HIGH TEMPERATURE CREEP IN AN Al-8.5Fe-1.3V-1.7Si ALLOY PROCESSED BY RAPID SOLIDIFICATION

JOSEF ČADEK, KVĚTA KUCHAŘOVÁ, SHIJIE ZHU

Creep behaviour of an Al-8.5Fe-1.3V-1.7Si alloy processed by rapid solidification is investigated at three temperatures ranging from 623 to 723 K. The measured minimum creep strain rates cover seven orders of magnitude. The creep behaviour is associated with the true threshold stress decreasing with increasing temperature more strongly than the shear modulus of aluminium. The minimum creep strain rate is controlled by the lattice diffusion in the alloy matrix and the true stress exponent is close to 5. The apparent activation energy of creep depends strongly on both applied stress and temperature and is generally much higher than the activation enthalpy of lattice self-diffusion in aluminium. Also the apparent stress exponent of minimum creep strain rate depends on applied stress as well as on temperature and is generally much higher than the true stress exponent. This behaviour of both the apparent activation energy and apparent stress exponent is accounted for by the strong temperature dependence of the threshold-stress-to-shear-modulus ratio. The true threshold creep behaviour of the alloy is interpreted in terms of athermal detachment of dislocations from fine incoherent $Al_{12}(Fe,V)_3Si$ phase particles admitting a temperature dependence of the relaxation factor characterizing the strength of the attractive dislocation/particle interaction.

Key words: Al-8.5Fe-1.3V-1.7Si alloy, creep, true threshold stress, true stress exponent, rate controlling process

VYSOKOTEPLOTNÍ CREEP SLITINY Al-8,5Fe-1,3V-1,7Si PŘIPRAVENÉ TECHNIKOU RYCHLÉ SOLIDIFIKACE

Creepové chování slitiny Al-8,5Fe-1,3V-1,7Si (vyrobené rychlou solidifikací) je studováno při třech zkušebních teplotách v rozmezí 623 – 723 K. Rozsah měřených minimálních rychlostí creepu zahrnuje sedm řádů. Creepové chování slitiny je spojeno s existencí prahového napětí klesajícího se vzrůstající teplotou rychleji než smykový modul pružnosti hliníku. Rychlost creepu je kontrolována mřížkovou difuzí matrice, skutečný napěťový exponent je blízký pěti. Zdánlivá aktivační energie creepu silně závisí jak na aplikovaném

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napětí, tak i na teplotě a je mnohem vyšší než aktivační entalpie mřížkové autodifuze hliníku. Také zdánlivý napěťový exponent minimální rychlosti creepu závisí na napětí a teplotě a je rovněž mnohem vyšší než skutečný napěťový exponent. Vysoké hodnoty zdánlivé aktivační energie i zdánlivého napěťového exponentu jsou důsledkem silné teplotní závislosti poměru prahového napětí ke smykovému modulu. Skutečné prahové napětí je interpretováno jako napětí atermického odpoutávání dislokací od jemných nekoherentních částic fáze $Al_{12}(FeV)_3Si$ za předpokladu teplotní závislosti relaxačního faktoru, charakterizujícího pevnost interakce dislokací s částicemi.

1. Introduction

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

High temperature creep behaviour of the Al-8.5Fe-1.3V-1.7Si alloy was investigated by a number of authors, specifically by Carreño et al. [1] and Peng et al. [2, 3]. Carreño et al. applied the strain-rate change technique in tension at temperatures ranging from 523 to 823 K and the strain rates ranging from 10^{-6} s^{-1} to 10^{-2} s^{-1}. Every tensile test was carried out in less than 9 ks. Peng et al. [2, 3] applied the constant-load-compression creep-test technique in the temperature range from 573 to 723 K. Most of the creep strain rates were obtained from the stress incremental tests: at any applied stress the specimen was crept to about 0.02 strain, at which the authors assumed the steady state was attained, and then the applied stress was increased and the procedure repeated. The measured creep strain rates did not fully cover three orders of magnitude. Neither Carreño et al. [1] nor Peng et al. [2, 3] have presented any evidence on the attainment of steady state deformation.

Both groups of authors, Carreño et al. [1] and Peng et al. [2, 3], found linear relations between their measured strain rates and applied stress in double logarithmic co-ordinates; the slopes of the straight lines (values of the apparent stress exponent m_c) were found much higher than ~ 5 – the value of m_c for minimum creep strain rate in aluminium (e.g. ref. [4]). The apparent activation energy of creep, Q_c, estimated from the temperature dependence of the measured strain rates were found much higher than the value of the activation enthalpy of the lattice self-diffusion in aluminium, i.e. 142 $kJ \cdot mol^{-1}$ [5]. Such high values of m_c and Q_c are typical for dispersion strengthened aluminium alloys (e.g. refs. [6, 7]).

An interpretation of the creep behaviour of the alloy was attempted [2, 3] alternatively in terms of the true threshold stress concept and in terms of the concept of thermally activated detachment of dislocations from incoherent $Al_{12}(Fe,V)_3Si$ particles. The analyses of creep data have not been found conclusive as to one of the above concepts. This is because the authors [2, 3] did not account for the temperature dependence of the true threshold stress being much stronger than that of the shear modulus G of the alloy matrix. Another, still more important reason should be sought in the narrow interval of the measured strain rates and a questionable definition of these strain rates. This is why the present investigation was undertaken. Thus, the aim of the present work is to reinvestigate the creep behaviour of an Al-8.5Fe-1.3V-1.7Si alloy at three temperatures ranging from 623 to 723 K, measuring the well-defined minimum tensile creep strain rates covering not less than six orders of magnitude. In the forthcoming paper, results of a similar study of creep behaviour of this alloy reinforced with 15 vol.% silicon carbide particulates – an Al-8.5Fe-1.3V-1.7Si-15SiC$_p$ composite (the subscript p means particulates) – will be reported.

2. Material and experimental procedures

The Al-8.5Fe-1.3V-1.7Si (the 8009 Al type) alloy was processed by rapid solidification – by planar flow casting into a strip about 100 μm in thickness. The strip was crushed into flakes by mechanical milling. The flakes were consolidated by hot pressing and then hot extruded into a rod 12 mm in diameter. The temperature of both pressing and extrusion was $\sim$ 830 K.

The structure of the as-extruded alloy was found to be sufficiently homogeneous exhibiting only weak texture. It consisted of particles of an $Al_{12}(Fe,V)_3Si$ intermetallic phase embedded in the alloy matrix of grain size estimated to $\sim$ 0.5 μm. The particle size was an order of magnitude smaller than the grain size, i.e. $\sim$ 50 nm in diameter. The $Al_{12}(Fe,V)_3Si$ phase particles made up a volume fraction of $\sim$ 0.27 and were found to be thermally stable. A detailed study of the structure of a similar alloy prior to creep as well as in crept conditions was performed by Peng et al. [2, 3].

From the alloy rod, specimens for tensile creep tests 4.0 mm in diameter and 25.0 mm in gauge length were machined. The constant tensile-stress creep tests were performed at temperatures 623, 673 and 723 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 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}. All the creep tests were run well into tertiary stage and interrupted. Generally, no steady stage was observed, only the minimum creep strain rate $\dot{\varepsilon}_m$ could be defined.

3. Results

In Fig. 1, the minimum creep strain rates $\dot{\varepsilon}_{\rm m}$ measured at various temperatures T and applied stresses σ are plotted against applied stress in double logarithmic co-ordinates. For any temperature T under consideration, the $\dot{\varepsilon}_{\rm m}(\sigma)$ relation clearly demonstrates the true threshold creep behaviour. In fact, the apparent stress exponent of minimum creep strain rate, $m_{\rm c} = (\partial \ln \dot{\varepsilon}_{\rm m}/\partial \ln \sigma)_T$, increases with decreasing applied stress reaching extremely high values at the applied stresses, at which the minimum creep strain rates decrease to $\sim 10^{-9}$ s^{-1} (Fig. 2).

Because of very strong applied-stress dependence of $\dot{\varepsilon}_{\rm m}$, it is not possible to estimate the apparent activation energy of creep, $Q_{\rm c} = [\partial \ln \dot{\varepsilon}_{\rm m}/\partial(-1/RT)]_\sigma$ and its

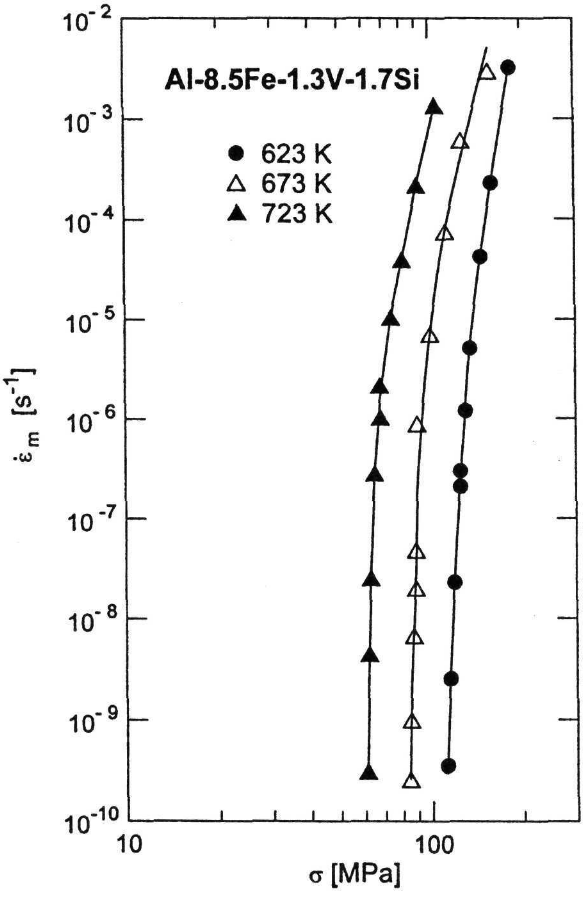

Fig. 1. The minimum creep strain rates $\dot{\varepsilon}_{\rm m}$ plotted against applied stresses σ for 623, 673 and 723 K in double logarithmic co-ordinates.

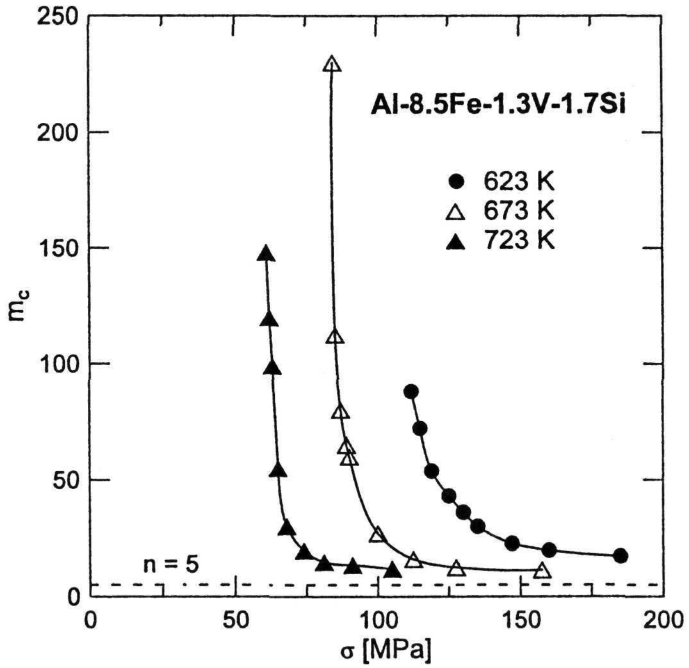

Fig. 2. Relations between apparent applied stress exponent of minimum creep strain rate, m_c, and applied stress for the temperatures under consideration.

applied-stress and temperature dependence from the $\dot{\varepsilon}_{\mathrm{m}}(\sigma, T)$ creep data obtained in the present work. However, it was shown [7] that the values of Q_{c} calculated by means of a relation given later in this section – Eq. (3) – are in remarkable agreement with the values of Q_{c} estimated in a conventional way from the temperature dependences of $\dot{\varepsilon}_{\mathrm{m}}(\sigma)$. Thus, the calculated $Q_{\mathrm{c}}(\sigma, T)$ relations will be presented later on in this section.

To determine the true threshold stress for the testing temperatures under consideration the conventional linear extrapolation technique (e.g. refs. [6, 7]) has been applied. Thus, in Fig. 3, $\dot{\varepsilon}_{\mathrm{m}}^{1/n}$ is plotted against σ in double linear co-ordinates for the true stress exponent n equal to 5. The values of the true threshold stress σ_{TH} obtained extrapolating the $\dot{\varepsilon}_{\mathrm{m}}^{1/n}$ vs. σ relations to zero creep strain rate are given in the figure together with the correlation coefficients R. These data suggest the true stress exponent n close to 5 and the true threshold stress σ_{TH} depending strongly on temperature. However, the $\dot{\varepsilon}_{\mathrm{m}}^{1/5}$ vs. σ relation for 623 K requires a discussion. At this temperature, the true threshold stress determined by the linear

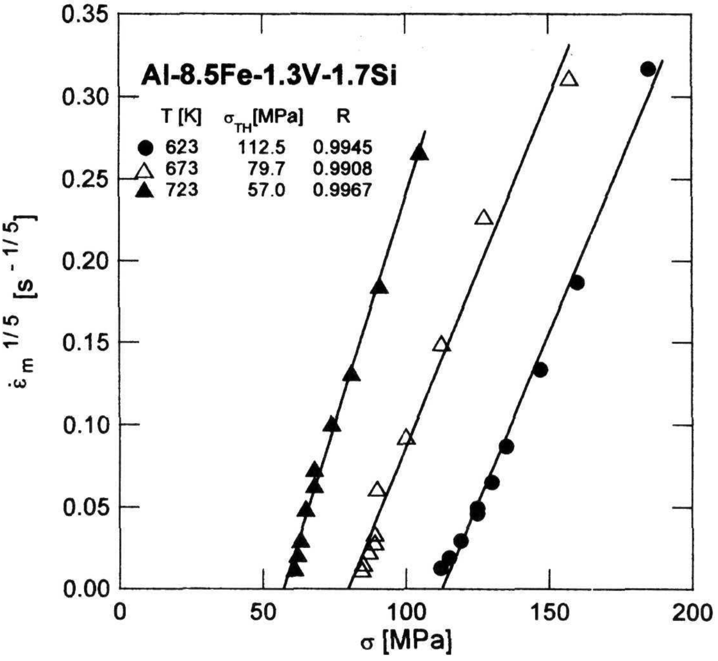

Fig. 3. $\dot{\varepsilon}_{\mathrm{m}}^{1/5}$ plotted against applied stress in double linear co-ordinates. R means the correlation coefficient.

extrapolation technique is higher than the applied stresses of 112.0 MPa, and also the $\dot{\varepsilon}_{\mathrm{m}}^{1/5}$ at 115.0 MPa is higher than that following from the linear extrapolation. Thus, for $\sigma = 112.0$ MPa the effective stress $(\sigma - \sigma_{\mathrm{TH}})$ is negative (which is absurd) and for $\sigma = 115.0$ MPa the value of the effective stress is unrealistically low. Thus, the value of $\dot{\varepsilon}_{\mathrm{m}}^{1/5}$ for 112.0 MPa cannot be taken into account and that for 115.0 MPa should be omitted in the further analysis, although in this analysis the threshold stress $\sigma_{\mathrm{TH}} = (112.45 \pm 4.44)$ MPa will be accepted. The higher creep strain rates at these two applied stresses can be, perhaps, in reality lower than those plotted in Fig. 3. In fact, to measure $\dot{\varepsilon}_{\mathrm{m}}$ close to and especially below 10^{-9} s^{-1} at the lowest testing temperature was found difficult. In the following analysis, the value of the true stress exponent n equal to 5 is accepted.

As pointed out referring to Fig. 3, the true threshold stress σ_{TH}, estimated accepting $n = 5$, depends on temperature. The values of σ_{TH} given in this figure are plotted against temperature in Fig. 4, in which also the σ_{TH}/G ratio is plotted against temperature; G is the shear modulus of aluminium [8]. It can be seen that not only σ_{TH} but also σ_{TH}/G decrease rather strongly with increasing temperature.

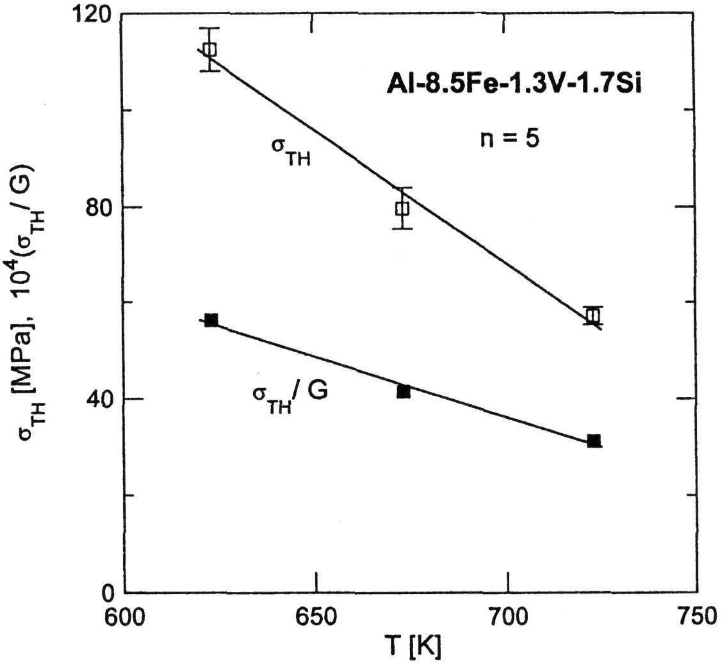

Fig. 4. Values of the true threshold stress σ_{TH} and values of the σ_{TH}/G ratio plotted against temperature.

This phenomenon is common to dispersion strengthened alloys, specifically to ODS Al alloys [9] (ODS means oxide dispersion strengthening) as well as to aluminium alloys processed by powder metallurgy [10, 11] and, of course, to discontinuous aluminium and aluminium alloy matrix composites fabricated by powder metallurgy route [6, 7].

In Fig. 5, σ_{TH} as well as σ_{TH}/G are plotted against reciprocal temperature in half logarithmic co-ordinates. The relations between $\ln \sigma_{\mathrm{TH}}$ as well as $\ln(\sigma_{\mathrm{TH}}/G)$ and $1/T$ can be well approximated by straight lines. Thus, the expression

$$\frac{\sigma_{\mathrm{TH}}}{G} = B \exp\left[\frac{Q_{\mathrm{o}}}{RT}\right] \tag{1}$$

approximately holds. In Eq. (1), B is a dimensionless constant, R is the gas constant and $Q_{\mathrm{o}} = 22.3\ \mathrm{kJ \cdot mol^{-1}}$ is the energy that will be briefly discussed in Section 4.

In Fig. 6, the minimum creep strain rates $\dot{\varepsilon}_{\mathrm{m}}$ are plotted against the effective stress $(\sigma - \sigma_{\mathrm{TH}})$ in double logarithmic co-ordinates. According to Li and Langdon

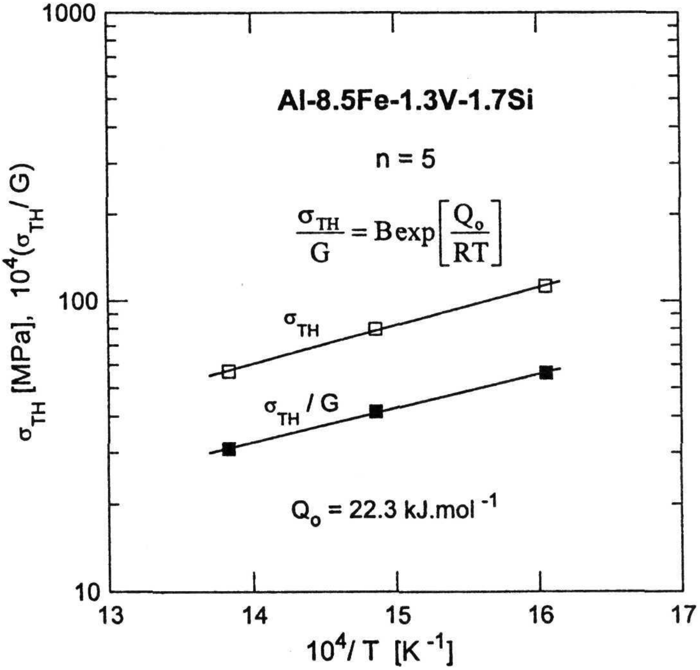

Fig. 5. The threshold stress σ_{TH} and the σ_{TH}/G ratio plotted against the reciprocal temperature in half logarithmic co-ordinates.

[12] this representation of creep data may be significant for supporting the assumption on the value of the true stress exponent n. From the figure it can be seen that the values of n_e, i.e. the slopes of $\dot{\varepsilon}_m$ vs. $(\sigma - \sigma_{TH})$ relations in the co-ordinates under consideration, range from 5.0 to 5.8 depending on temperature, in average being ~ 5.5. The cause of the scatter of n_e is not clear and the same holds for the average $n_e = 5.5$ higher than the true stress exponent $n = 5$ accepted for the determination of the values of σ_{TH} applying the linear extrapolation technique. Nevertheless, the average value of n_e supports the assumed value of the true stress exponent n close to 5 rather than that close to 8. The value of 5, accepted in the present work for the true stress exponent n, is equal to that following from the results of Peng et al. [2, 3]. Thus, the true threshold stress values given in Figs. 3 and 4 are considered to be realistic.

Therefore, it is justified to plot $\dot{\varepsilon}_m b^2/D_L$ vs. $(\sigma - \sigma_{TH})/G$ assuming that the minimum creep strain rate is lattice diffusion controlled and accepting $n = 5$; D_L is the coefficient of lattice self-diffusion in aluminium [5] and b is the length of the

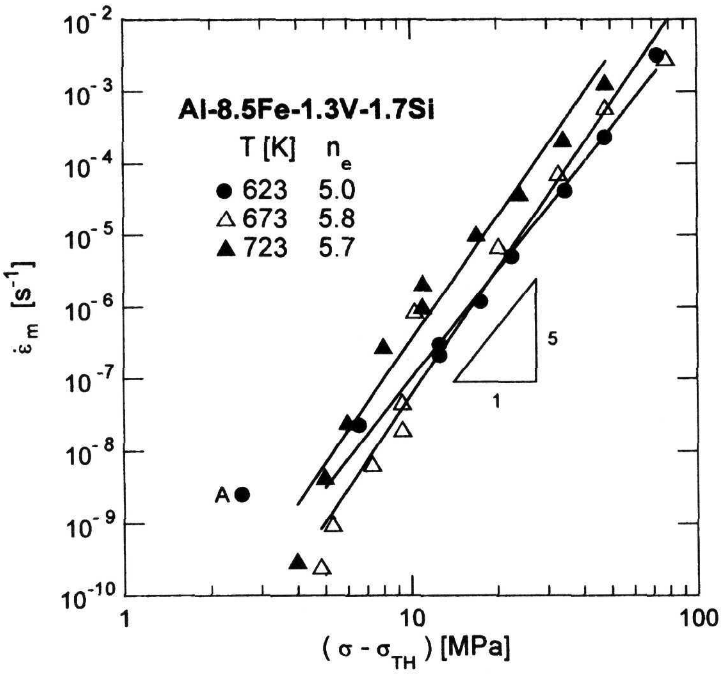

Fig. 6. The minimum creep strain rates $\dot{\varepsilon}_m$ plotted against the effective stresses $(\sigma - \sigma_{\mathrm{TH}})$; the point denoted by A was omitted.

Burgers vector of aluminium. Such a plot, shown in Fig. 7, can be approximated by a single straight line. This justifies the assumption of the lattice diffusion in the matrix as the creep strain rate controlling process as well as the choice of n close to 5. Thus, the minimum creep strain rate as a function of temperature and applied stress can be expressed as (see e.g. refs. [7, 13])

$$\frac{\dot{\varepsilon}_{\mathrm{m}} b^2}{D_{\mathrm{L}}} = A \left(\frac{\sigma - \sigma_{\mathrm{TH}}}{G} \right)^n, \qquad n = 5, \tag{2}$$

where A is a dimensionless constant. Combining this creep equation with the definition equation of the apparent activation energy of creep, $Q_{\mathrm{c}} = [\partial \ln \dot{\varepsilon}_{\mathrm{m}} / \partial(-1/RT)]_\sigma$ on one side and the definition equation of the apparent applied stress exponent $m_{\mathrm{c}} = (\partial \ln \dot{\varepsilon}_{\mathrm{m}} / \partial \ln \sigma)_T$ on the other one, the following expressions for the apparent activation energy Q_{c} and the apparent stress exponent m_{c} are obtained:

$$Q_{\mathrm{c}} = \Delta H_{\mathrm{L}} - \frac{nRT^2}{G} \left(\frac{G}{\sigma - \sigma_{\mathrm{TH}}} \frac{\mathrm{d}\sigma_{\mathrm{TH}}}{\mathrm{d}T} + \frac{n-1}{n} \frac{\mathrm{d}G}{\mathrm{d}T} \right) \tag{3}$$

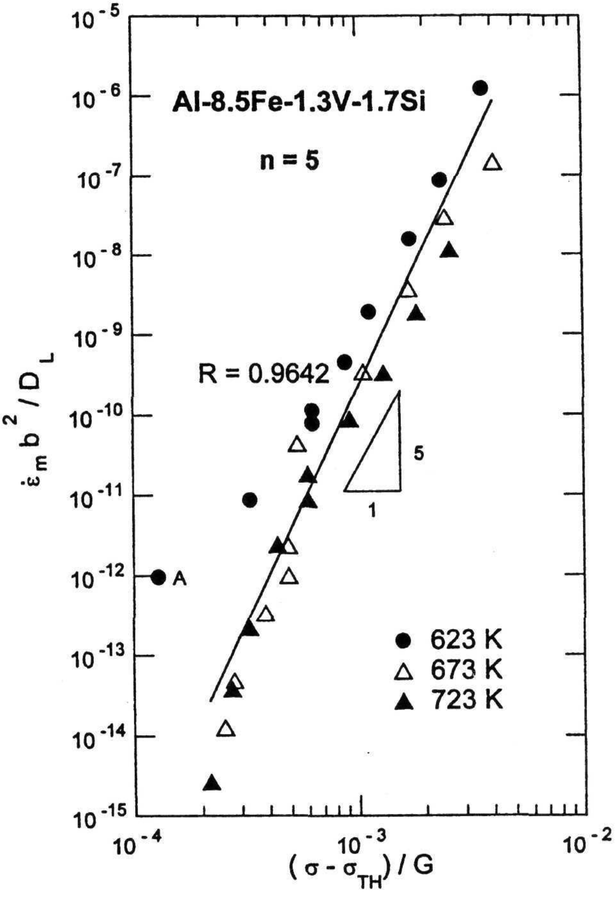

Fig. 7. The normalized minimum creep strain rates, $\dot{\varepsilon}_{\rm m} b^2/D_{\rm L}$, plotted against the normalized effective stresses, $(\sigma - \sigma_{\rm TH})/G$; the data point denoted A was omitted.

and

$$m_{\rm c} = \frac{n\sigma}{\sigma - \sigma_{\rm TH}}. \tag{4}$$

In Eq. (3), $\Delta H_{\rm L}$ is the activation enthalpy of lattice self-diffusion in aluminium ($\Delta H_{\rm L} = 142$ kJ·mol^{-1} [5]). From Eqs. (3) and (4) it follows that if $\sigma_{\rm TH}/G$ depends on temperature, both $Q_{\rm c}$ and $m_{\rm c}$ depend on both the applied stress and temperature, which is the case of the alloy under consideration.

Now, values of the apparent activation energy $Q_{\rm c}$ can be estimated for various temperatures and applied stresses by means of Eq. (3). It should be emphasized

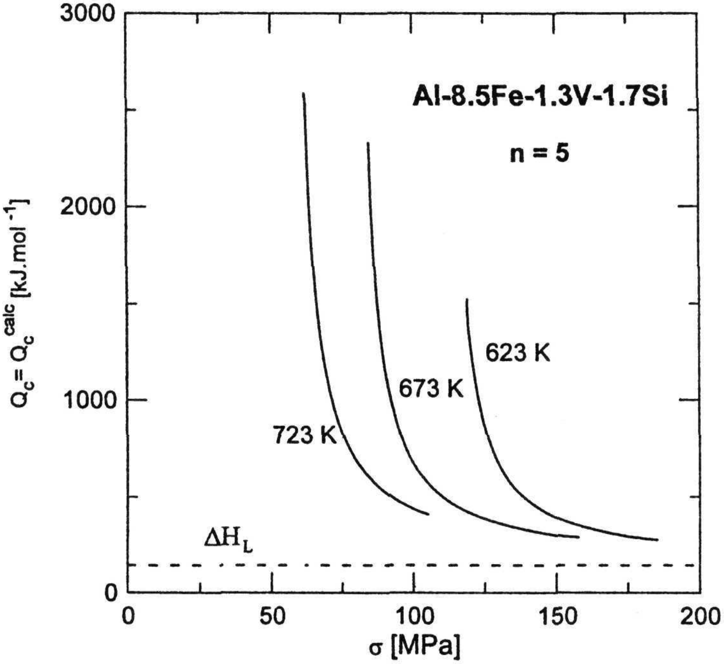

Fig. 8. The apparent activation energy of creep calculated by means of Eq. (3), $Q_{\mathrm{c}}^{\mathrm{calc}}$, as a function of applied stress and temperature.

Table 1. Al-8.5Fe-1.3V-1.7Si alloy. Values of $Q_{\mathrm{c}}^{\mathrm{calc}}$ and $m_{\mathrm{c}}^{\mathrm{calc}}$ for various temperatures and an unrealistically high applied stress of 500 MPa

$T = 623$ K		$T = 673$ K		$T = 723$ K	
$Q_{\mathrm{c}}^{\mathrm{calc}}$ [kJ.mol^{-1}]	$m_{\mathrm{c}}^{\mathrm{calc}}$	$Q_{\mathrm{c}}^{\mathrm{calc}}$ [kJ.mol^{-1}]	$m_{\mathrm{c}}^{\mathrm{calc}}$	$Q_{\mathrm{c}}^{\mathrm{calc}}$ [kJ.mol^{-1}]	$m_{\mathrm{c}}^{\mathrm{calc}}$
175.3	6.45	179.4	5.95	184.3	5.64

once again that the calculated values of $Q_{\mathrm{c}} = Q_{\mathrm{c}}^{\mathrm{calc}}$ were shown [14] to be in very good agreement with the values of $Q_{\mathrm{c}} = Q_{\mathrm{c}}^{\mathrm{exp}}$ determined from the temperature dependence of the minimum creep strain rate. For the Al-8.5Fe-1.3V-1.7Si alloy, the calculated values of the apparent activation energy, $Q_{\mathrm{c}}^{\mathrm{calc}}$, are plotted against applied stress σ for all the temperatures under consideration in Fig. 8. It can be seen that at any given temperature, the activation energy decreases with increasing stress. Even at an unrealistically high applied stress of 500 MPa the calculated values of the apparent activation energy are still significantly higher than that of ΔH_{L} for aluminium, which is illustrated in Table 1. At all the temperatures under consideration, the activation energy Q_c^{calc} reaches extremely high values at applied stresses only slightly higher than the respective value of the true threshold stress σ_{TH}. To give an example, at 673 K and $\sigma = 85$ MPa, Q_c^{calc} amounts to ~ 2125 kJ$\cdot$mol^{-1}. At any given applied stress the energy Q_c^{calc} increases with decreasing temperature. This is in qualitative agreement with the results obtained for an Al-30SiC$_p$ composite as well as an ODS Al-5Mg-30SiC$_p$ composite, for which the values of the apparent activation energy were determined performing the conventional analysis of the experimental $\dot{\varepsilon}_m(T,\sigma)$ creep data.

Similarly, Eq. (4) can be used to calculate the apparent stress exponent $m_c = m_c^{calc}$ as a function of applied stress and temperature. The relations between $m_c = m_c^{exp}$ and σ for 623, 673 and 723 K have been obtained from the experimental $\dot{\varepsilon}_m(T,\sigma)$ creep data, Fig. 1, and shown in Fig. 2. The values of m_c^{calc} for the temperatures under consideration are plotted against applied stress in Fig. 9. The comparison of m_c^{calc} and m_c^{exp} is most interesting since a good agreement between m_c^{calc} and m_c^{exp} would support the validity of the creep Eq. (2). Such a comparison

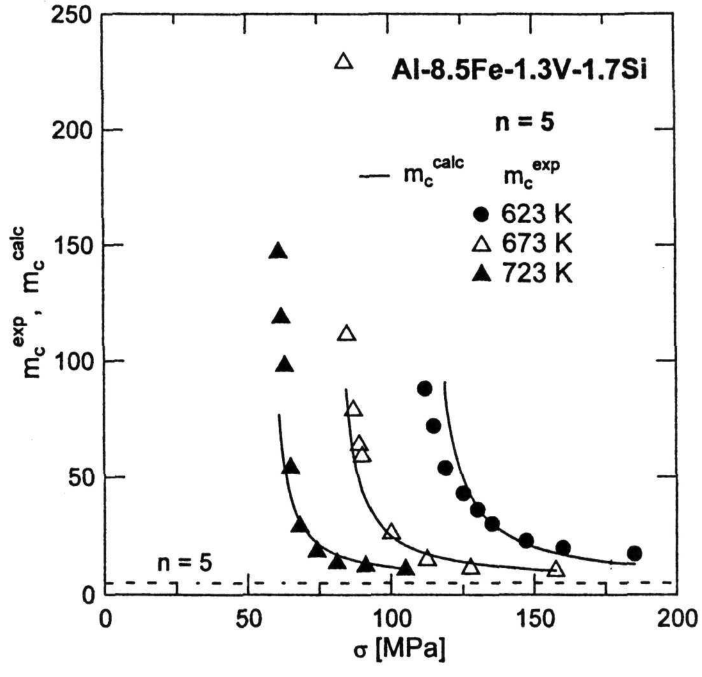

Fig. 9. The apparent applied stress exponent calculated by means of Eq. (4), m_c^{calc}, as a function of applied stress and temperature. The values of m_c obtained for various temperatures as the derivatives of $\dot{\varepsilon}_m$ vs. σ relations (Fig. 2), m_c^{exp}, are shown for comparison.

is made in Fig. 9, in which, beside m_c^{calc}, the values of m_c^{exp} for various temperatures are plotted against applied stress. The agreement between m_c^{calc} and m_c^{exp} is very good. The values of m_c^{calc}, which are still slightly higher than that of the accepted true stress exponent $n = 5$ (Table 1), correspond to an unrealistically high applied stress of 500 MPa.

4. Discussion

The minimum creep strain rate $\dot{\varepsilon}_{\mathrm{m}}$ of the Al-8.5Fe-1.3V-1.7Si alloy at three temperatures ranging from 623 to 723 K was found to be matrix-lattice-diffusion controlled and to depend on approximately the fifth power of the applied stress. Hence, $\dot{\varepsilon}_{\mathrm{m}}(T, \sigma)$ can be described by the well-known creep equation, in which the applied stress is replaced by the effective stress $(\sigma - \sigma_{\mathrm{TH}})$, Eq. (2), where σ_{TH} is the true threshold stress. As it follows from Fig. 1, the true threshold stress really exists. This stress depends strongly on temperature; the temperature dependence of σ_{TH}/G ratio is expressed by Eq. (1). Since σ_{TH}/G depends strongly on temperature, the apparent activation energy $Q_{\mathrm{c}} = [\partial \ln \dot{\varepsilon}_{\mathrm{m}}/\partial(-1/RT)]_{\sigma}$ is generally much higher than the activation enthalpy of lattice diffusion in the alloy matrix and depends strongly on both the applied stress and temperature (Fig. 8). The same holds for the apparent applied stress exponent $m_{\mathrm{c}} = (\partial \ln \dot{\varepsilon}_{\mathrm{m}}/\partial \ln \sigma)_T$, Fig. 9. Such a creep behaviour is typical for dispersion strengthened alloys, e.g. for ODS Al alloy investigated by Čadek et al. [9].

The true threshold creep behaviour of the Al-8.5Fe-1.3V-1.7Si alloy is associated with the presence of fine $\mathrm{Al_{12}(Fe,V)_3Si}$ phase particles in this alloy. Since these particles are incoherent with the matrix, they are expected to attract the moving dislocations under creep conditions [15, 16]. Then the controlling step of climb of a dislocation past a particle is represented by the detachment of the dislocation from the particle after the climb process had been finished. The stress necessary to detach a dislocation from a particle – the detachment stress – expressed as

$$\sigma_{\mathrm{d}} = \sigma_{\mathrm{OB}}\sqrt{1 - k_{\mathrm{R}}^2} \tag{5}$$

is then identified with the true threshold stress [15, 16]. In Eq. (5), σ_{OB} is the Orowan bowing stress and k_{R} is the relaxation factor characterizing the strength of the attractive dislocation/particle interaction. The Orowan bowing stress can be expressed by the well-known simple formula [17]

$$\sigma_{\mathrm{OB}} = \frac{0.84MGb}{\lambda - d}, \tag{6}$$

in which M is the Taylor factor, λ is the mean interparticle spacing and d is the mean particle diameter. The Orowan bowing stress is thus proportional to the shear modulus G, and, consequently, σ_{OB}/G does not depend on temperature. Of course, the same holds for σ_{d}, Eq. (5), if the relaxation factor is temperature independent. The experimentally determined true threshold stress σ_{TH} depends on temperature more strongly than the shear modulus G. Therefore, σ_{TH} cannot be identified with the detachment stress σ_{d} unless a temperature dependence of the relaxation factor k_{R} is admitted. At the present time, the relaxation factor k_{R} cannot be calculated from the first principles even for an idealized case of a fine interacting particle embedded in a pure metal. Arzt and Wilkinson [15] modelled the attractive dislocation/particle interaction in a simple way. Starting with the model of these authors and taking into account possible role of impurities [10, 18], it has been shown [14] that a temperature dependence of the relaxation factor can be realistically expected.

An idea on possible values of k_{R} and especially on temperature dependence of this factor can be obtained analysing proper experimental $\dot{\varepsilon}_{\mathrm{m}}(T, \sigma)$ creep data. This was illustrated for an ODS Al-5Mg-30SiC$_{\mathrm{p}}$ composite [14] and also for 2124Al-20SiC$_{\mathrm{p}}$ composite [19]. To get such an idea for the Al-8.5Fe-1.3V-1.7Si alloy under consideration, Eq. (5) is written in the form

$$\frac{\sigma_{\mathrm{d}}}{G} = C\sqrt{1 - k_{\mathrm{R}}^2}, \tag{7}$$

where $C = \sigma_{\mathrm{OB}}/G$ is a temperature independent constant equal to $0.84Mb/(\lambda - d)$, since $\sigma_{\mathrm{OB}} \propto G$, see Eq. (6). Accepting the value of 0.85 for k_{R} at 673 K (see refs. [10, 20]) and setting $\sigma_{\mathrm{d}}/G = \sigma_{\mathrm{TH}}/G = 79.7/19232 = 4.144 \times 10^{-3}$ for this temperature, the constant $C = \sigma_{\mathrm{TH}}/(G\sqrt{1 - k_{\mathrm{R}}^2})$ is obtained equal to 7.87×10^{-3}. Accepting this value of C, values of k_{R} equal to 0.70, 0.85 and 0.92 are obtained for 623, 673 and 723 K, respectively. These values of the relaxation factor k_{R} and, consequently, also the temperature dependence of this factor seem reasonable. The factor k_{R} seems to approach the „critical“ value of ~ 0.94 with the temperature approaching ~ 775 K. The „critical“ value of k_{R} followed from the analyses of Arzt and Wilkinson [15] and Arzt and Rösler [16], Fig. 10.

As pointed out previously by the present authors [14], the above assumption on temperature dependence of the relaxation factor k_{R} seems to be the only, although perhaps still a somewhat speculative, way out of difficulty regarding the observed temperature dependence of the normalized true threshold stress σ_{TH}/G as obtained analysing the experimental $\dot{\varepsilon}_{\mathrm{m}}(T, \sigma)$ creep data for dispersion strengthened alloys. This conclusion is supported by extensive considerations on the possible role of impurities on dislocations by-passing „interacting“ particles by localized climb and final detachment [10, 18].

Regarding the process of detachment of a dislocation from a particle, Eq. (1) should be mentioned briefly. The energy Q_{o} in this equation characterizing the

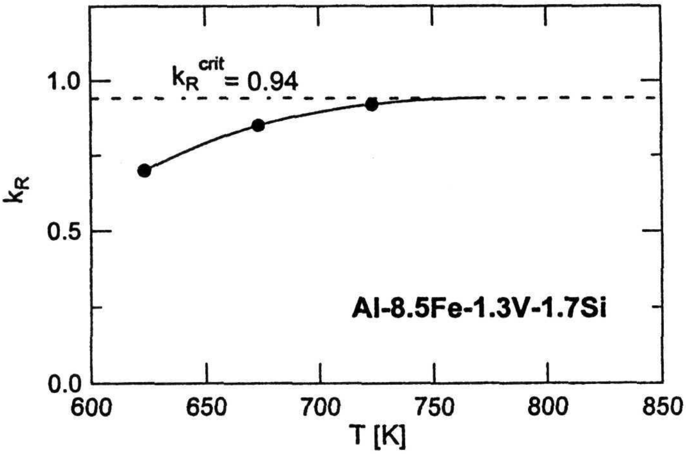

Fig. 10. Calculated values of the relaxation factor k_{R} plotted against temperature. $k_{\mathrm{R}}^{\mathrm{crit}}$ is the value of this factor that followed from the analysis of Arzt and Wilkinson [15] and Arzt and Rösler [16].

temperature dependence of σ_{TH}/G ratio is sometimes (e.g. ref. [11, 20]) compared to the energy of interaction of a dislocation with impurities on the departure side of the interacting particle. However, there is no sound evidence for such an interpretation. The present authors believe that the energy Q_{o} is a useful quantity, which, however, has not yet been given any clear physical meaning.

It should be emphasized that (as to the present authors' belief at least) further development of the concept of athermal detachment of dislocations from interacting particles can be hardly contributed to by further experimental $\dot{\varepsilon}_{\mathrm{m}}(T, \sigma)$ creep data and their analysis along the contemporary conventional line. However, the true threshold stress may be expected to disappear at high testing temperatures. This is, in fact, the case of a 2124Al-20SiC$_{\mathrm{p}}$ [21] and ODS Al-30SiC$_{\mathrm{p}}$ [22] composites. Of course, at temperatures too high for *the true threshold stress* to be observed, the attractive dislocation/particle interaction still acts (c.f. Fig. 10). At these temperatures, the detachment of dislocations from interacting particles may be expected to be thermally activated [23]. Therefore, the present authors believe (c.f. ref. [14]) that careful investigations of the true threshold stress disappearance at high creep testing temperatures and/or investigation of the transition from the athermal to the thermally activated detachment of dislocations from interacting particles as manifested by the $\dot{\varepsilon}_{\mathrm{m}}(T, \sigma)$ creep behaviour, may contribute to better understanding of the strongly temperature-dependent true threshold stress. In Fig. 11, values of the true threshold stress obtained by Peng et al. [2, 3] using the linear extrapolation technique are compared with those obtained in the present

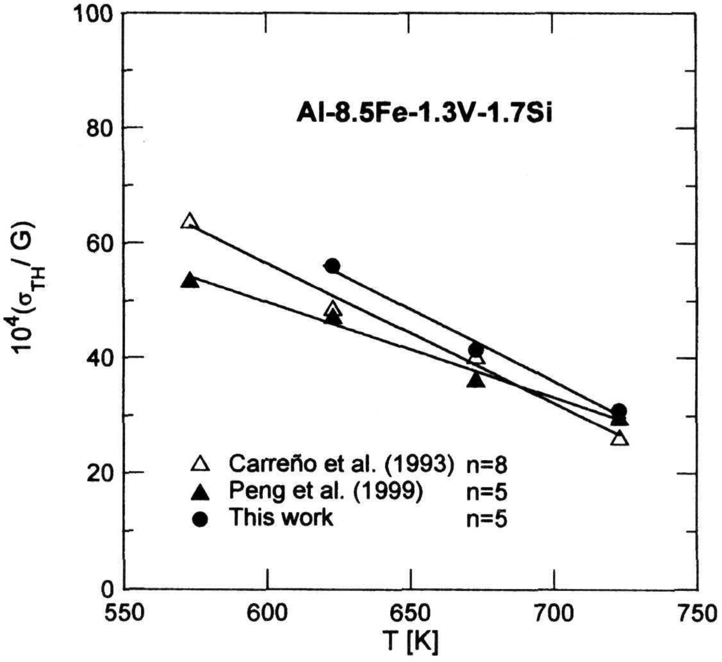

Fig. 11. Values of the threshold-stress-to-shear-modulus ratio, σ_{TH}/G, obtained in the present work and plotted against temperature compared with the σ_{TH}/G vs. T relations reported by Carreño et al. [1] and Peng et al. [2, 3].

work. In the figure also the values of σ_{TH} reported by Carreño et al. [1] for temperatures ranging from 573 to 723 K are shown, although these authors accepted the substructure invariant model of creep [24] and thus the true stress exponent $n = 8$. The difference in the temperature dependence of σ_{TH}/G ratio reported by Peng et al. [2, 3] and that following from the present $\dot{\varepsilon}_{\mathrm{m}}(T, \sigma)$ creep data analysis is not dramatic. This difference is due to higher values of σ_{TH} obtained for 623 and 673 K in the present work. In a previous paper [25] it was shown that to estimate accurately the true stress exponent n, the measured minimum creep strain rates should cover not less than five orders of magnitude (see also ref. [7]) and the same, of course, holds for the true threshold stress determination when the true stress exponent n is chosen considering a specific creep model.

5. Conclusions

In the present work, the creep behaviour of an Al-8.5Fe-1.3V-1.7Si alloy processed by rapid solidification was investigated at three temperatures ranging from

623 to 723 K using the isothermal constant-tensile-stress creep-test technique. The measured minimum creep strain rates covered seven orders of magnitude. The main results can be expressed as follows.

1. The creep behaviour is associated with the true threshold stress σ_{TH} decreasing with increasing temperature more strongly than the shear modulus G of aluminium.

2. The minimum creep strain rate $\dot{\varepsilon}_{\mathrm{m}}$ is controlled by lattice diffusion in the alloy matrix and the true stress exponent n is close to 5.

3. The apparent activation energy of creep, Q_{c}, depends strongly on both the applied stress and temperature and is generally much higher than the activation enthalpy ΔH_{L} of lattice self-diffusion in aluminium. Also the apparent stress exponent of the minimum creep strain rate, m_{c}, depends strongly on the applied stress as well as on temperature, generally being much higher than the true stress exponent n. This behaviour of Q_{c} and m_{c} is accounted for by the strong temperature dependence of the σ_{TH}/G ratio.

4. The threshold creep behaviour of the alloy is interpreted in terms of athermal detachment of dislocations from fine incoherent $\mathrm{Al_{12}(Fe,V)_3Si}$ phase particles admitting a temperature dependence of the relaxation factor k_{R} that characterizes the strength of the attractive dislocation/particle interaction.

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