Original Article

Thermal-mechanical fatigue behaviour and life prediction of P92 steel, including average temperature and dwell effects

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\textbf{A B S T R A C T}

This study was devoted to an investigation of the effects of the average temperature and dwell on the thermal-mechanical fatigue (TMF) behaviour of P92 steel at elevated temperatures in the range of 350–650 °C. The results revealed that increased average temperature decreased the in-phase TMF (IPTMF), out-of-phase TMF (OPTMF), and isothermal fatigue (IF) life values, and its effect was more significant in the range of 400-450 °C. The effect of a short dwell time on the IPTMF life was dependent on the form of dwell, because tensile dwell decreased the IPTMF life and symmetric dwell increased the IPTMF life. Both creep stress relaxation (CSR) and dynamic strain ageing (DSA) were detected under IPTMF cycling. The application of dwell enhanced the CSR phenomenon. In contrast, the DSA effect was restrained, which was beneficial for fatigue resistance. The appearance of crack tip blunting and crack branching because of symmetric dwell indicated the retardation of crack propagation under the combined effects of enhanced CSR and the disappearance of the DSA phenomenon. In all cases, the transformation of lath structures into substructures was the dominant deformation mechanism. In addition, the equiaxial subgrain grew with increases in the strain amplitude and dwell time. Finally, the application of the current life models to P92 steel fatigue life prediction under both IF and TMF cycling were evaluated. Because of the insufficiency of the current models, a modified Coffin–Manson model is proposed by incorporating the mean stress and temperature into power law forms.

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1. Introduction

Because of its excellent tensile and creep strengths, high thermal conductivity, and low thermal expansion coefficient, P92 steel, which is a type of 9–12%Cr steel alloy, has been widely used in the boilers and piping systems of power plants. Under long-term service loading, both mechanical strain and additional thermal strain are caused by the frequent start-up and shut-down of the components, which is called thermal-mechanical fatigue (TMF). In recent decades, many studies have focused on the elevated-temperature low-cycle fatigue behaviour of 9–12%Cr steels [1–8]. The relation between the

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accumulated inelastic strain and the temperature, as well as strain amplitude-dependent cyclic softening behaviour, has been determined [1–3]. The cyclic softening of 9–12%Cr steels has been attributed to the recovery of the low-angle boundary dislocation structure, decrease in the global dislocation density, and transformation of the lath microstructure into a subgrain microstructure with a lower energy configuration and grain rotation, as well as carbide coarsening [4–8]. Fournier et al. [9,10] analysed the influence of the holding times on the cyclic plastic behaviours of martensitic steels using an enhanced stress partitioning method. The much faster deterioration of the mechanical properties under creep-fatigue interactions in 9–12%Cr steel was observed, which was a result of the microstructural coarsening and decrease in the dislocation density [11,12]. Wang et al. [13–15] conducted a comprehensive investigation of the low cycle fatigue behaviour of P92, including the compressive hold damage mechanism, life prediction model, and influence of prior low-cycle fatigue on the microstructure evolution. In addition, they established a flexible thermo-metalo-mechanical model to simulate the multi-pass circumferential welding of P92 steel pipes [16,17].

Compared with the isothermal fatigue (IF) behaviour of 9–12%Cr steels at elevated temperatures, less attention has been given to the TMF behaviour because the testing process is time-consuming and complex. Generally, the fatigue life and cyclic damage evolution under TMF cycling can be evaluated using an IF test at the peak temperature. However, the derived fatigue life may be non-conservative because of the activation and interaction of more damage processes because of the cyclic temperature field, which is normally absent in an IF test. Nagesha et al. [18–20] evaluated the TMF behaviours of different types of steels considering the impacts of the phase angle and temperature range. For different materials, the fatigue life of the out-of-phase thermal-mechanical fatigue (OPTMF) in air was always lower than that of the in-phase thermal-mechanical fatigue (IPTMF) as a result of oxidation effects coupled with a tensile mean stress in the OPTMF [18–21]. At the identical maximum temperature, the loading condition (i.e., IF, IPTMF, and OPTMF) for the lowest fatigue life was dependent on the type of material [18–21]. Shankar et al. [22] performed IF, IPTMF, and OPTMF tests in air and vacuum to compare the effects of various time-dependent damage processes on the fatigue life. These revealed that the deleterious effect of oxidation played an important role in the lower fatigue life of the OPTMF in comparison with that of the IPTMF when tests were conducted in air. With the elimination of the oxidation effect in vacuum, a creep-related damage mechanism caused a greater life reduction under the IPTMF condition compared to the OPTMF condition [22]. With the introduction of a tensile dwell of 120 s, the IPTMF life was lower than that of the OPTMF as a result of the microstructure evolution and accelerated oxide scaling at the peak temperature [23]. In addition to the above creep, fatigue, and oxidation effects, another important factor affecting the TMF life of steels is the appearance of a dynamic strain ageing (DSA) phenomenon at a high temperature range. Nagesha et al. [18] observed the occurrence of DSA associated with serrated flow in a hysteresis loop of P91 steel under TMF cycling at the regime of 300–400 °C. The harmful effect of DSA on the fatigue resistance was realised by reducing the crack initiation and propagation life, as a result of the DSA-induced inhomogeneity of the deformation and strong interaction between the oxidation and DSA [24–26].

Because the abovementioned studies concentrated on the effects of the phase angle, more factors affecting the TMF behaviour need to be considered. Therefore, the objective of this study was to provide a better understanding of the TMF behaviour of P92 steel under different mean temperatures, along with the superposition of the dwell effect. Finally, a modified strain amplitude model that includes the mean stress and temperature effects is proposed.

2. Materials and experimental procedure

A normalised and tempered P92 steel pipe with an outer diameter of 350 mm and a wall thickness of 90 mm was obtained from Suzhou Nuclear Power Research Institute. It had the following chemical composition (wt%): 0.121C-0.191Si-0.418Mn-1.60W-0.98Cr-0.048N-0.074Nb-0.339Ni-0.462Mo-0.007Ti, with the balance consisting of Fe. An optical micrograph of the as-received material is shown in Fig. 1(a). It reveals the different martensite lath directions within the lath packets inside the prior austenite grain boundaries (PAGBs). Kimura et al. [27] pointed out that there were M23C6 carbides distributed along the martensite lath boundaries and PAGBs. The triple points can be considered the interfaces of different PAGBs with a higher density of precipitates. The initial microstructure of P92 consisted of a typical tempered martensitic structure, PAGBs, blocks, lath boundaries, packets, and subgrains, as shown in the schematic drawing of Fig. 1(b) [13–15,28–30]. Further, a transmission electron microscopy (TEM) micrograph revealed the distribution of the rod-like M23C6 carbide along the lath boundaries, as shown in Fig. 1(c). In addition, a considerable number of dislocations are found in the lath structure, leading to the formation of a dislocation network (DN). Previous studies [28–30] have confirmed the presence of a coarse type of M23C6 and nanoscale type of MX precipitates at the PAGBs and in the matrix, as illustrated in Fig. 1(d). The interactions between these precipitates, boundaries (i.e., PAGBs and lath boundaries), and DN, as well as the fine-grain strengthening effect at the low-angle lath boundaries, contribute to the excellent fatigue and creep resistance of P92 steel at high temperature [31].

Cylindrical specimens with a 20 mm gauge length and 6 mm gauge diameter were prepared using wire electrical discharge machining. Before conducting fatigue tests, the specimen surface was polished in three steps using 400, 800, and 1200 grit emery paper and buff-finished to eliminate the surface roughness. The detailed dimensions of a specimen are shown in Fig. 2.

Both IF and TMF tests were conducted on MTS 809 using a computer-controlled thermal-mechanical fatigue system. The test device and complete TMF test programme diagrams are shown in Fig. 3. Three thermocouples were attached along the gage length of a specimen (NI-TC1, NI-TC2, and NI-TC3), as shown in Fig. 3(a). NI-TC2 was used to control the temperature, and the symmetrically arranged NI-TC1 and NI-TC3 were responsible for monitoring the temperature deviation within
Fig. 1 – Initial microstructure of P92: (a) optical micrograph, (b) schematic microstructure, (c) TEM micrograph, and (d) distribution of precipitates.

Fig. 2 – Geometry of test specimen.

Fig. 3 – Schematic drawing of (a) test device and (b) test procedure.
the gage length. The complete TMF test procedure included the setup and test, as shown in Fig. 3(b). During the TMF setup, to ensure a temperature gradient along the gage length of the specimen and a temperature deviation between the command and feedback, the temperature gradient was adjusted and the temperature was optimised to make sure that it was in the range of 2% of the maximum cyclic temperature. Then, five thermal cycles without a force load were conducted to allow the free thermal expansion and contraction of the specimen. An extensometer was used to record the thermal strain data during each cycle. The dependence of the thermal strain ($\varepsilon_{\text{th}}^\text{th}$) on the temperature could be calculated using this procedure. Finally, the mechanical strain ($\varepsilon_{\text{m}}^\text{th}$) could be accurately controlled during the TMF test using the function $\varepsilon_{\text{m}}^\text{th} = \varepsilon - \varepsilon_{\text{th}}^\text{th}$. A zero stress test was conducted to verify the accuracy of the thermal strain calculation. After the TMF setup, only one thermocouple was used to control the temperature during the TMF test.

The specific loading conditions are listed in Table 1. Loading cases 1 to 11 were designed to investigate the effects of the mean temperature on the IF, IPTMF, and OPTMF behaviours. For each case, the temperature amplitude was kept constant and identical to 200 °C. Loading cases 12 to 15 were used to study the effect of the tensile dwell on the IPTMF behaviour. Here, the acronym T-IPTMF is used to refer to the IPTMF case with a tensile dwell. Further, IPTMF tests with a symmetric dwell were conducted for cases 16 to 24 to compare the effects of a symmetric dwell time on the IPTMF behaviour, where S-IPTMF refers to the IPTMF case with a symmetric dwell. The corresponding loading waveforms are displayed in Fig. 4. The determination criterion for the fatigue life is illustrated in Fig. 5. The failure life is defined as the crossover point between the parallel black line (20% away from the red one) and the cyclic peak stress curve [32]. Based on this criterion, all the experiment results are listed in Table 2.

### 3. Results and discussion

#### 3.1. Effects of temperature on IF and TMF results

Fig. 6 shows the variation of the cyclic stress amplitude with the normalised fatigue life under different loading conditions. Under all the loading conditions, the cyclic softening of specimens was shown by the continuous decrease of the cyclic stress amplitude during the entire fatigue process. The value of the cyclic stress amplitude during the IPTMF test was close to that of the OPTMF test in the identical temperature range. The cyclic stress amplitude curve decreased with an increase in the mean temperature. This decrease could be attributed to temperature-induced softening. However, the decreasing trend of the cyclic stress amplitude with the increase in the average temperature became less pronounced in the TMF tests, Fig. 6(a), which was also true for the IF tests, Fig. 6(b). Further, the dependences of the stress amplitudes of the IPTMF, OPTMF, and IF on the test temperature at the half-life points are depicted in Fig. 7. For the IPTMF and OPTMF tests, the mean temperature ($T_m = (T_{\text{max}} + T_{\text{min}})/2$) was adopted as the average temperature experienced during cyclic deformation. It was found that the decreases in the stress amplitudes of the IPTMF, OPTMF, and IF with an increase in the test temperature had two-stage variations. At the mean temperature range of 450–500 °C, the stress amplitude showed a significant decrease with an increase in temperature. Beyond 500 °C, the decrease in the stress amplitude was less significant, indicating that the effect of the test temperature on the

### Table 1 – Loading conditions.

<table>
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<th>Loading case</th>
<th>Specimen No.</th>
<th>Phase angle/°</th>
<th>Dwell time/s</th>
<th>Temperature/°C</th>
<th>Strain amplitude/%</th>
<th>Strain ratio</th>
<th>Period/s</th>
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<td>0</td>
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<td>−1</td>
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<td>−1</td>
<td>120</td>
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<td>−1</td>
<td>120</td>
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Fig. 4 – Loading waveforms of different tests: (a) IPTMF60, (b) OPTMF60, (c) IF, (d) IPTMF30, (e) T-IPTMF30, (f) IPTMF80, (g) S-IPTMF80 (20 s), and (h) S-IPTMF80 (40 s).
fatigue behaviour under all loading conditions became less pronounced.

**Fig. 8** provides the mean stress curves under different loading conditions. In the IF test loading case, the P92 steel showed cyclic asymmetry, because a compressive mean stress was produced during the cyclic deformation, indicating that the plastic deformation resistance in the compression direction was greater than that in the tension direction at the same temperature. With the application of the cycling temperature, the cyclic asymmetry became more predominant. In the IPTMF tests, the tension part of the hysteresis loop endured a higher temperature than the compression part. Thus, the tensile peak stress was much lower than the compressive peak stress, leading to the production of a compressive mean stress. In contrast, a tensile mean stress was produced during the OPTMF tests. From the value of this mean stress, it can be expected that the cyclic asymmetry caused more damage in the OPTMF tests than that in the IPTMF tests, because the tensile mean stress was harmful to the fatigue resistance by accelerating the crack opening and increasing the cumulative fatigue damage. In contrast, the compressive mean stress in the IPTMF tests could have retarded the crack growth and improve the fatigue life of the material.

The low cycle fatigue behaviour at elevated temperatures is affected by multiple time-dependent processes such as oxidation, creep, and DSA, depending upon the applied strain rate and temperature. Although various deformation and damage mechanisms may be activated simultaneously during cyclic deformation, one dominant mechanism develops after the complex interaction and competition between different factors [26]. In order to study the additional deformation process during cyclic deformation, hysteresis loops for the P92 steel at the second cycle and half-life cycle in the IF, IPTMF, and OPTMF tests are plotted in **Fig. 9**. It should be noted here that the strain rates applied in the IF and TMF tests were $1.5 \times 10^{-3}$ and $2 \times 10^{-4}$ s$^{-1}$, respectively. DSA is mainly the result of the pinning effect of solute atoms and mobile dislocations, and it leads to serrated flow in stress–strain hysteresis loops on

<table>
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<tr>
<th>Loading case</th>
<th>Total strain amplitude $\Delta \varepsilon$, %</th>
<th>Plastic strain amplitude $\Delta \varepsilon_p$, %</th>
<th>Maximum tensile stress $\sigma_{max}$, MPa</th>
<th>Mean stress $\sigma_m$, MPa</th>
<th>Stress amplitude $\Delta \sigma/2$, MPa</th>
<th>Fatigue life $N_f$</th>
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<td>188.06</td>
<td>337</td>
</tr>
<tr>
<td>23</td>
<td>0.8</td>
<td>0.71</td>
<td>148.85</td>
<td>−36.71</td>
<td>185.56</td>
<td>364</td>
</tr>
<tr>
<td>24</td>
<td>0.8</td>
<td>0.69</td>
<td>175.26</td>
<td>−36.86</td>
<td>212.12</td>
<td>375</td>
</tr>
</tbody>
</table>
the macroscale [18]. In the IF tests, the DSA was absent at the temperature range of 450–650 °C because of the higher strain rate, as shown in Fig. 9(a) and (b). In the TMF tests, the DSA phenomenon could occur at the lowest temperature range (350–550 °C), regardless of the variation of the phase angle. In Fig. 9(c), the material exhibits DSA, because serrations in the tension part of the hysteresis loops can be observed. In addition, the tensile peak stress does not correspond to the maximum tensile mechanical strain. After the tensile peak stress is reached, the tensile stress gradually decreases with an increase in the applied tensile mechanical strain, indicating the occurrence of CSR. In contrast to the results of the IPTMF test, the DSA and CSR of the OPTMF occurred during compressive deformation because of cyclic deformation in the compressive direction at higher temperature, as shown in Fig. 9(d). With an increase in the mean temperature, both the DSA and CSR phenomena became more prominent, as seen in Fig. 9(e) and (f). Comparing the relaxation stress (i.e., $\Delta \sigma_{\text{RC}}$ and $\Delta \sigma_{\text{CC}}$) values in Fig. 9(e) and (f), it is shown that the initial creep relaxation stress of the IPTMF test is higher than that of the OPTMF test, even though the average temperatures ($T_m$) are identical (600 °C). The more significant creep phenomenon during tensile deformation was caused by the lower deformation resistance along the tensile direction at the same temperature, which was already shown in the IF tests. With an increase in the cyclic number, the value of stress relaxation decreased as a result of the exhaustion of the creep effect. Therefore, the difference between the $\Delta \sigma_{\text{CC}}$ value of the IPTMF test and the $\Delta \sigma_{\text{CC}}$ value of the OPTMF test decreased.

Comparison of the relaxation stress ($\Delta \sigma_{\text{RC}}$ and $\Delta \sigma_{\text{CC}}$) with the normalised cyclic number. Because the value of the relaxation stress quickly decreased to a minor value before 0.1Nc, it can be inferred that the effect of creep on the IP and OPTMF was rather limited.

It has been proposed that the softening ratio ($\alpha$) is related to the microstructure evolution during cyclic deformation [1], which is defined as follows:

$$\alpha = \frac{\sigma_0 - \sigma_{\text{sat}}}{\sigma_0}$$

Fig. 11 plots the temperature-dependent fatigue life and softening ratio. In the IF test, the fatigue life showed a continuous decrease with an increase in the temperature, as shown in Fig. 11(a). The increase in the softening ratio indicated that a higher extent of recovery was caused by the increase in temperature, which was consistent with previous studies.
Fig. 9 – Typical hysteresis loops of P92 steel: (a) IF at 450 °C, (b) IF at 650 °C, (c) IP TMF at 350–550 °C, (d) OP TMF at 350–550 °C, (e) IPTMF at 450–650 °C, and (f) OPTMF at 450–650 °C.

Fig. 10 – Variation of relaxation stress with normalised fatigue life.

Fig. 11 – Variation of (a) isothermal fatigue life and (b) thermomechanical fatigue life with temperature and softening ratio.

[1,3,4]. The softening ratio showed less dependence on the test temperature beyond 500 °C when compared to the 450–500 °C range, indicating the weak effect of the temperature on the microstructure evolution. Consequently, the degradation of the fatigue resistance caused by the increase in temperature also became less prominent. Similar trends for the softening ratio and fatigue life with an increase in temperature could be found in the TMF tests. A previous study on P91 steel also found that the microstructural evolution of the material during IP and OP cycling showed no significant difference at the temperature range of 450–650 °C, and the fatigue life values for OPTMF and IPTMF tests also showed a lower difference [18], which was consistent with the results shown in Fig. 11(b). In addition, the failure cycles of the IPTMF tests at the average temperatures of 500 to 550 °C were close, which was also true for the OPTMF tests. It has generally been recognised that the DSA caused by the interaction between solute atoms and mobile dislocations enhances the resistance to plastic deformation and causes a reduction in ductility. A deterioration of fatigue resistance due to DSA has been reported by many researchers [24–26]. In Fig. 9, the more prominent DSA and
Fig. 12 – SEM fractographs of specimens: (a) IPTMF60 at 350–550 °C, (b) IPTMF60 at 450–650 °C, (c) OPTMF60 at 350–550 °C, and (d) IF at 450 °C.

CSR at higher temperatures can be seen from the variation of the hysteresis loops. In the IF tests, the fatigue life at $T_{\text{max}}$ (650 °C) was also higher than those of the IF and OPTMF tests at the temperature range of 450–650 °C. Compared with the IF tests, the decreased fatigue resistance in the TMF tests could be attributed to the occurrence of DSA and CSR. In addition, the fatigue life in the IPTMF test was higher than that in the OPTMF test at all temperatures. Although the initial relaxation stress of IPTMF was higher than that of OPTMF, the difference between $\Delta \sigma_{\text{PC}}$ and $\Delta \sigma_{\text{CC}}$ dropped quickly. Therefore, the direction of the mean stress played an important role in the fatigue life difference between the IPTMF and OPTMF.

Because the initiation and growth of microcracks are considered to make significant contributions towards cyclic damage and life reduction, fractographic examinations of the specimens after the IF and TMF tests were conducted using scanning electron microscopy (SEM). Multiple potential crack initiation sites could be observed and are indicated by the yellow arrows in Fig. 12(a)–(d). Comparing Fig. 12(a) and (b), it can be found that, with an increase in temperature, more crack initiation and growth traces appear in the fracture surface, leading to a decrease in the fatigue life at higher temperatures. At the same temperature range, the secondary cracks were more pronounced in the OPTMF test than those in the IPTMF test. Compared with Fig. 12(a)–(c), the fracture surface after the IF test shown in Fig. 12(d) is flatter, indicating that fewer plastic deformation traces were produced in the IF test because of the absence of the CSR effect. Most of the microcracks from the crack initiation sites in Fig. 12(a)–(c) propagated to a considerable depth into the material. Consequently, some microcracks could join and propagate as the primary crack or coalesce with the propagating primary crack, leading to a higher crack propagation rate and lower fatigue life in the TMF tests compared to the IF tests. Related research has pointed out that the DSA-induced inhomogeneity of deformation reduces the crack initiation and propagation life through the initiation of multiple cracks [24]. Magnified fractographs of the crack initiation areas are shown in Fig. 13. In relation to this aspect, a deeper main crack initiating at the surface and propagating into the matrix could clearly be seen in the OPTMF test, around which more secondary cracks appeared and tended to coalesce with the main crack, as shown in Fig. 13(b). The tensile mean stress in the OPTMF test prevented contact between the crack surfaces and facilitated crack opening, resulting in a rapid crack growth rate. Thus, the shortest fatigue life for the P92 steel was found in the OPTMF test.
the increase in the stress amplitude of a specimen with dwell was more significant at a higher strain amplitude, which can clearly be seen from Fig. 14a. Because the tensile dwell caused an increase in the compressive stress, the absolute value of compressive mean stress increased, as shown in Fig. 14b.

Fig. 15 plots typical hysteresis loops for the P92 steel at a strain ratio of zero. IPTMF tests at strain amplitudes of 0.30% and 0.45% with a tensile dwell were also conducted. Compared with the IP and OPTMF results in Fig. 9, the serrated feature of the hysteresis loops in Fig. 15(a) and (c) are inconspicuous, which may indicate the suppression of the DSA effect at a strain ratio of R = 0. For the IPTMF results with a tensile dwell shown in Fig. 15(b) and (d), the serrated cyclic loop further weakens. In addition, the CSR phenomenon can clearly be seen from the decrease in the stress amplitude. A comparison of the values of stress relaxation (ΔσTC) in Fig. 15(b) and (d), found that the CSR was enhanced by an increase in the strain amplitude. In addition, the stress relaxation gradually decreased because of the exhaustion of the creep effect. As listed in Table 2, plastic strain amplitude increases because of the tensile dwell, leading to a decrease in the fatigue life at an identical strain amplitude. Compared with Fig. 13, the oxidation phenomenon can clearly be seen in the crack initiation area after TMF cycling with dwell, as shown in Fig. 16(a). Although the tensile dwell decreases the fatigue life, the crack in the fracture surface still propagates with a transgranular mode. No intergranular crack caused by creep damage can be found in Fig. 16(a) and (b) because of the limited dwell time. The breakage of fatigue striations in the crack propagation area of Fig. 16(b) also indicates that an enhanced oxidation phenomenon resulted from the tensile dwell.

3.3. Effects of symmetrical dwell on TMF results

Fig. 17 shows the cyclic stress response of the P92 steel in IPTMF tests with a symmetric dwell versus the normalised fatigue life. From Fig. 17(a), it can be found that both the tensile peak stress and compressive peak stress increase with the symmetric dwell time. Therefore, the stress amplitude increases with the symmetric dwell time at the same strain amplitude in Fig. 17(b). In addition, the value of compressive mean stress shows a slight increase with the dwell time (Fig. 18). Because plastic deformation of the P92 steel in the tensile direction occurred more easily than that in the compressive direction in the IPTMF test, the increase in the
compressive peak stress caused by the tensile creep strain was more significant than that of the tensile peak stress that resulted from the compressive creep strain. Consequently, the symmetric dwell slightly increased the compressive mean stress, which can be seen in Fig. 18.

In order to investigate the effects of the symmetric dwell on the DSA and CSR produced in the IPTMF tests, the hysteresis loops at the second and half-life cycles are provided in Fig. 19. Compared with the IPTMF results in Fig. 9(c) and (e), the cyclic temperatures of the compressive parts of the hysteresis loops in Fig. 19(a) and (c) are higher. Therefore, the serrated flow of the hysteresis loops occurs not only during tensile deformation, but also during compressive deformation in Fig. 19(a) and (c), which is absent in Fig. 9(c) and (e). For the IPTMF tests with the symmetric dwell, the CSR can be seen in Fig. 19(b) and (d). In addition, the tensile creep effect is more prominent than the compressive creep effect, which can be found from the values of $\Delta \sigma_{TC}$ and $\Delta \sigma_{CC}$. The CSR value also decreases with the cyclic number and increases with the applied strain amplitude, which is consistent with the results shown in Fig. 9. Compared with the hysteresis loops of the IPTMF tests without dwell (Fig. 9(c), (e) and Fig. 19(a), (c)), the DSA effect in Fig. 19(b) and (d) is significantly weakened by the introduction of the symmetric dwell. The serrated flow
Fig. 17 – Cyclic stress response of P92 steel in IPTMF tests with symmetric dwell: (a) peak stress and (b) stress amplitude versus normalised fatigue life.

In the compressive part of the cyclic stress–strain hysteresis loop disappears in Fig. 19(b) and (d). In addition, the serration in the tensile part of the cyclic stress–strain hysteresis loop in Fig. 19(b) and (d) is not as significant as that in Fig. 19(a) and (c), which is also the manifestation of the restrained DSA phenomenon.

The dependence of the fatigue life on the dwell time at different strain amplitudes is illustrated in Fig. 20(a). At a constant strain amplitude, the fatigue life increased with the dwell time, which may be the result of the gradual disappearance of the DSA phenomenon after the dwell. The improvement in the fatigue life under the symmetric dwell effects is clearly shown in Fig. 20(b). From the decreasing trend of the fatigue life with the strain amplitude, it can be observed that the effect of the symmetric dwell on the fatigue life is more significant at a low strain amplitude.

In order to investigate a potential mechanism for the beneficial effect of a short dwell time on the TMF life, the fracture surfaces of specimens with and without dwell were analysed. In Fig. 21(a) and (b), multiple potential crack initiation sites can be observed, which are consistent with Fig. 12. Because cyclic deformation occurs at elevated temperatures, the application of dwell enhances the oxidation phenomenon. The crisp oxide layer produced during the tensile dwell was easy to crush, because the oxide layer covering the fracture surface was subjected to more compressive strain than the matrix material. Similarly, the crisp oxide layer produced during the compressive dwell tended to be broken during the reverse loading. Evidence of the fragile nature of the oxide layer in S-IPTMF was provided by the appearance of crack fissures, as shown in Fig. 21(d). In contrast, fatigue striations could be distinguished in IPTMF, as shown in Fig. 21(c). The fatigue life may be expected to be lower under S-IPTMF cycling because of the assistance of the secondary cracks produced by the oxide layer. To further investigate the crack propagation behaviour, longitudinal sections near fracture surfaces under the IPTMF and S-IPTMF were observed, as shown in Fig. 22. Distinct differences could be detected in the crack morphologies between the IPTMF and S-IPTMF tests. A crack with a sharp form without any branch in the IPTMF test could propagate into the material with a vertical length of approximately 285 μm (Fig. 22(a)), although crack deflection could be observed in the later stage of crack growth. The enlarged view of the crack front in the top right corner of Fig. 22(a) shows the non-significant oxidation phenomenon for the IPTMF. In contrast, a crack with a blunt shape in the S-IPTMF test propagated into the material with a vertical length of approximately 130 μm (Fig. 22(b)). Then, this crack split into two short cracks coated with a thick oxide layer. The oxide layer at the tips of the two cracks and the junction between the initial crack and crack branch was more remarkable. The enlarged view of the crack front in the top right corner of Fig. 22(b) shows the detail of crack tip blunting. From this perspective, the transgranular manner of crack propagation can clearly be seen. Therefore, creep damage was not the main damage mechanism because of the limited dwell time. In addition, crack tip blunting can also be detected in another crack of Fig. 22(b). More cracks could be produced in S-IPTMF with the assistance of oxidation, which has already been shown in Fig. 21(d). Compared with the IPTMF test, the DSA effect was weakened and the oxidation phenomenon was enhanced as a result of the dwell in the S-IPTMF test. A strong DSA effect caused a stress concentration and increased the driving force for the crack propagation, leading to the long and sharp crack that

Fig. 18 – Evolution of mean stress with normalised fatigue life in IPTMF tests with symmetric dwell.
has been shown in Fig. 22(a). Previous studies [6,33] found that the crack propagation rate was significantly increased under the influence of DSA and retarded blunting of the propagating crack. This increase resulted in a higher rate of crack propagation compared to that without the DSA effect. In the S-IPTMF test, the opening crack tip in the air was more easily oxidised because of the dwell, leading to the crack tip blunting that has been shown in Fig. 22(b). Moreover, the breakage of the oxide layer around the initial crack tip provided potential sites for crack initiation, leading to the appearance of crack branching and the release of stress. Consequently, the effective stress attainable near the propagating crack front decreased. In contrast, the stress concentration produced by the DSA near the propagating crack front ensured a more rapid crack propagation rate in the IPTMF test when compared to that in the S-IPTMF test. Therefore, the IPTMF fatigue life could be improved by a short symmetric dwell time.

3.4. TEM study

The microstructure evolution of the P92 steel after pure fatigue was related to the applied strain amplitude, temperature, and cyclic numbers [8]. It was concluded that the annihilation of mobile dislocations, growth of a martensitic lath, and transformation from a lath structure in equiaxed subgrain structure were the predominant microstructural evolutions during LCF [15]. Nagesha et al. [18–20] and Shankar et al. [22] found that the recovery of the lath structures into the substructures was still the main deformation mechanism under TMF cycling. Moreover, there was little difference in the substructure evolution between IP and OP cycling. Fig. 23(a) and (b) shows microstructure changes after IPTMF and IPTMF with a tensile dwell at a strain amplitude of 0.3% and R = 0. Compared with the initial microstructure in Fig. 1(c), it can easily be found that the coarsening of the martensite lath occurs as the width of the lath structure increases, as indicated by the red arrows in Fig. 23(a). Meanwhile, the dislocation density inside the lath structures decreases. In addition, the decomposition of the lath structure into substructures is observed. However, these substructures with an elongated shape are ill-defined. Comparing Fig. 23(a) and (b), it seems that there is little difference in the microstructure at the strain amplitude of 0.3% with a tensile dwell. Because the dwell is just a hold in the tension part and the dwell time is short, the modification of the microstructure produced by the tensile dwell during cyclic deformation is rather limited.

Fig. 24(a)–(b) and (c)–(d) shows the microstructures of the P92 steel after IPTMF and IPTMF with a symmetric dwell at a strain amplitude of 0.4% and R = −1, respectively. Fig. 24(a) provides evidence of the dissolution of the initial lath structure. The approximately aligned carbides depict the position of prior lath boundaries. However, those lath boundaries disappear after TMF cycling. With an increase in the strain amplitude, the recovery of substructures is more predominant, as indicated by the white triangles. However, a considerable number of dislocations are still stored at the boundary of the substructure. The total dislocation density is further decreased by the application of a symmetric dwell. The evolution of the substructure is more accentuated. The
number of substructures is obviously increased, as shown in Fig. 24(c). Moreover, there are also some equiaxed grains with a relatively large size in Fig. 24(d). Comparing Fig. 23 and Fig. 24, it can easily be seen that the transformation of the lath boundaries into equiaxed subgrain structures is more predominant at the loading condition of IPTMF with a symmetrical dwell at the strain amplitude of 0.4%, because of increases in the strain amplitude and dwell time. Fournier et al. [34] and Shankar et al. [35] also pointed out that fully developed cells can be observed under a creep-fatigue test. Because of the CSR effect, the increased plastic strain can increase the size of the cell or subgrain [34,35].

3.5.  Fatigue life prediction

In order to design for the fatigue of structural components operating under TMF cycling, a reliable estimation of the fatigue life has to be made. This section shows how different life prediction models were employed to predict the fatigue life under IF and TMF cycling.

3.5.1.  Coffin–Manson model

Fig. 25 provides the dependence of the fatigue life on the strain amplitude and plastic strain amplitude. From Fig. 25(a) and (b), it can be seen that the fatigue life of the IF at $T_{\text{max}}$ is still higher than those of IPTMF and OPTMF at identical strain amplitudes. Therefore, the traditional design practice for TMF that relies on the IF life obtained at $T_{\text{max}}$ of the expected TMF cycle is non-conservative for P92 steel and will cause adverse implications for the structural integrity. In addition, the slight reduction in the IPTMF fatigue life caused by the tensile dwell and extended IPTMF life resulting from the short symmetric dwell can be clearly seen in Fig. 25(a). Compared with Fig. 25(a), the data points versus fatigue life in Fig. 25(b) show less dispersion. The classic Manson–Coffin–Basquin law describes the dependence of the fatigue life on the plastic strain amplitude parameter as follows:

$$\frac{\Delta \varepsilon_p}{2} = \varepsilon_p' (N_f)^c \quad (2)$$

where $\Delta \varepsilon_p/2$, $\varepsilon_p'$, and $c$ represent the plastic strain amplitude, fatigue ductility coefficient, and exponent, respectively. Based on a regression analysis, the final expression for this model is as follows:

$$\frac{\Delta \varepsilon_p}{2} = 20.236 (N_f)^{-0.550} \quad (3)$$

3.5.2.  Energy model

The plastic strain energy density can be calculated by integrating the cyclic stress–plastic strain hysteresis loop, which represents the accumulated damage per cycle during cyclic deformation. In comparison with either the stress or strain alone, the energy-related method has a potential advantage, because it includes the constitutive response of the material. The value of plastic strain energy calculated from the half-life cycle was adopted as the damage parameter for predicting the fatigue life, as expressed by the following:

$$\Delta W_p = k (N_f)^m \quad (4)$$

where $\Delta W_p$, $k$, and $m$ are the plastic strain energy and material parameters. Based on the experimental data, the energy model is expressed as follows:

$$\Delta W_p = 9119.371 (N_f)^{-0.512} \quad (5)$$

3.5.3.  Ostergren model

A damage function that included a range of strain and stress was proposed by Ostergren [36]. The basic assumption of this model was that the net tensile hysteresis energy dominated the cyclic damage, because the driving force for crack initiation and growth was mainly caused by the tensile stress. Fatigue life versus $\Delta \varepsilon_p \sigma_{\text{max}}$ can be described by the power law relationship:

$$N_f = \alpha \left( \frac{\Delta \varepsilon_p}{2} \sigma_{\text{max}} \right)^b \quad (6)$$

where $\Delta \varepsilon_p$, $\sigma_{\text{max}}$, and $C$ are the plastic strain amplitude, peak tensile stress of the half-life cycle, and material constant,
respectively. Using regression analysis, the final expression for this model can be written as follows:

$$N_f = 60451.522 \left( \frac{\Delta \varepsilon_p}{2 \sigma_{max}} \right)^{-0.997}$$  \hspace{1cm} (7)

Fig. 26 displays a comparison of the observed fatigue life and the values predicted by the Coffin–Manson model, energy model, and Ostergren model. Scatter bands with a factor of two are shown to evaluate the accuracy of the data predicted by the different models. Data points above the upper bound represent overestimated predictions for the fatigue life, indicating non-conservative fatigue life predictions. In contrast, data points below the lower bound represent underestimated predictions for the fatigue life. According to the predicted results, the plastic strain energy method exhibits the worst accuracy in Fig. 26(b), because many of the data points predicted by this model are located outside the upper and lower bounds. Compared with the energy approach, almost all the data points of Fig. 26(a) are located within the scatter bands, indicating a more accurate fatigue life prediction by Coffin–Manson equation. However, the majority of the data points predicted by the Coffin–Manson model is higher than the experimental values, indicating an overestimation of the
3.5.4. Modified Coffin–Manson model

It has been shown that a tensile mean stress and an increase in the cycling temperature decrease the TMF life. The effects of these two factors have not been included in the current fatigue life by this model. In Fig. 26(c), one data point is located outside the scatter bands. Thus, the current models cannot satisfactorily predict the IF and TMF life values of P92 steel.

Fig. 23 – TEM micrographs of microstructure of P92 steel after (a) IPTMF and (b) IPTMF with tensile dwell at strain amplitude of 0.3% and $R = 0$.

Fig. 24 – TEM micrographs of microstructure of P92 steel after (a)–(b) IPTMF and (c)–(d) IPTMF with symmetrical dwell at strain amplitude of 0.4% and $R = -1$. 

fatigue life by this model. In Fig. 26(c), one data point is located outside the scatter bands. Thus, the current models cannot satisfactorily predict the IF and TMF life values of P92 steel.
models. Therefore, it is necessary to modify the current model by considering the mean stress and cycling temperature effects. It has been pointed out that mean stress effects can be realised in the conventional strain–life models by introducing the mean stress term in the power law form \((1 + \frac{\sigma_m}{\sigma_{max}})^n\) [37–39]. Similarly, the average cycling temperature effects of the TMF at a constant \(T_{max}\) can be described by the introduction of a temperature term in the power law form \((\frac{T_{min}}{T_{max}})^\theta\). Thus, a modified Coffin–Manson equation can be expressed as follows:

\[
\frac{\Delta k}{2} = \varepsilon_f(N_f)^c \left(1 + \frac{\sigma_m}{\sigma_{max}}\right)^a \left(\frac{T_{min}}{T_{max}}\right)^\beta \tag{8}
\]

where \(\sigma_m, T_{min}\), and \(T_{max}\) denote the mean stress, minimum temperature during TMF cycling, and maximum temperature during TMF cycling, respectively. \(a\) and \(\beta\) can be regarded as the mean stress sensitivity factor and temperature sensitivity factor, respectively. The values of \(a\) and \(\beta\) should be positive, because increasing the mean stress and temperature will lead to a decrease in the fatigue life. At a constant \(T_{max}\), the value of \(T_{min}\) is higher, and the TMF life is lower. It should be noted here that in the IF test, the decrease in the fatigue life with the applied temperature could be partly reflected by the increase in the plastic strain amplitude, because \(T_{min}\) was equal to \(T_{max}\) in the IF test. Because the fatigue ductility coefficient \(\varepsilon_f\) and

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**Fig. 25** – Dependence of fatigue life on (a) strain amplitude and (b) plastic strain amplitude.

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**Fig. 26** – Comparison of observed fatigue life and values predicted using (a) Coffin–Manson model, (b) energy model, and (c) Ostergren model.
exponent $c$ were determined in Eq. (3), only the mean stress sensitivity factor $\alpha$ and temperature sensitivity factor $\beta$ need to be fitted. Using regression analysis, the final expression for the modified Coffin–Manson model is as follows:

$$\frac{\Delta N_f}{2} = 20.236(N_f)^{0.559}(1 + \frac{\sigma_m}{\sigma_{max}})^{0.196}(\frac{T_{min}}{T_{max}})^{0.559}$$

(9)

Fig. 27 shows a comparison between the fatigue life predicted by the modified Coffin–Manson model and observed fatigue life. With the introduction of the mean stress and cycling temperature items, all the data points can be located between scatter bands with a factor of two. Moreover, almost all the predicted data points are located within narrow scatter bands with a factor of 1.5 and show a better data distribution around the prediction line of $y = x$. Therefore, compared with Fig. 26(a)-(c), the prediction accuracy of the modified Coffin–Manson model shows a significant improvement.

4. Conclusions

The IF and TMF behaviours and lifetime prediction of P92 were investigated in a temperature range of 350–650 °C. The main conclusions are as follows:

1. At a constant temperature amplitude, the fatigue life values of the IPTMF and OPTMF decreased with an increase in the mean temperature. The harmful effect of the cycling temperature on the fatigue life was also seen in the IF test. Compared with a higher temperature range, the IF and TMF results were more sensitive to the cycling temperature at a range of 450–500 °C. The DSA and CSR phenomena could be observed in the IPTMF and OPTMF tests. At identical average temperatures, the fatigue life values of P92 steel under different loading conditions obeyed the following sequence: IF > IPTMF > OPTMF.

2. The CSR was enhanced by the application of dwell, and the corresponding fatigue life of the IPTMF was decreased. Because the tensile dwell time was limited, the failure of the material was still dominated by fatigue damage. A crack in the fracture surface still propagated with a transgranular mode, and no intergranular crack could be found.

3. A short symmetric dwell had beneficial effects on the fatigue resistance of the TMF. The extended fatigue life of the IPTMF with dwell could be attributed to the following two factors. First, compared to the TMF behaviour without dwell, the DSA effect was weakened. Second, the opening crack tip in air was more easily oxidised as a result of the dwell, leading to crack tip blunting. In addition, the formation of a crisp oxide layer around the primary crack front could provide a new crack initiation site, leading to the appearance of crack branching and stress release. The absence of the DSA effect, and the formation of crack tip blunting and crack branching, could prevent a stress concentration, retard the propagation of the primary crack, and increase the fatigue life.

4. The transformation of lath structures into substructures was the dominant deformation mechanism under different loading conditions. Moreover, with increases in the strain amplitude and dwell time, the growth of equiaxial subgrains could be observed.

5. Compared to the current Coffin–Manson, energy, and Oster-gren models, the prediction accuracy of the Coffin–Manson equation that was modified by incorporating the mean stress and temperature was significantly improved.

Conflicts of interest

The authors declare no conflict of interest in this work.

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