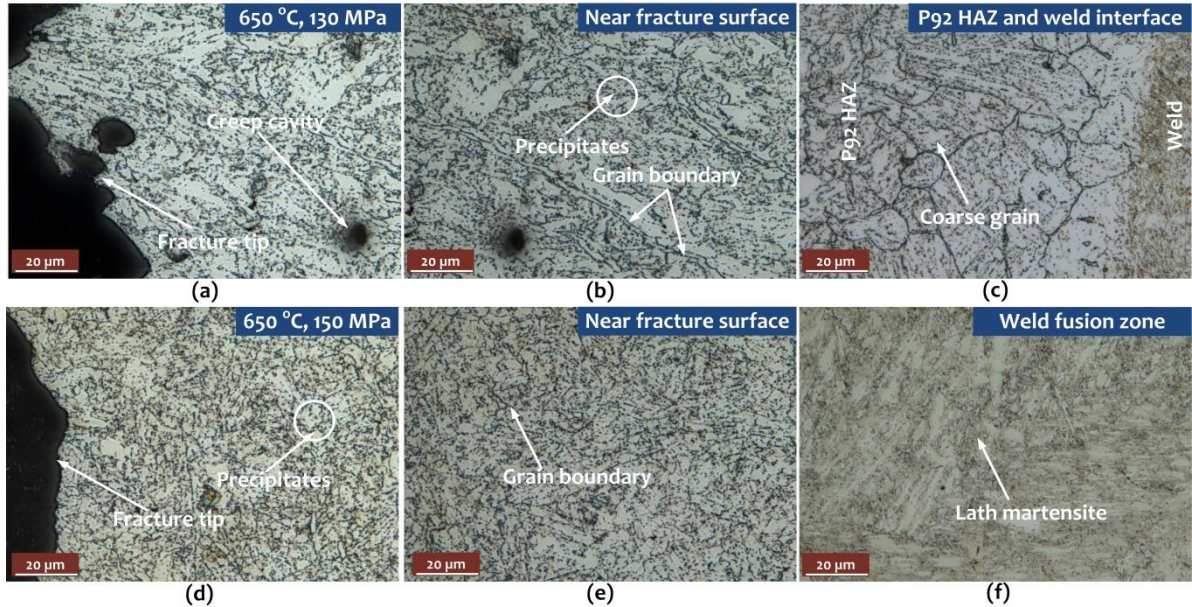


Figure 8.36: Optical microstructure of the creep ruptured specimen tested at 650 °C, 130 MPa: (a) fracture surface, voids can be seen (b) microstructure near the fracture surface (c) microstructure of the weld and P92 HAZ interface (d) fracture surface microstructure of the creep ruptured specimen tested at 650 °C, 150 MPa (e) microstructure near the fracture surface and (f) microstructure of the P92 filler weld fusion zone

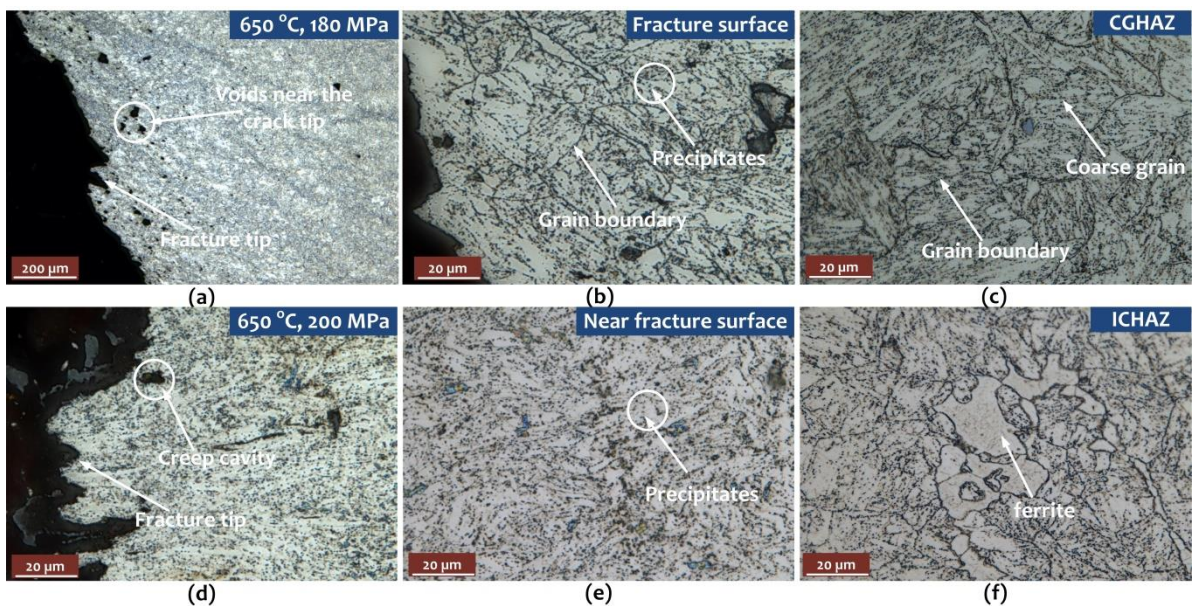


Figure 8.37: Optical microstructure of the creep ruptured specimen tested at 650 °C, 180 MPa: (a) fracture tip surface, voids can be seen (b) microstructure of the fracture tip surface at high magnification (c) P92 CGHAZ region (d) fracture tip surface microstructure of the creep ruptured specimen tested at 650 °C, 200 MPa (e) microstructure near the fracture surface and (f) ICHAZ region microstructure

The FESEM microstructural analysis was further conducted to identify the mechanism of the voids/cavity formation. The FESEM microstructural examination was conducted over the cross-section of the fractured creep specimen tested at 650 °C under the stress range of 130-200 MPa. The FESEM micrograph of the creep fractured specimen near the fracture tip surface and HAZ region is shown in Figure 8.38(a-d). The extensive void formation along the PAGBs can be observed from the FESEM micrographs. Compared to the necking region, fewer creep cavities were formed in the CGHAZ and FGHAZ regions. The large number of voids generated near the fracture surface indicates sufficient plastic deformation. During the creep test under high-stress and temperature conditions, high-stress concentration is caused at the grain boundaries. The grain boundary movement and high-stress concentration by the precipitates causes the micro-voids near the fracture surface. As time passes, these micro-voids combine and form a crack due to continuous loading. The cavities formation was observed along the grain boundaries. This causes the transgranular fracture from the P92 BM. The segregation of the key alloying elements such as W weakens the strength of the material due to the loss of the solid solution and precipitation strengthening mechanism. To determine the segregation of the elements near the fracture region, FESEM EDS and area map analysis was conducted, as shown in Figure 8.38(c) and Figure 8.39. The composition of these particles determined by EDS point analysis is listed in Table 8.9. The EDS point analysis results showed a strong spectrum of the W and Cr element. The combination of the EDS and area mapping analysis result showed that the particles present near the creep cavity were rich in W and Cr and poor in Mo, which are type Fe_2W laves phase. Thus, the dominating particles near the creep cavity are Cr_{23}C_6 and Fe_2W laves phase. The formation of the voids, secondary phase particles, and reduced dislocation density is the main reason behind the softening of the P92 steel. The FESEM microstructure analysis for fractured creep specimen tested at 650 °C under the applied stress range of 150-200 MPa is shown in Figure 8.40(a-i). The FESEM-EDS point analysis (Figure 8.40(h)) revealed a strong peak of Fe and Cr and the low peak of W. Thus, it can be said that particles around the micro-void are $(\text{Fe}, \text{Cr})_{23}\text{C}_6$. The formation of the laves phase was not observed for this test condition.

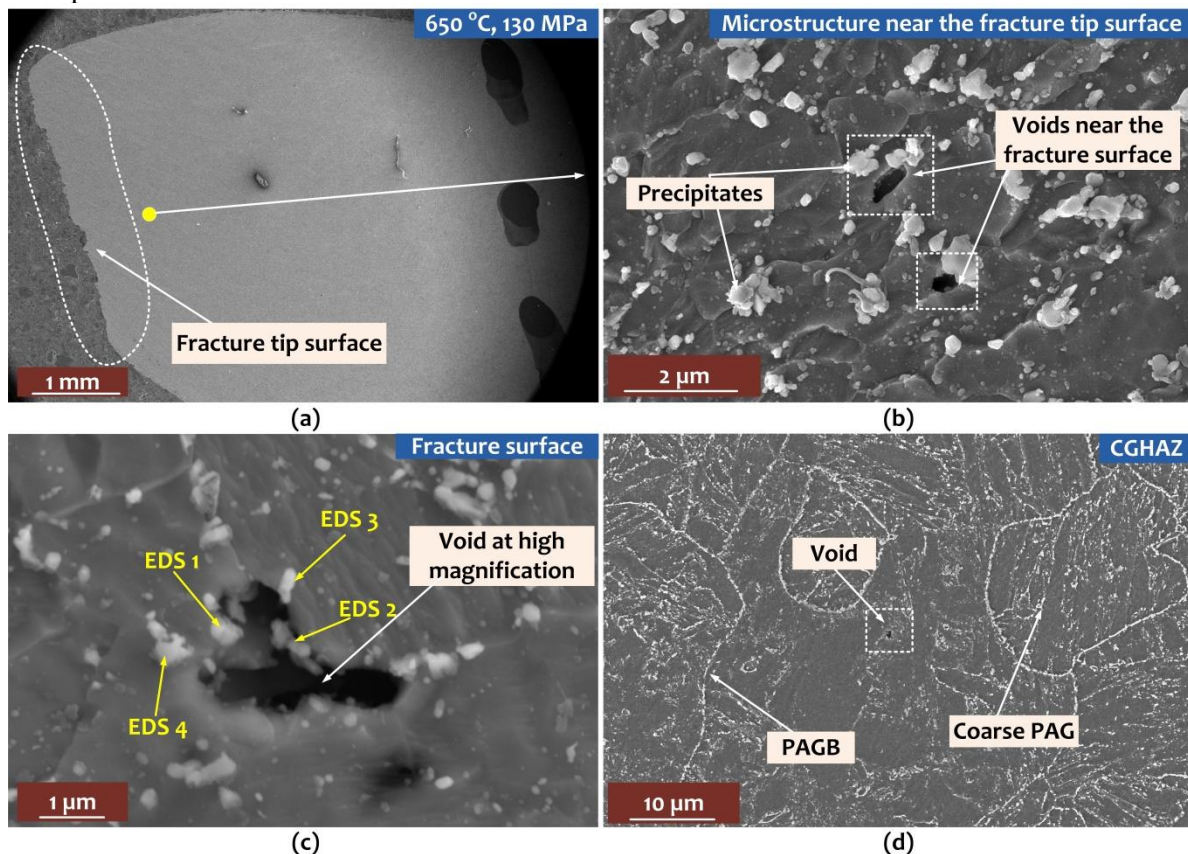


Figure 8.38: FESEM microstructure of the fracture creep specimen at 650 °C, 130 MPa (a) fracture surface (b, c) microstructure near the fracture surface and (d) microstructure at the P92 CGHAZ region, voids can be seen

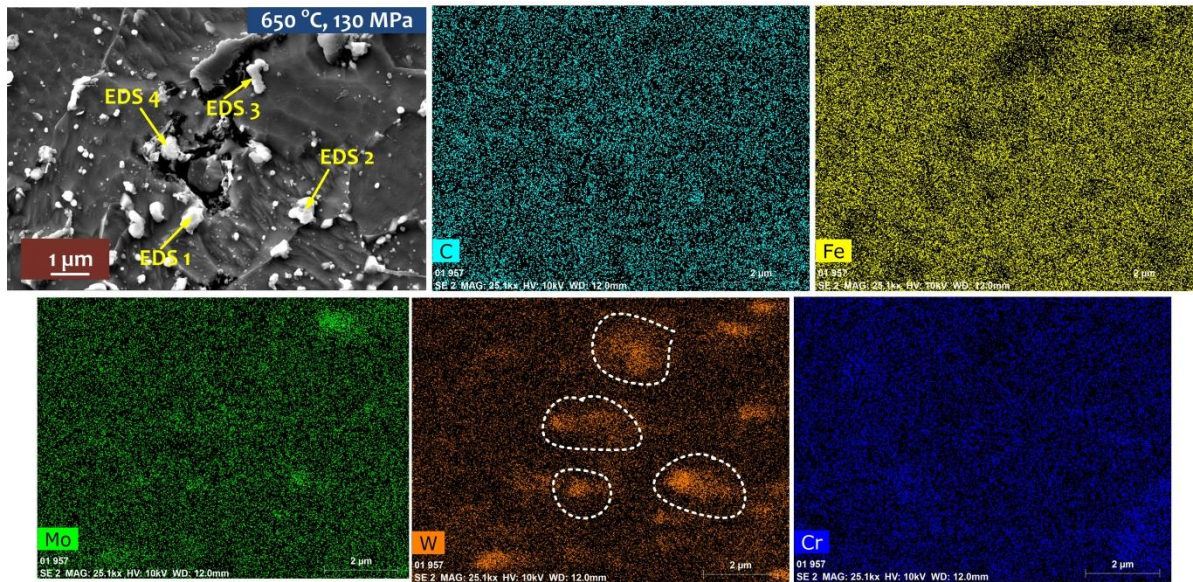


Figure 8.39: EDS elemental map of creep cavity present near the fracture tip surface for the specimen tested at 650 °C, 130 MPa (few W and Cr elements spots are marked by circle)

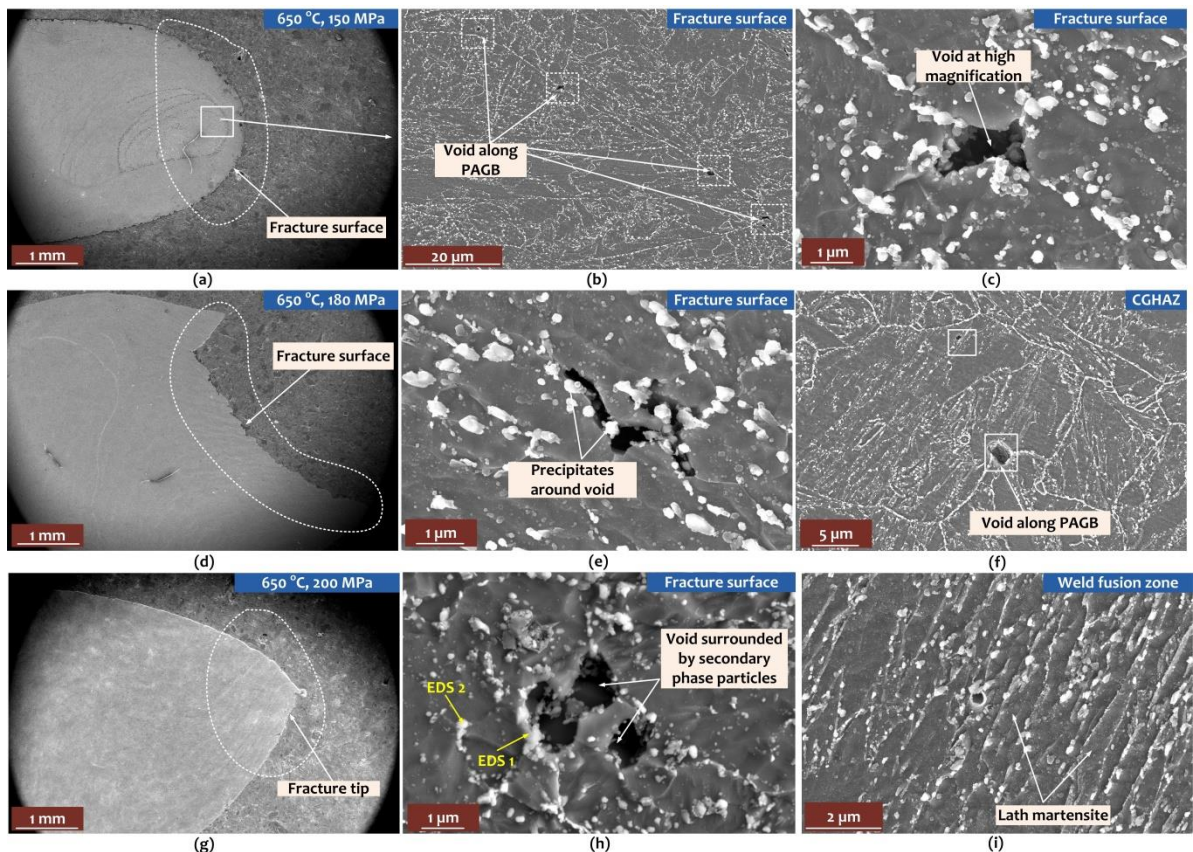


Figure 8.40: FESEM microstructure of the fracture creep specimen at 650 °C, 150 MPa: (a) fracture surface (b, c) microstructure near the fracture surface at low and high magnification. FESEM microstructure of the fracture creep specimen at 650 °C, 180 MPa: (d) fracture surface (e) microstructure near the fracture tip surface (f) microstructure at the P92 CGHAZ region, voids can be seen along the PAGBs. FESEM microstructure of the fracture creep specimen at 650 °C, 200 MPa; (g) fracture tip surface (h) microstructure near the fracture surface and (i) microstructure of the weld fusion zone

Table 8.9 EDS quantitative analysis results of the precipitate particles near the fracture surface

	Location	C	Fe	Cr	Mo	Nb	Ni	W
Figure 8.38(c) 650 °C, 130 MPa	1	17.11	64.86	6.13	0.17	1.36	0.17	6.15
	2	14.83	62.60	7.79	0.23	3.42	0.13	11.41
	3	11.15	42.09	34.60	0.0	5.35	1.05	5.45
	4	19.23	69.21	5.96	0.0	1.46	0.12	2.59
Figure 8.39 650 °C, 130 MPa	1	9.12	60.39	8.50	1.28	4.48	1.07	15.88
	2	17.61	57.51	5.74	0.0	2.54	0.59	4.27
	3	15.22	41.57	20.88	0.27	3.75	0.45	8.58
	4	22.06	40.50	15.06	0.0	4.04	0.79	2.66
Figure 8.40(h) 650 °C, 200 MPa	1	11.04	54.09	16.35	0.0	5.41	0.90	2.08
	2	23.65	63.90	6.14	0.0	2.56	0.48	1.21

8.4 Creep test for P92-ERNiCr-3-304L dissimilar weld joint

The appearance of the specimens after creep test is shown in Figure 8.41(a). The cross-section macrograph of the fractured creep specimen is shown in Figure 8.41(b). The macrograph demonstrates that the ERNiCr-3 weld fusion zone and 304L was free from the voids and deformation. It was seen from the macrograph that the P92/304L DWJ failed from the P92 BM, when tested at temperature 650 °C and stress 100-180 MPa. The weld fusion line was taken as the reference point and distance of fracture tip surface from weld fusion line was measured. For all the creep test condition, the fracture site was ~ 4 mm away from the weld fusion line, suggesting a fracture from the P92 BM. As discussed in previous section, the span of the HAZ is 2.35 mm. The fracture point can be identified as P92 BM because the distance of the fracture tip surface from the weld fusion line exceeds the width of the HAZ (2.35 mm).

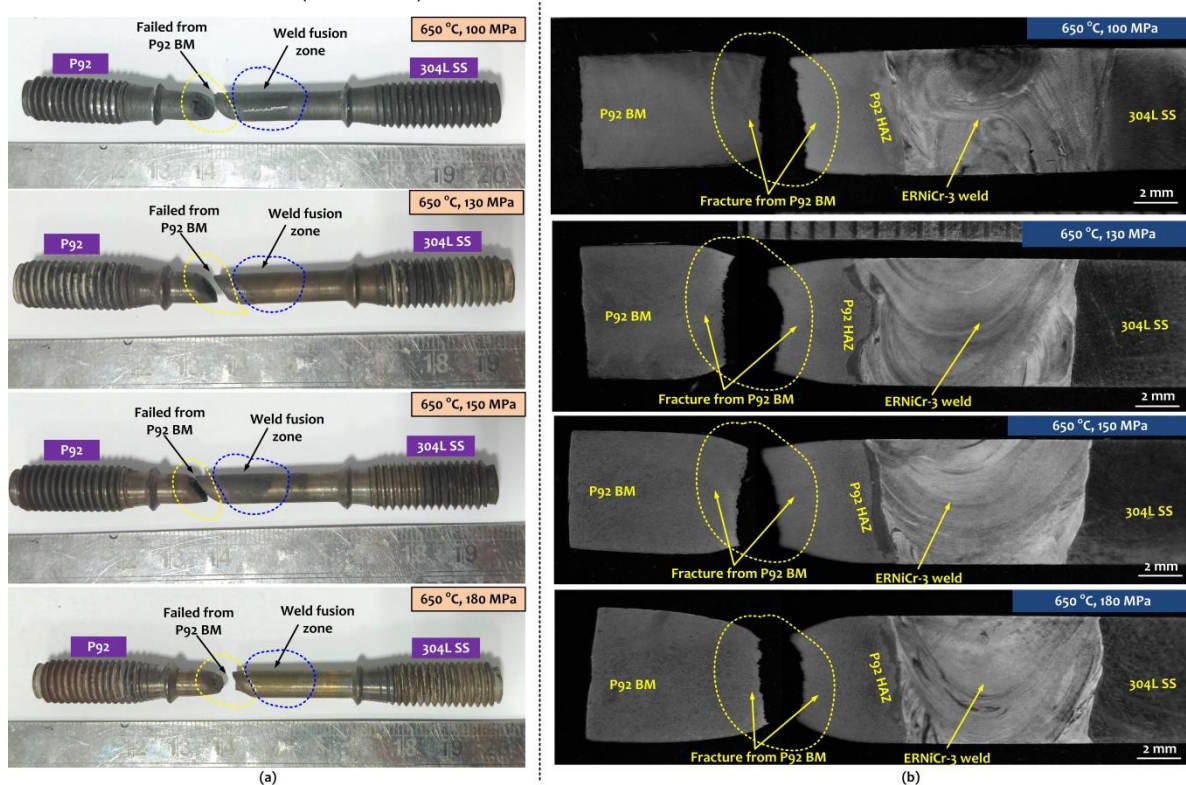


Figure 8.41: (a) Appearance of the creep specimens after fracture and (b) macrograph of the fractured creep specimen

The creep curve obtained at 650 °C temperature and at a stress range of 100-180 MPa is shown in Figure 8.42 (a, b). Furthermore, the measured data have been plotted in the form of a log-log scale between creep rate and time in (Figure 8.42(c, d)) to measure the MCR at each applied stress. The creep test result is summarized in Table 8.10.

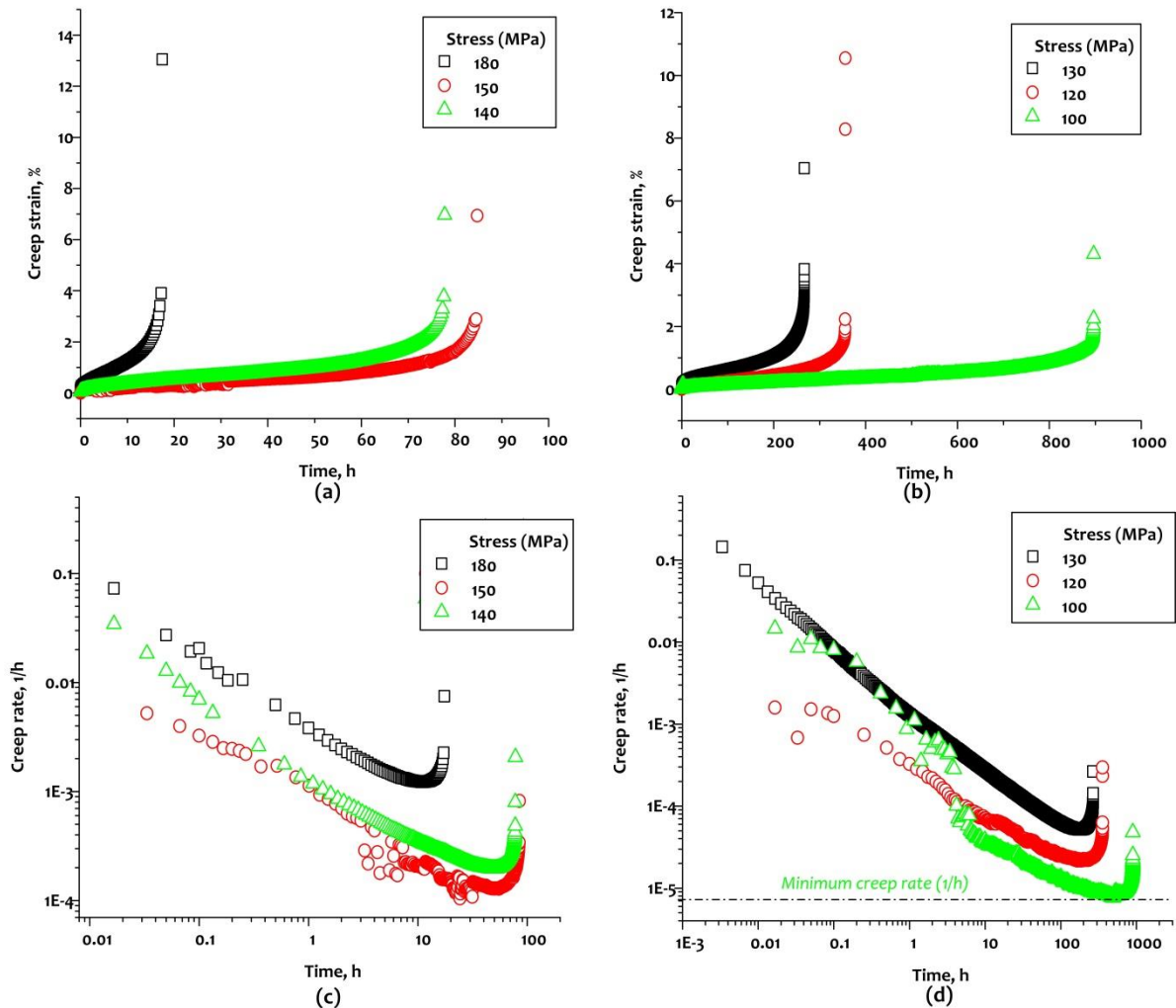


Figure 8.42: Creep curves of ERNiCr-3 filler tested P92/304L DWJ at 650 °C under the stress range of 100-180MPa. (a) Short term creep and (b) long-term creep and their creep rate plots were given in (c) and (d), respectively

Table 8.10 Creep tests results of P92/304L DWJ (ERNiCr-3 filler metal)

Temperature	Stress, MPa	Time to rupture, h	Creep strain, %
650°C	180	17.7	13.05
	150	84.8	6.94
	140	77.85	6.96
	130	267.03	7.04
	120	356.15	10.55
	100	897	4.31

The obtained creep data have also been plotted in the form of log-log scale (Figure 8.43) using the power law equation ($\dot{\epsilon}_m = A \sigma^n$), where A is the material constant. The stress dependence of minimum creep rate ($\dot{\epsilon}_m$) data is linearly fitted and results in the best fit. The slope of the fitted line is known as stress exponent (n) and found to be 8.6.

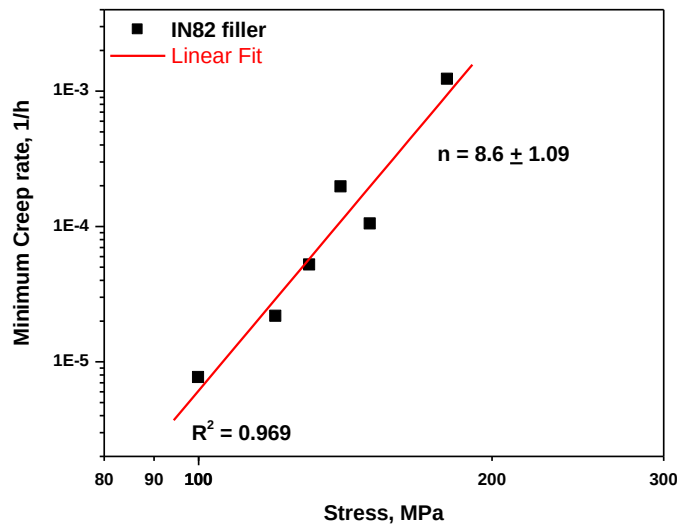


Figure 8.43: Stress dependence of minimum creep rate has been plotted to determine stress exponent (n) at 650°C for ERNiCr-3 filler. The obtained ‘ n ’ value is approximately 8.6 in the stress range of 100-180MPa

8.4.1 Microstructure changes after creep test

The FESEM micrograph of the creep fractured specimen near the fracture tip surface and HAZ region is shown in Figure 8.44(a-f) and Figure 8.45(a-f). All the joints were failed from the P92 BM irrespective of the creep test condition selected in current investigation. The Type IV cracking which occurs from the FGHAZ or ICHAZ of P92 steel was not observed in this investigation. The FESEM microstructure showed the micro-crack and precipitates near the fracture tip surface. The W and Mo element is added to improve the creep resistance of the P92 steel by solid solution strengthening. The V, Nb and Fe, Cr elements provide the precipitation strengthening. The coarsening of the $M_{23}C_6$ precipitates take place during the thermal exposure. Due to the coarsening of the $M_{23}C_6$ precipitates, the stress required for dislocations to climb over precipitates decrease. During the high temperature service, the structure of the lath martensite becomes inconsistent, the width of the lath martensite increases. This results into the reduction in the dislocation density of the lath martensite. The small size laths with large numbers give the desirable mechanical properties at elevated temperature service. Also, the $M_{23}C_6$ precipitates gather and grow at the PAGBs. Furthermore, the formation of the lath phase take place at the cost of consumption of the solid solution strengthening W and Mo elements. This overall microstructural changes result into the loss of the strength of the P92 steel at elevated temperature and stress service. The creep cavities grow with the applied stress and convert into the creep crack with the progress of the test, which ultimately lead to the creep fracture of the joint. The FESEM EDS analysis and area mapping analysis showed the similar kinds of results, as discussed in previous section.

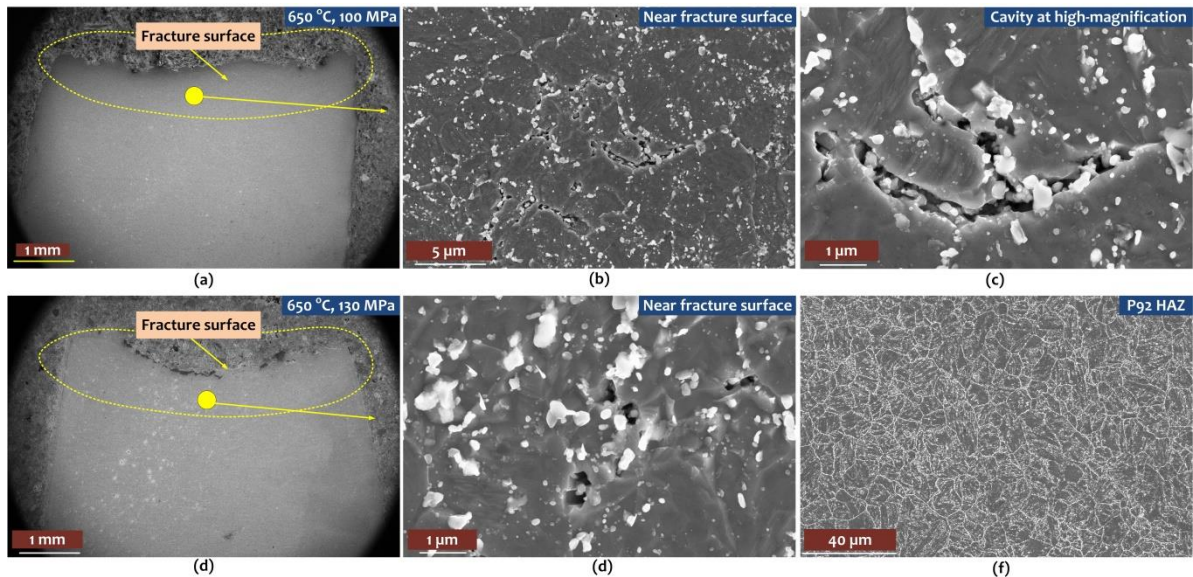


Figure 8.44: FESEM characterization of the fracture tip surface (a) fracture surface of the creep specimen tested at 650 °C, 100 MPa (b) microstructure showing the cavities near the fracture tip surface (c) creep cavity at high magnification (d) fracture surface of the creep specimen tested at 650 °C, 130 MPa (e) microstructure showing the creep cavities near the fracture tip surface at high magnification (f) microstructure of the HAZ region

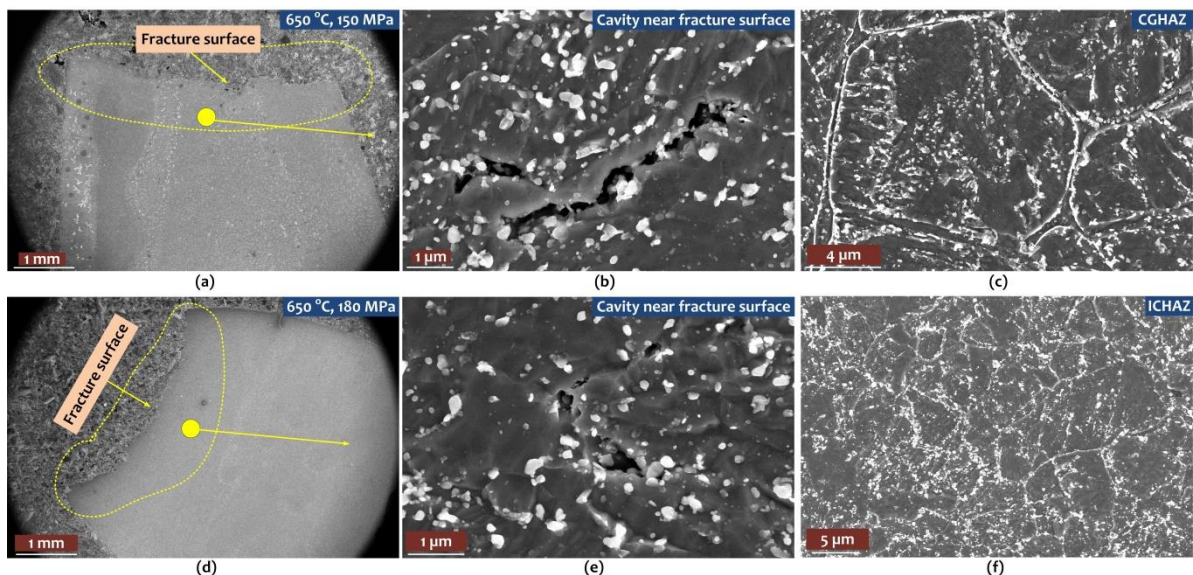


Figure 8.45: FESEM characterization of the fracture tip surface (a) fracture surface of the creep specimen tested at 650 °C, 150 MPa (b) microstructure showing the creep cavities near the fracture tip surface at high magnification (c) microstructure of the CGHAZ region (d) fracture surface of the creep specimen tested at 650 °C, 180 MPa (e) microstructure showing the creep cavities near the fracture tip surface at high magnification (f) microstructure of the ICHAZ region

8.5 Microhardness distribution after creep test

The Vickers micro-hardness distribution over the cross-section of the creep tested specimen (P92-ERNiCrMo-3-304L DWJ) is shown in Figure 8.46(a-e). The micro-hardness across the different zones of the fractured specimen was measured, i.e., fracture region, HAZ, BM, and weld fusion zone. Each location has a different micro-hardness value depending upon its microstructural characteristics. With the fracture tip surface set as a reference point ($X=0$), the hardness measurement was performed. The fracture location was observed in the P92 BM for all the test condition. The hardness was measured at the fracture region and found to be decreased with the lowest value (~ 173 HV) compared to the hardness of the as-received P92 steel (~ 235 HV). Thus, the hardness of the P92 steel was decreased after the creep test. The minimum hardness value was observed near the

fracture tip surface due to the softening mechanism occurring at the breaking point because of the extensive plastic deformation. The coarsening of the $M_{23}C_6$ precipitates is mainly responsible for the drop in the hardness near the fracture tip surface, as well as in the loss of the carbon and alloying elements, which are responsible for the precipitation strengthening of the P92 steel. The micro-hardness value is affected by the microstructural evolution during the creep exposure. The drop in the hardness value was due to the loss of the solid solution strengthening and precipitation strengthening due to the coarsening of the precipitates and formation of the Laves phases. Also, the hardness value significantly drops at the location of the large numbers of voids present across the fracture surface. As the distance from the fracture point increases, the hardness gradually increases. After welding, the maximum hardness was observed in the CGHAZ and FGHAZ regions. After the creep test, the hardness of the CGHAZ and FGHAZ regions was decreased due to the tempering effect. After welding, 304L BM has a hardness value of ~185HV. After the creep test, the hardness of the 304L BM barely changed as the obtained hardness value was similar to the value obtained after the welding.

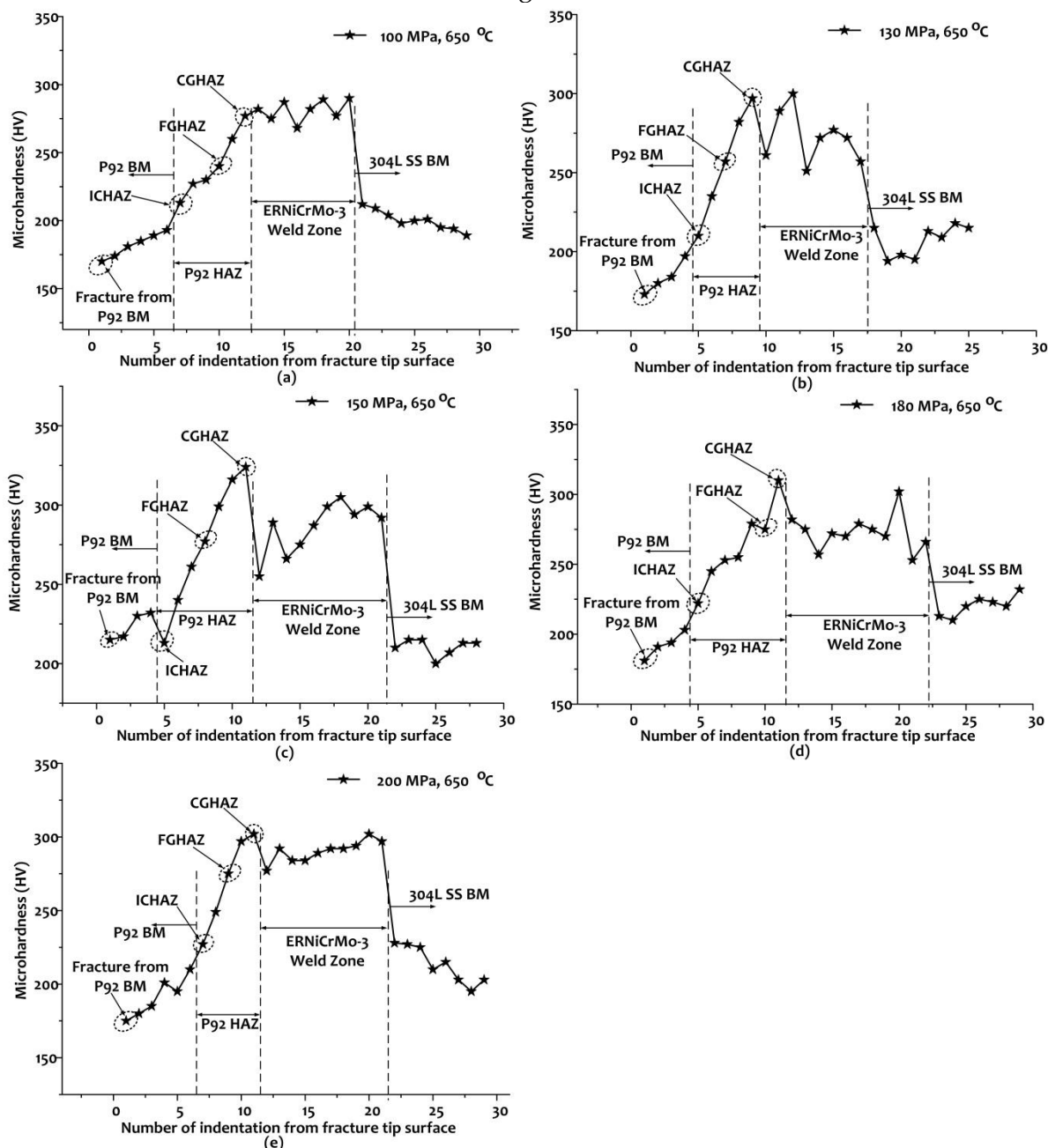


Figure 8.46: Vickers hardness profile across the creep tested specimen at 650 °C under different stresses: (a) 100 MPa (b) 130 MPa (c) 150 MPa (d) 180 MPa and (e) 200 MPa (P92-ERNiCrMo-3-304L DWJ)

The Vickers micro-hardness distribution over the cross-section of the fractured creep specimen (P92-ERNiFeCr-2-304L DWJ) is shown in Figure 8.47(a-e). The fracture location of all the creep test specimens was observed in the P92 BM based on the micro-hardness value at the fracture point. The hardness was measured at the fracture region and found to be decreased with a lowest value (~ 154 HV) compared to the hardness of the as-received P92 steel (~ 235 HV). The hardness drop was observed at the fracture point. The region with low hardness region exhibits a low creep service life (Sakthivel et al., 2019). As the distance from the fracture point increases, the hardness gradually increases. Due to the elevated temperature service at 650°C for a longer duration, the hardness of the CGHAZ and FGHAZ regions of P92 was decreased due to the tempering of the martensite. The ERNiFeCr-2 weld metal and 304L survived well during the creep service for all the test condition. The rise in the micro-hardness value of the ERNiFeCr-2 weld metal is due to the precipitation hardening effect after long-term service at elevated temperature. For 304L, no obvious change in the hardness value was observed. It showed good resistance to high-temperature creep.

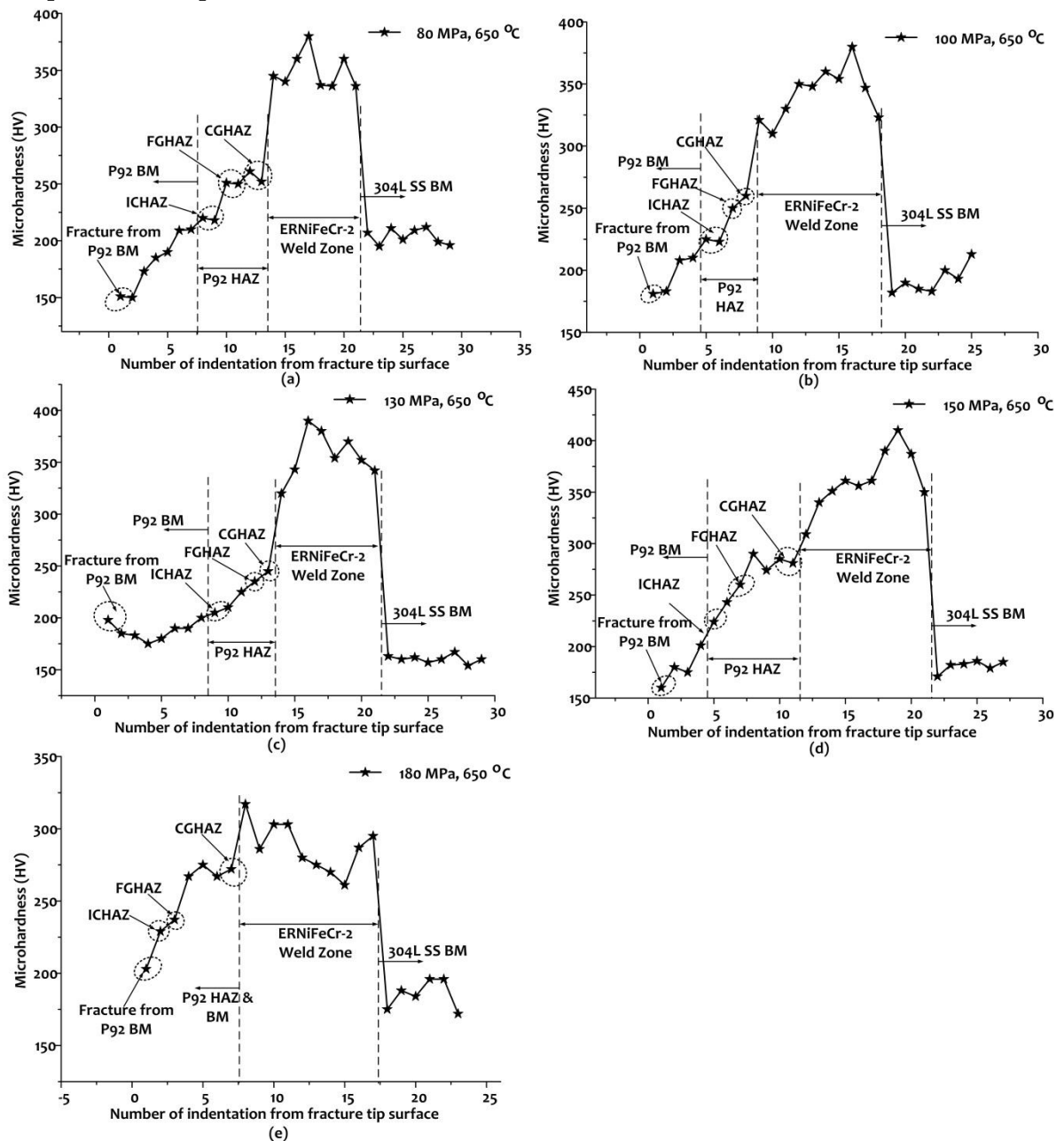


Figure 8.47: Vickers hardness profile across the creep tested specimen at 650°C under different stresses: (a) 80 MPa (b) 100 MPa (c) 130 MPa (d) 150 MPa and (e) 180 MPa (P92-ERNiFeCr-2-304L DWJ)

The Vickers micro-hardness distribution over the cross-section of the fractured creep specimen (P92-Thermanit MTS 616-304L DWJ) tested at (650 °C, 130 MPa), (650 °C, 150 MPa), (650 °C, 180 MPa), and (650 °C, 200 MPa), is shown in Figure 8.48(a-d). The micro-hardness inhomogeneity across the P92 steel can be noticed from the micro-hardness plot. The fracture location of all the creep test specimens was observed in the P92 BM based on the value of the micro-hardness at the fracture point and distance from the weld fusion line. The hardness dip was observed at the fracture point. The hardness at the fracture point was decreased, and it was lowest (as low as ~154 HV) compared to the hardness of the P92 BM (~235 HV). The decrease in the hardness value at the fracture location was due to the significant softening of the P92 steel. The decrease in the micro-hardness value towards the P92 steel side is mostly caused by the decline in the precipitation hardening mechanism and solid solution strengthening, as was covered in the previous section. At the location where a significant number of voids occurred close to the fracture surface, such a low micro-hardness magnitude was noticed. The hardness steadily rises with increasing distance from the fracture tip. Due to the elevated temperature service at 650 °C for a longer duration, the hardness of the P92 CGHAZ and P92 FGHAZ region was decreased because of the tempering of the martensite. The P92 filler metal weld fusion zone and 304L survived well during the creep service for all the test conditions. The hardness of the P92 filler weld fusion zone decreased during the creep test due to the tempering effect. The hardness barely changed toward the 304L side after the creep service. It showed good resistance to high-temperature creep.

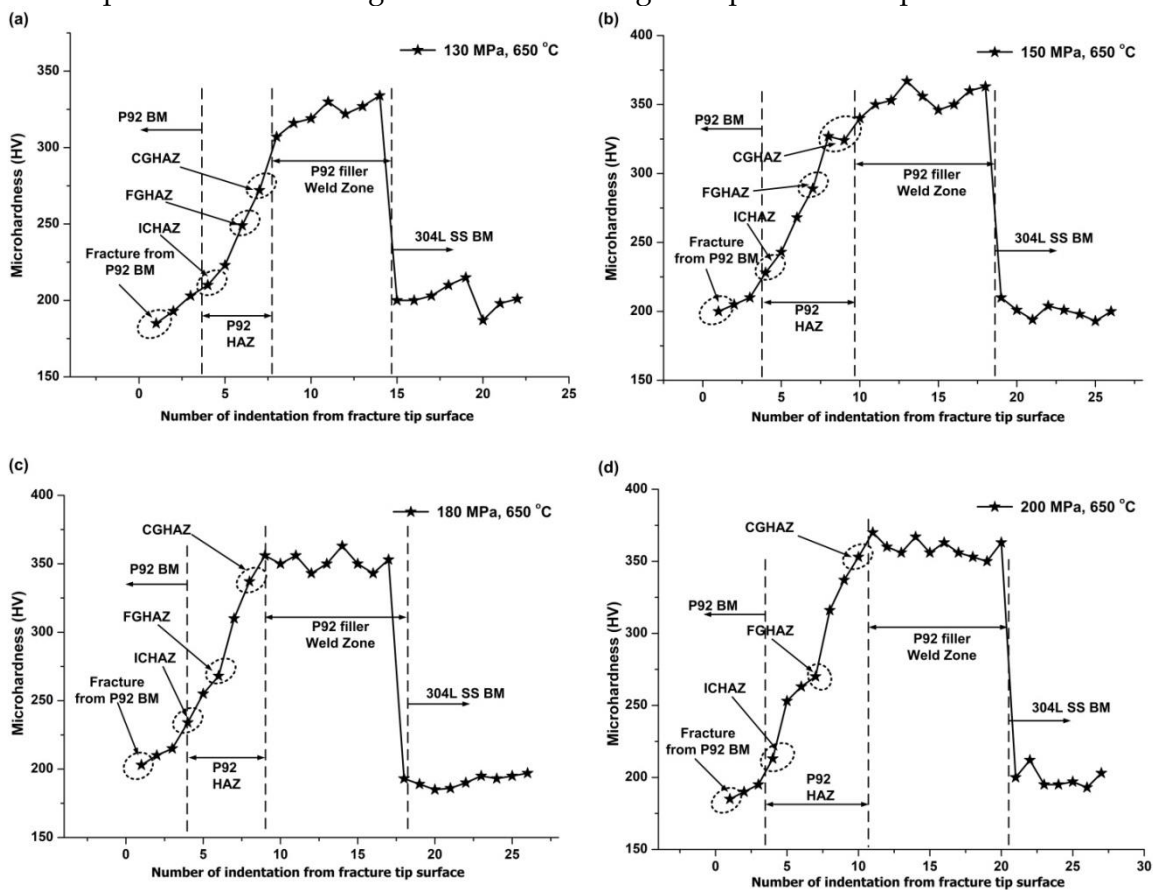


Figure 8.48: Vickers hardness profile across the creep tested specimen at 650 °C under different stresses: (a) 130 MPa (b) 150 MPa (c) 180 MPa and (d) 200 MPa (P92-Thermanit MTS 616-304L DWJ)

8.6 Fracture surface analysis of creep specimen

Figure 8.49(a-e) shows the SEM images of the fracture surface microstructure of the creep specimen (P92-ERNiFeCr-2-304L DWJ) tested at 650 °C under the various applied stresses. The fracture surface images taken at low magnification is shown in Figure 8.49(a-e). The low magnification (at 50X) image shows the overall appearance of the fracture surface and the same image was also used to measure the diameter of the fracture surface (final diameter/reduced diameter). The reduced diameter value was further used to measure the reduction in cross-sectional area. The fractograph obtained for all the creep-tested specimens with higher magnification (at 1500X and 2500X) is shown in subsequent images (figures adjacent to the low magnification image). The fracture surface was slant at low stress and converted into a flat at high temperature. The flat fracture surface indicates the brittle mode of the fracture. From the appearance of the edges of the fracture surface, shearing also occurred during the fracture. All the specimens ruptured from the P92 steel. The small size voids on the fractured surface were absent. Only secondary phase particles were observed for the specimen tested at 80 MPa and 100 MPa. The EDS result of the secondary phase particle was C: 12.40%, Cr: 8.54%, Fe: 76.51%, W: 2.40%, Nb: 0.07%, Mo: 0.09%. The particles were rich in Fe and Cr, indicating the carbide particles at the fracture surface. The very small micro-voids and cracks were observed in the center of the fracture surface. The small size micro-voids combine with each other and form a crack. As discussed in the microstructure section, the grain boundaries of the P92 steel weaken during the creep service due to the stress concentration and precipitates coarsening. Thus, failure takes place through the grain boundaries. Such type of failure is known as intergranular fracture (brittle fracture) (Sklenička et al., 2016). At high-temperature, grain boundaries are generally weak. Hence intergranular fracture mode takes place. However, grains are weak at low-temperature under high stress condition also, so the transgranular fracture mode takes place. Figure 8.49(a, b) shows the fractograph of the creep specimen at 650 °C under 80 MPa and 100 MPa. The small dimples and macro-voids were observed at the fracture surface. The large dimples were observed for the specimen tested at 650 °C within the stress range of 130-180 MPa, Figure 8.49(c-e). From the fractographs obtained, it can be concluded that the ductile mode of the fracture occurred for the specimen tested at 650 °C for the stress range of 130-180 MPa. With the rise of the applied stress, the size of the voids was increased. Thus, rupture mode changes from brittle at low stress to ductile at high stress.

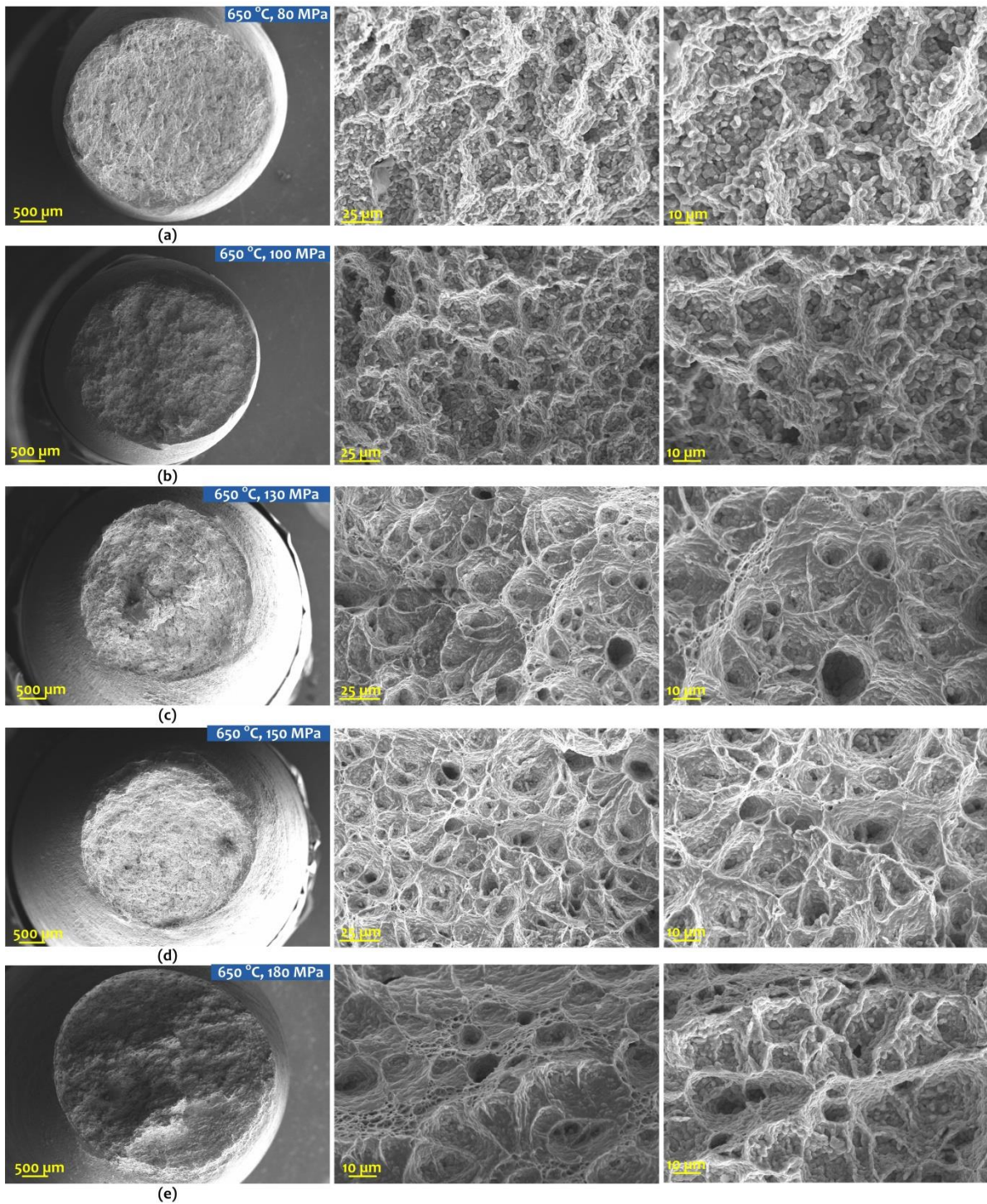


Figure 8.49: The fracture surface of the creep ruptured specimens tested at 650 °C under various stresses: (a) 80 MPa (b) 100 MPa (c) 130 MPa (d) 150 MPa and (e) 180 MPa (P92-ERNiFeCr-2-304L DWJ)

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