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Micromechanics of Cyclic Deformation in SSME Turbopump Blade Materials

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INTRODUCTION

Current and candidate SSME turbopump blade materials are anisotropic, both in their elastic and plastic response. The major objective of this study is to characterize the plastic deformation behavior of a typical single crystal nickel-base superalloy, PWA 1480, and to use this information to help guide the development of anisotropic constitutive models.

In order to predict structural response in the blade, and therefore in order to predict fatigue lives in the structure, a constitutive model is necessary. For ordinary isotropic materials, viscoplastic constitutive models are fairly advanced. However, due to the plastic anisotropy of the single crystal superalloys such as PWA 1480, constitutive models for these materials are still in the early stages of development. The two models which currently have the most promise are both based on the crystal plasticity approach. In this framework, the plastic shear strains on each individual slip system are computed from a micromechanical model, and then the strains are added to calculate the total strain in the material. Anisotropy is taken into account from a fundamental metallurgical point of view, and incorporated into the crystal plasticity formulation. Unfortunately, the current understanding of deformation in these alloys (from a metallurgical point of view) is not very complete. The current study is an attempt to extend this understanding.

To date, monotonic and cyclic deformation studies have been conducted on Mar-M 246, and <001> oriented single crystals of PWA 1480 over the temperature range of blade operation. Tests on <123> crystals, which are necessary to characterize the anisotropy, are being conducted presently.

THE CONSTITUTIVE MODELS

For ordinary isotropic materials undergoing deformation in the viscoplastic regime, the currently accepted models do an adequate job of correllating and sometimes predicting constitutive response under high temperature loading. These models are unified, in the sense that they treat "creep" and "plasticity" in the same analytical expression. This is very desirable for single crystal superalloys, because both "creep" and "plasticity" are the result of the same mechanism: thermally activated dislocation motion. The flow rule is of the following general form [1]:

$$\varepsilon_{ij} = \psi \{ (s_{ij} - \Omega_{ij}) / K \} \tag{1}$$

where ε_{ij} is the plastic strain rate, s_{ij} is the deviatoric stress, Ω_{ij} is a state variable usually referred to as the "back stress", which handles kinematic hardening, and K is another state variable referred to as the "drag stress", which handles isotropic hardening. The functional form of the flow rule is model-dependent, but is usually either an exponential function or a power-law function. Both state variables evolve with deformation.

For anisotropic materials, this general framework must be modified to account for the anisotropy. In particular, single crystal nickel base superalloys have an orientation-dependent yield strength which cannot be explained by a maximum shear stress (Schmid's Law) model. Many investigators have postulated that this anisotropy is due to the onset of cube plane slip within the strengthening γ' precipitates [2]. Since the anisotropy is due to variations in the type of active slip plane, the crystal plasticity approach was a natural choice. Several models of this type are currently being developed [3,4]. Both models assign a unified viscoplastic flow rule for slip on octahedral planes (the usual mode in FCC crystals), and a unified viscoplastic flow rule for slip on cube planes. The total plastic strain rate is calculated to be a sum of the two:

$$\varepsilon = \varepsilon_{octahedral} + \varepsilon_{cube} \tag{2}$$

Of course, the functional forms of the ψ's are orientation dependent, and the state variable for cube and octahedral slip can evolve independent of one another. This approach has been successful in correlating anisotropic behavior under limited testing for several alloys [3,4].

DEFORMATION BEHAVIOR

Since PWA 1480 is two-phase, and since the postulated cube slip occurs only in the strengthening γ' phase, the first question which needs to be addressed is how mobile dislocations interact with the strengthening precipitates. In two-phase systems with very large particles and high volume fractions (like PWA 1480), there are two possible mechanisms of deformation. As shown in Figure 1(a), the first mechanism is a classical shearing operation. Gliding dislocations under a stress high enough to penetrate the precipitate can continue to glide, thus shearing the particle. The second possible mechanism is illustrated in Figure 1(b). At high temperatures where diffusion is relatively easy, a gliding dislocation can by-pass the precipitate by a combination of climb and glide. In this case, the dislocation never enters the particle. Obviously, the behavior of the alloy under these two mechanisms will be very different.

As documented in earlier reports from this research [5,6], the deformation mechanism which is observed at yield in <001> oriented single crystals is a strong function of temperature and strain rate. At low temperatures (below 760^{o}C), the γ' is sheared by pairs of a/2<110> dislocations on {111} planes. Typical deformation microstructures in this temperature regime are shown in Figure 3(a), in which planar slip bands are seen to be shearing the precipitates. At high temperatures (above 815 to 927^{o}C depending on strain rate), the γ' is by-passed by the mechanism described above. Typical structures in this regime are shown in Figure 3(b), in which all dislocations are seen to be in the matrix and interfacial regions. This is indicative that no shearing has taken place, and further analysis [5] confirmed this conclusion. An analysis of current micromechanical models of both the shearing and by-pass mechanisms is contained in Reference [5].

In addition to the monotonic experiments described above, cyclic deformation tests have been conducted on <001> oriented single crystals over the same temperature range. The tests were strain controlled, fully reversed low cycle fatigue tests, and the plastic strain levels varied with test condition between 0.05% to 0.5%. In these tests, the exact same deformation mechanisms were observed as a function of tmeperature and strain rate as were observed in the monotonic experiments. At low temperatures, γ' shearing was observed, while at high temperatures γ' by-pass was observed. This is illustrated by Figure 4.

The anisotropy is currently being studied. However, data from another study [7] seems to indicate that the trends observed here will be general. As shown in Figure 5 (from Reference [7]), the yield strength is very anisotropic at temperatures up to and including 760^{o}C, while the anisotropy becomes much less severe at higher temperatures. Since γ' by-pass should be essentially isotropic, this indicates that by-pass probably is occurring regardless of orientation at higher temperatures.

IMPLICATIONS

Although cube slip has not been verified yet, the occurrence of γ' shearing at low temperatures indicates that cube slip is a possible source of anisotropy. Experiments which are currently being conducted should help clarify this shortly. If cube slip is observed in these tests, then the physical foundations of the constitutive models will be verified. In addition to identifying the type of deformation mechanism, these tests should lead to advanced micromechanical models which will be useful in the constitutive model foundation.

At high temperatures, however, no shearing occurs. Since by-pass appears to be isotropic, the anisotropic models need not be used. Again, future experiments will clarify these observations.

ACKNOWLEDGEMENTS

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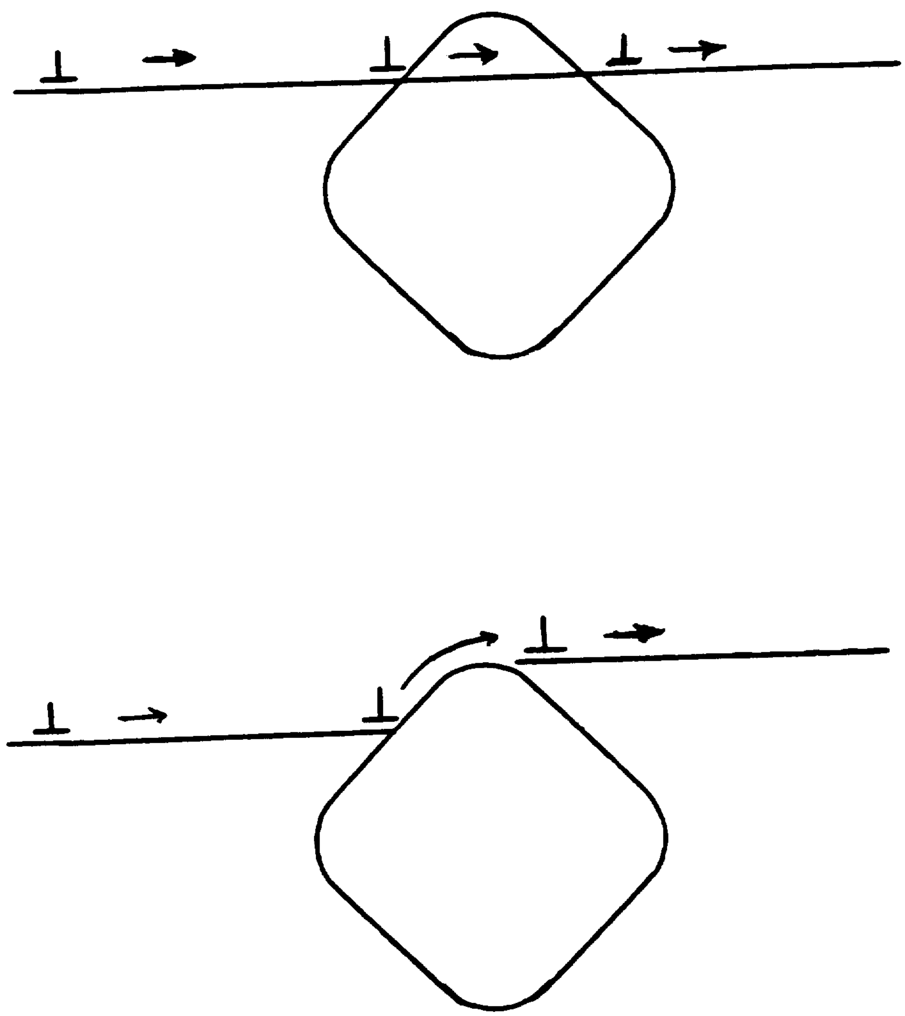

Fig. 1 Possible deformation mechanisms in PWA 1480. (a) γ' shearing. (b) climb-assisted γ' by-pass.

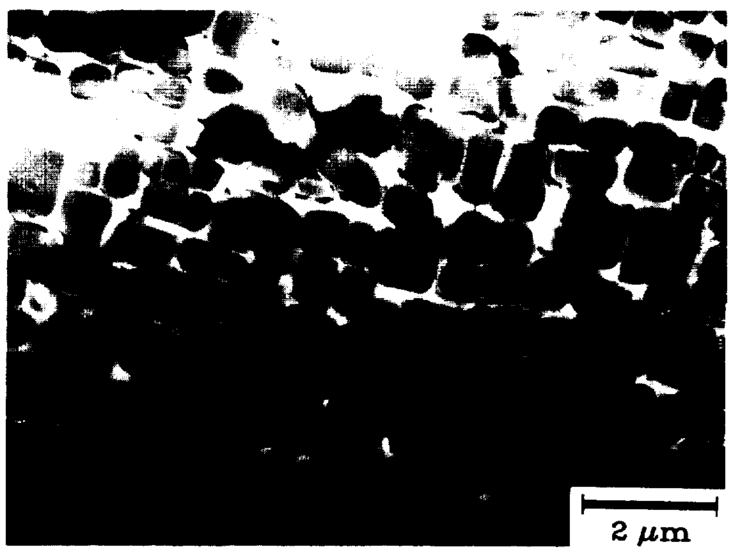

Fig. 2. As-received microstructure showing two-phase γ/γ' morphology.

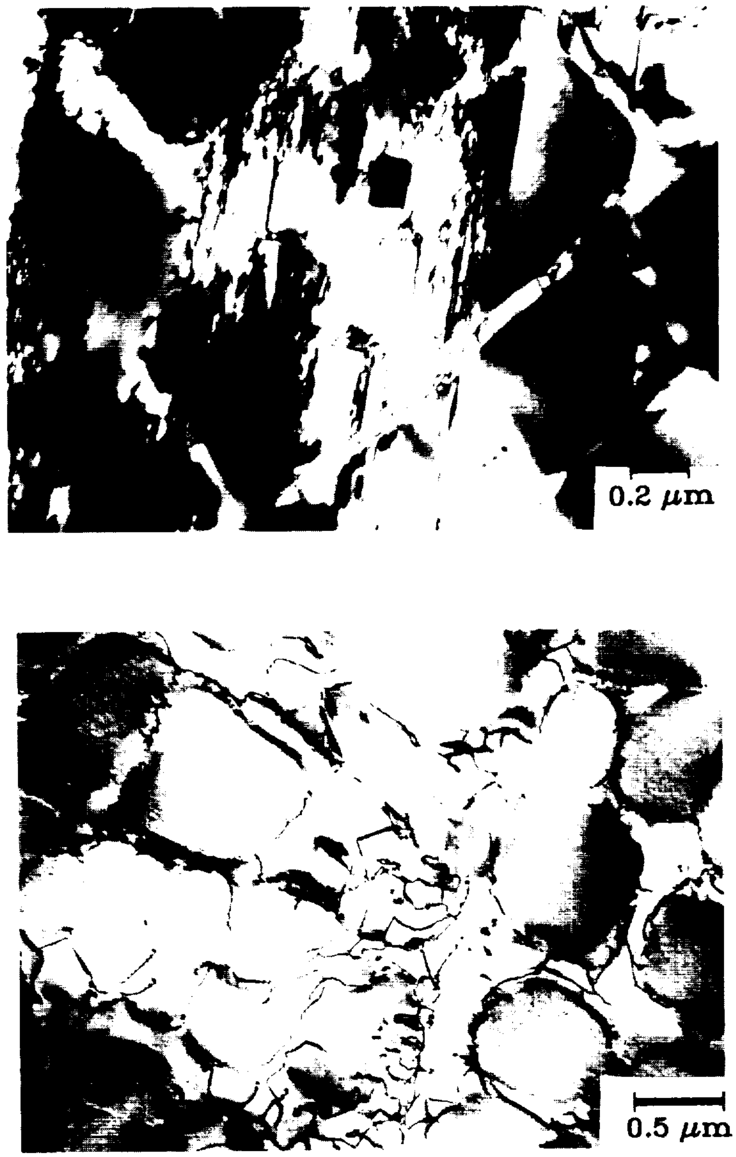

Fig. 3. Deformation structures at yield.
(a) Typical low temperature γ' shearing 705°C, 0.50%/min. (b) Typical high temperature γ' by-pass, 982°C, 0.5%/min.

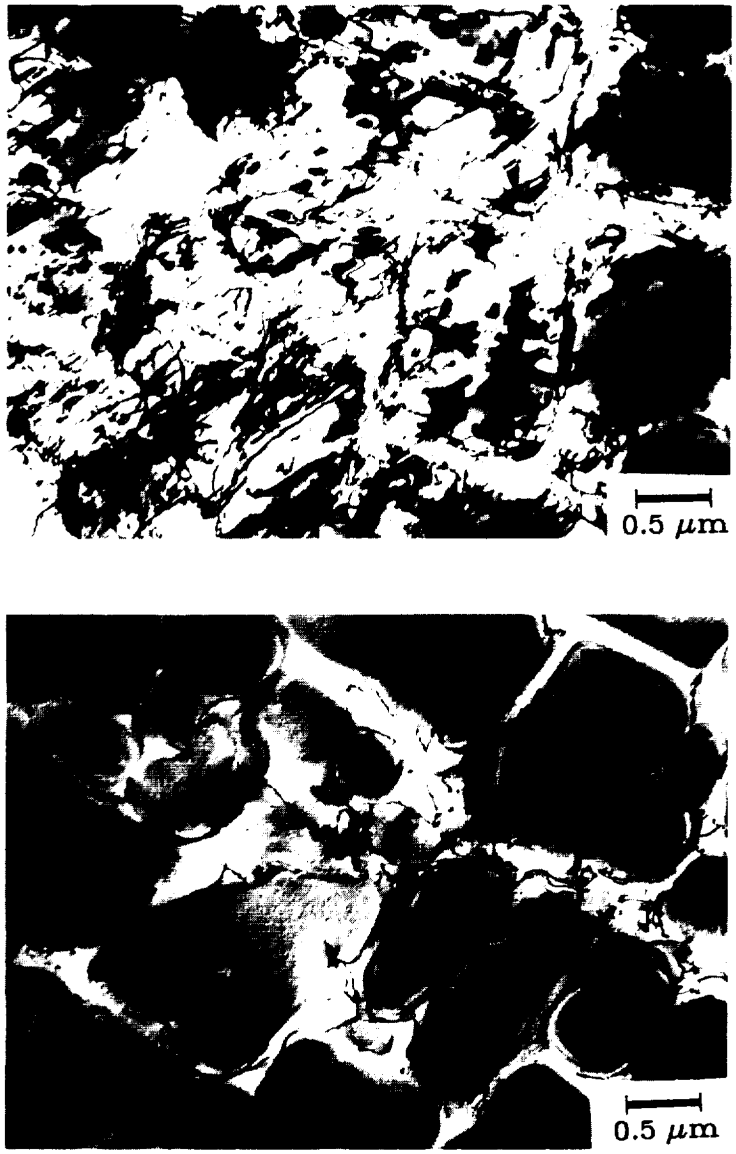

Fig. 4. Deformation structures after low cycle fatigue. (a) shearing at low temperatures, 20°C, 50%/min, $\Delta\varepsilon_p$ = 0.1% (b) By-pass at high temperatures, 915°C, 50%/min, $\Delta\varepsilon_p$ = 0.2%.

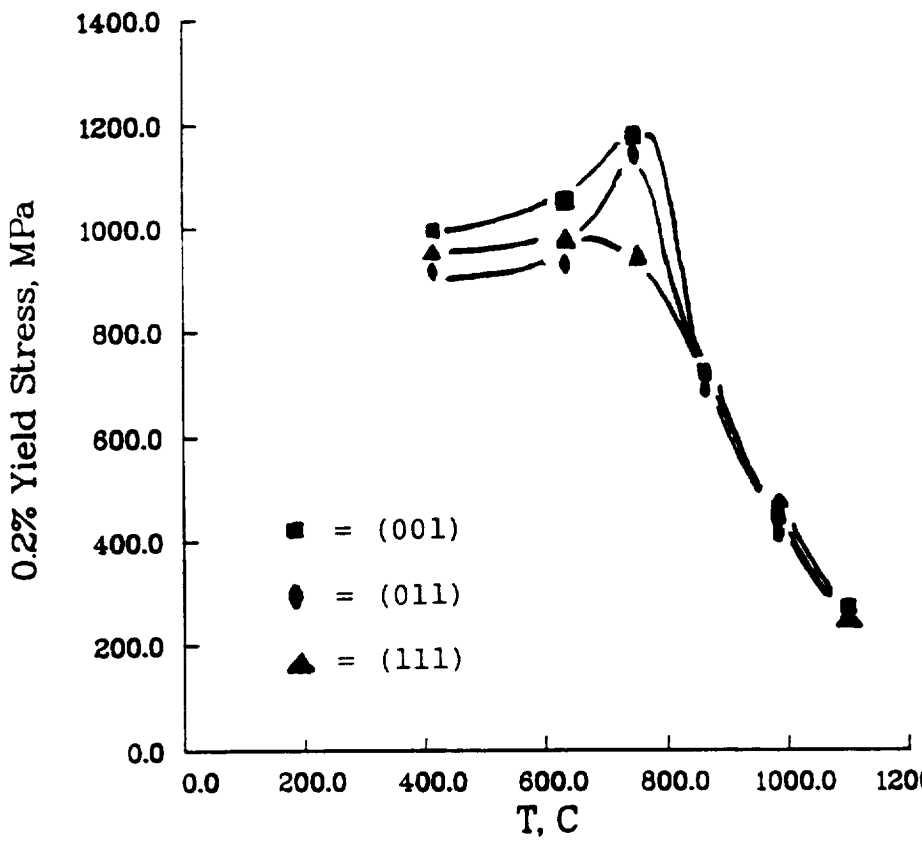

Fig. 5. Yield strength of PWA 1480 vs. Temperature for several orientations. Note diminishing anisotropy at high temperatures. (Reference 7)