Research Article

Lakshmiprasad Maddi*, Atul Ramesh Ballal, Dilip Ramkrishna Peshwe, and M. D. Mathew

Influence of normalizing and tempering temperatures on the creep properties of P92 steel

<https://doi.org/10.1515/htmp-2020-0033>
received April 2, 2019; accepted July 27, 2019

Abstract: P92 steel is used as a piping material in ultra super critical power plants that can be operated at steam temperatures up to 650°C. The changes in the martensitic microstructure of P92 steel must be evaluated thoroughly before it is put into actual service. In this study, indigenously developed P92 steel was used. The steel was subjected to normalizing and tempering heat treatments in the range of 1,040–1,060°C and 740–780°C. The changes in the microstructure were evaluated and creep-rupture properties were studied at test temperatures of 600 and 650°C. Although normalizing temperatures influenced the microstructure and creep strength marginally, the change in tempering temperatures led to significant changes. The creep rupture strength at 600°C was influenced largely by the changes in the dislocation substructure, while the precipitation of Laves phases was a significant observation made for 650°C test temperature. Proposed mechanisms for the microstructural evolution and its consequences on the rupture life are discussed in this study.

Keywords: P92 steel, creep, tempered martensite, pipe diffusion, dislocation knitting

1 Introduction

With the recent emphasis on climate change and global warming, it is important to decrease fossil fuel emissions.

Fossil fuel power plants are the principal source for CO₂ emission. Hence, the development of high-temperature materials for pressure vessels, boiler tubes, and pipes is the need of the hour [1,2]. In this context, it is necessary to develop newer grades of steel with creep resistance at temperatures of 600°C and higher [3–5]. P92 steel is one such material that has been developed for the usage in power plant piping due to its high thermal conductivity and low coefficient of thermal expansion [6]. A thorough understanding of the microstructural stability and long-term creep behavior of P92 steel is necessary before it is put into service. The initial microstructure of P92 steel is randomly oriented lath martensitic microstructure. The creep strength of the steel is attributed to the combined effects of solid solution hardening (due to Cr, W, V, Nb) and precipitation hardening (precipitates of M₂₃C₆ and MX [where M = Cr, W, V, Nb and X = C, N] at the lath boundaries) [6,7]. Fine lath morphology with high dislocation density and fine precipitates make the dislocation motion extremely difficult, resulting in high strength at room temperature [8–13]. However, the lath martensite structure is unstable and changes at high temperature during creep. With reference to the initial microstructure, a reduction in the dislocation density, an increase in the precipitate size, and an increase in the lath width were studied by various researchers [14–16]. Some studies also indicated the precipitation of the Laves phase (Fe₂Mo or Fe₂W) for high Cr steels (>10% Cr) during heat treatment [5,17]. The precipitation of Laves phases brings an interesting debate between solid solution strengthening and precipitation hardening. However, precipitation strengthening due to Laves phases was reported to be dominant [18–20].

In this study, indigenously developed P92 steel was used and an attempt was made to study the effect of heat treatment parameters on creep properties at 600 and 650°C. While the standard heat treatment for P92 steel is normalizing and tempering, the variation in the heat treatment parameters leads to a significant change in the mechanical properties [14–16]. Normalizing involves the

* **Corresponding author: Lakshmiprasad Maddi**, Department of Basic Sciences and Humanities, GMRIIT, Rajam 532127, India, e-mail: lakshmiprasad.m@gmrit.edu.in, tel: +91-9133095915
Atul Ramesh Ballal, Dilip Ramkrishna Peshwe: Department of Metallurgical & Materials Engineering, VNIT, Nagpur 440010, India
M. D. Mathew: Dean, Post Graduate and Research Studies, Saint Gits College of Engineering, Kottayam, Kerala 686532, India

dissolution of precipitates (carbides and nitrides) that help in homogenizing the microstructure, while simultaneously leading to an increase in the grain size. Hence, excess normalizing temperatures should be avoided. According to Goldshtein and Popov [21,22], V and Nb carbides dissolve readily in steel at temperatures of 1,000–1,050°C. Ennis et al. [11] showed that higher normalizing temperatures led to an increase in grain size and reported optimum normalizing temperature as 1,070°C. Normalizing of P92 steel results in martensitic microstructure, which is brittle and therefore, not suitable for the said application. Generally, normalized P92 steel is given tempering treatment, a process that leads to a reduction in dislocation density and an increase in toughness. However, excessive tempering needs to be avoided as it leads to an undesirable reduction in the tensile strength [23]. Accordingly, in this study, the maximum tempering temperature was fixed at 780°C, since tempering at 835°C resulted [11] in creep rupture times of two orders of magnitude less when compared with the creep rupture life at lower tempering temperatures. Although studies were reported on microstructural evolution and the mechanical properties of 9-Cr steel, a comprehensive picture on the structure – property correlation over a range of temperatures is missing. Therefore, in this study, P92 steel was studied for creep behavior under different normalizing temperatures in the range of 1,040–1,060°C and different tempering temperatures in the range of 740–780°C.

2 Experiments

The indigenously produced steel was procured in the form of rolled plates from Mishra Dhatu Nigam (MIDHANI), Hyderabad, India. The chemical composition of the material is given in Table 1.

The initial dimensions of the plate were $505 \times 123 \times 13 \text{ mm}^3$. The plate was cut into bars of $123 \times 10 \times 13 \text{ mm}^3$

along the rolling direction, using a wire electric discharge machine (WEDM-EZEEDCUT). A brass wire of diameter 0.6 mm was used for cutting. A tray was designed using heat-resistant steel 304 Stainless Steel rods on which a batch of bars was loaded simultaneously for the heat treatment. The heat treatment was performed in resistance furnaces (THERELEK). For normalizing treatment, a furnace with SiC heating elements with a maximum operating temperature of 1,400°C was used. For tempering treatment, where the maximum temperature was 780°C, muffle furnace with Nichrome (80% Ni + 20% Cr) heating coils was used. Bars were allowed to cool in steady air once they were removed from the furnace after the heat treatment. Heat-treated bars were turned on the lathe machine into test samples for creep. The gauge portion of the sample was further subjected to grinding by an abrasive wheel to obtain close dimensional tolerances. The accuracy of the gauge diameter of the specimen was maintained throughout the gauge length as $\pm 0.01 \text{ mm}$ (ASTM Standard E6). Care was taken that no tool marks were present on the finished sample surface, which otherwise act as stress concentration sites promoting rupture. Uniaxial creep testing machines (Star Testing Systems) were used to conduct stress rupture tests as per ASTM standard E139. The entire process of experimentation is shown in Figure 1.

Scanning electron microscopy (SEM) was used to investigate the effect of heat treatment and creep exposure on the microstructure of P92 steel. The samples were ground on a belt-grinder following polishing with 800, 1,500, and 2,000 grits of paper. Finally, they were cloth polished using alumina particles of 500 μm size to obtain mirror finish. Villela's reagent (1 g picric acid + 5 mL HCl + 100 mL ethanol) was used to etch the samples. Metallographic studies were carried using SEM (JEOL 6380 A) with tungsten filament as a source of electrons for imaging. Secondary electron images were taken at 15 kV accelerating voltage for magnifications in the range of 20 $\times$ to 5,000 $\times$ to study the topography.

Table 1: Chemical composition of P92 steel

Element	Cr	C	Mo	W	V	Nb	Si	Mn
Measured	9.38	0.126	0.499	1.91	0.201	0.067	0.257	0.379
ASTM A335	8.5–9.5	0.07–0.13	0.3–0.6	1.5–2.0	0.15–0.25	0.04–0.09	0.5 (max.)	0.3–0.6
Element	Ni	Al	P	S	N	B		
Measured	0.086	0.009	0.0087	0.0054	0.0521	0.0015		
ASTM A335	0.4 (max.)	0.02 (max.)	0.02 (max.)	0.01 (max.)	0.03–0.07	0.001–0.006

Figure 1: Flowchart of experimentation.

In addition, due to the characteristic depth of focus associated with SEM, fractography of the ruptured samples was also studied. While information regarding precipitate growth distribution can be obtained from the micrograph, fractographic studies were undertaken to correlate the associated microstructural changes with fractographic features. In general, laths and precipitate sizes are of the dimensions of few nanometers for the measurement of which transmission electron microscopy (TEM) was used. TEM (Philips CM-12) was used at 200 kV accelerating voltage for bright-field imaging. Thin slices of 0.5 mm thickness were cut longitudinally and mechanically thinned to 70 μm followed by twin jet electropolishing. Electrolyte comprising 10% perchloric acid and 90% methanol was used at -50°C .

3 Results and discussion

3.1 Dislocation substructure evolution

Reduction in dislocation density, increase in lath width, and change in orientation of martensitic laths manifest in dislocation substructure evolution. Under the action of creep stress, the dislocations at the grain boundary can be knit-in (a combination of opposite sign dislocations) resulting in a stress-free region/new boundary [24–28]. According to Blum

et al. [29] under the action of creep stress, dislocations in the same grain can interact in two ways: (a) Burger's vectors b_1 and b_2 subtract and b_3 vanishes (annihilation of dislocations) or (b) Burger's vectors b_1 and b_2 add to become b_3 (formation of new dislocation). It is to be noted here that lath boundaries are considered as an array of edge/screw dislocations. As a result, annihilation between mobile dislocations and boundary dislocations occurs in the gauge portion of the sample, resulting in a change in misorientation among the boundaries. This process is termed as stress-induced recovery (Figure 2).

Figure 2: Schematic diagram showing knit-in of mobile dislocations with the boundary dislocations (dotted lines indicate dislocation path [assumed] under creep stress).

3.2 Precipitate growth

Under the action of creep stress (during primary creep), the dislocations interact with carbides at boundaries. There are many “features” in the material (microstructure) that provide easier paths for diffusion, viz., grain boundaries and dislocation cores. These short circuit paths for diffusion are preferred over conventional volume or surface diffusion due to the low activation energy. The order of activation energies is $Q_{\text{surface}} < Q_{\text{grain boundary}} < Q_{\text{pipe}}$, which also means that the order of diffusion coefficients is $D_{\text{surface}} > D_{\text{grain boundary}} > D_{\text{pipe}}$. Pipe diffusion [30–32] involves the diffusion of atoms through the dislocation cores. The higher the number of dislocations, the more dominant the effect of pipe diffusion. Also, pipe diffusion is dominant at low temperatures and low stresses ($\sigma/G < 10^{-4}$). But, in this study, dislocation creep ($10^{-4} < \sigma/G < 10^{-2}$) is the predominant mechanism. According to Kaur et al. [33], grain boundary diffusion resulted in significant coarsening of precipitates. Aghajani et al. [17] also reported that precipitates on the boundary coarsened at a faster rate when compared with precipitates away from the boundary. Thus, the coarsening of precipitates in the

Figure 3: Pipe diffusion leading to precipitate coarsening (assuming the spherical shape of precipitates).

present study can be explained based on the assumption [33] that grain boundaries are considered to be isotropic, high-diffusivity slabs (of thickness δ) surrounded by dislocation pipes in all possible directions. Also, small-angled grain boundaries may be considered to be analogous to an array of parallel dislocation pipes [30], where diffusion parallel to the pipes is maximum and perpendicular to the pipes is minimum. A schematic of the process of precipitate coarsening is shown in Figure 3.

Earlier work of the authors confirmed the coarsening of existing $M_{23}C_6$ carbides and the growth of Laves phases adjacent to $M_{23}C_6$ carbides [34]. Figure 4 represents a schematic of the microstructure of P92 steel after creep. The analysis of growth kinetics [35] revealed that the Laves phase precipitates reach equilibrium volume fraction only after 13,000 h of creep, while $M_{23}C_6$ carbides reached equilibrium already during tempering. Although the diffusivity of atoms is faster along the grain boundaries when compared with the bulk, the shape of the Laves phases did not indicate such phenomena (which would otherwise result in lens-like morphologies along grain boundaries).

3.3 Stress rupture studies at 600°C

As mentioned earlier, normalizing results in uniform distribution of carbides throughout the material. However, reducing the pinning effect of carbides increases the grain size. Hence, higher normalizing temperatures need not necessarily be beneficial toward the mechanical strength of the material. Tempering results in the ejection of

excess dissolved C, thereby again forming carbide precipitates. In short, higher normalizing temperatures combined with low tempering temperatures exhibit high tensile strength and are assumed to show high rupture strength. Higher tempering temperatures lead to a reduction in tensile strength and are predicted to show less rupture strength.

3.3.1 Effect of normalizing temperature

Figure 5 shows the rupture data for various normalizing temperatures. With the decrease in stress level, longer rupture times were observed for all normalizing temperatures. For constant tempering temperature of 740°C (Figure 5(a)), the normalizing temperature of 1,060°C exhibited the highest rupture time over the complete stress range. The slope of the rupture curve of 1,060°C normalizing temperature was marginally steep when compared with 1,040°C normalizing temperature. The slope of the stress rupture curve is considered as a measure of the rate of microstructure evolution. In general, high normalizing temperatures are associated with high dislocation densities. Higher normalizing temperatures also help in retaining greater amounts of carbon in the matrix. The dissolved carbon ejects out in the form of carbides during tempering as well as creep deformation. The high percentage of carbon in the initial microstructure implies a higher driving force for the ejection of carbon during creep. This is validated by the slope of the rupture curves, which is observed to be steeper for higher normalizing temperatures.

Slopes of both normalizing temperatures were the same in the case of 760°C tempering temperature

Figure 4: Schematic showing the growth of the Laves phases around $M_{23}C_6$ precipitates.

(Figure 5(b)), indicating that the rate of microstructural evolution was the same. However, less rupture time was noted in the case of normalizing at $1,060^{\circ}\text{C}$ when compared with $1,040^{\circ}\text{C}$. This may be attributed to the fact that higher grain size along with increased precipitate size was observed in the room temperature microstructure of $1,060^{\circ}\text{C}/760^{\circ}\text{C}$ [23].

For constant tempering temperature of 780°C (Figure 5(c)), the steep slope in the rupture curve of samples normalized at $1,040^{\circ}\text{C}$ was observed when compared with the higher normalizing temperature of $1,060^{\circ}\text{C}$. At high stresses, the normalizing temperature of $1,040^{\circ}\text{C}$ resulted in higher rupture times. At low stresses, $1,060^{\circ}\text{C}$ normalizing temperature yielded higher rupture times. This was due to the dominant effect of precipitate hardening over solid solution strengthening. Higher area fraction of precipitates at lower normalizing temperatures [23] helps in pinning the boundaries, thereby resisting creep deformation at higher stresses. In the case of long-term tests, the same existing $M_{23}C_6$ carbides coarsen, leading to reduced pinning, thereby decreasing in rupture times. On the contrary, a higher percentage of dissolved carbon at higher normalizing temperatures provides precipitation hardening by the ejection of carbides during long-term exposure.

Hence, higher long-term rupture times were observed for the normalizing temperature of $1,060^{\circ}\text{C}$. Ennis *et al.* [11] carried out similar investigation over the temperatures 970 – $1,145^{\circ}\text{C}$ and observed fine precipitation of $M_{23}C_6$

at higher normalizing temperatures, which was responsible for better creep properties.

3.3.2 Effect of tempering temperature

Figure 6 shows the stress rupture data for various tempering temperatures. At a constant normalizing temperature of $1,040^{\circ}\text{C}$ (Figure 6(a)), tempering at 740°C resulted in the highest rupture time among the three tempering temperatures. This may be due to the high initial dislocation density, fine lath width, and solid solution strengthening associated with lower tempering temperatures.

The accelerated microstructural evolution is reflected by the steep slope of the rupture curve for lower tempering temperatures. Conversely, changes in the microstructures for high tempering temperatures (780°C) were less when compared with the initial microstructure. This can be seen from the shallow slope associated with the rupture curve of 780°C tempering temperature. Less dislocation density and coarse carbides may be responsible for inferior creep life at higher tempering temperature. The steep slope of the rupture curve of 740°C tempering temperature also hints at the possibility that tempering at 760°C could result in better rupture times at low stresses. The rupture data of $1,040^{\circ}\text{C}$ and $1,060^{\circ}\text{C}$ normalizing temperatures (Figure 6(b)) displayed a similar trend.

Figure 5: Summary of the stress rupture data revealing the effect of normalizing temperature for creep tests at 600°C, for tempering temperatures of (a) 740°C (b) 760°C (c) 780°C.

Figure 6: Summary of the stress rupture data revealing the effect of tempering temperature for creep tests at 600°C, for normalizing temperatures of (a) 1040°C (b) 1060°C.

Figure 7: Reduction in dislocation density and increase in lath width observed for 1060°C/740°C (a and c) after heat treatment (b and d) after creep rupture for 250 MPa/2,509 h (markings indicate lath width, enclosed area indicates dislocation density [qualitative]).

Figure 7 reveals dislocation substructure evolution and change in lath width with creep exposure time. No Laves phase precipitation was observed for the entire rupture tests at 600°C.

3.4 Stress rupture studies at 650°C

3.4.1 Effect of normalizing temperature

Higher normalizing temperatures and lower tempering temperatures result in high dislocation densities. Hence, high dislocation density is predicted for the combination of 1,060°C normalizing and 740°C tempering. Comparatively lower dislocation density might be responsible for the shallow slope for a normalizing temperature of 1,040°C. For the same reason, the rupture time of 1,060°C was observed to be high (Figure 8(a and c)). For a tempering temperature of 780°C, the shallow slope of 1,040°C normalizing (in Figure 8(c)) may result in the superior

rupture time for 1,040°C at low stresses. This may be attributed to the possible precipitation of Laves phases [34] at 1,040°C normalizing temperature (in otherwise weak microstructure – coarse laths, less dislocation density, coarse precipitates).

In the case of 760°C tempering temperature (Figure 8(b)), the steep slope was observed for 1,040°C while 1,060°C showed a relatively shallow slope. A cross-over was observed in the rupture curves and normalizing at 1,060°C resulted in higher rupture times at low stresses.

3.4.2 Effect of tempering temperature

Figure 9 represents the rupture curves for all three tempering temperatures for the constant normalizing temperatures.

For a normalizing temperature of 1,040°C (Figure 9(a)), tempering at 740°C resulted in the highest rupture time in the high-stress region, whereas a tempering temperature of 780°C may exhibit better rupture time at

Figure 8: Summary of the stress rupture data revealing the effect of normalizing temperature for creep tests at 650°C, for tempering temperatures of (a) 740°C (b) 760°C (c) 780°C.

Figure 9: Summary of the stress rupture data revealing the effect of tempering temperature for creep tests at 650°C, for normalizing temperatures of (a) 1040°C (b) 1060°C.

Figure 10: (a–d) Proposed mechanism for precipitate coarsening and fracture. SEM fractographs obtained for ruptured specimens of 1060°C/740°C at (e) 325 MPa/199 h and (f) 235 MPa/4,321 h.

lower stresses. Identical slopes of rupture curves of 740 and 760°C tempering temperatures were observed when compared with the shallow slope of 780°C. The shallow slope of the rupture curve of 780°C tempering temperature was also responsible for the crossover rupture curves of 740 and 760°C tempering temperatures. Since the microstructure after tempering at 780°C comprises less density of mobile dislocations and coarse laths, the dislocation substructure evolution during tempering was less. In this case, the precipitation of Laves phases was found to be the responsible factor for the increase in rupture life [34].

In the case of 1,060°C normalizing temperature (Figure 9(b)), the slopes of stress rupture curves for 760 and 780°C tempering temperatures were found to be

identical. This may be due to the higher initial dislocation density, which might have resulted in the rapid evolution of microstructure of 1,060°C/740°C when compared with tempering temperatures of 760 and 780°C.

3.5 Fractography

Fractographs of the long-term (~5,000 h) and short-term ruptures (<100 h) can also be correlated with the above phenomena. For short-term ruptures at high stresses, fine precipitates were observed, resulting in small cavities [11]. High stresses were responsible for the

formation of dimples between the cavities where the material is separated (fractured). In the case of long-term ruptures at low stresses, precipitate coarsening was observed [36]. Ostwald ripening is the established mechanism for precipitate coarsening, meaning one precipitate grows at the expense of its surrounding finer precipitates [9,17]. Thus, huge cavities are observed in the fractograph, while the separating material between the cavities (precipitates) gets transformed from fine dimples to ligaments. Figure 10 represents the proposed mechanism of fracture [37,38]. While the short-term rupture is preceded by stress-dominant necking, long-term rupture is due to metallurgical necking.

4 Conclusions

The effect of normalizing and tempering heat treatments on the microstructure and creep properties of P92 steel was studied.

From creep tests at 600°C, it was observed that the changes in microstructure during creep are more extensive for low tempering temperatures. This gave rise to steep slopes of stress rupture curves. It is suggested that a tempering temperature of 760°C may be preferred for the long-term usage of P92 steel for components that operate at 600°C, delaying microstructural instability.

The creep tests at 650°C showed precipitation of Laves phases at higher tempering temperatures, which contributed to precipitation hardening and increased the rupture strength. From this study, a tempering temperature of 780°C is recommended for the usage of P92 steel at 650°C for longer service times.

Acknowledgments: The authors sincerely thank the Board of Research in Nuclear Sciences (BRNS) (Sanction No. 2010/36/66-BRNS) for providing financial assistance in carrying out this work.

References

- [1] Klueh, R. L., and D. R. Harries. *High-chromium ferritic and martensitic steels for nuclear applications*, ASTM International, Pennsylvania, United States, 2001, Chapter 2 and 3, pp. 5–38.
- [2] Kaybyshev, R. O., V. N. Skorobogatykh, and I. A. Shchenkova. New martensitic steels for fossil power plant: creep resistance. *Physics of Metals and Metallography*, Vol. 109, No. 2, 2010, pp. 186–200.
- [3] Bhadeshia, H. Advances in physical metallurgy and processing of steels. Design of ferritic creep-resistant steels. *ISIJ International*, Vol. 41, No. 6, 2001, pp. 626–640.
- [4] Abe, F., M. Taneike, and K. Sawada. Alloy design of creep resistant 9Cr steel using a dispersion of nano-sized carbonitrides. *International Journal of Pressure Vessels and Piping*, Vol. 84, No. 1–2, 2007, pp. 3–12.
- [5] Rojas, D., J. Garcia, O. Prat, G. Sauthoff, and A. R. Kaysser-Pyzalla. 9%Cr heat resistant steels: alloy design, microstructure evolution and creep response at 650°C. *Materials Science and Engineering A*, Vol. 528, No. 15, 2011, pp. 5164–5176.
- [6] Klueh, R. L., and A. T. Nelson. Ferritic/martensitic steels for next-generation reactors. *Journal of Nuclear Materials*, Vol. 371, No. 1–3, 2007, pp. 37–52.
- [7] Yin, F., L. Tian, B. Xue, X. Jiang, and L. Zhou. Effect of carbon content on microstructure and mechanical properties of 9 to 12 pct Cr ferritic/martensitic heat-resistant steels. *Metallurgical and Materials Transactions. A, Physical Metallurgy and Materials Science*, Vol. 43A, No. 7, 2012, pp. 2203–2209.
- [8] Fedorova, I., A. Kipelova, A. Belyakov, and R. Kaibyshev. Microstructure evolution in an advanced 9 pct Cr martensitic steel during creep at 923 K (650 °C). *Metallurgical and Materials Transactions. A, Physical Metallurgy and Materials Science*, Vol. 44A, No. S1, 2013, pp. 128–135.
- [9] Hald, J., and L. Korcakova. Precipitate stability in creep resistant ferritic steels-experimental investigations and modelling. *ISIJ International*, Vol. 43, No. 3, 2003, pp. 420–427.
- [10] Maruyama, K., K. Sawada, and J. Koike. Advances in physical metallurgy and processing of steels. Strengthening mechanisms of creep resistant tempered martensitic steel. *ISIJ International*, Vol. 41, No. 6, 2001, pp. 641–653.
- [11] Ennis, P. J., and A. Czyrska-Filemonowicz. Recent advances in creep-resistant steels for power plant applications. *Sadhana*, Vol. 28, No. 3–4, 2003, pp. 709–730.
- [12] Kimura, K., K. Sawada, H. Kushima, and K. Kubo. Effect of stress on the creep deformation of ASME Grade P92/T92 steels. *International Journal of Materials Research*, Vol. 99, No. 4, 2008, pp. 395–401.
- [13] Vaillant, J. C., B. Vandenberghe, B. Hahn, H. Heuser, and C. Jochum. T/P23, 24, 911 and 92: new grades for advanced coal-fired power plants – properties and experience. *International Journal of Pressure Vessels and Piping*, Vol. 85, No. 1–2, 2008, pp. 38–46.
- [14] Pandey, C., A. Giri, and M. M. Mahapatra. Effect of normalizing temperature on microstructural stability and mechanical properties of creep strength enhanced ferritic P91 steel. *Materials Science and Engineering A*, Vol. 657, 2016, pp. 173–184.
- [15] Saini, N., C. Pandey, and M. M. Mahapatra. Characterization and evaluation of mechanical properties of CSEF P92 steel for varying normalizing temperature. *Materials Science and Engineering A*, Vol. 688, 2017, pp. 250–261.
- [16] Saini, N., C. Pandey, and M. M. Mahapatra. Effect of normalizing temperature on fracture characteristic of tensile and impact tested creep strength-enhanced ferritic P92 steel. *Journal of Materials Engineering and Performance*, Vol. 26, No. 11, 2017, pp. 5414–5424.

- [17] Aghajani, A., C. Somsen, and G. Eggeler. On the effect of long-term creep on the microstructure of a 12% chromium tempered martensite ferritic steel. *Acta Materialia*, Vol. 57, No. 17, 2009, pp. 5093–5106.
- [18] Abe, F. Precipitate design for creep strengthening of 9% Cr tempered martensitic steel for ultra-supercritical power plants. *Science and Technology of Advanced Materials*, Vol. 9, No. 1, Mar. 13, 2008, p. 013002.
- [19] Shibuya, M., Y. Toda, K. Sawada, H. Kushima, and K. Kimura. Effect of precipitation behavior on creep strength of 15%Cr ferritic steels at high temperature between 923 and 1023 K. *Materials Science and Engineering A*, Vol. 592, No. 1, 2014, pp. 1–5.
- [20] Zhu, J. H., C. T. Liu, L. M. Pike, and P. K. Liaw. A thermodynamic interpretation of the size-ratio limits for laves phase formation. *Metallurgical and Materials Transactions. A, Physical Metallurgy and Materials Science*, Vol. 30, No. 5, 1999, pp. 1449–1452.
- [21] Goldshtein, M. I., and V. V. Popov. Thermodynamic calculations of the solubility of carbonitrides in the austenite of constructional steels. *Metal Science and Heat Treatment*, Vol. 31, No. 11, 1989, pp. 840–848.
- [22] Popov, V. V., and M. I. Goldshtein. Dissolution of carbides and nitrides during austenitizing of steel. *Metal Science and Heat Treatment*, Vol. 33, No. 7, 1991, pp. 480–483.
- [23] Barbadikar, D. R., G. S. Deshmukh, L. Maddi, K. Laha, P. Parameswaran, A. R. Ballal et al. Effect of normalizing and tempering temperatures on microstructure and mechanical properties of P92 steel. *International Journal of Pressure Vessels and Piping*, Vol. 132–133, 2015, pp. 97–105.
- [24] Maddi, L., D. Barbadikar, M. Sahare, A. R. Ballal, D. R. Peshwe, R. K. Paretkar, et al. Microstructure evolution during short term creep of 9Cr–0.5 Mo–1.8 W steel. *Transactions of the Indian Institute of Metals*, Vol. 68, 2015, pp. 259–266.
- [25] Yan, W., W. Wang, Y. Y. Shan, and K. Yang. Microstructural stability of 9–12%Cr ferrite/martensite heat-resistant steels. *Frontiers of Materials Science*, Vol. 7, No. 1, 2013, pp. 1–27.
- [26] Dudko, V., A. Belyakov, D. Molodov, and R. Kaibyshev. Microstructure evolution and pinning of boundaries by precipitates in a 9 pct Cr heat resistant steel during creep. *Metallurgical and Materials Transactions. A, Physical Metallurgy and Materials Science*, Vol. 44A, No. S1, 2013, pp. 162–172.
- [27] Armaki, H. G., R. P. Chen, K. Maruyama, M. Yoshizawa, and M. Igarashi. Static recovery of tempered lath martensite microstructures during long-term aging in 9–12% Cr heat resistant steels. *Materials Letters*, Vol. 63, 2009, pp. 2423–2425.
- [28] Li, S., Z. Eliniyaz, X. Dong, Y. Shen, L. Zhang, and A. Shan. Effect of stress on microstructural evolution and mechanical properties of 12Cr3W3Co steel during aging and short-term creep. *Materials Science and Engineering A*, Vol. 580, 2013, pp. 51–57.
- [29] Blum, W., and P. Eisenlohr. Dislocation mechanics of Creep. *Materials Science and Engineering*, Vol. A510–511, 2009, pp. 7–13.
- [30] Hoffman, R., and D. Turnbull. The effect of relative crystal and boundary orientations on grain boundary diffusion rates. *Acta Metallurgica et Materialia*, Vol. 2, 1954, pp. 419–426.
- [31] Love, G. R. Dislocation pipe diffusion. *Acta Metallurgica et Materialia*, Vol. 12, No. 6, 1964, pp. 731–737.
- [32] Eggeler, G. The effect of long-term creep on particle coarsening in tempered martensite ferritic steels. *Acta Metallurgica et Materialia*, Vol. 37, No. 12, 1989, pp. 3225–3234.
- [33] Kaur, I., Y. Mishin, and W. Gust. *Fundamentals of grain and interphase boundary diffusion*, 3rd edn, John Wiley & Sons Ltd, New York, United States, 1995, Chapter 3, pp. 217–234.
- [34] Maddi, L., G. S. Deshmukh, A. R. Ballal, D. R. Peshwe, R. K. Paretkar, K. Laha, et al. Effect of Laves phase on the creep rupture properties of P92 steel. *Materials Science and Engineering A*, Vol. 668, 2016, pp. 215–223.
- [35] Prat, O., J. Garcia, D. Rojas, G. Sauthoff, and G. Inden. The role of Laves phase on microstructure evolution and creep strength of novel 9%Cr heat resistant steels. *Intermetallics*, Vol. 32, 2013, pp. 362–372.
- [36] Maddi, L., A. R. Ballal, D. R. Peshwe, R. K. Paretkar, K. Laha, and M. D. Mathew. Effect of tempering temperature on the stress rupture properties of Grade 92 steel. *Materials Science and Engineering A*, Vol. 639, 2015, pp. 431–438.
- [37] Das, A., N. Roy, and A. K. Ray. Stress induced creep cavity. *Materials Science and Engineering A*, Vol. 598, 2014, pp. 28–33.
- [38] Lee, J. S., H. G. Armaki, K. Maruyama, T. Muraki, and H. Asahi. Causes of breakdown of creep strength in 9Cr–1.8W–0.5Mo–VNb steel. *Materials Science and Engineering A*, Vol. 428, No. 1–2, 2006, pp. 270–275.