A model for steady state stage III creep regime at low-high stress/temperature range

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RIASSUNTO. Nonostante il creep diffusionale sia stato ritenuto, in passato, di scarso interesse per le applicazioni ingegneristiche, la sempre crescente domanda di nuovi e più affidabili strumenti di progettazione a creep per lunghe durate, che superino le 100.000h, richiede lo sviluppo di nuovi modelli che tengano in contro sia dei fenomeni di creep dislocazionale sia diffusivo relativamente ai regimi di sforzo elevato-bassa temperatura e di basso sforzo-alta temperatura. In questo lavoro viene proposto una formulazione per il creep che tiene in conto del cambio di meccanismi attraverso un'esplicita dipendenza dell'esponente di creep dalla temperatura e dal livello di sforzo applicato. Un esempio di applicazione del modello è presentata e discussa relativamente al caso di alluminio ad elevato grado di purezza (Al 99.999%).

ABSTRACT. Although diffusional flow creep is often considered out of practical engineering applications, the need for a model capable to account for the resulting action of both diffusional and dislocation type creep is justified by the increasing demands of reliable creep design for very long lives (exceeding 100.000h), high stress-low temperatures and high temperature-low stress regimes. In this paper, a creep model formulation, in which the change of the creep mechanism has been accounted for through an explicit dependence of the creep exponent n on stress and temperature, has been proposed. An application example of the proposed approach to high purity aluminum is given.

KEYWORDS: creep, dislocations, diffusional, pure aluminum

1 INTRODUCTION

Increasing demand for higher operative temperatures and longer design life requires new materials and advanced design tools for more accurate life predictions based on shorter duration laboratory tests. In order to achieve this challenging task, material modeling needs to be broaden down to the microstructural material scale where irreversible processes, such as deformations and damage, take places.

For a pure polycrystalline metal, subjected to constant tensile stress, the shape of the curve describing the accumulation of viscoplastic elongation as a function of time can be resolved fairly clearly into four stages: a virtually instantaneous extension also indicated as stage I; a decelerating Andrade creep, or stage II, well described by the following relationship,

$$\varepsilon = \beta t^{1/3} \tag{1}$$

almost steady-state Andrade creep,

$$\varepsilon = \kappa t \tag{2}$$

followed by an acceleration towards failure or stage IV creep, [1].

The attention is commonly focused on the steady state stage III creep regime because most of the strain and duration of laboratory test or in the applications occurs in this stage.

According to the range of stress and temperature, creep may occur by dislocation glide, or glide-plus-climb (limited by glide processes, limited by lattice-diffusion controlled climb, limited by core-diffusion controlled climb; power-law breakdown, Harper-Dorn creep; creep accompanied by dynamic re-crystallization), and diffusional flow (Nabarro-Herring creep; Coble creep and interface-reaction controlled diffusional flow).

For each of these mechanisms, steady state creep stage specific models, based on dislocation mechanics theory, are available in the literature. The majority of them leads to a n-power-law solutions for the creep rate,

$$\dot{\varepsilon} = A\sigma^n \tag{3}$$

where A temperature dependency follows Arrhenius' type law,

$$A = A_0 \exp\left(-\frac{Q}{kT}\right) \tag{4}$$

where Q is the activation energy and k the Boltzmann's constant.

The literature relative to stage III modeling is extensive and a detailed discussion and references for each model can be found elsewhere [2, 3].

As far as concerns diffusional flow creep, theoretical models predict a dependence of the strain rate on stress to the power of 1:

$$\dot{\varepsilon} \propto \sigma \tag{5}$$

while for dislocation creep type, the strain rate is,

$$\dot{\varepsilon} \propto \sigma^n \tag{6}$$

where n ranges from 3 to 5 for moderate and high stress. At higher stresses, the power-law description is no longer verified and several metals show an exponential dependence of the creep rate on stress,

$$\dot{\varepsilon} \propto \exp(\sigma) \tag{7}$$

Usually the application of eqn. (6) in this regime leads to very high values of the creep exponent n (>10) and very small values, not physically sound, of the power law amplitude constant A.

Although it is often said that diffusional flow creep is out of the practical engineering applications, the need for a unified model capable to account for the resulting action of both diffusional and dislocation type creep is justified by the increasing demands of reliable creep design for very long lives (exceeding 100.000h), high stress-low temperatures and high temperature-low stress regimes.

In this perspective, a creep model formulation, in which the change of the creep mechanism has been accounted for through an explicit dependence of the exponent n on stress, is proposed.

2 CREEP MODEL FOR STEADY STATE CREEP

According to the Orowan's equation, steady state creep rate for dislocation type creep is given by,

$$\dot{\varepsilon} = \rho_m b \overline{v} \tag{8}$$

where ρ_m is the mobile dislocation density, b the Burgers vector and $\overline{v}$ the average dislocation velocity. Both dislocation density and average velocity depend on stress and temperature according to:

$$\rho_m = \alpha \frac{1}{b^2}\left(\frac{\sigma}{\sigma_0}\right)^L \tag{9}$$

and,

$$\overline{v} = \beta \frac{D_V}{b}\left(\frac{\sigma}{\sigma_0}\right)^N \tag{10}$$

where α and β are constants, and D_V is the lattice diffusion coefficient. Usually it is assumed that the mobile dislocation density is a quadratic function of the applied stress (L=2) and that the velocity is a linear function of stress (N=1). With these assumptions a creep rate power law with n=3 is usually obtained. Although L=2 is commonly assumed or postulated in the literature, there are experimental evidences in support of the fact that L is neither constant nor necessarily equal to 2. Taleff et al. [4], for instance, found at 450°C L=1.0 and L=0.6 for Al–5.5 at.% Mg and AA5083 aluminum alloy, respectively.

Similarly, there are also compelling evidences that the stress dependence of the mobile dislocation velocity is larger than 1 as predicted by classical models, [5]. In particular Caillard and Martin [6] reported for N an exponential function of the applied stress. Evidence of the variability of the creep exponent n over a wide stress range can also be found in [7] where, for ½Cr½Mo¼V steel, the $\partial \ln(\dot{\varepsilon}) / \partial \ln(\sigma)$ increases with stress. Thus, in a very general form the creep exponent should be given as,

$$n = n(\sigma, T) \tag{11}$$

A possible evolution law for n can be derived from the deformation mechanism map (DMM) [8] where each region on the map is associated with a specific, dominating deformation mechanism for which the creep exponent is known.

In order to derive a plausible expression for eqn. (6), two paths on the DMM can be ideally followed: varying stress at constant temperature, $\sigma|_T$, or varying the temperature at constant stress, $T|_\sigma$.

In general, for the latter, $T|_\sigma$, the creep exponent n exponent does not change along the path. In fact, at low stress, the mechanism is diffusional (n=1) at low temperatures and it changes to lattice diffusion at high temperature but always with n=1 (eventually changing in Coble or Nabarro-Herring creep).

Increasing the stress there is only a very narrow stress band for which the creep mechanism may change, going from low to high temperature, from diffusional to dislocation climb.

At higher stress, the dominating creep mechanism is

dislocation climb at both low and high temperatures (n=3-5, n>5-8 in the power law breakdown, PLB, regime). According to this, for a given stress, eqn. (6) should show a weak dependence on temperature.

On the contrary, a change in n always occurs any $\sigma|_T$ path. For a given temperature, at low stress n=1 since the creep mechanism is always diffusional, while increasing the stress level, it switches to dislocation climb and, at even higher stress, PLB may also take place.

Since the change in the creep mechanism as a function of stress occurs rapidly as soon as threshold stress is exceeded, the following expression is proposed:

$$n = 2\sinh\left[\left(\frac{\sigma}{\hat{\sigma}_0}\right)^m\right] + 1 \tag{12}$$

where m is a constant and $\hat{\sigma}_0$ is the stress required to move a dislocation in the absence of other dislocations that can arise as a result of solutes, Peierls-type stresses, grain-size strengthening, etc. This value is temperature dependent according to,

$$\hat{\sigma}_0 = A_1 \exp\left(-T^*\right) + A_2 \tag{13}$$

where T* is the homologous temperature.

In its general form eqn. (8) becomes,

$$\dot{\varepsilon} = \kappa \frac{D_V}{b^2}\left(\frac{\sigma}{\hat{\sigma}_0}\right)^L \left(\frac{\sigma}{\hat{\sigma}_0}\right)^N = \kappa \frac{D_V}{b^2}\left(\frac{\sigma}{\hat{\sigma}_0}\right)^{N+L} \tag{14}$$

where, n=N+L and κ is a constant.

From eqn. (12), expressing the *sinh* in exponential terms, N and L may be given according to the following expressions:

$$L = 1 - \exp\left[-\left(\frac{\sigma}{\hat{\sigma}_0}\right)^m\right] \tag{15}$$

$$N = \exp\left[\left(\frac{\sigma}{\hat{\sigma}_0}\right)^m\right] \tag{16}$$

For the mobile dislocation density, eqn. (15) results in a function which saturates as confirmed by experiments, [9] while Eqn. (16) gives an increasing N with stress, with a lower bound of N=1 for stress going to zero.

In analogy with the Orowan's equation, the steady state creep rate for diffusional flow can be given as,

$$\dot{\varepsilon} = \phi_V C_V \frac{\Omega}{d} \tag{17}$$

where ϕ_V is the vacancy diffusion flux, C_V is the vacancy concentration, Ω is the atomic volume and d is the average grain size. Since,

$$\phi_V \sim 2\sigma\Omega D_V / kTd \tag{18}$$

and $D_V C_V = \overline{D}_V / \Omega$, where D_V is the vacancy diffusion coefficient, $\overline{D}_V$ is the atom diffusion coefficient, it follows [10],

$$\dot{\varepsilon} = \alpha \frac{\overline{D}_v}{d^2}\frac{\Omega}{kT}\sigma \tag{19}$$

which shows a linear dependence of the creep rate on stress (i.e. n=1).

Eqn. (14) can be also used to describe diffusional flow creep because it is formally analogous to eqn. (19) posing:

$$\hat{\sigma}_0 = \frac{kT}{\Omega} \tag{20}$$

and κ = α. Since the diffusional flow creep takes place at very low stress, from eqn. (8) N+L=1 is obtained.

Therefore, the proposed approach allows the derivation of a single equation for the steady state creep rate, derived in the framework of dislocation based models, that can be used over a wide range of temperature and stress covering most of the DMM.

3 APPLICATION TO HIGH PURITY ALUMINUM

The proposed creep model has been applied to high purity aluminum. Experimental data have been collected from [11] and integrated with data from [12]. The constants in eqn. (12) and (14) have been found as follows: b= 2.86E-10 m, D_v=D_0*exp(-Q/RT)* where D_0=1.7E-04 m^2/s [8], Q=116 kJ/mol, for the reference temperature T=533K, κ=7.8288E-12, $\hat{\sigma}_0$ is function of temperature and it has been derived from [8] (for T=533K, $\hat{\sigma}_0$=3.4 MPa).

The coefficient m has been identified on the $\ln(\dot{\varepsilon})$ – $\ln(\sigma)$ plot for T=533K and kept constant for all other temperatures. In Fig. 1, the DDM for pure aluminum is reported, [13] while in Fig. 2 the comparison with experimental data of the predicted creep rates as a function of stress, at several temperatures, is given. It is worth to underline how the agreement is very good even at very low stress.

A confirmation of the fact that the proposed approach is physically sound can be found in the calculation of the activation energy that can be determined as the slope of linear fit on the $\ln\left(\kappa D_V / b^2\right)$ vs *1/T* plot.

If the Norton type creep power law is used, this fit usually results in unrealistic values of Q. On the contrary, with the proposed model, the Q found for pure

aluminum is 116 kJ/mol, see Fig. 3, in the temperature range of 400-700K, that is very close to the value measured in [14].

Similar good agreement is found for also for single crystal high purity aluminum. In Fig. 4 the comparison of the present model with experimental data compensated for temperature is given.

4 CONCLUSIONS

In this work a creep law which allows to account for both diffusional flow and dislocation climb creep mechanisms has been derived. In the proposed creep equation, that result in a n-power-law type formulation, the creep exponent n is function of both stress and temperature, even though this latter dependence is weak.

The proposed model has the following major features: a) it is mechanism based and allows one to follows the mechanism evolution according to the DMM; b) it is able to predict the non linearity in the at low stress, that is a fundamental requirements for long live predictions; c) the activation energy values derived with

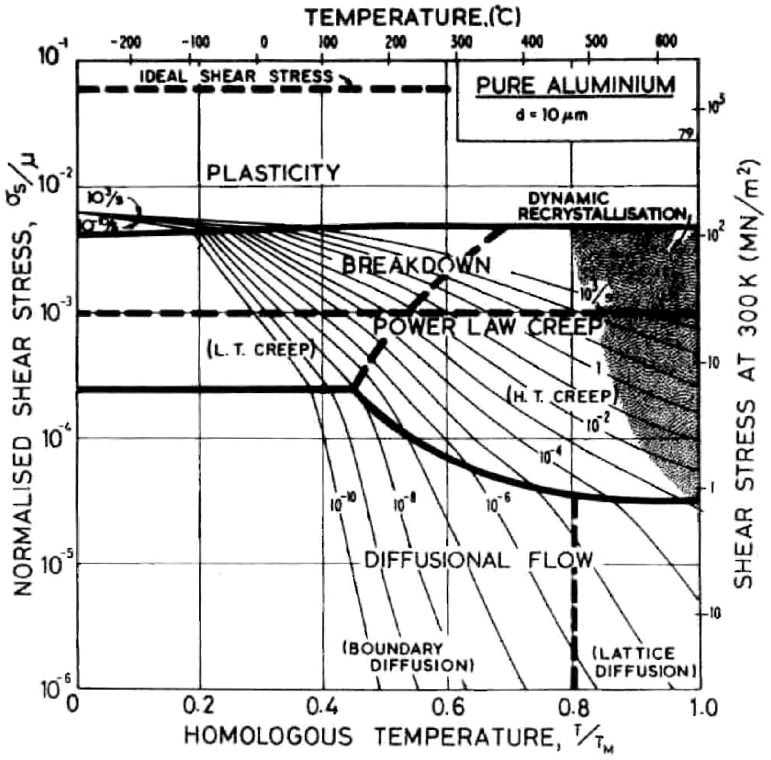

Figure 1: Deformation mechanism map (DMM) for pure aluminum average grain size d=10 μm.

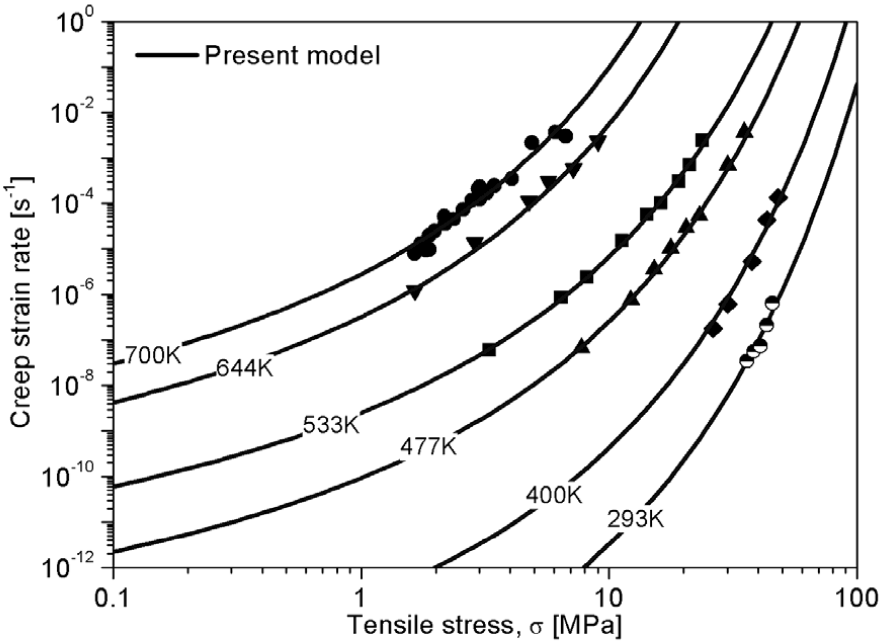

Figure 2: Comparison of the predicted creep rates for variable temperature, as a function of stress.

the proposed model are in agreement with experimental measurements.

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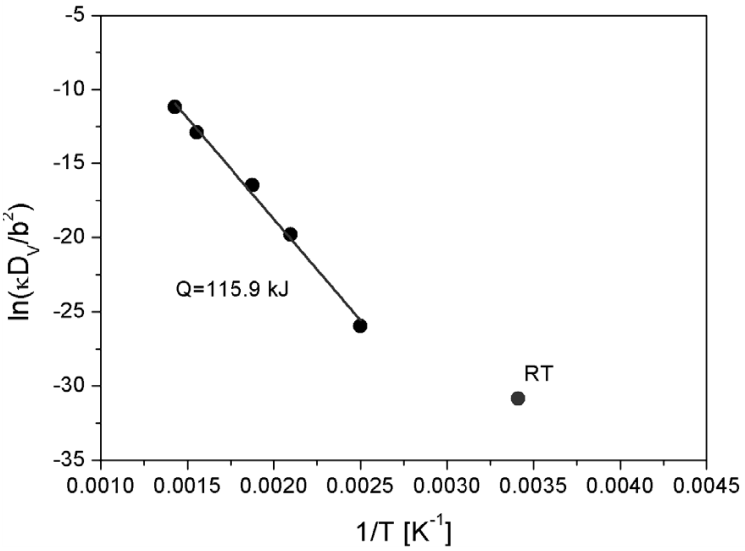

Figure 3: Determination of the activation energy for pure aluminum with the present model.

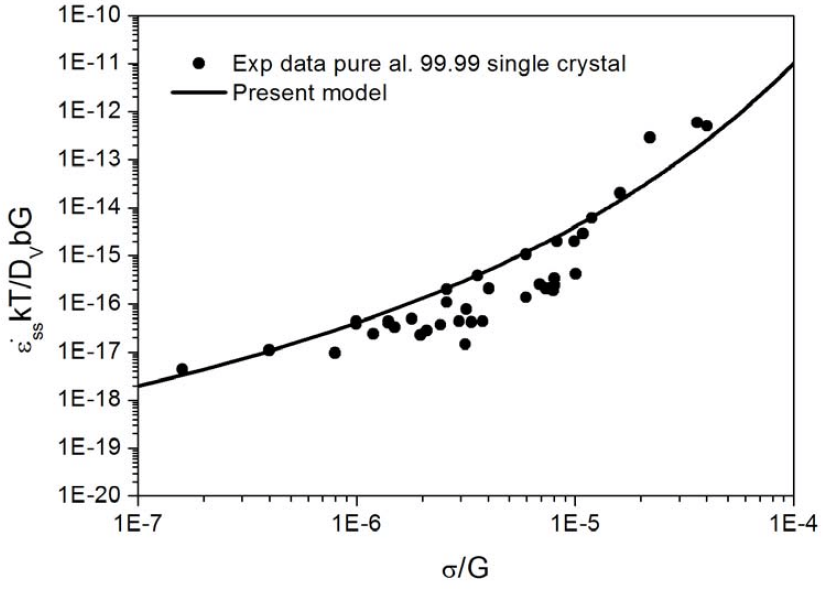

Figure 4: Comparison of predicted normalized steady state creep rate as function of normalized stress for high purity Al with experimental data.