THRESHOLD CREEP BEHAVIOUR OF AN Al-8.5Fe-1.3V-1.7Si ALLOY REINFORCED WITH SILICON CARBIDE PARTICULATES[†]

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Creep behaviour of an Al-8.5Fe-1.3V-1.7Si (8009Al type) alloy reinforced with 15 vol. % SiC particulates – Al-8.5Fe-1.3V-1.7Si-15SiC$_p$ composite – is investigated at temperatures ranging from 623 to 823 K. At temperatures 623–723 K the creep is associated with a true threshold stress decreasing with increasing temperature more strongly than the shear modulus. The minimum creep strain rate is controlled by the lattice diffusion in the composite matrix and the true stress exponent is close to 5.

At temperatures 748–823 K the true threshold stress is not observed, the true activation energy of creep is higher than the activation enthalpy of lattice diffusion in the composite matrix and the true stress exponent of minimum creep strain rate increases with increasing applied stress. The creep behaviour at these temperatures is interpreted in terms of thermally activated detachment of dislocations from fine incoherent particles.

Key words: Al-8.5Fe-1.3V-1.7Si-15SiC$_p$ composite, creep, true threshold stress, disappearance of the true threshold creep behaviour

PRAHOVÉ CREEPOVÉ CHOVÁNÍ SLITINY Al-8,5Fe-1,3V-1,7Si VYZTUŽENÉ PARTIKULEMI KARBIDU KŘEMÍKU

Creepové chování slitiny Al-8,5Fe-1,3V-1,7Si (známé jako slitina 8009Al) vyztužené 15 obj. % partikulí SiC – kompozitu Al-8,5Fe-1,3V-1,7Si-15SiC$_p$ – je studováno při teplotách od 623 do 823 K. Při teplotách 623–723 K je creep spojen se skutečným prahovým napětím klesajícím se vzrůstající teplotou silněji nežli smykový modul. Minimální rychlost creepu je řízena mřížkovou difuzí v matrici kompozitu a skutečný napěťový exponent minimální rychlosti creepu je blízký 5.

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Při teplotách 748–823 K není prahové napětí pozorováno, skutečná aktivační energie creepu je vyšší nežli aktivační entalpie difuze v mřížce matrice kompozitu a skutečný napěťový exponent minimální rychlosti creepu vzrůstá se vzrůstajícím aplikovaným napětím. Creepové chování při těchto teplotách je interpretováno tepelně aktivovaným odpoutáváním dislokací od jemných nekoherentních částic v matrici kompozitu, především částic intermetalické fáze $Al_{12}(Fe,V)_3Si$.

1. Introduction

An Al-8.5Fe-1.3V-1.7Si (numbers indicate wt. %) alloy processed by fast solidification and powder metallurgy route exhibits remarkable creep resistance up to temperatures of 700 K (e.g. ref. [1]). This resistance is due to high volume fraction (~ 0.27) of fine incoherent particles of the intermetallic $Al_{12}(Fe,V)_3Si$ phase and low coarsening rate of these particles at high temperatures. For this alloy, the true threshold creep behaviour is characteristic.

It is well known that the Young's modulus of an aluminium alloy can be increased significantly by discontinuous reinforcement with hard unshearable ceramic particulates, short fibres or whiskers, at least at temperatures up to 700 K [2]. Beside increasing the Young's modulus, the discontinuous reinforcement generally introduces the load transfer effect (e.g. ref. [3]) that enhances the creep strength.

It should be reminded that the creep strength of the composite like the present Al-8.5Fe-1.3V-1.7Si-15SiC_p (the subscript p stands for particulates) is generally determined by that of the matrix (e.g. ref. [4]), though, also the above mentioned load transfer can play a role.

In our previous paper [5], the creep behaviour of an Al-8.5Fe-1.3V-1.7Si alloy at three temperatures ranging from 623 to 723 K was investigated. In the present paper, the results of an investigation of creep behaviour of an Al-8.5Fe-1.3V-1.7Si alloy reinforced with 15 vol. % silicon carbide particulates (SiC_p) are reported. It is shown that at temperatures ranging from 623 to 723 K the measured minimum creep strain rates can be satisfactorily interpreted in terms of athermal detachment of dislocations from fine interacting $Al_{12}(Fe,V)_3Si$ phase particles (and also fine alumina particles, appearing during the composite fabrication). This detachment is associated with the threshold stress behaviour. On the other hand, at temperatures 773 to 823 K no true threshold stress is observed and the creep behaviour at these temperatures is interpreted in terms of thermally activated detachment of dislocations from fine interacting particles accepting a slightly modified model of Rösler and Arzt [6].

2. Material and experimental procedures

The Al-8.5Fe-1.3V-1.7Si alloy was processed by fast solidification and powder metallurgy [5, 7]. This alloy was atomised and the powder was mixed with nominally 15 vol. % silicon carbide powder of the mean particulate diameter of ~ 4.5 μm.

The mixed powders were consolidated and extruded at a temperature of ~ 830 K to a rod 12 mm in diameter. The resulting mean grain diameter of the composite was found close to 1 μm. The structure of the as extruded composite was found to be reasonably homogeneous, although not only the intermetallic phase particles but also silicon carbide particulates were mostly aligned to the extruding direction. The composite matrix consisted of an intermetallic $Al_{12}(Fe,V)_3Si$ phase particles of slightly less than 50 nm in mean diameter embedded in the matrix solid solution. Also small volume fraction of fine alumina particles was present in the composite matrix as a result of fabrication of the composite by powder metallurgy route.

The tensile creep tests were performed at temperatures ranging from 623 to 823 K in purified argon. The testing temperatures were controlled to within 0.5 K. The creep elongation was measured by means of linear variable differential transformers coupled with a digital data acquisition system. The measured minimum creep strain rates covered seven orders of magnitude, the lowest of them were well below 10^{-9} s^{-1}. Generally, no steady state stage was observed, only the minimum creep strain rate $\dot{\varepsilon}_{\rm m}$ could be defined.

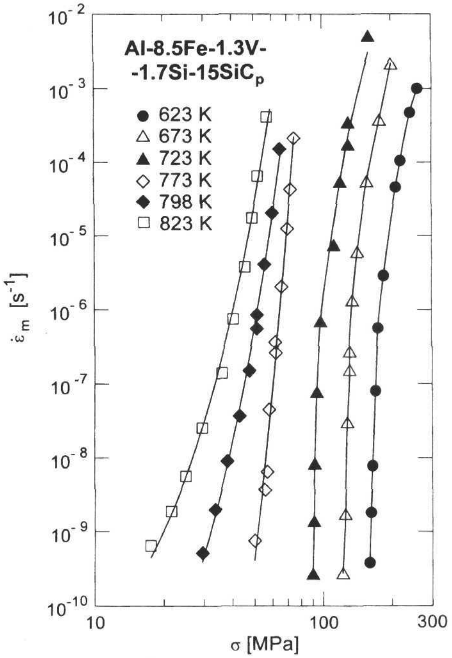

Fig. 1. Minimum creep strain rates plotted against applied stresses in double logarithmic co-ordinates.

3. Results and analysis

In Fig. 1, the minimum creep strain rates are plotted against applied stress in double logarithmic co-ordinates for all the temperatures under consideration, i.e. temperatures ranging from 623 to 823 K. At temperatures 623 to 723 K the apparent stress exponent $m_{\rm c} = (\partial \ln \dot{\varepsilon}_{\rm m} / \partial \ln \sigma)_{\rm T}$ obviously increases with decreasing applied stress σ indicating the true threshold creep behaviour. Plotting $\dot{\varepsilon}_{\rm m}^{1/n}$ vs. σ in double linear co-ordinates for the true stress exponent $n = 5$ (Fig. 2) the values of the true threshold stress values $\sigma_{\rm TH}$ are obtained extrapolating the $\dot{\varepsilon}_{\rm m}^{1/n}$ vs. σ relations to $\dot{\varepsilon}_{\rm m} = 0$. At the same time, these relations strongly suggest that the true stress exponent is really close to 5. The values of the true threshold stresses are plotted against temperature in Fig. 3, from which it can be seen that not only

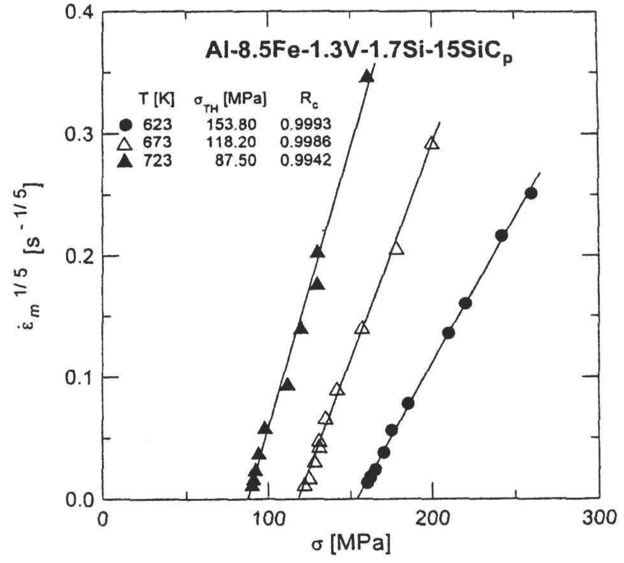

Fig. 2. $\dot{\varepsilon}_{\mathrm{m}}^{1/n}$ plotted against σ for the true stress exponent $n = 5$. The true threshold stress σ_{TH} is defined as an applied stress at which the minimum creep strain rate $\dot{\varepsilon}_{\mathrm{m}} = 0$. R_{c} is the correlation coefficient.

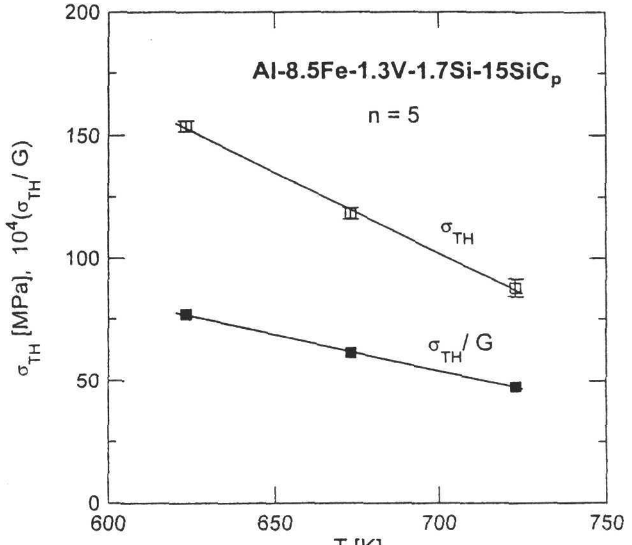

Fig. 3. Temperature dependences of the true threshold stress σ_{TH} and the ratio of true threshold stress to the shear modulus G, i.e. σ_{TH}/G.

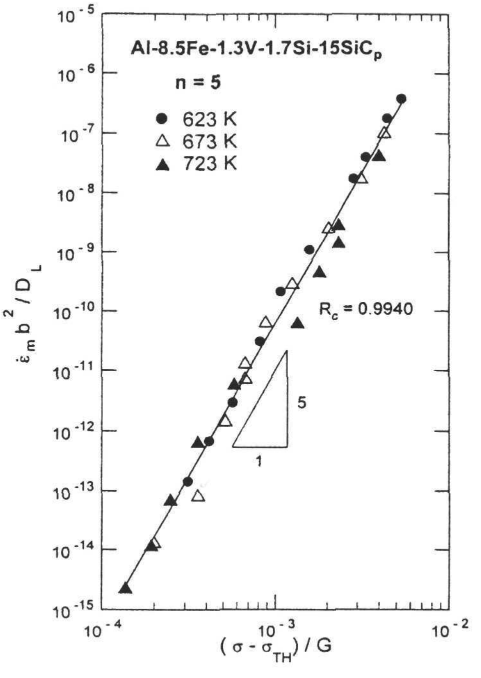

Fig. 4. Relations between $\dot{\varepsilon}_{\mathrm{m}}b^2/D_{\mathrm{L}}$ and $(\sigma - \sigma_{\mathrm{TH}})/G$ in double logarithmic co-ordinates for 623, 673 and 723 K. R_{c} is the correlation coefficient.

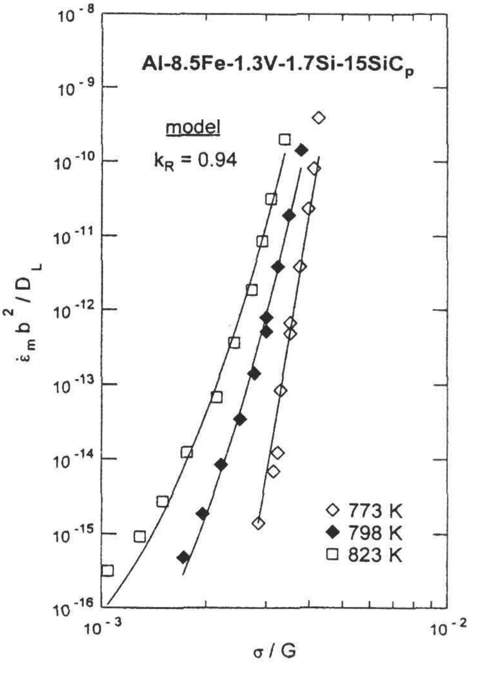

Fig. 5. Relations between $\dot{\varepsilon}_{\mathrm{m}}b^2/D_{\mathrm{L}}$ and σ/G in double logarithmic co-ordinates for 773, 798 and 823 K.

σ_{TH} but also the σ_{TH}/G ratio (G is the shear modulus of aluminium [8]) decrease significantly with increasing temperature.

Plotting the normalized minimum creep strain rates $\dot{\varepsilon}_{\mathrm{m}}b^2/D_{\mathrm{L}}$ (b is the length of the Burgers vector and D_{L} the coefficient of self-diffusion in composite matrix – aluminium [9]) against normalized effective stress $(\sigma - \sigma_{\mathrm{TH}})/G$, in double logarithmic co-ordinates, these creep strain rates fit a single straight line with the slope close to the true stress exponent n, i. e. to 5 (Fig. 4). Thus, the minimum creep strain rate is lattice diffusion controlled and the true stress exponent is close to 5. Consequently, the creep strain rate of the composite at temperature ranging from 623 to 723 K can be described by the equation (e.g. ref. [10])

$$\dot{\varepsilon}_{\mathrm{m}}b^2/D_{\mathrm{L}} = A[(\sigma - \sigma_{\mathrm{TH}})/G]^n; \qquad n = 5, \tag{1}$$

where A is a dimensionless constant. It can be easily shown (see ref. [5]) that the values of the apparent activation energy $Q_{\mathrm{C}} = [\partial \ln \dot{\varepsilon}_{\mathrm{m}}/\partial(-1/RT)]_{\sigma}$ (where R is the gas constant) as well as the apparent stress exponent m_c defined above are generally higher, or even much higher, than the activation enthalpy of the diffusion in the composite matrix (aluminium) $\Delta H_L = 142$ kJ·mol^{-1} [9] and the true stress exponent than that for pure aluminium, which is close to 5 [11].

On the contrary, at temperatures 773, 798 and 823 K the apparent stress exponent m_c decreases with decreasing applied stress (Fig. 1) strongly indicating an absence of the true threshold stress. In Fig. 5 the normalized minimum creep strain rates $\dot{\varepsilon}_m b^2/D_L$ are plotted against normalized applied stresses σ/G in double logarithmic co-ordinates. The $\dot{\varepsilon}_m b^2/D_L$, σ/G data points do not fit a single curve, strongly suggesting the *true* activation energy (activation enthalpy) of creep ΔH_c higher than the activation enthalpy of lattice self-diffusion in aluminium. Evidently, the true threshold stress disappears at temperatures higher than 723 K.

Such a disappearance of the true threshold stress behaviour can be interpreted in terms of transition from the athermal to the thermally activated detachment of dislocations from small interacting (incoherent) particles [12] applying the creep model developed by Rösler and Arzt [6].

4. Discussion

Arzt and Wilkinson [13] and Arzt and Rösler [14] developed a theory of athermal detachment of dislocations from small incoherent (and thus interacting) particles at high temperature creep conditions. The stress required to detach a dislocation from an interacting particle can be expressed as

$$\sigma_d = \sigma_{OB}\sqrt{1 - k_R^2}, \tag{2}$$

where σ_{OB} is the Orowan bowing stress and k_R is the relaxation factor characterizing the strength of attractive dislocation/particle interaction. The detachment stress σ_d represents a true threshold stress. However, this detachment stress cannot be identified with the true threshold stress following from the experimental $\dot{\varepsilon}_m(T, \sigma)$ creep data (see Section 3) unless the relaxation factor k_R is assumed to increase with increasing temperature. This is because σ_d is proportional to σ_{OB} (which in turn is proportional to the shear modulus) if $k_R \neq k_R(T)$, while σ_{TH}/G decreases rather strongly with increasing temperature. Assuming a value of $k_R = 0.85$ for 673 K [4, 15] and using a simple procedure [7] it was shown that k_R approaches 0.94 (c.f. refs. [7, 16]) at temperatures above about 750 K. Accepting the obtained temperature dependence of k_R (Fig. 6), the stress σ_d can be interpreted as the stress required to detach a dislocation from an interacting particle athermally.

The creep strain rate at temperatures 773–823 K at which the true threshold stress is absent, can be described in terms of thermally activated detachment of

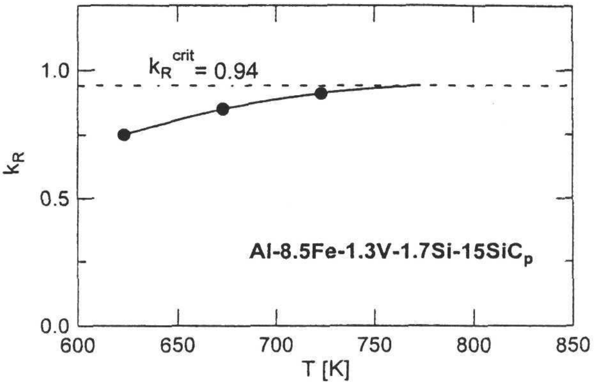

Fig. 6. Calculated values of the relaxation factor k_R plotted against temperature. k_R^{crit} is the "critical" value of this factor (see Section 4).

dislocations from interacting particles as [6]

$$\frac{\dot{\varepsilon}_m b^2}{D_L} = C \exp\left\{-\frac{Gb^2 d[(1-k_R)(1-\sigma/\sigma_d)]^{\frac{3}{2}}}{2kT}\right\}, \tag{3}$$

where C is a structure factor expressed as

$$C = 6\lambda\rho b; \tag{4}$$

ρ is the density of mobile dislocations and λ the mean interparticle spacing. In Eq. (3) d is the mean particle diameter. Rösler and Arzt [6] proposed a procedure to estimate k_R and σ_d. Taking into account the above-mentioned temperature dependence of the relaxation factor k_R (Fig. 6), the present authors [7, 16] modified this procedure and for the composite under consideration obtained an excellent agreement of the present creep data with the Rösler-Arzt model, Fig. 5. However, the structure factor C following from the fitting the $\dot{\varepsilon}_m(T, \sigma)$ creep data with the creep Eq. (3) was found to differ from the factor calculated from the structure data $C = 6\lambda\rho b = 1.63 \times 10^{-3}$ by two orders of magnitude in average. This should be considered not a serious deficiency of the creep model under consideration.

5. Summary

In the present work, the creep behaviour of an Al-8.5Fe-1.3V-1.7Si alloy reinforced with 15 vol. % silicon carbide particulates (SiC$_p$) is investigated at temperatures ranging from 623 to 823 K. It is found that, in contrast to the creep behaviour at temperatures 623–723 K, the creep at temperatures 773 to 823 K is not associated with a true threshold stress. The disappearance of the true threshold creep behaviour of the Al-8.5Fe-1.3V-1.7Si-15SiC$_p$ composite at temperatures above 723 K is interpreted in terms of transition from the athermal to the thermally activated detachment of dislocations from small interacting, predominantly Al$_{12}$(Fe,V)$_3$Si phase particles, in the composite matrix.

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