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Creep modelling of 316H stainless steel over a wide range of stress

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Abstract

Investigation of material creep behaviour in the diffusion controlled creep regime is often unfeasible because of the long duration associated with low stress levels. On the other side, extrapolation from higher creep rates usually provides inaccurate results because of the sharp change in the data trend as a result of the change in the governing deformation mechanism from dislocation to diffusion type controlled creep. Similarly, extrapolation based on creep models, which have been formulated and validated mainly for dislocation type creep (such power law creep with the creep exponent ranging from 6 to 9), underestimates the actual creep rate significantly. Recently, Bonora and Esposito (2010) developed a mechanism-based model (BE model) capable to account for deformation and damage mechanism occurring in creep. In this work the BE model was applied to AISI 316H stainless steel for which considerable creep data in both dislocation and diffusion temperature/stress controlled regime were available. Using the same data set, the predictive capabilities of several models were compared.

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1. Introduction

Most of the components operating under creep conditions undergo inelastic strain accumulation due to several creep deformation mechanisms. Each mechanism is characterized by a different activation energy and rules for a specific stress and temperature range as extensively stated, Ashby and Frost (1982). Usually experimental tests are conducted at high stresses to reduce the laboratory time. For the same reason, to explore lower stress behavior, higher

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temperatures are often imposed. A challenging practice allows to analyze a greater data-set by normalizing the minimum creep rate respect to the temperature over a wide range of stress. In this case, the main difficulty is to recognize the correct activation energy for the considered stress and temperature range. Data analysis on wide range of stress reveals that the creep rate shows different stress dependences at high and low stress regimes. The extrapolation of creep rupture times, based on the minimum creep rate evaluated from short time creep tests is a common practice, Monkman and Grant (1956), Abe (2013), Bonora et al. (2014). Consequently, the risk of creep life over-estimation is related to the reliable determination of the minimum creep rate, Dimmler et al. (2008). In this paper is proposed an enhanced formulation for the minimum creep rate prediction over a wide range of stress.

2. Mechanism-based model

From a general point of view, the steady-state or minimum creep rate can be seen as the result of linear superposition of two contributions:

$$\dot{\epsilon}_m = \dot{\epsilon}_m^\perp + \dot{\epsilon}_m^d \quad (1)$$

where \perp and d superscript symbols are related to dislocational and diffusional contribution, respectively. Usually the minimum creep rate and the steady-state creep rate are terms used as synonyms even though they don't have the same meaning. In fact, the steady state is characterized by a dynamic equilibrium among the strain-rate accelerating and decelerating processes, while the experimentally determined minimum creep rate may indicate when the microstructural damage starts to overweight the work hardening. The formulation proposed in this paper is useful to predict both the conditions. A creep model formulation for the dislocational contribution to the total creep rate over a wide range of stress and temperature has been developed and proposed in Bonora and Esposito (2010). The different stress dependences of the creep rate in the dislocational regime was justified through the contribution of both mobile dislocations density and average dislocations velocity resulting in an explicit dependence of the exponent n on stress, $n = \exp[(\sigma/\sigma_0)^m]$. Here the formulation is extended to account for the diffusional contribution and validated on AISI 316 experimental data.

2.1. Diffusional theory

Diffusional theory of creep leads to the well-known equation for diffusion controlled visco-plastic strain accumulation, Burton (1977):

$$\dot{\epsilon}_m^d = \phi_v C_v \frac{\Omega}{d} \quad (2)$$

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,

$$\phi_v \sim \frac{2\sigma \Omega D_v}{kTd} \quad (3)$$

and $D_v C_v = \bar{D}_v / \Omega$, where D_v is the vacancy diffusion coefficient, \bar{D}_v is the atom diffusion coefficient, it follows,

$$\dot{\epsilon}_m^d = \kappa \frac{\bar{D}_v}{d^2} \frac{\Omega}{kT} \sigma \quad (4)$$

where κ is a material constant. Eqn. (4) shows a linear dependence of the creep rate on stress (i.e. $n=1$ in a general power-law form). Since kT/Ω has the dimension of a stress, we can write eqn. (4) in a more convenient form:

$$\dot{\epsilon}_m^d = \kappa' \frac{\bar{D}_V}{d^2} \left(\frac{\sigma}{\sigma_0} \right) \quad (5)$$

where $\kappa' = \kappa \Omega \sigma_0 / kT$. Diffusion may occur not only in the lattice but also at the grain boundaries. Coble (1963) firstly recognized the role of grain boundaries and calculated the contribution to the rate of creep. Since both lattice and grain boundary diffusion contribute independently to creep, a convenient representation is given as follows:

$$\dot{\epsilon}_m^d = \kappa \frac{\bar{D}_V}{d^2} \frac{\sigma}{kT} \Omega \left(1 + \frac{\kappa_{gb}}{\kappa} \frac{w D_{gb}}{d \bar{D}_V} \right) \quad (6)$$

where w is the grain boundary width and D_{gb} is the grain boundary self-diffusion coefficient. The diffusional contribution to creep is always present. It dominates the inelastic accumulation process at very low stress for those materials having numerous sources of vacancies such as large grain boundaries and fine particle precipitates.

2.2. Combined dislocation plus diffusion type creep

The total minimum creep rate accounting for both the dislocation and diffusion contribution, leads to,

$$\dot{\epsilon}_m = \xi \frac{D_v}{b^2} \left(\frac{\sigma}{\sigma_0} \right)^{\exp\left[\left(\frac{\sigma}{\sigma_0}\right)^m\right]} + \kappa' \frac{\bar{D}_V}{d^2} \left(\frac{\sigma}{\sigma_0} \right) \quad (7)$$

For stress going to zero, the exponent of the dislocation creep contribution $n \rightarrow 1$ and eqn. (7) can be approximated to the sum of two terms, linear in stress, which leads to

$$\dot{\epsilon}_m \Big|_{\sigma \rightarrow 0} = \left(\xi \frac{D_v}{b^2} + \kappa' \frac{\bar{D}_V}{d^2} \right) \cdot \left(\frac{\sigma}{\sigma_0} \right) \quad (8)$$

A threshold stress below which the dislocation motion or the diffusion process is inhibited, due to the lattice drag resistance, may also be considered. The threshold stress value for the two mechanisms not necessarily is equal and it also should show a weak dependence on temperature. In the following, the diffusion threshold stress was neglected because usually extremely low and out of practical applications.

Therefore eqn. (7) should be modified as follows:

$$\dot{\epsilon}_m = \xi \frac{D_v}{b^2} \left(\frac{\sigma - \sigma_{th}^\perp}{\sigma_0} \right)^{\exp\left[\left(\frac{\sigma}{\sigma_0}\right)^m\right]} + \kappa' \frac{\bar{D}_V}{d^2} \left(\frac{\sigma}{\sigma_0} \right) \quad (9)$$

with this modification it is possible to predict the apparent increase of the current creep exponent n at very low stress, as observed for instance in pure Al, Straub and Blum (1990).

The creep rate temperature dependence, descending from the diffusive coefficients, can be explicitated in Eqn. (9) by the Arrhenius term, $\exp(-Q/RT)$, with $R=8.314\text{J/molK}$. Finally, for practical purpose under the hypothesis of constant grain size, all the constants can be grouped as follow:

$$\dot{\epsilon}_m = A_0 \exp\left(-\frac{Q^\perp}{RT}\right) \left(\frac{\sigma - \sigma_{th}^\perp}{\sigma_0} \right)^{\exp\left[\left(\frac{\sigma}{\sigma_0}\right)^m\right]} + B_0 \exp\left(-\frac{Q^d}{RT}\right) \left(\frac{\sigma}{\sigma_0} \right) \quad (10)$$

where Q^\perp and Q^d are the activation energy for dislocational and diffusional term, respectively.

For stress values equal or below of σ_0 the resulting creep exponent is in the order of 3 for dislocation glide and

dislocation glide plus climb, (power-law type creep).

Two possible situations of interest can be highlighted:

- at low stress, the dislocation contribution is greater than the diffusion contribution. In this case the diffusional contribution is small and can be neglected in the calculation. On the $\ln(\dot{\epsilon})$ vs $\ln(\sigma)$ plot the creep strain rate shows a smooth transition from low to high stress. That trend is qualitatively reported in figure 1. This is the case, for instance, of pure metals with large grain size with limited source of vacancies. The dislocational deformation mechanism proposed by Harper and Dorn (1957) produces this situation too.
- the diffusional flow contribution is equal or greater than the dislocation term. In this case, at low stress, diffusion mechanism dominates resulting in a rate equation with a creep exponent $n=1$ and the presence of a sharp change of slope in the $\ln(\dot{\epsilon})$ vs $\ln(\sigma)$ plot, where the local creep strain exponent usually jumps from 1 to 3 and higher, figure 2. This is the case of pure metals with very fine grain or for precipitation hardened alloys with extensive sources of vacancies.

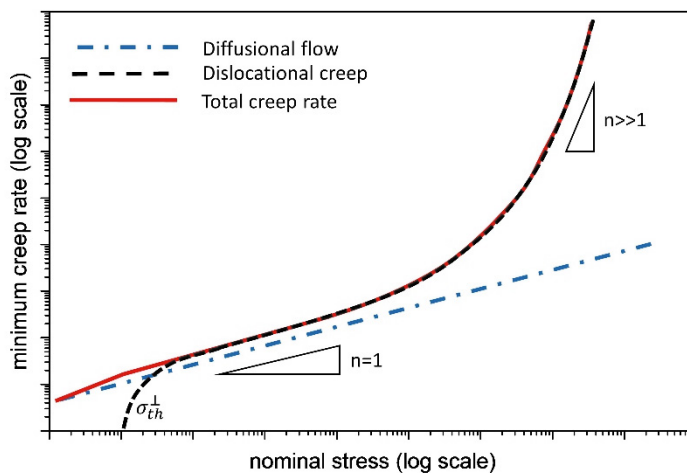


Fig. 1: Qualitative trend of the creep rate on wide range of stress if diffusional contribution is negligible.

A third case may be theorized:

- under particular condition of low stresses, the effect of the dislocational threshold stress may be only partially masked by the viscous flow and, at the transition between the dominating deformation mechanisms, an apparent increase of the current creep exponent appears, figure 3.

Since both diffusion and dislocation contribution experienced $n \approx 1$ approaching very low stresses, understanding which has a grain size dependence is the principal reliable way to distinguish them.

3. AISI 316H experimental data set

The capability of the proposed model to predict the minimum creep rate over a wide range of stress and temperature, for the AISI 316H stainless steel, was tested. Type 316H (18Cr-12Ni-Mo) is a higher carbon content variant of 316 making the steel more suitable for use in applications at elevated temperatures. The $\dot{\epsilon}_m$ values ranging from about 1 up to $1e-9$ h^{-1} , were collected from literature, Boerman and Zhu (1984), Kloc and Fiala (2005), Whittaker et al. (2012), and from the database of the National Institute for Materials Science (NIMS). On that range the dominating deformation mechanisms, and consequently the activation energies useful to normalize the data, are not the same.

Thus, the product $\dot{\epsilon}_m \cdot \exp(Q/RT)$, with a specific activation energy, allows to aggregate only homogeneous data (i.e. data that experienced the same deformation mechanism). The grain size was assumed identical for the entire data-set.

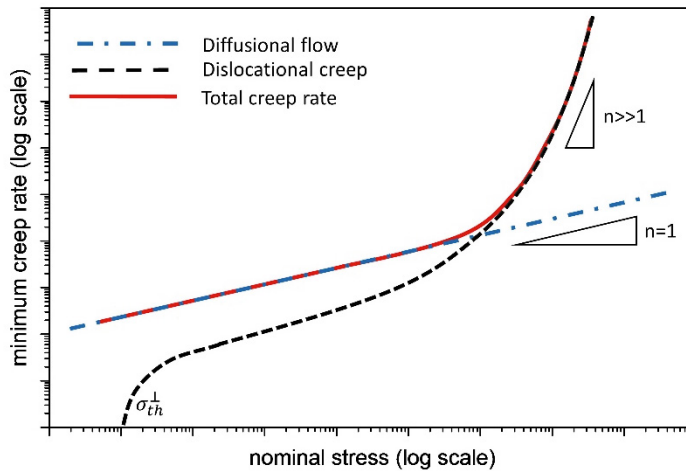


Fig. 2: Qualitative trend of the creep rate if diffusional contribution is greater than the dislocational creep at stresses of interest.

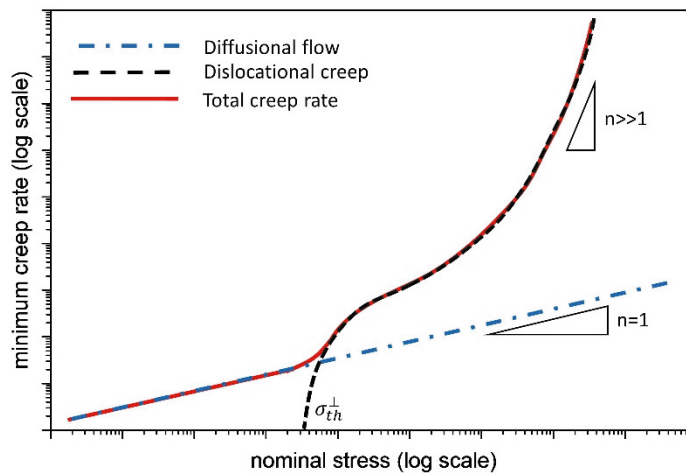


Fig. 3: Qualitative trend of the creep rate if diffusional contribution only partially masks the effect of a dislocational threshold stress.

4. Results and conclusions

The creep-rate values that experienced the same deformation mechanism were recognized assuming the activation energy and plotting the $\dot{\epsilon}_m \cdot \exp(Q/RT)$ value versus the applied stress. Since the activation energies were unknown, an iterative procedure was performed. In Figure 4a and 4b the aggregated data, after the activation energy identification, are shown. Be noted that the low-stress data, (less than 10 MPa), from 923 up to 1073K appears ruled by the diffusional processes because resulting aggregated and linearly dependent on stress after the normalization with a low value of the activation energy (about the self-diffusion activation energy). On the contrary, at higher temperature the dislocation-creep mechanism seems to prevail. The identified parameters for AISI 316H are summarized in table 1. The proposed model correctly predicts the minimum creep rate on the entire stress range by two temperature-

independent values of the activation energy. An activation energy for diffusive processes equal to 140kJ/mol appears adequate, while a dislocational activation energy about three times bigger is needed. The comparison between model prediction and experimental data is given in figure 5.

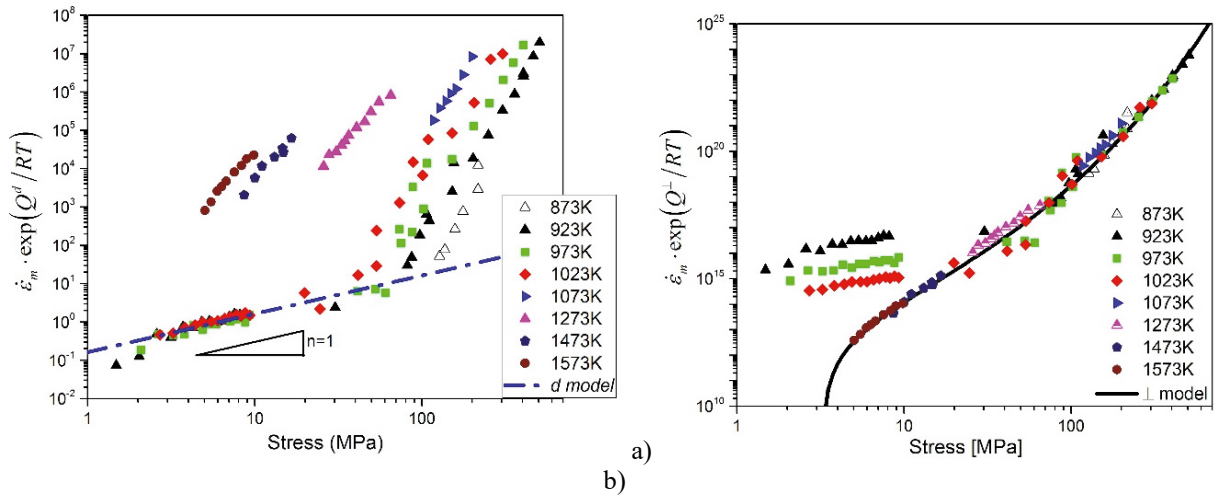


Fig. 4: Aggregated data plot of the minimum creep rate after normalization by the activation energy.

Table 1: Model parameters for AISI 316H.

A_0 [h ⁻¹]	Q^- [J/mol]	σ_{th}^- [MPa]	m	σ_0 [MPa]	B_0 [h ⁻¹]	Q^d [J/mol]
8.7267E+13	431250	3.2	0.112	6.1	1	140000

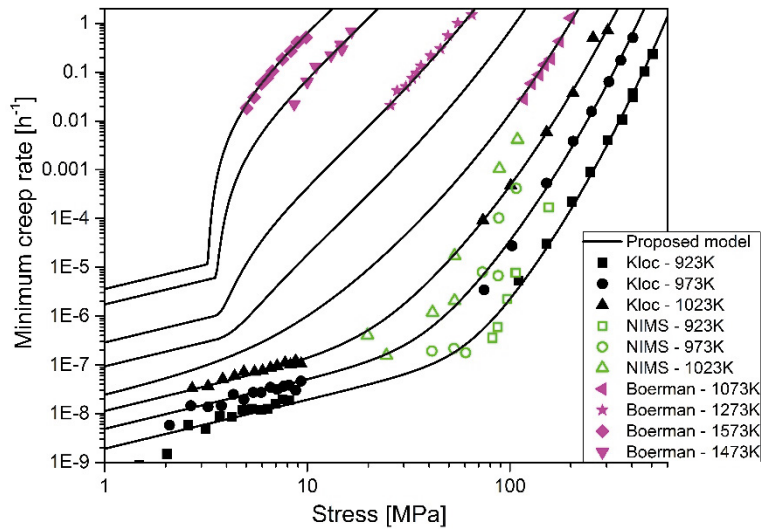


Fig. 5: Model prediction of the AISI 316H minimum creep rate on wide stress range.

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