FATIGUE AND DEFORMATION BEHAVIOR OF DIRECTIONALLY

SOLIDIFIED RENE 80

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Summary

The low cycle fatigue (LCF) behavior of directionally solidified (DS) Rene 80 having standard and Hf modified chemistries was studied over the temperature range 25-981°C. Specimens were cut from the ingot axis in the longitudinal, transverse and 45° orientations. When represented on the basis of plastic strain, the life decreased with increasing temperature and reached a minimum in the range 550-760°C, depending on the composition and orientation. In general, longitudinal specimens exhibited the longest lives with transverse and 45° specimens having the shortest lives. Hafnium additions produced a more benign carbide structure, increased the strength but did not improve the LCF life. The experimental results are interpreted in terms of factors such as γ' strength and stability, deformation mode, Young's modulus, Schmidt factor and environmental attack.

Introduction

Rene 80 is a cast Ni base superalloy that is used for fabricating first stage gas turbine blades in jet engines. It has been extensively studied in the conventionally cast form (1-3) and it has been shown that the low cycle fatigue (LCF) behavior can be understood by considering effects such as preferential oxidation along grain boundaries and γ' coarsening.

Earlier work (4, 5) demonstrated that directionally solidified (DS) columnar structures and single crystal (XL) product forms result in improved mechanical properties, chiefly due to the low modulus and small number of transverse grain boundaries. Their superior properties have made them attractive for applications in modern jet engines and Hf modified Rene 80 is being used in the B1-B bomber.

While the general superiority of these anisotropic materials has been established, detailed information on factors that affect fatigue life is relatively scarce. For these reasons a study was undertaken to document the effects of orientation, microstructure, deformation mode and temperature on the LCF behavior of DS Rene 80 having both standard and Hf modified chemistries.

Experimental Procedure

Materials, Heat Treatment and Specimen Fabrication

1. Composition. The composition of both the standard and Hf modified Rene 80 used in this study is shown in Table I. Previous work on DS Rene 80 (6) has shown that during casting a Chinese script morphology of Ti rich MC carbides forms in the middle section of a DS bar. This morphology increased the rate of crack growth during LCF and reduced ductility. In order to reduce (if not eliminate) this undesirable characteristic, Hf was added and the C level was reduced. Other reasons for adding Hf were to improve the castability and to increase the resistance to environmental attack. However, the effects of Hf on the volume fraction and physical properties of γ' have not been well documented in the open literature.

Table I Composition of Rene 80 (wt. %)

Element	C	Cr	Co	Mo	W	Ti	Al	B	Zr	Hf	Mg	Ni
Standard	0.16	13.8	9.2	4.0	4.0	4.9	3.04	0.015	0.02	0.05	0.07	Bal
Hf Modified	0.073	12.9	9.6	4.0	4.9	4.48	3.02	0.015	0.01	0.74	--	Bal

2. Solidification. The master heat was remelted in a directional solidification apparatus. Melting was carried out at a melt/solid interface temperature gradient of 55°C/cm and at a linear solidification rate of 25 cm/hr. The final ingots were 4.1 cm in diameter and 20-25 cm long.

3. Heat Treatment and Specimen Fabrication. In order to preserve material, gage section blanks 0.69 cm in diameter and 3.94-4.45 cm long were machined from the DS ingots at 0° (longitudinal), 90° (transverse) and at 45° to the growth axis. Details of the cutup are given in (6). The blanks were then brazed to IN 718 end pieces. All brazing and heat treatment was carried out in vacuum, the details of which are given below:

1. Solution DS blanks at 1204°C for 2 hrs and He quench to RT
2. Ni plate gage blanks with 0.0013 cm thick layer
3. Braze end blanks to gage section using alloy B-28 at 1204°C for 15 min. Cool to 1093°C in 6-10 minutes.
4. Hold at 1052°C for 4 hr. to simulate coating cycle
5. Age at 843°C for 16 hr. and cool to RT

In some cases the brazed ensemble was aged at 718°C for 8 hrs., cooled to 621°C at 56°C/hr., held at 621°C for 8 hrs. and furnace cooled. This last sequence was carried out to harden the end blanks. The heat treatment results in a duplex γ' structure.

After heat treating, the specimens were finish machined into standard button head LCF specimens and the gage section was low stress ground to a surface finish of RMS 8.

LCF Testing

LCF testing was carried out in accordance with the requirements of ASTM E-606 (7) over the temperature range 24°C-871°C. Heating was done using well-calibrated induction coils for each test temperature. Prior

to testing at elevated temperatures, the specimens were allowed to come to thermal equilibrium for 30 minutes at zero load. All testing was done with an MTS universal test machine in the strain control mode using a balanced tension/compression triangle waveform and a strain rate of 50%/min. In most of the tests small adjustments were made periodically to maintain an approximately constant plastic strain range. The specimen temperature did not vary by more than $\pm 3^{\circ}\text{C}$ over the gage length of the specimen. Frequent calibration checks were carried out to ensure that a high degree of temperature control was maintained. The load and strain ranges were continually monitored and hysteresis loops were periodically recorded. Further testing details and tabulated results are given in (6).

Metallography

Optical, scanning (SEM) and transmission (TEM) electron microscopy were used to characterize both the initial and as-tested microstructures. The misorientation between adjacent grains was determined using TEM diffraction patterns of a thin foil normal to the tensile axis. Using techniques described elsewhere (8), stereographic projections were constructed and the operations necessary to bring one grain into coincidence with another allowed the relative tilt, twist and rotations to be determined.

Results and Discussion

The structure of a typical DS ingot of standard composition was found to vary along the length of the bar. The major differences were a pronounced tendency to form Chinese script-like carbides near the middle of the ingot with much less at the bottom and somewhat less at the top. In addition there was an increase in the grain size and a decrease in the misorientation from [001] from bottom to top. In many instances only a few grains occupied the cross section of the longitudinal LCF specimens. Similar observations apply to the Hf modified alloy except the carbide structure was more benign over the length of the bar. In spite of these variations, there was not a large difference in life at the different locations for the plastic strain range used in this investigation (6). At higher strain ranges differences in life were found to depend somewhat on location (6) with the lowest lives for specimens taken from the center (script-like carbides). The carbide morphology for both standard and Hf modified Rene 80 is shown in Fig. 1 and it can be seen that Hf addition reduced the tendency to form script-like carbides. The dendrite and grain structure is seen in Fig. 2 perpendicular to the [001] growth axis. It was possible to use photos such as Fig. 2 to determine the relative rotation between adjacent grains. That and other data is given in Table II. The structure of the γ' is shown in Fig. 3.

Figure 1 - Longitudinal section showing the carbide morphology in (a) standard Rene 80 and (b) Hf modified Rene 80.

TABLE II

Microstructural Characterization
of selected DS Rene 80 Specimens

Specimen Number	Number of Grains in Cross Section	Interdendritic Spacing (mm)	Average Misorientation from Stress Axis	Relative Rotation about stress Axis	Relative Tilt	Relative Twist
LT1	2	.25	--	45°	--	--
LT2	3	.23	5°	45°-40°-3°	2°	3°
LT3	3	.21	6°	20°-10°-35°	12°	--
LT4	3	.27	7°	35°-35°-35°	8°	4°
LB1	3	.22	--	30°-15°-35°	--	--
LB2	3	.21	--	35°-35°-30°	--	--
LB3	7	.22	--	30°-35°-40°	4°	--
LM1	3	.22	3°	45°-20°-25°	2°	4°

Figure 2 - Transverse section of standard Rene 80 tested at 871°C and etched to reveal the dendrite structure.

Figure 3 - Dark field TEM micrographs showing the γ' structure which was similar for both compositions.

The LCF life as a function of temperature is shown in Fig. 4 while the stress range and maximum stress are also shown as a function of temperature in Fig. 5. The most striking feature of Fig. 4 is that regardless of the chemical composition or orientation, the life at 0.1% plastic strain reaches a minimum in the temperature range 550°C-760°C and increases rapidly thereafter. Similar behavior was observed for the LCF of conventionally cast Rene 80 (9). Ductility minima have been observed in the same temperature range for pure Ni and other Ni base superalloys (10). The ductility minima were assumed to be related to oxygen diffusion to boundaries and subsequent pinning which became less effective at higher temperatures as the boundaries became more mobile. Due to the grain structure of the present alloys, this explanation seems improbable. Instead such behavior may be related to the strength and stability of the γ' , changes in deformation mode, the nature of the carbides and environmental effects. These factors are discussed below.

Figure 4 - Cycles to failure as a function of temperature for standard and Hf modified Rene 80 tested at 0.1% plastic strain range.

Figure 5 - Maximum stress and stress range as a function of temperature for standard and Hf modified Rene 80 tested at 0.1% plastic strain range.

Fractographic studies were carried out for 45° and longitudinal specimens in the Hf modified alloys and for all orientations of the standard composition. Generally, crack initiation and/or early propagation was associated with carbides, especially for the standard composition. Up to 871°C, both chemistries exhibited strongly crystallographic features. However, in the range 425°C-650°C, Hf modified specimens were clearly more crystallographic in appearance. Typical SEM micrographs are shown in Fig. 6 and the crystallographic nature of the fracture surfaces is evident. At high magnifications a cellular network on the crystallographic faces was seen as shown in Fig. 7. This was observed for both chemistries but appeared to be more uniform for the Hf modified alloys and may be an indication that there are subtle differences in the plastic deformation processes.

For the 45° specimens, fracture frequently initiated at grain

Figure 6 - SEM fractographs of standard (left column) and Hf modified Rene 80 (right column). All specimens were longitudinal and tested at 25°C (a and b), 760°C (c and d) and 871°C (e and f) at 0.1%. Similar features were observed for the other orientations with more boundary cracking.

Figure 7 - Cellular network observed on "cleavage" faces such as these shown in Fig. 6.

Figure 8 - SEM micrograph showing carbide-initiated boundary cracking in a standard Rene 80 45° specimen tested at 760°C at a plastic strain range of 0.1%

boundaries, especially when carbides were present as illustrated in Fig. 8 where a large initiating carbide can be seen on the specimen gage surface. The differences between the standard and Hf modified Rene 80 may be explained in part using information given in Figs. 1 and 5 as well a simple fracture mechanics analogy. In Fig. 5 we saw that the strength of the Hf modified alloy is higher than that of the standard composition. This may be due to either an increased volume fraction of γ' , an increased hardness of γ' (due to higher antiphase boundary energy or increased mismatch) or both. The concept of an intrinsic flaw being formed on the first loading cycle may now be introduced. It may further be assumed that the flaw size is the maximum carbide plate size which is $2C_s$ and $2C_H$ for the standard and Hf modified versions respectively. The stresses for the strain controlled cycle are σ_s and σ_H respectively. For earlier crack initiation in Hf modified Rene 80 the following conditions are required:

$$\text{If } N_i^H < N_i^S \quad (1)$$

then

$$\sigma_H \sqrt{C_H} > \sigma_s \sqrt{C_s} \quad (2)$$

$$\left(\frac{\sigma_H}{\sigma_s}\right)^2 > \frac{C_s}{C_H} \quad (3)$$

For the material tested, C_s/C_H was approximately 2, as seen in Fig. 1. Using data for the longitudinal orientations we see that the lives were approximately equal at room temperature and 760°C. However, at 871°C, the life of the standard composition is somewhat longer. The stress ratio computed at this temperature was in the range 2-2.75 and the inequality expressed in eq. 3 is satisfied. It thus appears that the higher stress associated with the Hf modified version offsets the advantage of a more benign carbide morphology and accounts, at least in part, for its lower life at 871°C.

The deformation behavior was also studied and results are shown in Figs. 9-11. There were no obvious differences in dislocation morphology for either the standard or Hf modified alloys and all TEM photomicrographs are for the standard composition. At room temperature, well-defined slip bands were formed, regardless of orientation as shown in Fig. 9. The high γ' volume fraction, the occurrence of dislocation pairs and the well-defined slip bands all suggest that precipitate shearing was the operative deformation mechanism at this temperature. For the longitudinal orientation, multiple slip bands were observed as expected since the Schmidt factor on 4 slip planes is the same in the ideal orientation. Even when [001] and the tensile axis do not coincide, hardening on a given system will cause other {111} planes to become active. On the other hand, for the specimens cut at 45° to the ingot axis, the tensile axis relative to any given columnar grain will be on arc AB of the standard triangle shown in Fig. 12 and slip on a single system is generally favored. In agreement with this, a single set of slip bands is seen for the 45° specimen shown in Fig. 9b. Around 550° - 760°C, the minimum fatigue life was observed and it is known that the γ' reaches a maximum hardness at this temperature (11). The

Figure 9 - TEM bright field micrographs for standard Rene 80 tested at 0.1% plastic strain range at 25 C. In (a) multiple slip bands are shown for a longitudinal specimen using $g=[020]$ while in (b) multibeam diffracting conditions show a single set of slip bands for a 45° specimen. Note the paired dislocations in the lower left corner of (a) indicative of precipitate shearing.

Figure 10 - Same as Fig. 9 except tested at 760°C. In (a) a longitudinal specimen exhibits linear segments, dislocation pairs and loops around small γ' . In (b) a 45° specimen is shown and similar features can be seen along with dislocations stored on the faces of the large γ' particles.

Figure 11 - Same as Fig. 9 except tested at 871°C. In both longitudinal (a) and 45° (b) orientations, numerous examples of loops and interfacial dislocations can be seen. The γ' structure has also coarsened somewhat resulting in much less fine γ' . For both micrographs $g = [002]$.

strength of the alloys tested remained relatively constant up to 760°C. Typical deformation substructures are shown in Fig. 10 for 760°C. In comparison with Fig. 9, the dislocation substructure is much more diffuse. For both orientations, interfacial dislocations are observed on the large γ' faces. These are probably glide dislocations which were unable to penetrate the large γ' particles. Numerous dislocation loops (about the same size as the small γ' particles), dipoles and tangles can also be seen in the channels between the large γ' precipitates along with some indication of pairing. These observations, taken as a whole, show that along with hardening of the γ' , thermally activated deformation was increasingly important at 760°C. It is also clear that the γ' precipitate structure remained stable during testing. At 871°C, both the dislocation substructure and γ' precipitate morphology were strikingly different as seen in Fig. 11. There is no evidence of slip band formation nor can the linear segments observed in Fig. 10 be seen. Instead, the dislocation substructure is very diffuse and there are numerous loops and tangles in the channels between the large γ' particles. Furthermore, there is virtually no evidence of dislocations penetrating the large γ' particles and the γ' has coarsened leaving relatively few small γ' particles in the interprecipitate channels. These observations suggest that the deformation process is predominantly thermally activated and along with the γ' coarsening are completely consistent with the drop-off in stress seen in Fig. 5. The drop-off in stress is also consistent with the increase in life (when represented on the basis of plastic strain) seen in Fig. 4. At these higher temperatures it is probable that the life is determined by a trade-off between a decreased stress range (beneficial) and an increase in environmental degradation, similar to what occurs in conventionally cast materials (1, 12).

Misorientation effects can also play a significant role in these alloys. For example, two different grains in the same ingot may have slightly different growth directions and there will be a relative misorientation. Another possibility is that an over-all misorientation is introduced during various steps (i.e. sampling, brazing, final machining) of the fabrication process. Data for various types of misorientation are shown in Table II. Even when there is no difference in tilt, the relative rotation between adjacent grains is unrestricted and may be as high as 45°. Grain misorientations can be inferred from Fig. 13 which shows etched slip bands on a plane perpendicular to the growth direction. If all grains had [001] normal to the plane of polish the slip traces on the various {111} planes would appear as perpendicular lines, as they do for the large grain in the lower right but not for the other grains.

Quantities such as Young's modulus and Schmidt factor are known to play a significant role in the fatigue behavior of DS alloys (13). Differences in Young's modulus were seen and for a given specimen type are a reflection of misorientation effects. For longitudinal specimens, Fig. 14 shows that a misorientation of 10° from the stress axis can increase the modulus by about 10%. Similarly, as shown in Fig. 12, the Schmidt factor can increase by approximately 6%. For these relatively small changes, the LCF life should not vary appreciably from specimen to specimen for the longitudinal orientation and the results tend to reflect that fact. For the 45° specimens, the situation is quite different. As mentioned previously the position of the tensile axis may

Figure 12 - Contours of constant Schmidt factor for $\{111\} \langle 110 \rangle$ slip.

Figure 14 - Variation in modulus of pure Ni as a function of orientation.

Figure 13 - Optical micrograph of longitudinal specimen sectioned normal to $[001]$ growth axis. Specimen was tested at 25°C at 0.1% plastic strain range. Note the slip bands and grain boundaries.

lie anywhere on the arc AB shown in Fig. 12. Along with misorientation effects discussed previously, this means that one grain could have an orientation near $[1\bar{3}3]$ while its neighbor had an orientation near $[1\bar{1}1]$. A rotation of 20° and a tilt of 10° would bring the two crystals into correspondence. For the $[1\bar{3}3]$ grain, the Schmidt factor is about 60% higher than for the $[1\bar{1}1]$ grain. Slip tends to occur easily on one plane for the $[1\bar{3}3]$ grain while it doesn't occur at all for the $[1\bar{1}1]$ grain. This situation should lead to a stress concentration at the boundary, and a relatively low life compared to other possibilities. Boundary initiation and decohesion were in fact frequently seen for 45° specimens but not for the longitudinal specimens. Similar inhomogeneity of plastic deformation was observed by other investigators (14) for polycrystalline alloys. Of 63 grains studied, 50% of the deformed grains were near $\langle 110 \rangle$ while no grains near $\langle 111 \rangle$ were deformed even for plastic strains as high as 0.2%.

Summary and Conclusions

1. The LCF life of DS Rene 80 depends on orientation with the life generally increasing progressively from transverse to 45° to longitudinal.
2. Regardless of orientation, a minimum in the plastic strain vs. life was observed in the range $550 - 760^{\circ}\text{C}$. This minimum can be attributed, in part, to a combination of high stress and environmental damage.

3. Hafnium additions resulted in both a more benign carbide morphology as well as higher maximum hysteretic stresses. The high stresses (all other factors being equal) appear to offset the beneficial carbide morphology and Hf additions did not result in longer LCF lives for the materials used in this investigation.
4. Both Young's modulus and Schmidt factor show more variation for 45° specimens compared to longitudinal ones for equivalent misorientations. Since there is complete rotational freedom between adjacent grains in DS alloys, this means that significant differences in LCF life and other mechanical properties will occur for even the most precise sampling and testing procedures.

Acknowledgments

The authors wish to express their appreciation to the Air Force Office of Scientific Research for financial support of this project. (Grant No. AFOSR-80-0065, Dr. A. R. Rosenstein, Program Manager). They would also like to thank Mr. C. Wukusick of the General Electric Company, Evendale, Ohio for supplying the material and for many helpful discussions.

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