Simulation of Creep and Life Prediction of Life Time by Thermal Fatigue Using a Viscoplastic Model

A. Rosselet

A viscoplastic model is presented which permits the description of the mechanical behaviour of a material including microstructural changes. This model is based on a constitutive equation system with a number of internal state variables. This system can be used in a user subroutine for standard Finite-Element programs. To describe the evolution of damage and to predict the life, additional internal state variables have been introduced. Up to now the damage model has been used to describe the low cycle fatigue behaviour of martensitic steel E911 in a COST 501 /III WP 11 project. First results are presented which permit to expect that the method allows a simple and reliable life time prediction under creep-fatigue interaction conditions.

1 Introduction

Non-stationary temperature fields generate thermal stresses which can be higher than the yield strength of the material and produce local plastic deformations. If repeated a number of times during the life time of the component, these deformation cycles lead to cracking in the stressed zones. To predict the life time of such a component, it is necessary to know on one hand the service conditions and on the other hand the material behaviour. The better both are known, the more confident the life time prediction will be. The argument of the service conditions being often not known exactly, and therefore an exact knowledge of the material behaviour being useless, is not true. With an exact knowledge of the material behaviour it is possible at least to optimize the service conditions or to specify service conditions that maximize the life of the component. For this reason the development of more exact material models is still an important task.

During thermal fatigue conditions the component is submitted to large temperature differences. Generally the temperature range extends from a minimum temperature at which the material hardens with the plastic deformation, up to a maximum temperature at which creep and softening are predominant. The standard Finite-Element codes used in industry already contain material models. At low temperature they are rate-independent plastic models and at high temperature rate-dependent creep models. The majority of these models are very primitive and not suitable for anisothermal calculations, particularly if the temperature range considered obliges one to pass from a low temperature to a high temperature model. The problem lies in the transition between the two. One approach is to use a single viscoplastic model for the whole temperature range considered. Physically this can be justified: Plastic deformation at low temperature is not instantaneous but rate-dependent. Figure 1 shows as an example the result of a test at room temperature with austenitic steel AISI 316. The test sequence was the following: loading up to 120 MPa (this stress is still in the elastic range) and maintaining it for a few minutes and then a rapid loading up to 240 MPa. This stress is beyond the yield strength. It can be clearly seen, in Figure 1, that the plastic strain does not follow instantaneously.

In standard Finite-Element codes it is possible to implement a user subroutine with an appropriate material model. For this reason a viscoplastic model has been developed which permits to write such a subroutine. It consists of a constitutive equation system with a number of internal state variables. This model takes into acount the effect of microstructural changes on the behaviour of the material by adding the appropriate equations and internal state variables.

Life time prediction is not available with the standard Finite-Element codes. Using the viscoplastic model described below, this is possible by adding internal state variables to describe a life reduction as long as the material contains only dispersed reversible microscopic damage. This type of damage (permanent slip band, cavities, etc.) is related, at low as well as at high temperature, to the plastic strain so that the description of the damage with internal state variable can be justified. The advantage of such a model is that it is, as the model for the deformation, valid independently of the temperature range considered.
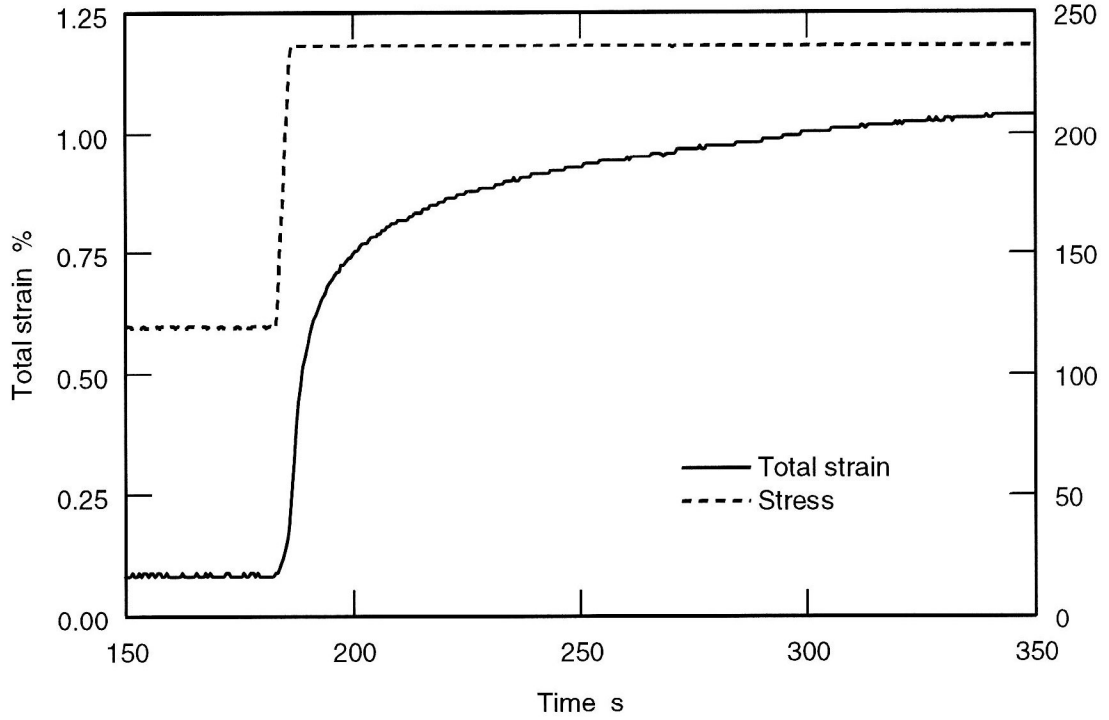
Figure 1. Result of a Tensile Jump Test with the Austenitic Steel AISI 316 at Room Temperature

2 The Model for the Simulation of Viscoplastic Deformation

The model used to describe the material behaviour considers two principal mechanisms for the viscoplastic strain rate due to the movement of dislocations: gliding of the dislocations which is predominant at low temperature, and climbing of the dislocations which is predominant at high temperature. These mechanisms are described by a set of differential equations. Strain hardening and recovery as well as microstructural effects are characterised by internal state variables.

The total viscoplastic strain rate $\dot{\varepsilon}_v$ is given by:

$$\dot{\varepsilon}_v = \dot{\varepsilon}_g + \dot{\varepsilon}_c \tag{1}$$

with $\dot{\varepsilon}_g$ the strain rate due to dislocation gliding and $\dot{\varepsilon}_c$ the strain rate due to dislocation climbing.

2.1 Strain Rate due to Dislocation Gliding

The basic equation for the strain rate is:

$$\dot{\varepsilon}_g = y_1\sigma_e\left[1 - \exp\left(-\left(\frac{\sigma_e}{y_2}\right)^{y_3}\right)\right] \tag{2}$$

with y_1, y_2 and y_3 the internal state variables and σ_e the true stress in a uniaxial stress state. When σ is the applied stress then σ_e is given by

$$\sigma_e = \sigma(1 + \varepsilon_v) \tag{3}$$

Equation (2) expresses the fact that the dislocation flux is due to a viscous gliding of the dislocations. The glide velocity is proportional to the effective stress, and the density of the moving dislocations follows a Weibull distribution. The stress y_2 represents a characteristic value for the threshold stress under which the dislocations cannot move. In a simple model that considers only obstacles to the movement of dislocations in the form of incoherent particles, it corresponds to the Orowan stress τ_0

$$\tau_0 = \frac{Gb}{\lambda} \tag{4}$$

with G the shear modulus, b the Burgers vector and λ the particle interspacing. In a technical material this interspacing is not a constant value but follows a certain distribution. Consequently the threshold stress follows also a distribution: a Weibull distribution has been assumed for them. The quantity y_3 is a scale factor which measures the scatter of the threshold stress in the material. The quantity y_1 is proportional to the density of movable dislocations.

The internal state variables y_1, y_2 and y_3 describe the change of the material behaviour as a function of deformation (strain hardening) and also of time (time recovery). These internal state variables can in turn be dependent on other internal state variables to describe for example the effect of microstructural changes or phase transformations. The number of internal state variables used is not limited, but the equations describing their evolution have the simple form

$$\dot{y}_i = A_1(A_2 - y_i)\dot{\varepsilon}_v + A_3(A_4 - y_i) \tag{5}$$

The parameters A_1 to A_4 can be constant or a function of the temperature, the stress or/and a function of another internal state variable.

2.2 Strain Rate due to Dislocation Climbing

Dislocation climbing is a thermally activated mechanism. For this reason it is predominant at high temperature. The model used to describe the strain rate is based on the classical thermodynamical approach (Kocks et al., 1985; Nicols, 1971). The equation for the strain rate due to dislocation climbing is (Rosselet, 1995)

$$\dot{\varepsilon}_c = y_4 \exp\left(-\frac{y_5 \chi(\sigma_e)}{T}\right)\left(1 - \exp\left(-\frac{\sigma_e}{B_1 T}\right)\right) \tag{6}$$

with

$$\chi(\sigma_e) = \left(1 - \left(\frac{\sigma_e}{y_6}\right)^{B_2}\right)^{B_3} \tag{7}$$

In this equation y_4 is proportional to the density of moveable dislocations by this mechanism, while y_5 is the free enthalpy of activation for the movement of dislocations at zero stress and the function $\chi(\sigma)$ describes the variation of this free enthalpy with stress. The parameters B_1 to B_3 are either constant or temperature dependent.

2.3 Cyclic Behaviour

To describe the cyclic behaviour, in particular during LCF tests, we consider that the first loading produces an initial dislocation structure and a damage state which are both decisive for the following cyclic behaviour. During the stress cycling the internal state variables also evolve in a cyclic manner. Figure 2 shows schematically the evolution of one internal state variable with time during a one dimensional stress cycle. The model considers not only the values of the internal state variable in the forward direction, but also at the same time the evolution of the variable in the reverse direction. The stress X is an equilibrium stress at which the motion of dislocations reverses. It is also an internal state variable. During loading, both parts y_f (forward) and y_b (backward) of the internal state variables change, depending on strain and time. However, only y_f is active

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and controls the behaviour of the material. During unloading when the applied stress becomes lower than X, the y_b part of the internal state variable becomes active and controls the behaviour of the material. During this time y_f is inactive but continues to evolve.
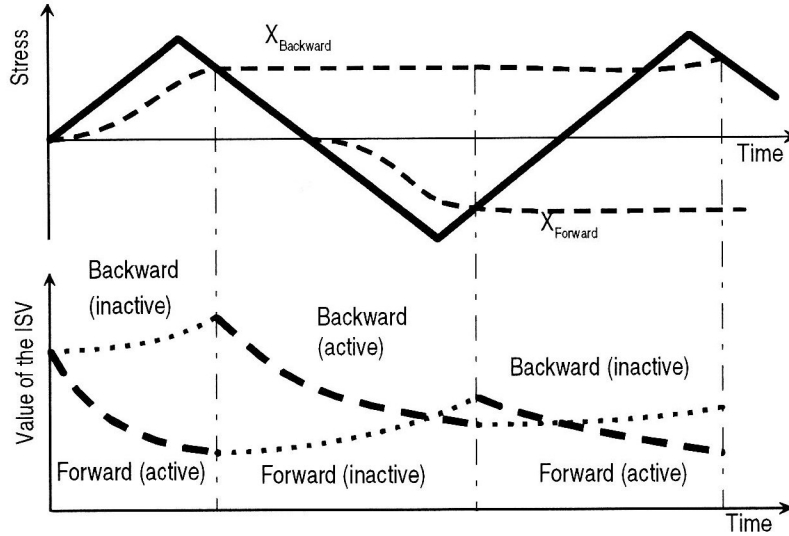

Figure 2. Evolution of One Internal State Variable (ISV) during Cyclic Loading

3 The Model for the Simulation of Damage Evolution

The equation system presented above only describes the viscoplastic behaviour of the material without damage mechanisms. However, to predict the component lifetime it is also necessary to consider the evolution of the microstructure and the damage. For this reason the viscoplastic model has been extended in order to describe a damage evolution using additional internal state variables. In this model we consider only the reversible, in the volume dispersed microscopic damage. Failure is defined as the instant where this microscopic damage becomes a localised irreversible macroscopic crack. Each increment of the cumulative plastic deformation and each increment of time produces an increment of the damage coefficient. This damage coefficient increases and reaches a critical value at failure. In the model two damage coefficients are considered, as for the viscoplastic deformation: one for fatigue damage which is associated with dislocation gliding and formation of permanent slip bands (PSB), and one for creep damage which is related to dislocation climbing and cavity formation. The total damage coefficient is the sum of both.

3.1 Evolution of Fatigue Damage

For dislocation gliding the equation for the evolution of the damage coefficient ω_g is

$$\dot{\omega}_g = \left(|\varepsilon_g|\right)^m |\dot{\varepsilon}_g| \tag{8}$$

The failure occurs when the coefficient ω_g reaches a critical value ω_f. The damage coefficient ω_g can also be normalised by dividing ω_g through ω_f so that

$$\int_0^{t_f} \frac{\dot{\omega}_g}{\omega_f} \, dt = 1 \tag{9}$$

with t_f a time to failure. The critical value ω_f is not necessarily a function of only the temperature but can also be a function of time and/or another internal state variable to describe microstructural changes. The critical value ω_f can be determined by isothermal LCF tests.

$$\omega_f = \frac{2\Delta\varepsilon_t}{\dot{\varepsilon}_t} \int_0^{N_f} \left(\left|\varepsilon_g\right|\right)^m \left|\dot{\varepsilon}_g\right| dN \tag{10}$$

where $\Delta\varepsilon_t$ is the total strain amplitude, $\dot{\varepsilon}_t$ is the total strain rate, and N_f the number of cycles to failure.

If some simplifications are made, equation (8) leads to the Manson-Coffin equation: The damage increment during one cycle is given by

$$\Delta\omega_g(N) = \int \left(\left|\varepsilon_g\right|\right)^m \left|\dot{\varepsilon}_g\right| dt \cong \frac{1}{m+1} \Delta\varepsilon_p^{m+1}(N) \tag{11}$$

with $\Delta\varepsilon_p$ the plastic strain amplitude. If we consider that $\Delta\varepsilon_p$ is constant over the whole test then

$$N_f = \omega_f (m + 1)\Delta\varepsilon_p^{-(m+1)} \tag{12}$$

which is a form of the Manson-Coffin equation.

3.2 Evolution of Creep Damage

At high temperature the principal damage mechanism is cavity formation. For the evolution of damage we can write a general equation:

$$\dot{\omega}_c = \Omega(T, \sigma, y_i) \tag{13}$$

The function Ω depends not only on the temperature and the stress, but also on internal state variables which describe the changes in the microstructure due to the stressing at high temperature.

As with ω_g, ω_c can also be normalised by ω_f

$$\int_0^{t_f} \frac{\dot{\omega}_c}{\omega_f} dt = 1 \tag{14}$$

The function Ω must be determined by performing isothermal LCF and creep tests or any other long-time tests. Isothermal LCF tests of short duration permit to determine the effect of the cumulative strain, i.e. ω_g.

From isothermal creep tests we get

$$\omega_f = \omega_g + \omega_c = \int_0^{t_f} \left(\left(\left|\varepsilon_c\right|\right)^m \left|\dot{\varepsilon}_c\right| + \Omega(T,\sigma,y_i)\right) dt \cong \frac{\varepsilon_f^{m+1}}{m+1} + \int_0^{t_f} \Omega(T,\sigma,y_i) dt \tag{15}$$

where ε_f is the rupture elongation. One possibility to find the function Ω is to consider a life fraction rule taking into account the microstructural changes (Rosselet, 1972)

$$d\omega = \frac{dt}{t_f(T,\sigma,y_i)} \tag{16}$$

By analogy, Ω can be written

$$\Omega(T,\sigma,y_i) = \frac{\omega_f - \dfrac{\varepsilon_f^{m+1}}{m+1}}{t_f(T,\sigma,y_i)} \tag{17}$$

In spite of its linear aspect, this equation does not lead to a linear damage accumulation rule for non-constant temperature and/or stress conditions. The presence of the internal state variables has the effect that the variation of Ω is dependent on the stress and temperature history.

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4 Experimental Results

The viscoplastic model for deformation has already been established for a number of materials. Particularly the short time properties required for processing simulation have been modelled for austenitic and ferritic steels, etc. taking into account phase transformations occurring during processing, for example the martensitic transformation in ferritic steels. Models have also been developed for cobalt and nickel based alloys.

The model for damage evolution has, at the moment, been developed only for one material: the martensitic steel E911 of a COST 501/III WP11 project. Only basic tests have been performed to establish the model and a few tests which permit a first validation of the model. For this reason, only results of the fit for this material will be presented and discussed.

4.1 Material Used

The COST 501/III WP 11 has the purpose to qualify the new martensitic steel, developed for boiler applications, E911. The chemical composition (in weight %) of the steel is the following:

C	Si	Mn	P	S	Cr	Mo	Ni	Nb	V	Al	W	N
0.105	0.20	0.35	0.007	0.003	9.16	1.10	0.23	0.068	0.23	0.007	1.0	0.072

The tests have been carried out with specimens from a pipe with the dimensions $\varnothing$ 285 mm x 55 mm fabricated by Dalmine (I). The material is in the standard full heat treatment condition.

4.2 Model for the Cyclic Deformation

A number of LCF tests have been made in the temperature range from room temperature up to 650°C with strain rates of 0.00005 to 0.02 s^{-1}, and strain amplitudes of 0.4, 0.8 and 1.8%. Also a few constant loading rate tensile tests have been performed. Based on the test results a viscoplastic model has been established. Figures 3 and 4 show, as an example, the comparison between experimentally measured peak stresses and the fitted values of the model during some LCF tests at 300 and 650°C.

4.3 Model for the Damage Evolution

Figure 5 shows the evolution of the damage coefficient ω_f as a function of the cycle number during the LCF tests at room temperature. The analysis of the results gives a value of the exponent m in equation (7) of 0.9. At 300°C (Figure 6) the value of m is 0.5. It decreases with temperature, and at high temperature, it should be equal to zero to correspond to equation (12). Figure 7 shows the change of m with temperature. The variation of the critical damage coefficient ω_f is plotted in Figure 8.

At 650°C, the analysis of the LCF data seems to lead to a function Ω which is proportional to the stress.

$$\Omega = 2.2 \cdot 10^{-7}\sigma \tag{18}$$

Figure 9 shows the evolution of the damage coefficient with the cycle number using this equation for Ω.

At 650°C the stress rupture data available for this steel leads to the following stress dependence of the rupture time for short testing times (100 - 1000 hours):

$$t_f = \left(\frac{\sigma}{200}\right)^{-11.7} \tag{19}$$

with σ in MPa and t_f in hours, so that from equation (16)

$$\Omega = B\sigma^{11.7} \tag{20}$$

The definitive equation for Ω is the sum or a combination of both equations (17) and (19). Also the introduction of one or more internal state variables might be necessary. A number of tests and also a detailed analysis of the data available are still to be performed in order to determine the form of this equation.
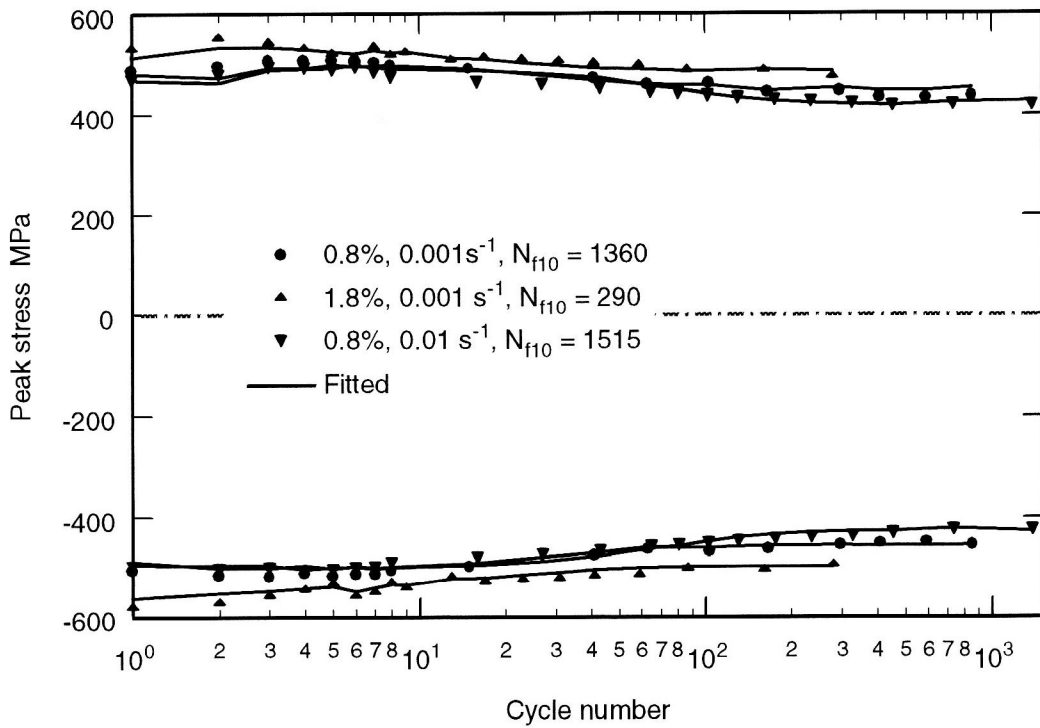

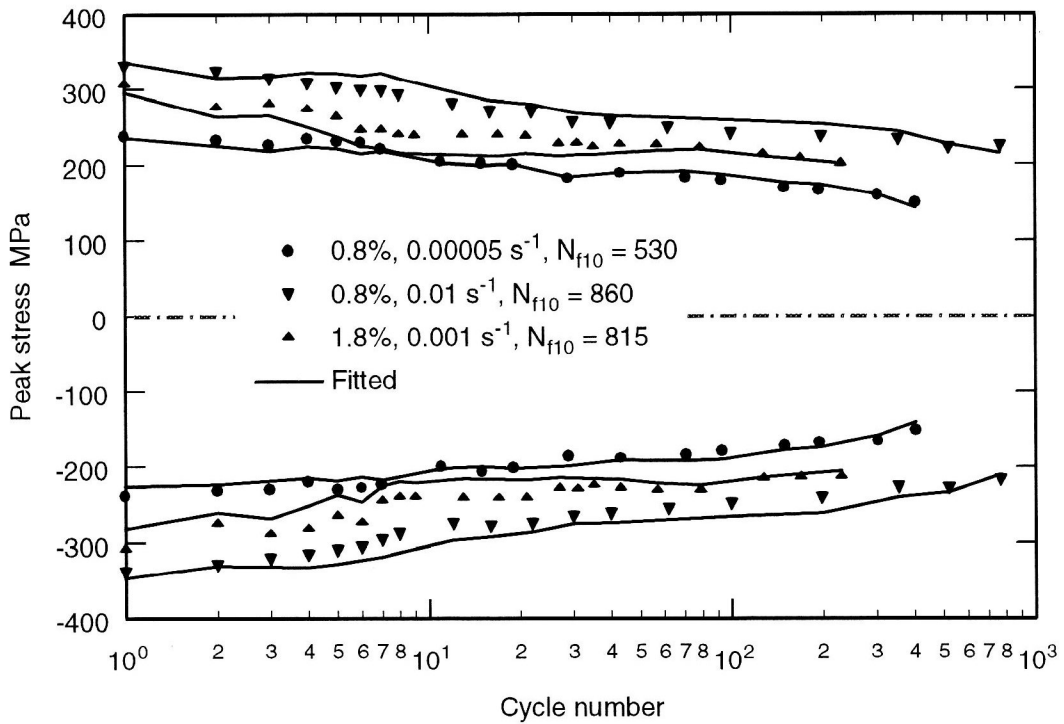Figure 3. Peak Stress Evolution during LCF Tests at 300°C

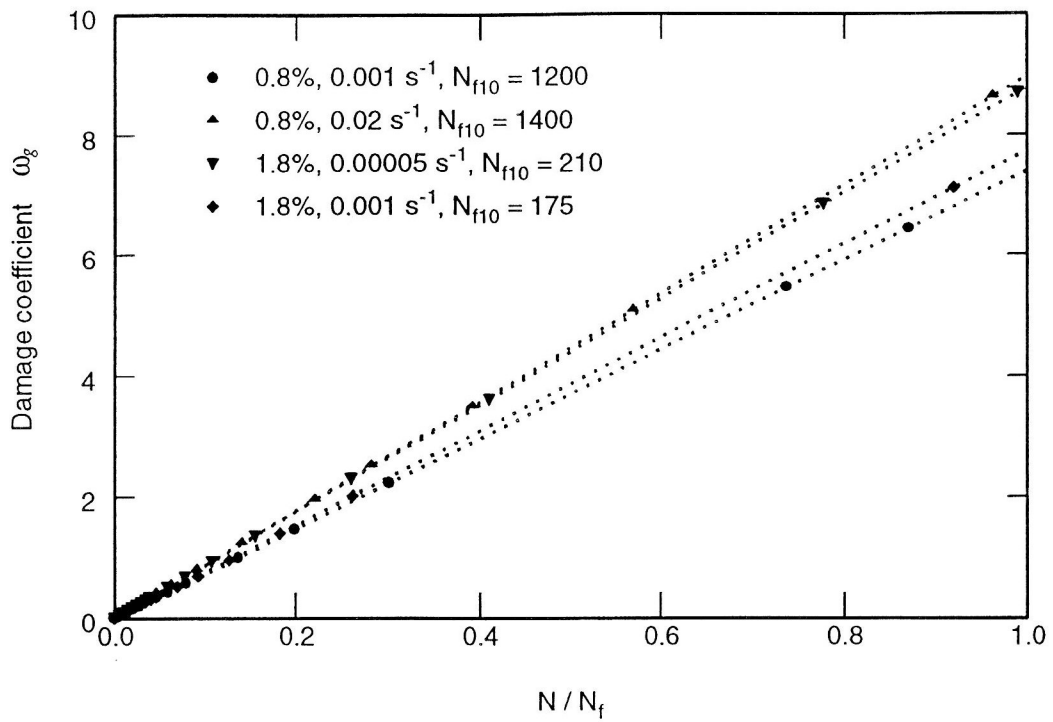Figure 4. Peak Stress Evolution during LCF Tests at 650°C

Figure 5. Evolution of the Damage Parameter during LCF Tests at Room Temperature
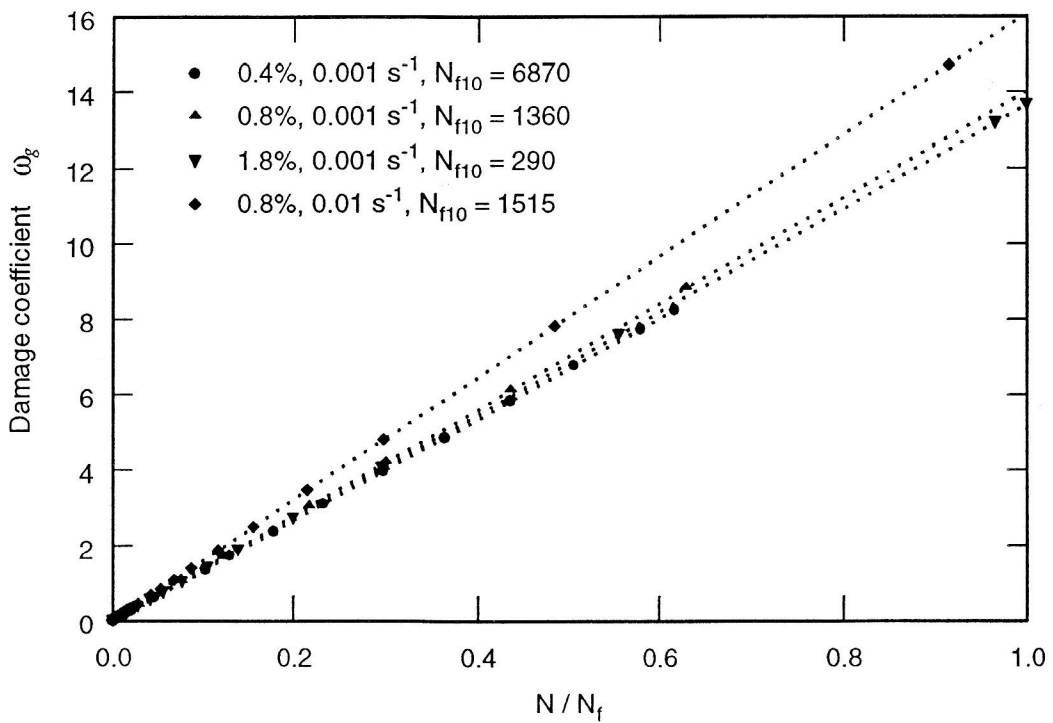
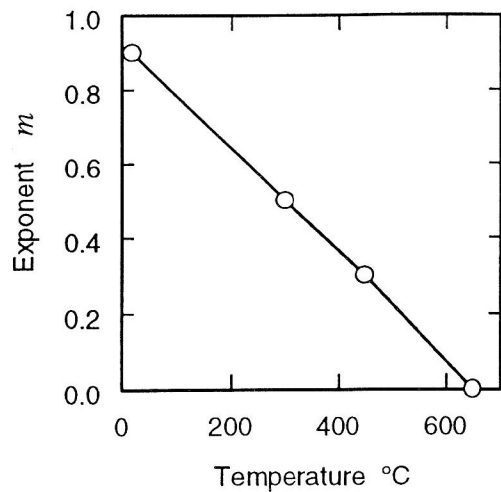Figure 6. Evolution of the Damage Parameter during LCF Tests at 300°C

Figure 7. Exponent m in Equation (8) as a
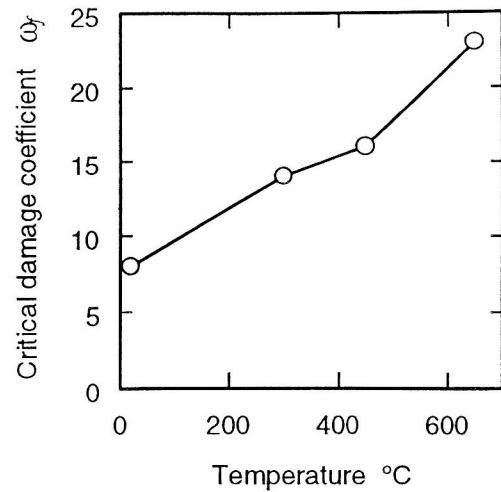Function of the Temperature

Figure 8. Critical Value of the Damage Coefficient
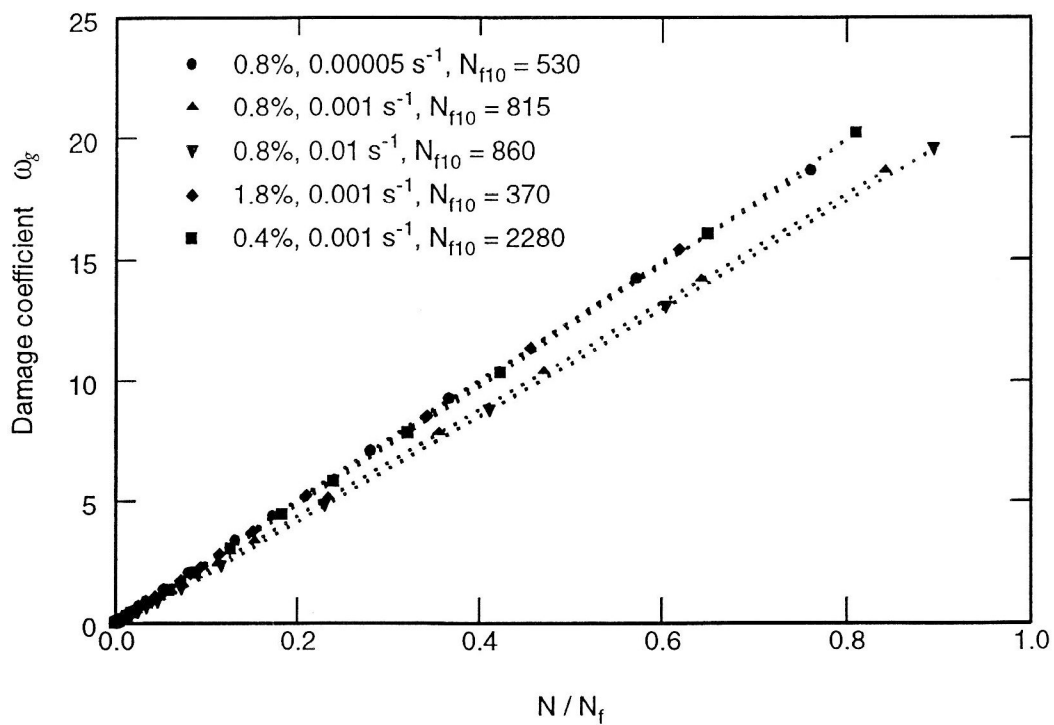as a Function of the Temperature

Figure 9. Evolution of the Damage Parameter during LCF Tests at 650°C

5 Conclusion

The present results of the analysis of LCF tests with the martensitic steel E911 show that it is possible to describe the evolution of damage as a parameter consistent with a two mechanism viscoplastic model of the

deformation process. At low temperature this damage parameter increases more or less linearly with the cycle number as with conventional fatigue data. This is expected because the parameter depends on cumulative plastic strain. In an isothermal LCF test, with constant strain amplitude, this cumulative plastic strain increases steadily during the test so that the damage parameter also evolves steadily. At high temperature the evolution of the damage parameter depends not only on the increase of the cumulative plastic strain but also on time. The increase of the parameter is also linear in the case of the isothermal LCF tests without hold times. This does not indicate that the damage accumulation for a real cycle with a complex temperature and stress history will be linear, particularly in the intermediate temperature range in which the creep-fatigue interaction is important. The accumulation of plastic strain, and consequently of the damage parameter, depends on a number of internal state variables which evolve differently according to the temperature and stress history.

Literature

1. Kocks, U. F.; Argon, A. S.; Ashby, M. F.: Thermodynamic and Kinetics of Slip, Pergamon Press, Oxford, (1985).

2. Nicols, F. A.: On the Stress Dependence and Activation Area for Creep, Material Sci. and Eng. 8 (1971), 108 - 120.

3. Rosselet, A.: Constitutive Viscoplastic Equations and their Use for the Prediction of Residual Stresses and Strains in Solidification Processes, COST 504, Round 3, Project CH2, Final Report (1995).

4. Rosselet, A.: Numerische und statistische Auswertung der Ergebnisse von Zeitstandversuchen, unter anderem mit Berücksichtigung der Gefügeveränderungen während der Prüfzeit, Proc. of the Int. Conf. on "Properties of Creep-Resistant Steels", Düsseldorf 1972.

Address: Dr. André Rosselet, Sulzer Innotec Ltd, P.O. Box, CH-8401 Winterthur, Switzerland