FATIGUE PROPERTIES OF NICKEL-BASE SUPERALLOY INCONEL 792-5A AT 800°C

Miroslav ŠMÍD¹, Karel OBRTLÍK², Martin PETRENEČKA, Jaroslav POLÁK³, Karel HRBÁČEK⁴

¹ Ústav fyziky materiálů, Akademie věd České republiky, v.v.i., Žižkova 22, 616 62 Brno
² PBS Velká Bíteš a.s., Vlkovská 279, 595 12, Velká Bíteš

Abstract

Smooth specimens were cyclically strained under strain control with constant strain amplitude and constant strain rate. Low cycle fatigue tests were conducted in servo-hydraulic pulsator MTS equipped with a three zone resistance furnace at temperature 800°C in air. Surface of the specimen gauge length was mechanically ground and polished to enable SEM observation. Fracture surface was studied in SEM after fatigue test termination. Selected specimens were used to prepare foils for the transmission electron microscope (TEM) observation of microstructure and dislocation arrangement.

Hysteresis loops were recorded for selected numbers of cycles. They were used to obtain cyclic hardening/softening curves, cyclic stress-strain curve and fatigue life curves in the representation of stress amplitude, total strain amplitude and plastic strain amplitude versus number of cycles to fracture. Experimental points can be approximated with the Manson-Coffin law and the Basquin law. Fracture surface examinations revealed fatigue crack initiation sites.

Keywords: Inconel 792-5A, low cycle fatigue, high temperatures, fatigue life curves, dislocation structure.

1. INTRODUCTION

Inconel 792-5A (In 792-5A) is a cast polycrystalline nickel base superalloy. The material is strengthened by precipitates γ’ (Ni₃Al). Therefore, it exhibits excellent mechanical properties at elevated temperatures. Its other advantage is very good high temperature corrosion resistance and hence it is used for production of the most thermally and mechanically stressed components, for example discs and blades of gas turbines and other power-producing units. In critical parts of these components cyclic elastic-plastic deformation occurs especially during warming up and cooling, e.i. start up and shut down cycles respectively [1]. From this reason, it is important to study low cycle fatigue properties at elevated temperatures and also to observe changes in the material microstructure after cyclic straining with electron microscopy. The aim of the paper is to report results of the stress-strain response and fatigue life of In 792-5A at temperature 800°C supplemented with electron microscope observation.

2. EXPERIMENT

Polycrystalline superalloy In 792-5A was provided by PBS Velká Bíteš, a.s. as conventionally cast rods. Cylindrical button-end specimens were machined parallel to the rod axis with gauge length and diameter of 15 and 6 mm, respectively. Chemical composition is presented in Table 1. The gauge length of specimens was mechanically ground and polished for further surface relief observation.

Structure of the superalloy consists of coarse dendritic grains, carbides, eutectics γ/γ’ and numerous shrinkage pores with diameter up to 0.7 mm. Typical structure is shown in Figure 1a. Microstructure of In 792-5A contain matrix γ and mostly cuboidal precipitates γ’ with the edge size approximately 600...
nm. It is also obvious that numerous fine precipitates $\gamma'$ are homogenously distributed in the matrix (Figure 1b). The linear intercept method revealed the average grain size of 3 mm, thus the gauge length can comprise several grains. This fact reflects in significant scatter of elastic modulus of individual specimens. Typical values were in interval from 151 to 193 GPa at 800°C.

Table 1. Chemical composition of IN 792-5A superalloy (wt. %)

<table>
<thead>
<tr>
<th>Cr</th>
<th>Co</th>
<th>Ti</th>
<th>Al</th>
<th>Ta</th>
<th>W</th>
<th>Mo</th>
<th>Nb</th>
<th>Fe</th>
<th>Zr</th>
<th>C</th>
<th>B</th>
<th>Ni</th>
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<td>12.28</td>
<td>8.87</td>
<td>3.98</td>
<td>3.36</td>
<td>4.12</td>
<td>4.1</td>
<td>1.81</td>
<td>0.1</td>
<td>0.16</td>
<td>0.031</td>
<td>0.078</td>
<td>0.015</td>
<td>rest</td>
</tr>
</tbody>
</table>

Fig. 1 Structure of IN 792-5A a) carbides, eutectics $\gamma/\gamma'$ in interdendritic areas b) TEM micrograph showing typical microstructure consisted of cuboidal $\gamma'$ precipitates and matrix $\gamma$ with very fine $\gamma'$ precipitates.

Low cycle fatigue tests were conducted on an electro-hydraulic computer controlled testing system MTS 810 in the strain control regime at temperature 800°C in air. The strain rate was held constant (0.002 s$^{-1}$) and a symmetric push-pull cycle ($R_{\varepsilon} = -1$) was used. The strain was measured and subsequently controlled by a sensitive extensometer with 12 mm base. Fatigue tests with constant total strain amplitude $\varepsilon_a$ were run to fracture. Heating on experimental temperature was provided by a 3-zone resistance furnace controlled by a 3-channel regulator. Actual temperature was measured by 3 thermocouples attached to both specimen ends and also to the gauge length.

Scanning electron microscope JEOL JMS6460 was used for surface relief and fracture surface observations. Dislocation structure observation was conducted in transmission electron microscope Philips CM-12 operating at 120 kV with a double tilt holder.

3. RESULTS

3.1 Stress-strain response

Figure 2a and 2b shows the cyclic hardening/softening curves of selected specimens in the representation of the stress amplitude versus the number of cycles and the plastic strain amplitude versus the number of cycles respectively. Each experiment was conducted with different total strain amplitude. Tests conducted at low amplitudes result in stable stress response with slight softening
during the whole fatigue life. Specimens cycled at high total strain amplitudes show initial hardening followed by softening which became more significant at the end of the fatigue life. It can be also seen that the specimen cycled at the highest total strain amplitude doesn’t yield the biggest stress and plastic strain amplitude. This fact is a result of modulus scatter of individual specimens.

Cyclic stress-strain curve of the material is shown in Figure 3. The diagram was plotted using the stress and plastic strain amplitudes at half-life. Experimental data were fitted by the power law

\[ \log \sigma_a = \log K' + n' \log \varepsilon_{ap} \]  \hspace{1cm} (1)

where \( K' \) is fatigue hardening coefficient and \( n' \) is fatigue hardening exponent. Their values are 1120 MPa and 0.146 respectively.

![Fig. 3 Cyclic stress-strain curve](image)

### 3.2 Fatigue life

The representation of the plastic strain amplitude \( \varepsilon_{ap} \) at half life vs. the number of cycles to fracture \( N_f \) is shown in Figure 4a. Experimental data were approximated by the Manson-Coffin law

\[ \varepsilon_{ap} = \varepsilon'_f \left( 2N_f \right)^c \]  \hspace{1cm} (2)

where \( \varepsilon'_f \) is the fatigue ductility coefficient and \( c \) is the fatigue ductility exponent. Their values are 0.732 and -0.747 respectively. It is obvious that this law describes sufficiently low cycle fatigue behaviour of the material. The great scatter in the elastic modulus can contribute to the great scatter of experimental data in the low amplitude domain. Figure 4b shows a fatigue life curve in the representation of the stress amplitude \( \sigma_a \) at half life vs. the number of cycles to fracture \( N_f \). Experimental data were fitted by the Basquin law

\[ \sigma_a = \sigma'_f \left( 2N_f \right)^b \]  \hspace{1cm} (3)
where $\sigma_f$ (1333 MPa) is the fatigue strength coefficient and $b$ (-0.140) is the fatigue strength exponent. Their values are in Table 2. It can be seen from Fig. 4b that the law describes the fatigue life satisfactorily.

3.3 Surface relief

Observation of surface relief was conducted on selected specimens with polished gauge length by SEM. Cyclic strain localisation was found in the vicinity of natural stress concentrators like shrinkage pores. A typical example is shown in Fig. 5a. The surface relief consists of short and wavy persistent slip markings (PSM) of the orientation which depends on the crystal orientation of the grain. Fatigue crack initiation along the markings is clearly visible in Fig. 5a.

3.5 Fracture surface

Selected specimens cycled until fatigue failure were used for fracture surface observation. Shrinkage pores that are the most frequent sites of the fatigue crack initiation are clearly visible in Fig. 5b. Those technological flaws were up to 0.7 mm in diameter. The first transgranular stage of the fatigue crack growth was predominantly found in the vicinity of shrinkage pores. Fields of striations (Figure 5b) were found in areas where the fatigue crack reached the second stage of the fatigue crack propagation. Secondary cracks were also observed in those areas. Similar features were already described on specimens cycled at different elevated temperatures. [2].

![Fig. 4](image1.png)

Fig. 4 a) Manson-Coffin fatigue life curve b) Basquin fatigue life curve

![Fig. 5](image2.png)

Fig. 5 SEM micrographs a) surface relief with persistent slip markings and fatigue cracks initiated along them b) fields of striation and shrinkage pore denoted with arrows
3.5 Dislocation structures

![Fig. 6 TEM micrographs](image)

Fig. 6 TEM micrographs a) stacking fault going through matrix and precipitate $\gamma'$ along $(\overline{1}11)$ plane b) a precipitate cut by two stacking faults and dislocation net and fine precipitates in the matrix.

Thin foils were cut from gauge length of a cycled specimen ($\varepsilon_a = 0.3\%$, $N_f = 170$) in the direction parallel with the stress axis. Typical feature of dislocation structure after cyclic loading is non-homogeneous distribution of dislocations. Cyclic plastic deformation concentrates both in the $\gamma$ channels and in the form of persistent slip bands (PSB). It was already documented that PSBs occur in superalloy In 792-5A after cyclic loading at elevated temperatures [3,4].

Figure 6a shows dislocations predominantly at the interface between matrix and precipitates. In the middle of Fig. 6a, a stacking fault is apparent going along the $(\overline{1}11)$ plane through the matrix and shearing precipitates $\gamma'$. Similar situation is also in Figure 6b where two stacking faults in the plane $(\overline{1}11)$ shear a precipitate $\gamma'$. The observed grain is oriented for multiple slip. PSBs were observed mostly in grains oriented for single slip [3,4]. Figure 6b also shows fine precipitates $\gamma'$ in the matrix. They are obstacles for the dislocation movement in the matrix. Thus, dislocations can be fixed and bent in their vicinity.

4. DISCUSSION

Cyclic hardening/softening curves, cyclic stress-strain curves and fatigue life curves were evaluated from the LCF tests of IN 792-5A. The LCF behaviour of the material is in agreement with the previous study [3]. However, a large scatter of experimental data manifest in the LCF parameters of IN 792-5A. Both large casting defects and comparatively large grain size contribute to the scatter.

Surface relief was investigated by SEM. PSMs were found short and wavy. Further research is needed to reveal their properties.

The fatigue crack initiation was found mostly in the vicinity of shrinkage pores both at the surface and in the bulk. This finding is in agreement with studies [2]. The first stage of the fatigue crack propagation is transgranular and parallel to the primary slip planes $(111)$. When the plastic zone at the crack tip is big enough, i.e. the crack is long enough, additional slip systems are activated. It can result
if the formation of striation fields on the fracture surface with possible occurrence of secondary cracks – see Fig. 5.

Shearing of precipitates γ´ by stacking faults (see Fig. 6) after cyclic loading and also after creep straining of superalloys is in accord with earlier studies [5-7]. Fine precipitates γ´ in the matrix (Fig. 6) can hinder the movement of dislocations and therefore they beneficially contribute to strength properties of the material.

5. CONCLUSIONS

Results of present study can be summarized as follows:

(i) High strain amplitude LCF tests are characterised by initial hardening followed by softening which is more pronounced at the end of fatigue life. Low strain amplitude test showed stable behaviour. Cyclic stress-strain curve can be approximated by power law.

(ii) Fatigue life experimental data can be approximated using the Basquin and the Manson-Coffin law.

(iii) The strain localisation into short and wavy persistent slip markings is documented. The fatigue cracks initiate along the markings later in the fatigue life.

(iv) Shrinkage pores prove to be significant stress concentrators. The fatigue cracks initiate predominantly in their vicinity. Dislocations are present both in γ channels and in γ´ precipitates. High dislocation density was found in the γ channels particularly at the γ/γ´ interface.

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LITERATURE


