

ANOMALOUS HIGH TEMPERATURE CREEP BEHAVIOUR OF AN Al-8.5Fe-1.3V-1.7Si (8009Al TYPE) ALLOY REINFORCED WITH ALUMINA SHORT FIBRES

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Creep in an Al-8.5Fe-1.3V-1.7Si (8009Al type) alloy dispersion strengthened with fine $Al_{12}(Fe,V)_3Si$ phase particles (Alloy in the following) at temperatures 798–848 K is not associated with a true threshold and the same holds for the Alloy reinforced with alumina short fibres (Composite in the following). While the true activation energy of creep in the Alloy is higher than the activation enthalpy of lattice self-diffusion in aluminium due to the energy of athermal detachment of dislocations from dispersed particles, this is not the case of the Composite. The true activation energy of creep in the latter is anomalously low and the creep behaviour cannot be described by the creep equation that holds for the Alloy, but contains the factor by which the flow stress in the dispersion strengthened matrix is reduced due to load transfer. Nevertheless, the creep strength of the Composite is, at the certain conditions at least, by a factor up to six orders of magnitude higher than that of the Alloy. This creep strength increment due to alumina short fibre reinforcement is accounted for in terms of the load transfer effect.

Key words: Al-8.5Fe-1.3V-1.7Si (8009Al type) alloy, alloy matrix composite, high temperature creep, load transfer, creep strength

ANOMÁLNÍ CREEPOVÉ CHOVÁNÍ SLITINY Al-8,5Fe-1,3V-1,7Si (TYPY 8009Al) VYZTUŽENÉ KRÁTKÝMI VLÁKNY OXIDU HLINÍKU PŘI TEPLOTÁCH VYŠŠÍCH NEŽLI 780 K

Creep slitiny Al-8,5Fe-1,3V-1,7Si (typu 8009Al) – složení v hm.% – disperzně zpevněné jemnými částicemi fáze $Al_{12}(Fe,V)_3Si$ při teplotách 798–848 K není spojen se skutečným prahovým napětím a totéž platí o téže slitině vyztužené krátkými vlákny oxidu hliníku – kompozitu. Zatímco je skutečná aktivační energie creepu slitiny vyšší nežli aktivační entalpie mřížkové autodifuze v hliníku o energii atermálního odpoutávání dislokací od disperzních částic, v případě kompozitu tomu tak není. Skutečná aktivační energie creepu kompozitu je anomálně nízká a jeho creepové chování nelze popsat rovnicí, která

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platí pro slitinu, obsahuje však faktor, jímž je sníženo deformační napětí v disperzně zpevněné matici v důsledku transferu zatížení z matrice na výztuhu. Nicméně creepová pevnost kompozitu je za určitých podmínek přinejmenším o pět řádů vyšší nežli creepová pevnost slitiny. Tento vzrůst creepové pevnosti je přičten transferu zatížení z disperzně zpevněné matrice na výztuhu (krátká vlákna oxidu hliníku) a z tohoto přenosu vyplývajícímu snížení deformačního napětí v matici kompozitu.

1. Introduction

The Al-8.5Fe-1.3V-1.7Si (8009Al type) alloy – composition in mass % – belongs to advanced aluminium alloys processed by rapid solidification and powder metallurgy route and dispersion strengthened with fine incoherent particles, specifically the $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles in the alloy under consideration. This alloy exhibits very good creep resistance at temperatures up to at least 700 K [1–6]. Also, up to this temperature its Young's modulus can be increased significantly by discontinuous reinforcement with mechanically strong particulates, whiskers or short fibres, specifically, e. g., with silicon carbide particulates of micrometer dimensions [4–6, 7]. Moreover, under specific conditions the discontinuous reinforcement may enhance the creep resistance of the alloy due to load transfer from the matrix to the reinforcement [7]. Such a load transfer reduces the flow stress of the matrix and just this effect may result in the enhanced creep resistance. The Al-8.5Fe-1.3V-1.7Si alloy reinforced with ceramic particulates, whiskers or short fibres represents a discontinuous composite with the dispersion strengthened matrix.

In creep of an Al-8.5Fe-1.3V-1.7Si alloy at temperatures ranging from 623 to 748 K the true threshold stress is observed [7–9], for which the apparent stress exponent of minimum creep strain rate

$$m_c = \left(\frac{\partial \ln \dot{\epsilon}_m}{\partial \ln \sigma} \right)_T \quad (1)$$

decreasing with increasing applied stress σ is characteristic; T is the temperature. The threshold is associated with the presence of fine incoherent $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles. Also in creep of this alloy reinforced with mechanically strong short fibres (to give an example of proper reinforcement) the true threshold stress is observed [9]. Hence, the detachment of dislocations from the interacting $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ particles is generally supposed to be athermal in the alloy as well as in the alloy matrix composite [10]. The apparent activation energy of creep

$$Q_c = \left[\frac{\partial \ln \dot{\epsilon}_m}{\partial (-1/RT)} \right]_\sigma \quad (2)$$

(R is the gas constant) is higher or even much higher than the activation enthalpy ΔH_L of lattice self-diffusion in aluminium [11]. Nevertheless, the *true* activation

energy (activation enthalpy) of creep, ΔH_c , is equal to ΔH_L and the true applied stress exponent n is close to 5. The difference between Q_c and ΔH_L on one side and m_c and n on the other is readily explained in terms of the true threshold stress and its temperature dependence [7], respectively.

However, in creep of the alloy under consideration at temperatures above ~ 780 K, the true threshold stress is not observed. The apparent stress exponent m_c generally increases with increasing applied stress, the true activation energy of creep ΔH_c is greater than the activation enthalpy of lattice self-diffusion in the matrix (aluminium) and the true stress exponent n is higher than 5 [9]. Hence, the detachment of dislocations from the interacting $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles at these temperatures can be supposed to be thermally activated [10]. For such creep strain rate controlling process Rösler and Arzt [10] developed an equation that can be in a slightly different form written as

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

where C is the structure factor, r_p is the particle radius, k_R is the relaxation factor, σ_d is the detachment stress, and k is the Boltzmann constant. The structure factor is expressed as

$$C = 6\lambda\rho b; \quad (4)$$

λ is the mean interparticle spacing and ρ is the density of mobile dislocations tacitly assumed to be stress-independent by Rösler and Arzt [10]. The detachment stress can be expressed as [12]

$$\sigma_d = \sigma_{\text{OB}} \sqrt{1 - k_R^2}, \quad (5)$$

where σ_{OB} is the Orowan stress [13]. The relaxation factor k_R characterizes the strength of the attractive dislocation/particle interaction.

In a previous paper [7], the creep behaviour of an Al-8.5Fe-1.3V-1.7Si alloy reinforced with alumina short fibres at temperatures ranging from 648 to 748 K was investigated. It was shown that at these temperatures the creep strain rate of the reinforced alloy is at least six orders of magnitude higher than that of the unreinforced alloy and that this creep strength increment is essentially due to the effect of load transfer on the true threshold stress [7].

The aim of this paper is to present some $\dot{\varepsilon}_m(\sigma, T)$ creep data for the same alloy matrix composite obtained at the testing temperatures ranging from 773 to 848 K and to compare these data with those for the matrix alloy, especially, to attempt to correlate the $\dot{\varepsilon}_m(\sigma, T)$ creep data for the composite with the prediction of the recent Rösler-Bäker theoretical concept of “dual-scale particle strengthening” [14].

2. Experimental materials and procedures

Materials and experimental procedures are described in detail in the previous paper [7] aimed at the load transfer effect in the true threshold creep behaviour of an Al-8.5Fe-1.3V-1.7Al alloy (Alloy in the following) and the same alloy reinforced with alumina short fibres (Composite in the following). It should be emphasized that the same powder processed by atomisation was used to produce both the Alloy and the Composite by the powder metallurgy route.

The constant tensile stress creep tests were performed at temperatures 773, 798, 823, and 848 K. The testing temperatures were controlled to within 0.5 K. The creep elongation was measured by means of linear variable differential transducers coupled to a digital data acquisition system.

The technique of incremental stress creep tests was applied to measure the minimum creep strain rate, although some tests at properly chosen conditions of applied stress and temperature were performed running them well into tertiary creep stage if not to fracture. The steady state creep was not observed and thus only the minimum creep strain rate $\dot{\epsilon}_m$ could be defined at any testing temperature under consideration.

3. Results and analysis

With respect to the fact that the $\dot{\epsilon}_m(T, \sigma)$ creep data considered here were presented in full elsewhere [7], only some of them will be shown here in proper representation.

3.1 Alloy

The minimum creep strain rates $\dot{\epsilon}_m$ measured at the temperatures 798, 823 and 848 K and normalized to the coefficient of lattice self-diffusion in aluminium [11], D_L , and the length of the Burgers vector, b , squared, $\dot{\epsilon}_m b^2 / D_L$, plotted against the applied stress σ normalized to the shear modulus G of aluminium [15], σ / G , are shown in Fig. 1. From this figure it can be seen that at the temperatures under consideration the true activation energy (activation enthalpy) of creep in the Alloy is higher than the activation enthalpy of lattice self-diffusion in aluminium. In fact, the $(\dot{\epsilon}_m b^2 / D_L, \sigma / G)$ data points cannot be fitted by a single curve for all the temperatures. This is in qualitative accord with the creep model of Rösler and Arzt [10]. In Fig. 1, also the relation between $\dot{\epsilon}_m b^2 / D_L$ and σ / G is shown for the temperature of 773 K. While for the temperatures ranging from 798 to 848 K the apparent stress exponent increases with increasing applied stress, for 773 K it is obviously applied-stress-independent. This strongly suggests that at this “transient” temperature the creep strain rate controlling mechanism at temperatures up to 748 K contributes to the measured creep strain rates in an approximately the same extent as the mechanism controlling creep strain rate at temperatures above

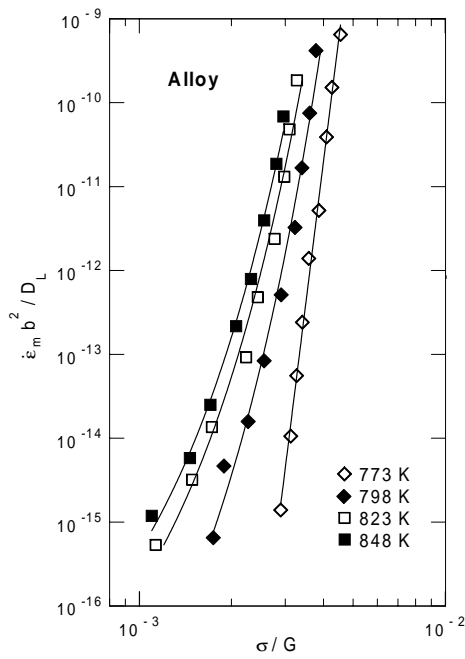


Fig. 1. Alloy. The minimum creep strain rates $\dot{\epsilon}_m$ normalized to the coefficient of lattice self-diffusion in aluminium, D_L , and to the length of the Burgers vector, b , squared, $\dot{\epsilon}_m b^2 / D_L$, plotted against the applied stress σ normalized to the shear modulus G , σ/G .

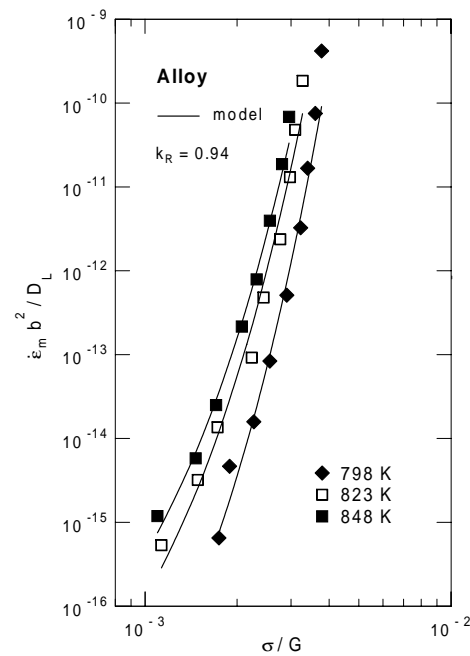


Fig. 2. Alloy. Correlation of the $(\dot{\epsilon}_m b^2 / D_L, \sigma/G)$ creep data with the model of Rösler and Arzt [10] accepting the relaxation factor $k_R = 0.94$ (cf. Eq. (8)).

798 K. This is why the temperature of 773 K – the “transition” temperature – will be mostly left aside in the following analysis.

In Fig. 2, the $(\dot{\epsilon}_m b^2 / D_L, \sigma/G)$ creep data are compared with the model creep equation (3), accepting the relaxation factor $k_R = 0.94$ [5] and the values of the detachment stress σ_d estimated using the equation developed by the present authors elsewhere (see further Eq. (8)). The agreement of the experimental $\dot{\epsilon}_m(T, \sigma)$ creep data with the prediction of the Rösler-Arzt model [10] is good. Of course, the values of the structure factor C following from the fitting procedure, namely 2.8, 4.5 and 1.6 for 798, 823 and 848 K, respectively, cf. [9], should be close to that calculated from the structure data, Eq. (4), before the correlation could be considered fully satisfactory. Accepting a mobile dislocation density $\rho = 2 \times 10^{14} \text{ m}^{-2}$ [9] and the interparticle spacing $\lambda = 181 \text{ nm}$ [7], the value of C obtained by means of Eq. (4) is equal to 6.2×10^{-2} . Thus the structure factor C following from the fitting

procedure differs from that calculated from the structure data by a factor ranging from ~ 72 to ~ 26 , in average equal to ~ 47 . This discrepancy can be hardly considered significant.

There is still another possibility to compare the present creep data with the prediction of the Rösler-Arzt model. According to Eq. (3), the activation energy of detachment of a dislocation from an attractive particle is expressed as

$$Q_d = Gb^2 r_p [(1 - k_R)(1 - \sigma/\sigma_d)]^{3/2}. \quad (6)$$

The values of the activation energy of detachment calculated by means of Eq. (6) can be compared with those following from the $\ln(\dot{\epsilon}_m b^2/D_L)$ vs. $1/T$ relations. For $\sigma/G = 2.25 \times 10^{-3}$, Q_d are estimated to 514, 434 and 322 $\text{kJ}\cdot\text{mol}^{-1}$ for temperatures 798, 823 and 848 K, respectively (cf. [9]). These values of Q_d differ from those predicted by Eq. (6) by a factor ranging from ~ 2.1 to ~ 0.3 depending on temperature, and in average equal to 1.33. This result is again quite satisfactory and supports the Rösler-Arzt model of creep in dispersion strengthened alloys.

3.2 Composite

Introducing mechanically strong reinforcement into the Alloy, specifically alumina short fibres of micrometer dimensions, the reinforcement strengthening can be accounted for by the creep equation (3) in which $\dot{\epsilon}_m$ is replaced by $\dot{\epsilon}_m(\text{RS})$, C by $C(\text{RS})$ and the applied stress σ by the matrix flow stress σ/Λ ; (RS) means reinforcement strengthening. The structure factor $C(\text{RS})$ is expressed by the equation of the form of Eq. (4). By the load transfer factor Λ expressed as [14]

$$\Lambda = 1 + 2[2 + (l_R/d_R)] f_R^{3/2} \quad (7)$$

the flow stress of the matrix is reduced due to the load transfer from the matrix to the reinforcement. In Eq. (7), l_R/d_R is the aspect ratio of the reinforcing short fibres (l_R is the mean length and d_R the mean diameter of fibres) and f_R is the volume fraction of the reinforcement. The value of Λ was estimated as the ratio of the true threshold stress of the Composite and the true threshold stress of the Alloy to 2.1 [7], see also [16].

For the Composite, the minimum normalized creep strain rates, $\dot{\epsilon}_m b^2/D_L$, are plotted against normalized applied stresses, σ/G , for the temperatures ranging from 773 to 848 K in Fig. 3. At the first glance, it may seem that all the data points can be rather well fitted by a single curve and, hence, that the creep strain rate is probably lattice diffusion controlled. However, the true stress exponent n estimated from $\dot{\epsilon}_m b^2/D_L$ vs. σ/G relations is as high as about 28. Such a high true stress exponent is not predicted by any known model of creep controlled by lattice

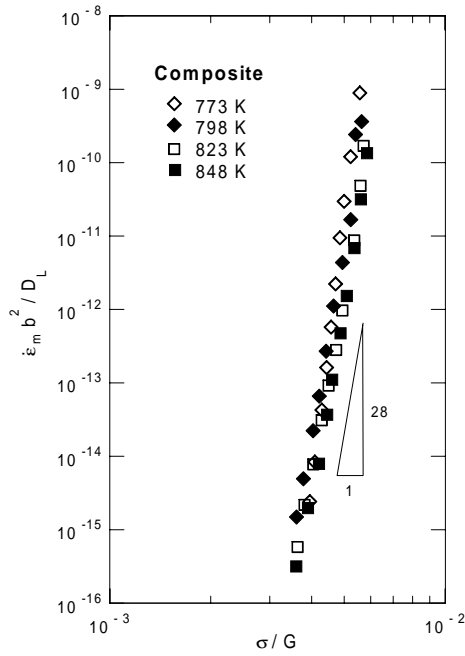


Fig. 3. Same as in Fig. 1, but for the Composite. The $(\dot{\epsilon}_m b^2 / D_L, \sigma / G)$ creep data for the “transient” temperature of 773 K are also shown in the figure.

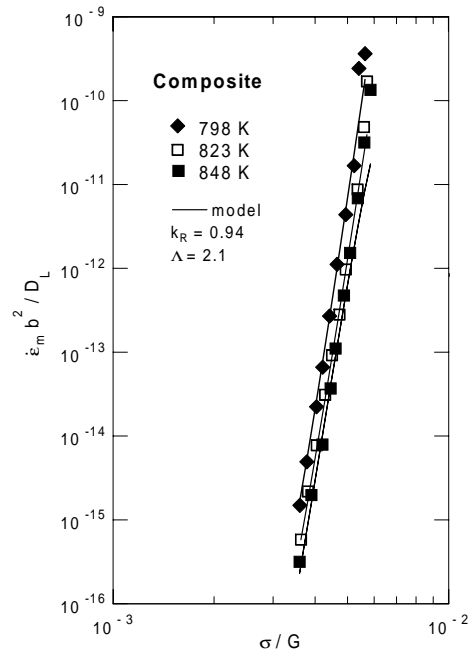


Fig. 4. Composite. An attempt to correlate the $(\dot{\epsilon}_m b^2 / D_L, \sigma / G)$ creep data with the Rösler-Arzt model creep equation (3), modified introducing the load transfer factor Λ . Again, the relaxation factor $k_R = 0.94$ is accepted; the factor $\Lambda = 2.1$.

self-diffusion. But the closer inspection of the figure reveals that the normalized minimum creep strain rates $\dot{\epsilon}_m b^2 / D_L$ for the temperature of 848 K are *lower* than the normalized minimum creep strain rates for the transient temperature of 773 K, especially, at higher normalized applied stresses σ / G . Such temperature dependences of $\dot{\epsilon}_m b^2 / D_L$ vs. σ / G relations are certainly anomalous. No similar anomaly is observed in creep of the Alloy (Fig. 1). This difference between the Composite and the Alloy cannot be explained by a difference in reliability of the minimum creep strain rate measurements. In fact, the $\dot{\epsilon}_m(T, \sigma)$ measurements were performed using the same technique and the same equipment in the same laboratory.

With regards to the creep data presented in Fig. 3, it can be hardly expected that they can be correlated with the Rösler-Arzt creep equation (3) as modified by introducing the load transfer factor Λ (see above). Nevertheless, in Fig. 4, the relations between $\dot{\epsilon}_m b^2 / D_L$ are plotted against σ / G for 798, 823 and 848 K accepting $k_R = 0.94$ and $\Lambda = 2.1$ and setting $\sigma = 70$ MPa. To estimate the detachment stress

values, the following equation was employed (cf. [5]):

$$\left(\frac{\sigma}{\Lambda\sigma_d}\right)^3 - \left(\frac{\sigma}{\Lambda\sigma_d}\right)^2 = -\frac{(2kTm_c/3Gb^2r_p)^2}{(1-k_R)^3}. \quad (8)$$

As it can be seen, the data points for various temperatures can be very well fitted by straight lines, but the normalized $\dot{\epsilon}_m b^2/D_L$ creep strain rates are higher for 798 K than for 848 K, which is anomalous. Using Eq. (8), the values of the detachment stress σ_d equal to 66.5, 65.0 and 50.5 MPa were obtained for the matrix flow stress $\sigma/\Lambda = 70/2.1 = 33.3$ MPa and the temperatures 798, 823 and 848 K, respectively. These values of σ_d are in average ~ 1.8 times higher than σ/Λ for the Composite, while for the matrix Alloy they are in average ~ 5.7 times higher than the applied stress of 30 MPa. In the matrix Alloy, the applied stress is identical with the flow stress. Thus the above comparison of the values of σ_d is relevant.

For the structure factor $C(\text{RS})$ very low values follow from the fitting procedure of the creep data to the modified creep equation (3), namely 4.2×10^{-7} , 5.3×10^{-8} and 5.9×10^{-11} for 798, 823 and 848 K, respectively, while for this factor calculated from the structure data a quite reasonable value of 5.5×10^{-2} is obtained accepting the volume fraction of the $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles of 0.05 [6] (see later) and the particle radius of 25 nm. Such differences in the structure factor are certainly not tolerable.

Thus, to close the present Section, it must be concluded that the attempt to correlate the *anomalous* $\dot{\epsilon}_m(T, \sigma)$ creep data for the composite at temperatures ranging from 798 to 848 K with the modified model creep equation (3) clearly shows that the correlation is not possible.

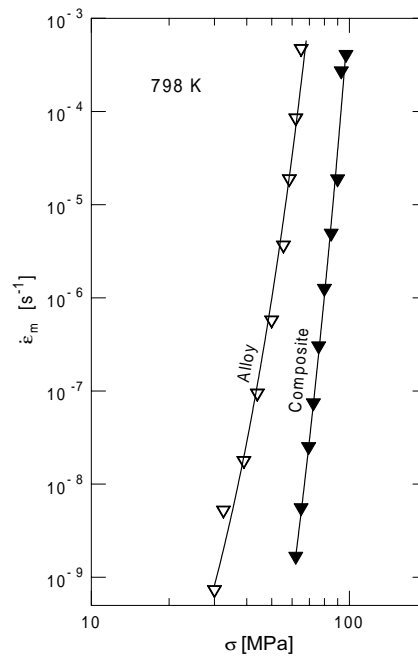
However, despite of this fact, the Rösler-Bäker [14] concept of the “dual scale particle strengthening” seems to preserve its validity despite of the above anomaly, Section 3.3.

3.3 Creep strengthening due to reinforcement

In the following, the creep strength (or creep resistance) is characterised by the minimum creep strain rate $\dot{\epsilon}_m$. The lower $\dot{\epsilon}_m$ at a given temperature and a given applied stress – the higher the creep strength.

To give an example of the effect of alumina short fibre reinforcement on creep strength of the matrix Alloy at 798 K, Fig. 5 is presented. From the figure it can be seen that at ~ 60 MPa the creep strength of the Composite ($\dot{\epsilon}_m \cong 10^{-9} \text{ s}^{-1}$) is about five orders of magnitude higher than that of the Alloy. Surprisingly enough, at a temperature of 848 K this difference is still almost by other two orders of magnitude greater [7].

Fig. 5. The relation between the minimum creep strain rate $\dot{\epsilon}_m$ and the applied stress σ at the temperature of 798 K for the Composite and a similar relation for the Alloy. At $\sigma = 60$ MPa, the creep strength (as characterized by the minimum creep strain rate $\dot{\epsilon}_m$) of the Composite is five orders of magnitude lower than that of the Alloy. At the above stress σ , the minimum creep strain rate $\dot{\epsilon}_m$ in the Composite is close to 10^{-9} s^{-1} .



4. Discussion

In the present investigation, the minimum creep strain rates were measured at four temperatures ranging from 773 to 848 K. This temperature interval is narrow and, besides, if these testing temperatures are compared with the melting temperature of aluminium, i.e. 933 K, the homologous temperatures range from ~ 0.83 to 0.91. At such high homologous temperatures the diffusional processes are already very fast and, consequently, at least two effects detrimental to the creep strength must be expected. First, a marked dimensional instability of the dispersed $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles in the Alloy matrix and, second, partial relaxation of the load transfer associated with the presence of the alumina short fibres in the Composite. This relaxation is due to the diffusional flow of matter along the matrix/alumina-fibre interfaces while the growth of the $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles occurs due to the lattice diffusion.

The first of the above effects certainly plays a role in creep of the Alloy as well as in the Composite at the testing temperatures considered in the present investigation. The volume fraction of the $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles in the alloy is approximately 0.27. Comparing the experimentally determined yield stress of the alloy of the composition nominally similar to that of the Alloy of interest here with the Orowan stress calculated from the structure data, Spigarelli [6] concluded that

from the nominal volume fraction only some of the $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles represent active obstacles to dislocation motion at room temperature. The author estimated the volume fraction of such active particles to about 0.04 in the alloy and to about 0.05 in the alloy reinforced with 15 vol.% silicon carbide particulates. The remaining particles, mostly located at grain boundaries [6], do not play any role as obstacles to dislocation motion in plastic deformation.

However, a question naturally arises whether the fraction of the active $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles is affected by their dimensional instability, and/or whether the above result of Spigarelli [6] is relevant for interpretation of the $\dot{\epsilon}_m(T, \sigma)$ creep data presented above. The answer to this question could be obtained using the procedure similar to that applied by Spigarelli, but at temperatures of interest here. Unfortunately, the present authors have not yet been in a position to perform relevant experiments. Hence, they accepted an assumption that the dimensional instability has no significant effect on the volume fraction of the “active” $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles. Under this assumption, at least one example can be given to illustrate the effect of dimensional instability of dispersed particles on the parameters of the modified creep equation (3). If at 848 K the mean $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particle radius increases from 25 to 50 nm and the volume fraction of the “active” particles remains the same, i. e. 0.05 for the Composite, the structure factor $C(\text{RS})$ following from the fitting procedure of the experimental $\dot{\epsilon}_m(\sigma)$ creep data increases from 5.9×10^{-11} to 4.2×10^{-7} , i.e. approximately by four orders of magnitude. On the other hand, the structure factor calculated by means of Eq. (4) is not lower than about 1.11×10^{-1} . This shows once again that the *modified* Rösler-Arzt equation (3) does not describe the present $\dot{\epsilon}_m(T, \sigma)$ creep data for the Composite adequately. In fact, the difference between the structure factor calculated from the structure data amounts to more than six orders of magnitude. Such a difference is certainly not tolerable.

The coarsening of the dispersed $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles in the composite acts in parallel with the relaxation of the load transfer due to diffusional flow of matter along the matrix/short-fibre interfaces. The results presented in Section 3.2 do not make it possible to separate the effects of these processes without their rather complicated analysis. Intuitively, the present authors believe that the “stability of the load transfer effect” plays a role superior to that of the “dimensional stability effect” of the dispersed $\text{Al}_{12}(\text{Fe},\text{V})_3\text{Si}$ phase particles. These authors’ belief is at least partly supported by a careful inspection of the creep data comparison illustrated in Fig. 5 (cf. Section 3.3).

5. Summary and conclusions

In the present investigation of creep behaviour of an Al-8.5Fe-1.3V-1.7Si (8009Al type) alloy and this Alloy reinforced with 15 vol.% alumina short fibres – the Composite – the minimum creep strain rates were measured at temperatures

ranging from 773 to 848 K. At these temperatures, the true threshold creep behaviour was not observed neither in the composite or its matrix alloy. In the latter, the true stress exponent increases with increasing applied stress and the true activation energy (activation enthalpy) of creep is higher than the activation enthalpy of lattice self-diffusion in aluminium. For the matrix alloy, the experimental creep data were very well correlated with the creep model presented by Rösler and Arzt.

On the other hand, the creep behaviour of the Alloy reinforced with alumina short fibres – the Composite – was found anomalous as to the temperature dependence of minimum creep strain rates: values of the apparent activation energy of creep were found unrealistically low or even negative. Consequently, the creep behaviour of the Composite could not be described by the equation that holds for the Alloy but is modified introducing the load transfer factor by which the flow stress of the matrix of the Composite is reduced. However, despite of the above anomalous creep strain rate behaviour, the creep strength of the composite differs from that of the alloy significantly, which is accounted for solely by the load transfer. Introducing into the matrix Alloy 15 vol.% alumina short fibres (3 μm in diameter and $\sim 22 \mu\text{m}$ in length) enhances the creep strength significantly at all the temperatures under consideration: the minimum creep strain rate decreases at least by five orders of magnitude.

The factors that may be responsible for the anomalous creep behaviour of the Composite have not been identified. If this anomaly is confirmed in an independent way, such factors will have to be found and introduced into the relevant creep equation.

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