HIGH TEMPERATURE CREEP OF AN Al-8.5Fe-1.3V-1.7Si ALLOY

KVĚTA KUCHAŘOVÁ[1], SHIJIE ZHU[2], JOSEF ČADEK[1]*

A previous investigation of creep behavior in an Al-8.5Fe-1.3V-1.7Si (8009Al type) alloy processed by rapid solidification and strengthened with small incoherent $Al_{12}(Fe,V)_3Si$ phase particles (Metall. Mater. Trans. A, *31A*, 2000, p. 2229), is extended to temperatures ranging from 773 to 923 K using the isothermal constant-stress-creep test technique. While in the temperature interval 623–723 K the creep behavior of the alloy is associated with the true threshold stress, at 773–923 K the true threshold stress is absent: the apparent stress exponent of minimum creep strain rate, m_c, decreases with decreasing applied stress. Assuming a temperature independent value of the relaxation factor k_R characterizing the strength of attractive dislocation/$Al_{12}(Fe,V)_3Si$ particle interaction, an attempt is made to interpret the minimum creep strain rate behavior in terms of the Rösler-Arzt model (Acta Metall. Mater., *38*, 1990, p. 67). The model fits the experimental creep data very well. The structure factor C following from the fitting procedure differs from that calculated from the structure data by a factor in average equal to ~ 60. The activation energy of detachment of dislocations from interacting $Al_{12}(Fe,V)_3Si$ phase particles, following from the experimental creep data analysis, differs from that predicted by the model by a factor in average equal to 1.33. Thus, the thermally activated detachment of dislocations from the fine incoherent (interacting) particles is identified as the creep strain rate controlling process.

K e y w o r d s: Al-8,5Fe-1,3V-1,7Si alloy, creep behavior, true threshold stress, load transfer effect

VYSOKOTEPLOTNÍ CREEP SLITINY Al-8,5Fe-1,3V-1,7Si

Dřívější studie creepového chování slitiny Al-8,5Fe-1,3V-1,7Si připravené technikou rychlé solidifikace a zpevněné jemnými nekoherentními částicemi fáze $Al_{12}(Fe,V)_3Si$ (Metall. Mater. Trans. A, *31A*, 2000, s. 2229) je rozšířena na teploty od 773 do 923 K. Zatímco v teplotním intervalu 623–723 K je creepové chování této slitiny spojeno se skutečným prahovým napětím, při teplotách 773–923 K není skutečné prahové napětí pozorováno: zdánlivý napěťový exponent minimální rychlosti creepu m_c klesá s klesajícím napětím.

[1] Institute of Physics of Materials, Academy of Sciences of the Czech Republic, Žižkova 22, 616 62 Brno, Czech Republic
[2] Institute of Industrial Sciences, The University of Tokyo, 4-6-1 Komaba, Meguro-ku, Tokyo, 153-8904 Japan
* corresponding author

Za předpokladu teplotně nezávislé hodnoty relaxačního faktoru k_R, charakterizujícího pevnost atraktivní interakce dislokací s částicemi fáze $Al_{12}(Fe,V)_3Si$, je učiněn pokus interpretovat creepové chování přijetím Röslerova-Arztova modelu (Acta Metall. Mater., *38,* 1990, s. 671). Je ukázáno, že model popisuje experimentální creepová data velmi dobře. Strukturní faktor C, získaný korelací experimentálních dat s modelem, se liší od faktoru vypočítaného ze strukturních dat asi 60krát. Aktivační energie odpoutávání dislokací od interagujících částic fáze $Al_{12}(Fe,V)_3Si$, vyplývající z experimentálních creepových dat, se liší od této aktivační energie předpovězené modelem v průměru 1,33krát. Jako proces řídící minimální rychlost creepu je tudíž identifikováno tepelně aktivované odpoutávání dislokací od jemných nekoherentních (interagujících) částic.

1. Introduction

The Al-8.5Fe-1.3V-1.7Si (8009Al type) alloy processed by rapid solidification and powder metallurgy route exhibits remarkable creep resistance up to temperatures of around 700 K (e.g., Refs. 1, 2 and 3). This resistance is due to the high volume fraction (~ 0.27) of fine (less than 50 nm in diameter) incoherent particles of the $Al_{12}(Fe,V)_3Si$ phase and low coarsening rate of these particles at high temperatures [1, 2, 3, 4].

The high temperature creep behavior of the Al-8.5Fe-1.3V-1.7Si alloy was investigated by a number of authors, specifically by Carreño et al. [1], Peng et al. [2, 3], Carreño and Ruano [4] Ma and Tjong [5] and Zhu et al. [6]. The present work is aimed at explaining the creep behavior of the alloy under consideration at temperatures above 723 K.

Carreño et al. [1] investigated high temperature deformation behavior of an Al-8.5Fe-1.3V-1.7Si alloy at seven temperatures ranging from 523 to 823 K. The authors used the tensile strain rate change technique and presented the flow stresses σ for the strain rates $\dot{\varepsilon}$ ranging from 10^{-6} to 10^{-2} s^{-1}. The relations between $\dot{\varepsilon}$ and σ were found linear in double logarithmic co-ordinates. The apparent stress exponent m_c decreases with increasing temperature from about 24 at 523 K to about 13 at 823 K. In the same temperature interval the apparent activation energy Q_c increases from about 140 $kJ \cdot mol^{-1}$ to about 800 $kJ \cdot mol^{-1}$. In a more recent paper, Carreño and Ruano [4] using a similar tensile strain rate change technique, also found linear relations between $\dot{\varepsilon}$ and σ in double logarithmic co-ordinates. The authors performed strain rate decrements of one order of magnitude, the strain per one decrement was about 0.03. They stated that "after a change in strain rate the steady state was quickly developed after small deformation". However, Carreño and Ruano [4] have not presented any evidence on the attainment of steady state following each strain rate decrement at any temperature. The apparent activation energy Q_c and the apparent stress exponent m_c were, with the exception of the lowest testing temperature of 623 K, approximately independent of temperature at least up to 823 K and equal to ~ 400 $kJ \cdot mol^{-1}$ and ~ 25, respectively.

Very recently, Ma and Tjong [5] published rather extensive results of an investigation of creep in an Al-8.5Fe-1.3V-1.7Si-15SiC_p composite (p stands for particulates) together with the results for the unreinforced alloy. The authors applied the constant-tensile-load creep test technique. At temperatures 723, 748, 773, and 823 K the measured minimum creep strain rates for the matrix alloy ranged from $\sim 10^{-4}$ to $\sim 10^{-8}$ s^{-1}. The relations between the minimum (or "steady state" [5]) creep strain rate $\dot{\varepsilon}_m$ and applied stress σ are reproduced in Fig. 1. It is interesting to note that the $\dot{\varepsilon}_m\,(T, \sigma)$ relations for the alloy are practically identical with those of the composite. The apparent stress exponent m_c decreases with increasing stress, which clearly suggests the true threshold creep behavior at all temperatures under consideration. The temperature and stress dependence of the minimum creep strain rate $\dot{\varepsilon}_m$ can be described as [5, 6, 7, 8]

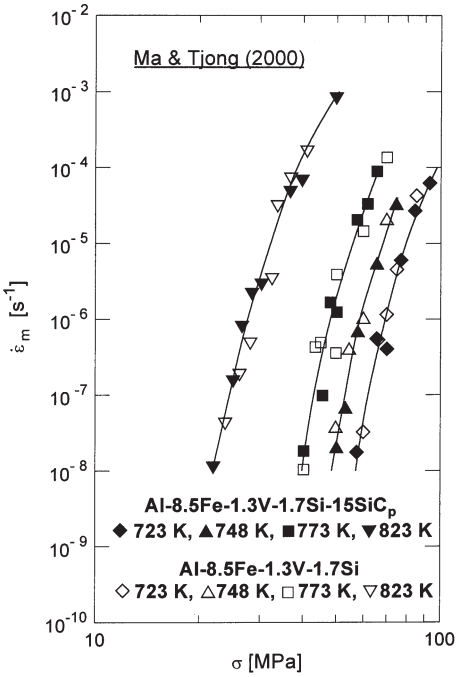

Fig. 1. An Al-8.5Fe-1.3V-1.7Si alloy and the same alloy reinforced with 15 vol.% silicon carbide particulates (SiC_p). Relations between minimum creep strain rates $\dot{\varepsilon}_m$ and applied stress σ (Ma and Tjong [5]).

$$\frac{\dot{\varepsilon}_m b^2}{D_L} = A\left(\frac{\sigma - \sigma_{TH}}{G}\right)^n, \qquad (1)$$

where D_L is the coefficient of self-diffusion in the matrix lattice (aluminum [9]), b is the length of the Burgers vector in aluminum, A is a dimensionless constant, σ_{TH} the true threshold stress, G the shear modulus of aluminum [10] and $n = 5$ is the true stress exponent. Thus, the creep strain rate is matrix-lattice-diffusion controlled and depends on the fifth power of the effective stress $\sigma - \sigma_{TH}$.

Almost simultaneously, Zhu et al. [6] reported similar results. The authors measured minimum creep strain rates covering seven orders of magnitude at the temperatures 623–723 K. They described the creep behavior by Eq. (1) with the true stress exponent $n = 5$ and the threshold creep behavior interpreted in terms of athermal detachment of dislocations from fine incoherent particles [11, 12], predominantly the $Al_{12}(Fe,V)_3Si$ phase particles.

At temperatures 623–723 K, the Al-8.5Fe-1.3V-1.7Si-15SiC$_p$ [5, 13] composite exhibits a similar creep behavior as the matrix alloy. However, at 773–823 K the true threshold creep behavior was not observed in the composite [14]. In fact, the apparent stress exponent m_c increased with increasing applied stress. Such a difference in the creep behavior was interpreted in terms of the thermally activated detachment of dislocations from small incoherent particles, predominantly the $Al_{12}(Fe,V)_3Si$ phase particles at 773–823 K [14]. A similar transition from athermal to thermally activated detachment of dislocations from small incoherent particles, specifically alumina particles, was also observed [15].

Taking into account all the results mentioned above, a question arises of whether the disappearance of the true threshold creep behavior at high testing temperatures is or is not associated with the discontinuous reinforcement in the composites or, in other words, whether a similar disappearance of the true threshold behavior takes place also in the matrix alloy, specifically in Al-8.5Fe-1.3V-1.7Si. To answer this question, an investigation of the creep behavior of Al-8.5Fe-1.3V-1.7Si alloy was undertaken. The minimum creep strain rates were measured at six temperatures ranging from 773 to 923 K and applied stresses chosen such that the minimum creep strain rates covered not less than four orders of magnitude, and the lowest of them were close to, or slightly lower than, 10^{-9} s^{-1}.

2. Material and experimental procedures

The Al-8.5Fe-1.3V-1.7Si alloy investigated in the present work was processed by rapid solidification and powder metallurgy route. The rapid solidification was performed by argon atomisation in contrast to the alloy investigated at 623–723 K [6], which was solidified by the planar flow casting technique. The alloy powder was consolidated at 823 K and extruded at the same temperature to a rod 12 mm in diameter. The resulting mean grain diameter was found close to 0.7 μm. The structure of the as-extruded alloy was reasonably homogeneous although the $Al_{12}(Fe,V)_3Si$ phase particles (also fine alumina particles are necessarily present as a result of atomisation) are mostly aligned to the extrusion direction as it is, as a rule, the case of this alloy processed by powder metallurgy route (c.f., Fig. 1 in Ref. 6). Detailed studies of the structure of alloys of this type were performed by Peng et al. [2, 3], and, especially, Carreño and Ruano [4].

From the composite rod, specimens for tensile creep tests 4.0 mm in diameter and 25.0 mm of gauge length were machined. The constant tensile stress creep tests were performed at six temperatures ranging from 773 to 923 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 transducers coupled to a digital data acquisition system. As a rule, the creep tests were run well into the tertiary stage and interrupted. Generally, no steady state was observed, only the minimum creep strain rate could be defined (c.f., Ref. 6).

3. Results

The minimum creep strain rates $\dot{\varepsilon}_{\mathrm{m}}$ measured at various temperatures T and applied stresses σ are listed in Table 1. The relations between the minimum creep strain rate and the applied stress at various temperatures are shown in Fig. 2. The apparent stress exponent $m_{\mathrm{c}} = (\partial \ln \dot{\varepsilon}_{\mathrm{m}}/\partial \ln \sigma)_{\mathrm{T}}$ increases with increasing applied stress with an exception of the lowest testing temperature of 773 K, Fig. 3. At temperatures 798–923 K the apparent stress exponent increases with applied stress linearly (from ~ 4.5 at 12.5 MPa to ~ 23 at 65 MPa) and does not depend on temperature. At the lowest testing temperature of 773 K the exponent m_{c} apparently does not depend on applied stress; its value is close to 29. The linearity of the $\ln \dot{\varepsilon}_{\mathrm{m}}$ vs. $\ln \sigma$ relation at 773 K indicates a transition from the temperature region where the creep behavior is associated with the true threshold stress to the temperature region where the true threshold stress is absent. A similar transition was also observed in creep of an Al-8.5Fe-1.3V-1.7Si-15SiC$_{\mathrm{p}}$ composite, [13, 14], Fig. 4.

Table 1. Minimum creep strain rates $\dot{\varepsilon}_{\mathrm{m}}$ at various temperatures T and applied stresses σ

$T = 773$ K		$T = 798$ K		$T = 823$ K	
σ[MPa]	$\dot{\varepsilon}_{\mathrm{m}}[\mathrm{s}^{-1}]$	σ[MPa]	$\dot{\varepsilon}_{\mathrm{m}}[\mathrm{s}^{-1}]$	σ[MPa]	$\dot{\varepsilon}_{\mathrm{m}}[\mathrm{s}^{-1}]$
80.0	3.5×10^{-4}	65.0	4.5×10^{-4}	55.0	3.8×10^{-4}
75.0	8.2×10^{-5}	62.0	8.1×10^{-5}	52.0	9.9×10^{-5}
72.0	2.1×10^{-5}	58.5	1.8×10^{-5}	50.0	2.7×10^{-5}
68.0	2.8×10^{-6}	55.5	3.5×10^{-6}	46.5	4.9×10^{-6}
63.0	7.5×10^{-7}	50.0	5.5×10^{-7}	41.0	9.9×10^{-7}
60.0	1.3×10^{-7}	44.0	9.0×10^{-8}	37.5	1.9×10^{-7}
57.5	3.0×10^{-8}	39.0	1.7×10^{-8}	29.0	2.8×10^{-8}
55.0	5.7×10^{-9}	32.5	5.0×10^{-9}	25.0	6.6×10^{-9}
51.0	7.5×10^{-10}	30.0	7.0×10^{-10}	19.0	1.1×10^{-9}
$T = 848$ K		$T = 873$ K		$T = 923$ K	
σ[MPa]	$\dot{\varepsilon}_{\mathrm{m}}[\mathrm{s}^{-1}]$	σ[MPa]	$\dot{\varepsilon}_{\mathrm{m}}[\mathrm{s}^{-1}]$	σ[MPa]	$\dot{\varepsilon}_{\mathrm{m}}[\mathrm{s}^{-1}]$
48.5	2.6×10^{-4}	43.5	5.7×10^{-4}	38.0	3.3×10^{-4}
46.0	7.1×10^{-5}	41.0	1.5×10^{-4}	34.0	7.1×10^{-5}
42.0	1.5×10^{-5}	37.5	2.5×10^{-5}	30.0	1.7×10^{-5}
38.0	3.0×10^{-6}	32.5	4.3×10^{-6}	26.0	4.5×10^{-6}
34.0	8.2×10^{-7}	27.5	8.5×10^{-7}	22.5	1.3×10^{-6}
28.0	9.5×10^{-8}	24.0	2.0×10^{-7}	19.0	3.3×10^{-7}
24.0	2.2×10^{-8}	20.0	5.6×10^{-8}	16.0	1.1×10^{-7}
18.0	4.5×10^{-9}	15.5	1.3×10^{-8}	12.5	3.4×10^{-8}

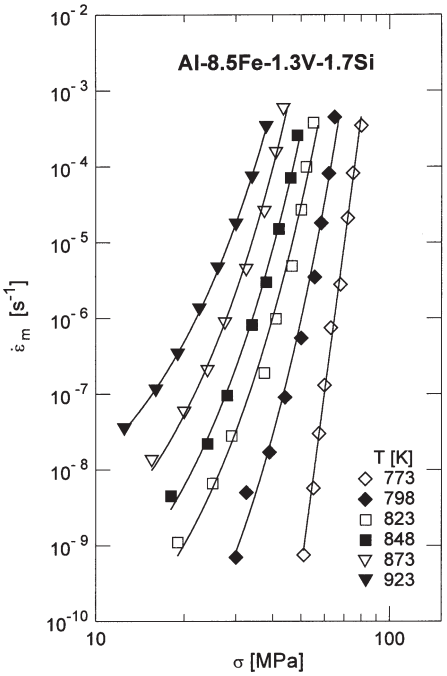

Fig. 2. Relations between minimum creep strain rates $\dot{\varepsilon}_\mathrm{m}$ and applied stress σ for temperatures ranging from 773 K to 923 K.

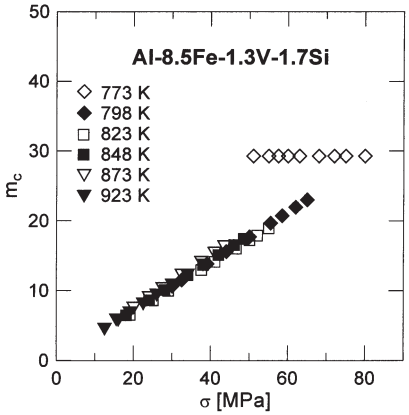

Fig. 3. The apparent stress exponent $m_\mathrm{c} = (\partial \ln \dot{\varepsilon}_\mathrm{m}/\partial \ln \sigma)_\mathrm{T}$ plotted against applied stress σ for six temperatures ranging from 773 to 923 K.

In Fig. 5, the minimum creep strain rates are plotted against temperature (in $\ln \dot{\varepsilon}_\mathrm{m}$, $1/T$ co-ordinates) for four applied stresses ranging from 20 to 50 MPa. The non-linearity of these plots indicates apparent activation energy of creep, $Q_\mathrm{c} = [\partial \ln \dot{\varepsilon}_\mathrm{m}/\partial(-1/RT)]_\sigma$ decreasing with increasing temperature. Values of Q_c could be estimated reliably enough from the $\ln \dot{\varepsilon}_\mathrm{m}$ vs. $1/T$ plots (Fig. 5) for two temperatures only, namely for 823 and 848 K. The values of Q_c obtained are plotted against applied stress in Fig. 6. While the energy Q_c apparently decreases rather strongly with increasing temperature (Q_c is equal to ~ 604 kJ·mol^{-1} at 823 K and to ~ 483 kJ·mol^{-1} at 848 K), it is apparently applied stress independent. According to the identity (e.g., Ref. 16)

$$\left[\frac{\partial(m_\mathrm{c}/\sigma)}{\partial(-1/RT)}\right]_\sigma = \left(\frac{\partial Q_\mathrm{c}}{\partial \sigma}\right)_\mathrm{T} \tag{2}$$

the apparent activation energy Q_c is independent of the applied stress once the apparent stress exponent m_c is independent of temperature. Hence, the result presented in Fig. 3 is consistent with that shown in Fig. 6.

From Fig. 6 it can be seen that the apparent activation energy Q_c at 823 K is approximately four times higher than the activation enthalpy of lattice self-diffusion in aluminum [9]. Without any doubt, this difference cannot be accounted for by the temperature dependence of the shear modulus G for aluminum. This is clearly demonstrated in Fig. 7, in which the normalized minimum creep strain rates $\dot{\varepsilon}_m b^2/D_L$ are plotted against the normalized applied stresses σ/G. In fact, the data points do not fit a single curve. At any normalized applied stress, the true activation energy (activation enthalpy) of creep, ΔH_c is much higher than the activation enthalpy of lattice self-diffusion, ΔH_L, in aluminum, i.e. 142 kJ·mol^{-1} [9].

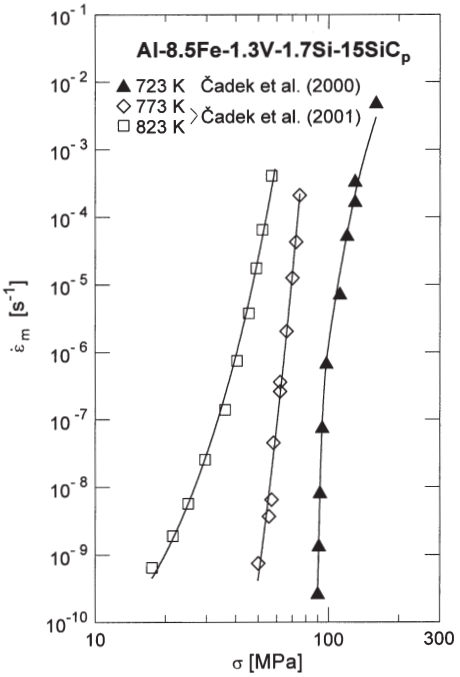

Fig. 4. An Al-8.5Fe-1.3V-1.7Si-15SiC$_p$ composite. Examples of $\dot{\varepsilon}_m$ vs. σ relations for three different temperatures (c.f., Refs. 13 and 14).

From the $\dot{\varepsilon}_m b^2/D_L$ vs. σ/G relations, Fig. 7, the true stress exponent n can be estimated in the usual way. As expected, the true stress exponent is found to increase with applied stress linearly and to be temperature independent if the temperature of 773 K is left aside. The $n(\sigma, T)$ relation is identical with the $m_c(\sigma, T)$ relation (Fig. 3) . Values of n range from ~ 4.5 at 923 K and $\sigma = 12.5$ MPa to ~ 24 at 798 K and $\sigma = 65$ MPa, similarly, as the apparent stress exponent m_c, Fig. 3. Apparently, the true stress exponent for the alloy is generally higher, or even much higher, than that for its matrix – aluminum – i.e. 4.4 [10] – ~ 4.9 [17].

In Fig. 8, values of $\dot{\varepsilon}_m b^2/D_L$ are plotted against the reciprocal temperature for two normalized applied stresses σ/G, namely 2.0×10^{-3} and 2.5×10^{-3}. The activation energy of detachment of dislocations from interacting particles, Q_d, is estimated as $[\partial \ln(\dot{\varepsilon}_m b^2/D_L)/\partial(-1/RT)]_{\sigma/G}$. The values of Q_d obtained are listed in Table 2. It can be seen that Q_d decreases rather strongly with increasing temperature (the temperature of 773 K is left aside); the values of Q_d are slightly higher at $\sigma/G = 2.0 \times 10^{-3}$ than at 2.5×10^{-3}. In the following, the average values of Q_d obtained for $\sigma/G = 2 \times 10^{-3}$ and $\sigma/G = 2.5 \times 10^{-5}$ are considered for any given temperature (e.g., 433 kJ·mol^{-1} for 823 K).

In Table 3, these average values are denoted Q_d^{av}. For the temperatures

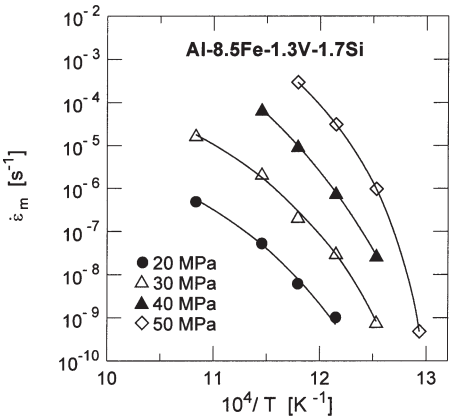

Fig. 5. Minimum creep strain rates $\dot{\varepsilon}_\mathrm{m}$ plotted against the reciprocal temperature $1/T$ for applied stresses ranging from 20 to 50 MPa.

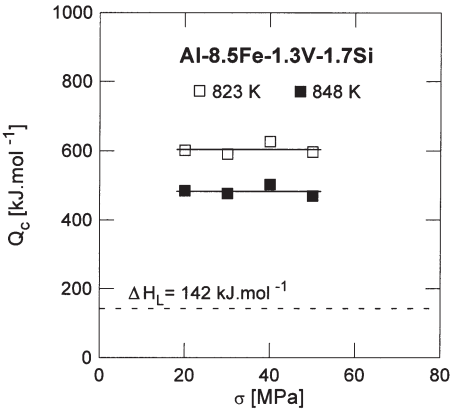

Fig. 6. Values of the apparent activation energy $Q_\mathrm{c} = [\partial \ln \dot{\varepsilon}_\mathrm{m}/\partial(-1/RT)]_\sigma$ following from Fig. 5 and plotted against applied stress σ. The value of the activation enthalpy of lattice self-diffusion in aluminum [9], ΔH_L, is shown for comparison.

Table 2. Values of the activation energy of detachment, Q_d, following from Fig. 8 for two normalized applied stresses σ/G and various temperatures

T [K]	Q_d [kJ·mol^{-1}] for	
	$\sigma/G = 2 \times 10^{-3}$	$\sigma/G = 2.5 \times 10^{-3}$
798	519	509
823	444	424
848	340	305
873	181	151
923	60	50

823 and 848 K, for which the apparent activation energy Q_c could be estimated, the sum $\Delta H_\mathrm{L} + Q_\mathrm{d}$ should be equal to Q_c corrected for the temperature dependence of the shear modulus. The corrected apparent activation energy $Q_\mathrm{c}^\mathrm{corr} = Q_\mathrm{c} + [m_\mathrm{c} R T^2/G](\mathrm{d}G/\mathrm{d}T)$ is equal to 547 and 421 kJ·mol^{-1} at 823 and 848 K, respectively. The quantity $Q_\mathrm{c}^\mathrm{corr}$ represents, in fact, the true activation energy (activation enthalpy ΔH_c) of creep. The values of $Q_\mathrm{c}^\mathrm{corr}$ can be compared with the sum of $\Delta H_\mathrm{L} + Q_\mathrm{d}$. Taking the average values of Q_d, i.e., Q_d^av (i.e., 434 kJ·mol^{-1} for 823 K and 322 kJ·mol^{-1} for 848 K, Table 3) $\Delta H_\mathrm{L} + Q_\mathrm{d}$ is equal to 576 and 464 kJ·mol^{-1} for 823 K and 848 K, respectively. The differences between $\Delta H_\mathrm{L} + Q_\mathrm{d}$

Table 3. Comparison of the experimental $\dot{\varepsilon}_m(T,\sigma)$ creep data with the predictions of the Rösler-Arzt model [18]. Values of the detachment stress σ_d calculated by means of Eq. (4) for various temperatures T, applied stress $\sigma = 30$ MPa and the relaxation factor $k_R = 0.94$. Values of the structure factor C obtained comparing the experimental $\dot{\varepsilon}_m(T,\sigma)$ creep data with the model prediction, Eq. (5), and the value of C calculated from the structure data, Eq. (6). Values of the activation energy of detachment, Q_d, calculated by means of Eq. (7) and its average values estimated from the $\ln \dot{\varepsilon}_m b^2/D_L$ vs. $1/T$ relations (Fig. 8). The true activation energy of creep, ΔH_c, estimated as $Q_d^{av} + \Delta H_L$ ($\Delta H_L = 142$ kJ·mol^{-1}) compared to $\Delta H_c = Q_c^{corr}$

T [K]	σ_d [MPa] Eq. (4)	C[a] Eq. (5)	C[b] Eq. (6)	Q_d [kJ·mol^{-1}] Eq. (7)	Q_d^{av} [c] [kJ·mol^{-1}] Fig. 8	$\Delta H_c = Q_d^{av} + \Delta H_L$ [kJ·mol^{-1}]	$\Delta H_c = Q_c^{corr}$ [d] [kJ·mol^{-1}]
798	181.1	2.79	3.26×10^{-2}	237	514	656	
823	170.5	4.55		228	434	576	547
848	160.5	1.61		218	322	464	421
873	151.0	0.95		208	166	308	
923	133.5	0.04		188	55	197	

[a] Fitting Eq. (5) to experimental $\dot{\varepsilon}_m(T,\sigma)$ creep data
[b] $\rho = 2 \times 10^{14}$ m^{-2}, $\lambda = 95$ nm
[c] Average values for $\sigma/G = 2 \times 10^{-3}$ and $\sigma/G = 2.5 \times 10^{-3}$ from Fig. 8; c.f., Table 2
[d] Corrected for temperature dependence of the shear modulus

and $Q_c^{corr}(\equiv \Delta H_c)$ (amounting 5% at 823 K and 9% at 848 K) are certainly not significant with respect to the fact that both Q_c and Q_d were obtained as the derivatives of the non-linear relations presented in Figs. 5 and 8, respectively.

4. Discussion

As in some previous reports (e.g., Refs. 14 and 15), also in the present investigation, the model of creep developed by Rösler and Arzt [18] is accepted for an attempt to interpret the experimental $\dot{\varepsilon}_m(T,\sigma)$ creep data. The Rösler-Arzt model is based on the idea that the detachment of dislocations from fine incoherent (and thus "interacting") particles is thermally activated. The strength of the attractive interaction of a dislocation with a particle is characterized by the relaxation factor k_R, to which the detachment stress σ_d is related as [11, 12]

$$\sigma_d = \sigma_{OB}\sqrt{1 - k_R^2}, \tag{3}$$

where σ_{OB} is the Orowan bowing stress. In several papers concerning the creep behavior of discontinuous aluminum alloy matrix composites specifically an Al-8.5Fe-1.3V-1.7Si-15SiC$_p$ composite [13], the well substantiated hypothesis [6] was

accepted, namely that at the testing temperatures higher than about 750 K the relaxation factor k_{R} is temperature independent and approximately equal to 0.94. For Al-8.5Fe-1.3V-1.7Si alloy this is illustrated in Fig. 9, taken from the paper of Zhu et al. [6]. From the Figure it follows that the relaxation factor k_{R} increases with increasing temperature and approaches the "critical value" of ~ 0.94 with the temperature approaching ~ 775 K. The critical value of $k_{\mathrm{R}} \cong 0.94$ is close to that following as an upper limit from the analysis of Arzt and Wilkinson [11] and Arzt and Rösler [12]. Thus, above this temperature the relaxation factor is temperature independent, and, consequently, the values of the detachment stress σ_{d} for various testing temperatures and a constant stress σ can be estimated solving the equation [14]

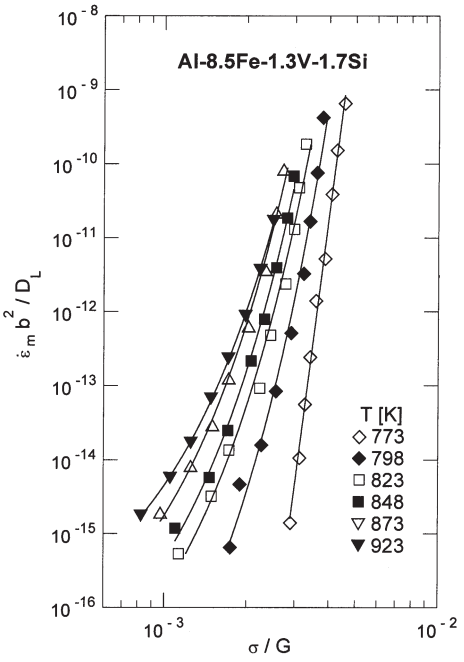

Fig. 7. Normalized minimum creep strain rates $\dot{\varepsilon}_{\mathrm{m}} b^2/D_{\mathrm{L}}$ plotted against normalized applied stresses σ/G for temperatures 773–923 K.

$$\left(\frac{\sigma}{\sigma_{\mathrm{d}}}\right)^3 - \left(\frac{\sigma}{\sigma_{\mathrm{d}}}\right)^2 = -\frac{K\,(T)^2}{(1-k_{\mathrm{R}})^3}, \quad (4)$$

where $K(T) = 4kTm_{\mathrm{c}}/3Gb^2d_{\mathrm{p}}$; k is the Boltzmann constant and d_{p} is the mean diameter of an interacting particle. The values of the detachment stress σ_{d} calculated for applied stress $\sigma = 30$ MPa and the temperatures 798–923 K are listed in Table 3.

According to the Rösler-Arzt model [18], the temperature and applied stress dependence of the minimum creep strain rate is described by the equation

$$\frac{\dot{\varepsilon}_{\mathrm{m}} b^2}{D_{\mathrm{L}}} = C \exp\left\{-\frac{Gb^2 d_{\mathrm{p}}[(1-k_{\mathrm{R}})(1-\sigma/\sigma_{\mathrm{d}})]^{3/2}}{2kT}\right\}, \quad (5)$$

where

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

is the structure factor; λ is the mean interparticle spacing and ρ is the density of mobile dislocations. Using the values of σ_{d} listed in Table 3 and $\mathrm{k_R} = 0.94$,

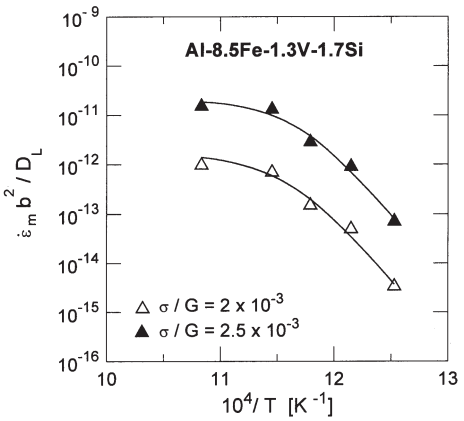

Fig. 8. Relations between $\dot{\varepsilon}_{\mathrm{m}} b^2/D_{\mathrm{L}}$ and $1/T$ for two values of the σ/G ratio.

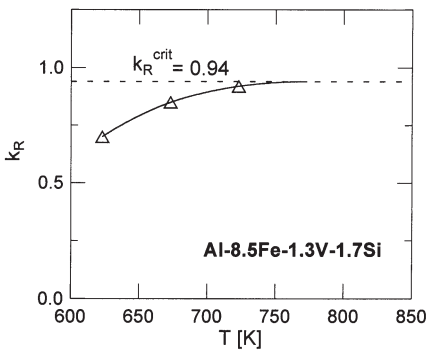

Fig. 9. Calculated values of the relaxation factor k_{R} plotted against temperature. The term $k_{\mathrm{R}}^{\mathrm{crit}}$ is the value of this factor following from the analysis of Arzt and Wilkinson [11] and Arzt and Rösler [12]. Taken from Zhu et al. [6].

the experimental $\dot{\varepsilon}_{\mathrm{m}}(T,\sigma)$ creep data plotted in $\dot{\varepsilon}_{\mathrm{m}} b^2/D_{\mathrm{L}}$ vs. σ/G co-ordinates are compared with the model-creep Eq. (5) in Fig. 10. The agreement of the experimental $\dot{\varepsilon}_{\mathrm{m}}(T,\sigma)$ creep data with the prediction of the Rösler-Arzt model – the creep Eq. (5) – is very good. However, the values of the structure factor C following from the fitting procedure, listed in Table 3, should be close to those calculated from the structure data, Eq. (6), before the correlation could be considered fully satisfactory. Accepting a mobile dislocation density $\rho = 2 \times 10^{14}$ m^{-2} [19] and taking $\lambda =$ $= 95$ nm [6], the value of C calculated by means of Eq. (6) is equal to 3.26×10^{-2}. Thus, the structure factor C following from the fitting procedure (Table 3) differs from that calculated from the structure data, by a factor ranging from 85.6 to 1.2, in average equal to ~ 60. Similar differences were also observed and discussed in some other papers of the present authors [13, 15, 20] with the conclusion that a certain deficiency of the Rösler-Arzt model has to be considered [18] in its original formulation (see also Ref. 21), although, as pointed out by these authors "the discrepancy amounting three orders of magnitude may not be considered serious".

However, there is still another possibility to compare the present results with the prediction of the Rösler-Arzt model. According to Eq. (5), the activation energy of detachment of a dislocation from an attractive particle is expressed as

$$Q_{\mathrm{d}} = (1/2) G b^2 d_{\mathrm{p}} [(1 - k_{\mathrm{R}})(1 - \sigma/\sigma_{\mathrm{d}})]^{3/2}. \qquad (7)$$

The values of the activation energy of detachment calculated by means of Eq. (7) can be compared with those following from the $\ln \dot{\varepsilon}_\mathrm{m} b^2/D_\mathrm{L}$ vs. $1/T$ relations, Fig. 8, and listed in Table 2. Such a comparison is presented in Table 3. From this table it can be seen that the values of Q_d obtained from the experimental $\dot{\varepsilon}_\mathrm{m}(T,\sigma)$ creep data analysis differ from those predicted by the model by a factor ranging from $\sim$ 2.1 to $\sim$ 0.3 depending on temperature and in average equal to 1.33. This result is consistent with that obtained comparing the structure factor following from fitting the $(\dot{\varepsilon}_\mathrm{m} b^2/D_\mathrm{L},\ \sigma/G)$ creep data to Eq. (5), Fig. 7, to the value of this factor calculated from the structure data, Eq. (6), Table 3.

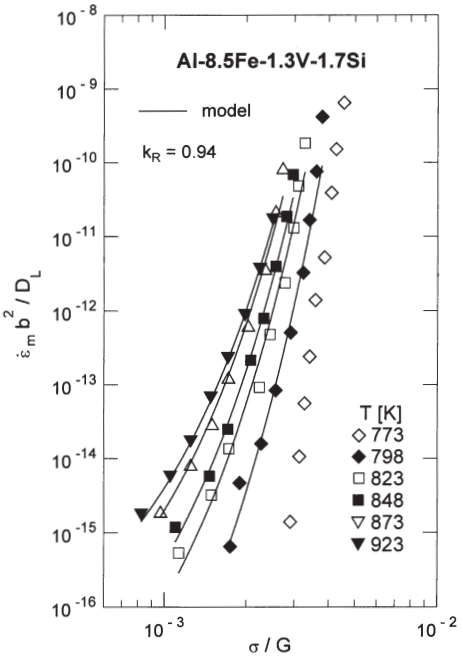

Fig. 10. The $(\dot{\varepsilon}_\mathrm{m} b^2/D_\mathrm{L},\ \sigma/G)$ creep data for 798–923 K compared with the prediction of the Rösler-Arzt-model creep, Eq. (5), accepting the relaxation factor $k_\mathrm{R} = 0.94$.

In a recent paper, Carreño and Ruano [4] rejected the Rösler-Arzt model because it "predicts unrealistically high mobile dislocation density" (in other words much higher values of the structure factor C than those following from the structure data). These authors have proposed a new approach – $\tilde{n}$ concept – to interpret the creep behavior of alloys strengthened with finely dispersed particles, specifically the creep behavior of an Al-8.5Fe-1.3V-1.7Si alloy. According to the $\tilde{n}$ concept, the apparent stress exponent

$$m_\mathrm{c} = h + \tilde{n}. \tag{8}$$

The component h is equal to the true stress exponent characteristic for the structure invariant creep mechanism: $h = 8$. The component $\tilde{n}$ raises the exponent m_c due to dislocations/dispersed particles interaction. The apparent activation energy expressed as

$$Q_\mathrm{c} = \Delta H_\mathrm{L} \frac{m_\mathrm{c}}{h} \tag{9}$$

is thus generally higher than the activation enthalpy ΔH_L of the lattice diffusion in the matrix – specifically in aluminum.

Hence, the creep strength of the alloy is contributed independently by the strength of the matrix and by the strength following from the dislocations/dispersed particles interaction. In other words, "the two parameters h and $\tilde{n}$ are related solely to the matrix behavior and the presence of dispersoids, respectively" [4].

The creep equation is then expressed as

$$\dot{\varepsilon} = A \left(\frac{\sigma}{E}\right)^{\tilde{n}+h} \exp\left[-\frac{\Delta H_{\mathrm{L}}}{RT}\frac{\tilde{n}+h}{h}\right], \tag{10}$$

where E is the Young's modulus. The authors' strain rate vs. flow stress data for temperatures 623–827 K and strain rates ranging from 10^{-2} to 2.5×10^{-6} s^{-1} fit this equation satisfactorily if $h = 8$ and $\tilde{n} = 14$ are accepted. Thus, the exponent m_{c} is predicted to be equal to 22, while from the experimental strain rate vs. flow stress data $m_{\mathrm{c}} \cong 25$ follows. The apparent activation energy Q_{c} close to 400 $\mathrm{kJ \cdot mol^{-1}}$ follows [4] from these data, while a value of 390.5 $\mathrm{kJ \cdot mol^{-1}}$ is obtained by means of Eq. (9). The constant A in Eq. (10) attains an extremely high value of 2.6×10^{83} s^{-1}.

Because of a relatively narrow interval of experimental strain rates and, probably, because of the technique of flow stress evaluation (see Section 1), the authors [4] could not detect a threshold behavior even at the lowest testing temperatures of 623 and 674 K, while the recently published experimental $\dot{\varepsilon}_{\mathrm{m}}(T, \sigma)$ creep data [6] provide a strong evidence of such behavior. Thus, the authors' evaluation of their strain rate vs. flow stress data should be taken with caution and the disagreement between the present authors' data and those of Carreño and Ruano need not be disquieting. The $\tilde{n}$ concept has been extended by Carreño and Ruano in another more recent paper [22]. This concept still remains to be a phenomenological one.

On the other hand, the discrepancy between the present authors' creep data and the data recently reported by Ma and Tjong [5] can be hardly explained without obtaining further experimental $\dot{\varepsilon}_{\mathrm{m}}(T, \sigma)$ creep data for an alloy of a precisely defined origin using the constant stress creep test technique and *measuring the minimum creep strain rates*. In fact, a reliable and independent verification of a transition from the athermal to the thermally activated detachment of dislocations from interacting particles, although reported in a number of papers (e.g., Refs. 13 and 14) is still desirable for further support of the concept of thermally activated detachment of dislocations from small interacting particles in creep of the alloys strengthened with such particles.

In a previous investigation [23] of an Al-8.5Fe-1.3V-1.7Si alloy reinforced with 15 vol.% silicon carbide particulates – Al-8.5Fe-1.3V-1.7Si-15$\mathrm{SiC_p}$ composite – the creep behavior at temperatures ranging from 873 to 948 K was found to be characterized by the true activation energy ΔH_{c} very close to the activation enthalpy ΔH_{B} of grain boundary diffusion in aluminum and the true stress exponent n close

to, but not higher than 2.5. At the same time, the creep behavior was found to be associated with the true threshold stress decreasing with increasing temperature more strongly than the shear modulus of aluminum. Thus, the creep behavior of the composite in the above temperature interval could be described [23] by the equation

$$\frac{\dot{\varepsilon}_{\mathrm{m}} b^2}{D_{\mathrm{B}}} = A_{\mathrm{B}} \left(\frac{\sigma - \sigma_{\mathrm{TH}}}{G} \right)^n , \tag{11}$$

where D_{B} is the coefficient of grain boundary diffusion in aluminum, A_{B} is a dimensionless constant and σ_{TH} is the true threshold stress. The origin of this threshold stress should be expected to be different from that associated with the creep at temperatures ranging from 623 to 723 K [13]. On the other hand, the creep behavior of the composite at temperatures 798 and 823 K is similar to that described in Section 3 of the present paper: the threshold stress is absent, the true activation energy ΔH_{c} is higher than the activation enthalpy ΔH_{L} of the lattice self-diffusion in aluminum, the true stress exponent n increases with increasing stress and is independent of temperature [14].

The creep equation, Eq. (11), suggests the grain boundary sliding as the mechanism controlling the creep strain rate. Of course, the grain boundary sliding must be accommodated by glide of lattice dislocations within grains. Under certain conditions, Eq. (11) with $n = 2$ and $A_{\mathrm{B}} = A_{\mathrm{s}}(b/d)^2$, where d is the mean grain diameter, describes the superplastic strain rate [24, 25], which may not be the case in the alloy composite under consideration. A question naturally arises why at temperatures 873, 798 and 923 K the creep strain rate in the unreinforced alloy can be satisfactorily described by Eq. (5) while that in the composite is well described by Eq. (11). The present authors suggest that for the strain rate behavior in an Al-8.5Fe-1.3V-1.7Si-15$\mathrm{SiC_p}$ composite described by Eq. (11) the reinforcement is responsible: The presence of large volume fraction of coarse silicon carbide particulates inhibits the grain boundary sliding much more efficiently (perhaps because the mean diameter of silicon carbide particulates being several times larger than the mean matrix grain diameter) than the almost twice as high volume fraction of fine $\mathrm{Al_{12}(Fe,V)_3Si}$ phase particles. Just this strong effect of particulates introduces the true threshold stress observed.

5. Summary and conclusions

In the present work, the creep behavior of an Al-8.5Fe-1.3V-1.7Si alloy processed by rapid solidification and powder metallurgy route is investigated at temperatures 773–923 K using the isothermal constant-tensile-stress creep test technique. The measured minimum creep strain rates cover four to six orders of magnitude. The main results can be expressed as follows:

1. With the exception of the lowest testing temperature of 773 K at which the apparent stress exponent m_c does not depend on applied stress, this exponent increases with applied stress linearly and is temperature independent. The temperature independence of the apparent stress exponent m_c is consistent with the applied stress independence of the apparent activation energy of creep, Q_c.

2. The values of the apparent activation energy corrected for the temperature dependence of shear modulus, Q_c^{corr}, amount to 547 and 421 kJ·mol^{-1} at 823 and 848 K, respectively. These values of Q_c^{corr} represent those of the true activation energy (activation enthalpy) ΔH_c of creep corresponding to these temperatures. The true activation energy of creep is thus significantly higher than the activation enthalpy of lattice diffusion in the alloy matrix and depends on temperature. The true stress exponent n of minimum creep strain rate is generally higher than the true stress exponent in the alloy matrix (aluminum).

3. Comparing the present experimental $\dot{\varepsilon}_m(T,\sigma)$ creep data with the model of creep developed by Rösler and Arzt, the structure factor predicted by this model is a factor in average 60 times greater than that calculated from the structure data. The values of the activation energy of detachment Q_d following from the $\dot{\varepsilon}_m(T,\sigma)$ creep data analysis differ from those predicted by the model by a factor in average equal to 1.33. Generally, the creep behavior of the alloy is well interpreted in terms of the thermally activated detachment of dislocations from fine and predominantly $Al_{12}(Fe,V)_3Si$ phase particles.

4. Consequently, the minimum creep strain rate is, most probably, controlled by the thermally activated detachment of dislocations from fine incoherent and, thus, with dislocation interacting particles.

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