FATIGUE DAMAGE BEHAVIOUR OF A COATED Ni-BASE SINGLE CRYSTAL SUPERALLOY

F. Gallerneau*, D. Nouailhas* and J.-L. Chaboche*

Fatigue creep tests have been performed on AM1 single crystal superalloy specimens, grown in <001> orientation and C_1A coated, at many temperatures in air and under various loading conditions. At high temperature, time dependent phenomena, such as creep or oxidation, have a great influence on the fatigue life. On the other hand, comparisons with bare specimens show that coating's effect on the fatigue life depends on the temperature. A phenomenological fatigue-creep-oxidation interaction model is proposed. Coating effect on the superalloy fatigue strength is taken into account through two distinct damage processes: a microinitiation phase and a micropropagation one. The interactions between the variables are supported by the observations of the damage mechanisms. Model predictions are shown for a large set of experimental data including thermomechanical fatigue tests.

INTRODUCTION

The design of aircraft gas turbine with improved performances has led to the development of blades made from directionnaly solidified single crystals. The mechanical behaviour of Ni-Base single crystal superalloys is today well known. The life time prediction of turbine blades requires to develop fatigue prediction models working at high temperature. The creep fatigue tests results, together with the main damage mechanisms, are presented in the first section. All the specimens are coated with C_1A chromizing-aluminizing thermal treatement, in order to be as representative as possible of the real blades. The next section describes a phenomenological fatigue-creep-oxidation interaction model and its application to the prediction of isothermal and anisothermal fatigue tests.

EXPERIMENTAL PROCEDURE

Fatigue tests results. Isothermal fatigue tests at high frequency (50 Hz) were conducted at 950°C and 1100°C in air and at zero mean stress. At this frequency, time effects are supposed to be negligible, and those tests are consequently near the pure fatigue reference. Tests involving time effects have also been

*ONERA, Division OR/ce, B.P. 72, 92322 Châtillon Cédex (France)

performed: fatigue tests at lower frequencies (5 and 0.5 Hz) and tests with hold time of 90s (rising and going down in load in 10s). For the same stress amplitude, the fatigue life decreases when the loading frequency decreases. This time dependent effect becomes predominant at 1100°C for the lowest stress ranges. Experimental Woehler's curves obtained at this temperature are plotted in figure 1. Comparisons with bare specimens show that C_1A coating has a favourable effect at both temperatures and particularly at 1100°C and 50 Hz as it can be seen on the same figure. On the contrary, at 600°C, similar comparisons at 50 Hz have shown that the coating has then an unfavourable effect on the fatigue life of the superalloy (Gallerneau et al (1)). This is due to the fact that C_1A coating is brittle at low temperature and exhibits a ductile/brittle transition temperature situated between 700°C and 750°C (Chataigner (2)).

Microscopic observations. Breaking surfaces and longitudinal sections of specimens have been systematically observed by using optical and scanning electron microscopes. Damage mechanisms are complex and depend on many parameters: temperature, loading frequency and stress range. In spite of all, at 1100°C, where the time effects are the most important, the damage mechanisms are easiest to analyse:
- At 50 Hz, in pure fatigue conditions, multiple cracks are initiated in surface of the specimens due to C_1A coating damage. Those cracks then progress into the alloy perpendicularly of the loading axis. When their size reaches about 75 μm, a macroscopic crack is formed which leads to the failure of the specimen.
- At 0.5 Hz, this surface damage process is accelerated by oxidation effects. Interrupted tests have shown that the initiation phase of the cracks is then reduced.
- For the tests with hold time, the damage mechanisms completely differ. As a matter of fact, the failure occurs by the formation, in the necking area, of large cracks initiated on internal micropores (due to the dendritic growth of the alloy). This damage mechanism is then related to creep phenomenon.

At 600°C, coating fragility induces deep cracks in the superalloy from the first fatigue cycles. The initiation period is then strongly reduced. The failure of the coated specimen occurs much earlier than the bare specimen for wich the cracks are initiated on micropores.

DAMAGE MODELLING

The continuous damage prediction model is a stress based model which assumes that the progressive deterioration processes can be described by scalar damage variables. A general multiaxial formulation for anisotropic materials has been detailed elsewhere (Gallerneau et al (3)). Since only the <001> orientation has been experimentally studied under uniaxial loading, we limit here the application of the model to uniaxial thermomechanical loading, for which the temperature T(t) can strongly vary during the stress cycle σ. Using a reduced stress $S=\sigma/\sigma_{uI}(T)$ or $S=\sigma/\sigma_{uP}(T)$ as parameter describing the thermomechanical fatigue cycle, where σ_{uI} and σ_{uP} are respectively the ultimate stresses in microinitiation and in micropropagation which are temperature dependent, the fatigue laws can then be supposed temperature independent. Only time dependent and thermically activited phenomena, such as creep and oxidation, are described by temperature dependent laws.

Fatigue-creep-oxidation interaction model. To take into account the baneful effect of the coating on the fatigue resistance of the superalloy at low temperature, the model differentiates two distinct damage processes: a microinitiation phase and a micropropagation phase respectively described by D_I and D_P variables. Interaction effects are introduced between fatigue and oxidation damage process (traduced by D_{ox} variable) during the microinitiation phase only. The variable D_c is related to creep damage, that can be developed during the microinitiation phase, but that interacts with the fatigue damage during the micropropagation phase.

Fatigue damage law for crack initiation is expressed by the relation:

$$dD_I = \left(\frac{1}{C}\right)\left\langle\frac{\Delta S/2 - S_{lI}(\bar{S})\,(1-D_{ox})}{(1-D_{ox}) - S_{Max}}\right\rangle^b dN \qquad (1)$$

where the oxidation damage is introduced through the effective stress concept. ΔS is the amplitude, S_{Max} and $\bar{S}$ the maximum and mean values of S during the cycle. S_{lI} is the fatigue limit in initiation. The oxidation law is given by a second differential equation which traduces the loading influence (waveform and modulus) on the oxidation kinetics and in which a delayed stress X_{ox} has been introduced:

$$dD_{ox} = \frac{1}{2}D_{ox}^{-1}\left(\frac{k^*}{e_o}\right)^2\left[1 + \frac{\langle X_{ox}(S) - S_{lox}(\bar{S})\rangle}{B}\right]^{2m} dt \qquad (2)$$

where S_{lox} is the stress above which the oxidation kinetics is accelerated, and k* the oxidation velocity, obeying to an Arrhenius's law, and calculated on the temperature cycle by:

$$(k^*)^2 = \int_0^{\Delta t}\left[k_o \exp\left(-\frac{Q}{RT(t)}\right)\right]^2 dt \qquad (3)$$

The micropropagation is described by the differential equation:

$$dD_P = F_1\left(\frac{\Delta S}{2}, \bar{S}, S_{Max}, D_P\right)dN$$

$$dD_P = \left[1-(1-D_P)^{\beta+1}\right]^{1-a}\left\langle\frac{\Delta S/2 - S_{lP}(\bar{S})}{1-S_{Max}}\right\rangle\left(\frac{\Delta S/2}{M(\bar{S})\,(1-D_P)}\right)^{\beta} dN \qquad (4)$$

where S_{lP} is the fatigue limit in propagation. Creep law generalizes Rabotnov's law in which tension and compression loadings are differentiated and a delayed stress X_c has also been introduced:

$$dD_c = F_2(X_c(\sigma), D_c)\,dt = \left(\frac{X_c(\sigma)}{A(T)}\right)^{r(T)}[1-D_c]^{-k(T)}\,dt \qquad (5)$$

The limit stresses S_{lI}, S_{lp}, S_{lox} and the parameter M depend on the mean value $\bar{S}$ by the relations $S_{li} = S_{lio}\left(1 - h_i\ \bar{S}\right)$ and $M = M_o\left(1 - h_M\ \bar{S}\right)$. The delayed stresses X_c and X_{ox} are introduced to annul the time dependent effects on fatigue damage process at high loading frequency:

$$\left(\frac{dX_i}{dt}\right) = \frac{(\sigma(t) - X_i)}{\tau_i} \qquad i = ox, c \tag{6}$$

where τ_{ox} and τ_c respectively define the frequencies above which oxidation and creep effects are neglected.

<u>Computation of the life time.</u> The number of cycles at microinitiation N_I is given when D_I reaches the value 1. The creep-fatigue interaction then starts. At each cycle, the total damage D_t is defined by:

$$dD_t = H(D_I - 1) F_1\left(\frac{\Delta S}{2}, \bar{S}, S_{Max}, D_t\right)dN + F_2(X_c(\sigma), D_t)\,dt \tag{7}$$

where $H(D_I\text{-}1)=0$ if $D_I<1$ and $H(D_I\text{-}1)=1$ if $D_I>=1$. The number of cycles at micropropagation N_P is reached when $D_t=1$ and the fatigue life is then obtained by the sommation $N_F=N_I+N_P$. For the creep law, the damage is computed on the cycle by integrating the equation (5) at every time step with the corresponding mean values of A, r and k which are temperature dependent.

<u>Model identification.</u> The proposed writing makes the model attractive for its identification on a large temperature domain. As a matter of fact, microinitiation and micropropagation laws evolve with temperature through the variations of the ultimate stresses $\sigma_{uI}(T)$ and $\sigma_{uP}(T)$. To take into account the brittleness of C_1A coating at low temperature, the ultimate stress in microinitiation $\sigma_{uI}(T)$ is taken, below 750°C, lower than the rupture stress $\sigma_{uP}(T)$ of the superalloy. Pure fatigue tests results (at 50 Hz) performed at one temperature are then sufficient to identify all the parameters of microinitiation and micropropagation laws. Fatigue creep tests results conducted at lower frequencies (5 and o.5 Hz) at the same or another temperature allow us to identify the oxidation law. Pure creep tests are nevertheless required to identify the damage creep law at several temperatures.

<u>Model prediction.</u> On figure 1 are plotted the calculated Woehler's curves in isothermal condition at 1100°C for R_σ=-1. When the loading frequency decreases, time effects increase, in agreement with the tests results. The gradient of the curves increase by the way from pure fatigue (50 Hz tests) to tests with hold time, up to reach the gradient characteristic of pure creep tests. Several thermomechanical fatigue (TMF) cycles have been studied on AM1 coated C_1A (J.C. Lautridou et al (4)) to simulate real loadings on blades. The variations of the temperature T(t) and of the mechanical strain $\varepsilon(t)$ are given figure 2, for two cycles of period equal to 180s. For each test, the stress $\sigma(t)$ has been recorded at the stabilized cycle. For a

same strain range, W cycle is much damaging than S cycle. As a matter of fact, for W cycle, the maximum strain is applied at 600°C where C_1A coating is brittle. As it is seen on figure 3, the model predictions are fairly good, as well as for many isothermal fatigue tests at several temperatures and loading ratios, than for TMF tests for which the influence of the coating is well taken into account.

CONCLUSION

Life durations comparisons with bare specimens and microscopic observations show that C_1A coating has an important influence on the fatigue strength of AM1 superalloy. A phenomenological fatigue-creep-interaction model is proposed for an application to the prediction of anisothermal tests. The brittleness of the coating below 750°C is taken into account by differentiating a microinitiation phase from a micropropagation phase in the fatigue processes. Predictions of the model are fairly good for the set of experimental data including TMF tests results. Only <001> crystallographic orientation has been investigated. The complete identification of the general multiaxial model (Gallerneau et al (3)) requires the realization of tests following different crystallographic orientations.

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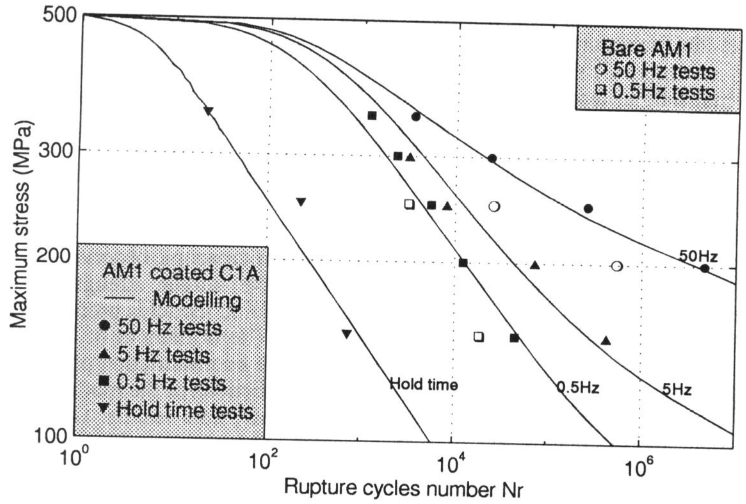

Fig. 1. Woehler curves at 1100°C for R_σ=-1.

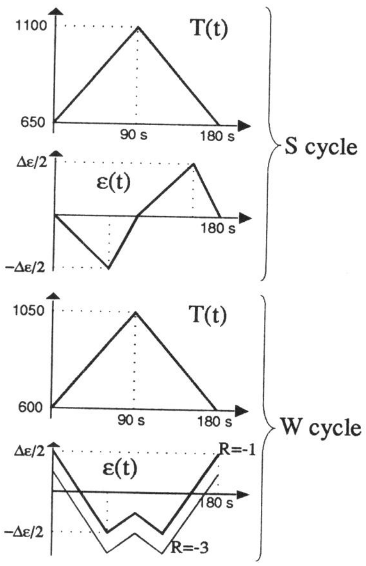

Fig. 2. Description of the thermo-
-mechanical fatigue cycles.

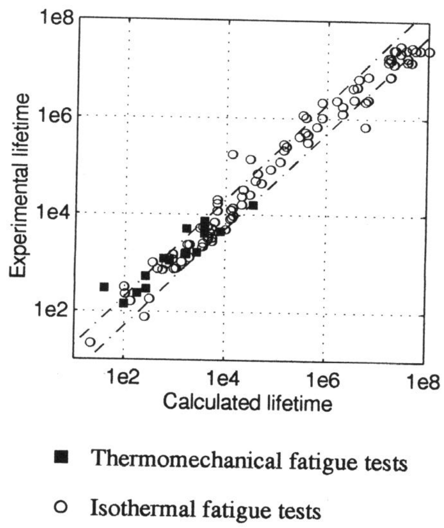

Fig. 3. Comparison between calculated and experimental lifetime.