Composites Part A 112 (2018) 346–355

Contents lists available at ScienceDirect

Composites Part A

journal homepage: www.elsevier.com/locate/compositesa

A new model for the description of creep behaviour of aluminium-based composites reinforced with nanosized particles

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A R T I C L E I N F O

Keywords:
A. Metal matrix composites
A. Nanoparticles
B. Creep
C. Analytical modelling
E. Powder processing

A B S T R A C T

A basic model developed to describe the creep response of fcc metals has been used to obtain a new constitutive equation valid for alloys and composites reinforced with a dispersion of nanosized particles. The new model gives an excellent description of the minimum strain rate dependence of applied stress and temperature, and gives reason for the reduced creep rate observed in these composites when compared with conventional alloys. The reason for the different behaviour is that the dislocation mean free path becomes equivalent to the distance between nanosized particles. The volume fraction and size of the nanosized particulate thus assumes a key-role in determining the creep response of these materials. In addition, the strengthening effect due to the interaction between particles and dislocation has been described by introducing a back stress, which is in general proportional to the Orowan stress and has the nature of a true threshold stress.

1. Introduction

Up to a decade ago, the creep response of Aluminium-based metal matrix composite was investigated in many studies [1–16]. In general, experimental data (minimum creep rate $\dot{\varepsilon}_m$, as a function of applied stress σ and temperature T) were described by a modified form of the well-known equation

$$\dot{\varepsilon}_m = A\frac{D_{0L}Gb}{kT}\left(\frac{\sigma}{G}\right)^n \exp\left(-\frac{Q_L}{RT}\right) \tag{1}$$

where A is a material parameter, k is the Boltzmann constant, G is the shear modulus, b is the length of the Burgers vector and R is the gas constant. D_{0L} and Q_L are, respectively, the pre-exponential factor and the activation energy in the Arrhenius equation describing the temperature dependence of the vacancy diffusion coefficient. In the case of Al-based metal matrix composites, Eq. (1) needed to be modified to describe the behaviour that was almost invariably observed experimentally, that is, a marked curvature of the strain rate *vs* stress curve, which, in double logarithmic coordinates, became almost vertical in the low stress regime. An example is clearly illustrated in Fig. 1, which shows the minimum strain rate dependence on applied stress for an oxide-dispersion strengthened Al [17] (ODS-Al in the following) and for the corresponding composite reinforced with 30% of SiC particulate (ODS-Al-30%SiC) [18,19]. The third material is an oxide dispersion strengthened Al-5%Mg alloy reinforced with 30%SiC (ODS-Al5Mg-30% SiC) [20]. The matrix of the ODS-Al-30%SiC composite and the ODS-Al alloy contained about 2.6% in volume of fine (25 nm) particles of alumina. The same content of alumina particles of similar size was present also in the matrix of the second composite (ODS-Al5Mg-30%SiC). These materials were specifically selected for the analysis of this study, since they were produced by the same processing route, and constitute an ideal database for modelling the different separate strengthening mechanisms, namely: i. The presence of finely dispersed nanosized oxide particles; ii. The effect of a relatively high volume fraction of coarse SiC particulate; iii. The effect of a second element in solid solution (Mg).

The presence of a fine dispersion of nanosized alumina particles is indeed typical of many metal-matrix composites produced by powder metallurgy (PM). There is a general consensus that these particles, a by-product of the processing route, exert a strong effect in reducing dislocation mobility, thus enhancing creep strength [5,6,8,15]. On the other hand, although no significant plastic flow occurs in the ceramic coarse particulate, also the reinforcements of the composites are considered to significantly improve creep response. This effect is usually attributed to the load transfer mechanism [11,12,14,15], which assumes that part of the external load is carried by the reinforcement, with a resultant reduction in the stress in the matrix. Park and Mohamed [21] expressed the combined effect of fine alumina particles and load-transfer by changing Eq. (1) into

$$\dot{\varepsilon} = A\frac{D_{0L}Gb}{kT}\left[\frac{(1-\beta_{LT})(\sigma-\sigma_p)}{G}\right]^n \exp\left(-\frac{Q_L}{RT}\right) \tag{2}$$

In this formulation, σ_p is a threshold stress representing the

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https://doi.org/10.1016/j.compositesa.2018.06.021
Received 29 January 2018; Received in revised form 29 May 2018; Accepted 16 June 2018
Available online 18 June 2018

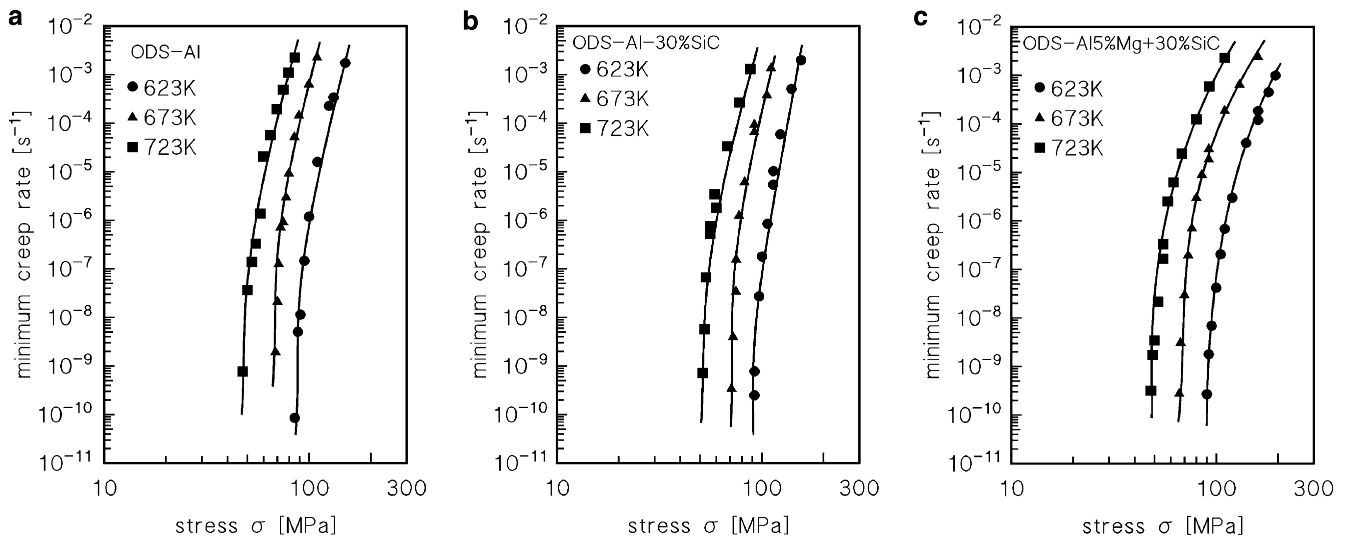

Fig. 1. Experimental minimum creep rate dependence on applied stress for ODS-Al [17], ODS-Al + 30%SiC [18,19], ODS-Al5Mg + 30%SiC [20].

strengthening effect due to the interaction of nano-sized alumina-particles with dislocations and β_{LT} is the load transfer coefficient. Eq. (2) can be rewritten in the form

$$\dot{\varepsilon} = A\,(1-\beta_{LT})^n \frac{D_{0L}Gb}{kT}\left[\frac{\sigma-\sigma_p}{G}\right]^n \exp\left(-\frac{Q_L}{RT}\right) \quad (3)$$

Eq. (3) shows that, in the presence of a load-transfer effect, Eq. (1) remains valid, once the applied stress and the A constant are replaced with the effective stress $\sigma_e = (\sigma - \sigma_p)$ and $A(1-\beta_{LT})^n$ respectively.

Li and Langdon [11] later proposed an alternative equation in the form

$$\dot{\varepsilon} = A\frac{D_{0L}Gb}{kT}\left[\frac{(1-\beta_{LT})\,\sigma-\sigma_p}{G}\right]^n \exp\left(-\frac{Q_L}{RT}\right) \quad (4)$$

which can be rewritten as

$$\dot{\varepsilon} = A\,(1-\beta_{LT})^n \frac{D_{0L}Gb}{kT}\left[\frac{(\sigma-\sigma_p^*)}{G}\right]^n \exp\left(-\frac{Q_L}{RT}\right) \quad (5)$$

being

$$\sigma_p^* = \frac{\sigma_p}{1-\beta_{LT}} \quad (6)$$

Eq. (5) is formally similar to Eq. (3), the main difference is that, in the latter, the apparent threshold stress σ_p^*, estimated by one of the conventional method summarised in [18], is substantially higher than its real value σ_p.

The main reason that led to the introduction of the load-transfer concept in the creep modelling of composites is well illustrated in Fig. 2, which plots the steady state creep rate as a function of applied stress for pure Al [22] (Fig. 2a) and Al-5 Mg [23] (Fig. 2b) and the creep rate as a function of effective stress $(\sigma - \sigma_p)$ for the materials considered in Fig. 1. Once the effective stress replaces the applied stress, the creep rates for the composites align on a straight-line that is roughly parallel, but lower, to that describing the corresponding unreinforced material. This peculiar response, which is a common feature of Al-matrix composites [11,24], is fully consistent with Eqs. (3) and (5), a fact that was interpreted as a direct confirmation that the load-transfer mechanism was operative in these materials [11]. Yet, the validity of this view is questioned by the response of the ODS-Al. In this material, load transfer is not operative, but the strain rate in Fig. 2 is still almost four orders of magnitude lower than that of pure Al, exactly as in the case of ODS-Al-30%SiC. In addition, the data-points for ODS-Al and ODS-Al-30%SiC overlap, and the threshold stress value is only marginally different. This observation led Čadek et al. [20] to conclude that load transfer did not play any significant role in these materials, that is, $\beta_{LT} \cong 0$. Thus Eqs. (3) and (5) become

$$\dot{\varepsilon} = A\frac{D_{0L}Gb}{kT}\left(\frac{\sigma-\sigma_p}{G}\right)^n \exp\left(-\frac{Q_L}{RT}\right) \quad (7)$$

The problem is that, once the load-transfer mechanism is ruled out, no other explanations for the dramatic reduction in the A parameter in ODS materials (four orders of magnitude in pure Al and two orders of magnitude in Al-5 Mg) can be provided by Eq. (7).

A last interesting feature worthy of mentioning is that, although the composites produced by powder metallurgy have usually a very fine grain size (few microns), no direct evidence of grain boundary sliding was reported in the studies mentioned above. By contrast, other works [25–27] clearly report the occurrence of the alleged “high-strain-rate superplasticity” in Al-based composites. In these studies, the elongation to failure of samples tested at high temperature (above 823 K in [25], 733 K in [26], 723 K in [27]) was usually close to 300%, while, in the experimental range where the maximum in ductility was observed, the strain rate sensitivity $(\partial \log \sigma / \partial \log \dot{\varepsilon})$ was in most cases reasonably close to 0.3. As a general rule, these cases could be more properly classified as a form of “extended plasticity”, not dissimilar to that observed in coarse grained Al-Mg alloys [28], where the strain rate sensitivity is close to 0.25–0.3. High strain rate sensitivity values slow-down strain localisation and make large tensile elongations possible. This observation does not imply that grain boundary sliding does not cause superplasticity in powder-metallurgy Al-based composites: several examples have been reviewed, for example, in [29]. In general, a true superplastic behaviour, with elongations well in excess of 500% and up to 1400%, was observed in materials with grain size well below 5 μm, at temperatures usually above 700 K. This does not seem to be the case of the materials shown in Fig. 1; although the grain size was very fine (1 μm) [20], elongation to fracture was well below 10% [17], a direct evidence that dislocation creep was the dominating mechanism.

The aim of this study is to reconsider the data on the ODS alloy and the composites [17–20,24] by applying a basic model for dislocation creep originally developed for Cu [30–33] and austenitic steels [24] and already successfully used for the description of pure Al [35] and Al-Mg alloys [36]. The effect of a fine grain size will be analysed for its consequences on dislocation mobility, while the effect of grain boundary sliding, which in principle could be operative only at very high temperature, will not be expressly considered.