Review paper

CREEP OF DISCONTINUOUS ALUMINIUM ALLOY MATRIX COMPOSITES PRODUCED BY POWDER METALLURGY ROUTE

JOSEF ČADEK, KVĚTA KUCHAŘOVÁ, KAREL MILIČKA, SHIJIE ZHU

Some recent results obtained investigating creep behaviour of discontinuous aluminium and aluminium alloy matrix composites, i.e. aluminium and/or aluminium alloys reinforced by silicon carbide particulates or short fibres and processed by powder metallurgy are overviewed with the aim to get better understanding of some characteristic features of this behaviour. The creep behaviour of the composites under consideration is primarily affected by fine alumina particles generally present in the matrix of any aluminium and aluminium alloy matrix composite processed by powder metallurgy. The known potential mechanisms of surpassing these fine incoherent particles by dislocations are briefly discussed. The most likely mechanism involving athermal detachment of dislocations from such particles is discussed with respect to the experimental fact that the true threshold stress decreases with increasing temperature more strongly than the shear modulus. In this context, disappearance of the true threshold stress at high creep testing temperatures is illustrated by some experimental results and interpreted in terms of a transition from the athermal to the thermally activated detachment of dislocations from interacting particles. Further, such topics are discussed as the relation between the true stress exponent and the true threshold stress, the role of simultaneously acting load transfer effect and the threshold effect and, especially, the possible origin of temperature dependence of the relaxation factor characterizing the strength of dislocation/fine alumina particle interaction.

Key words: discontinuous composites, aluminium matrix, fine alumina particles, creep behaviour, threshold stress, load transfer effect

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CREEP DISKONTINUÁLNÍCH KOMPOZITŮ S HLINÍKOVOU MATRICÍ PŘIPRAVOVANÝCH TECHNIKOU PRÁŠKOVÉ METALURGIE

Některé nedávné výsledky studia creepového chování diskontinuálních kompozitů s matricí tvořenou hliníkem či vhodnou slitinou hliníku (to je v nejjednodušším případě hliníku vyztuženého partikulemi či krátkými vlákny karbidu křemíku připraveného technikou práškové metalurgie) jsou analyzovány s cílem získat lepší představy o některých charakteristických rysech tohoto chování. Creepové chování uvažovaných kompozitů je primárně ovlivněno jemnými částicemi oxidu hliníku obecně přítomnými v matrici kteréhokoliv kompozitu tvořené hliníkem nebo vhodnou slitinou hliníku a připraveného technikou práškové metalurgie. Stručně diskutovány jsou v současné době známé potencionální mechanismy překonávaní těchto nekoherentních částic dislokacemi. Nejpravděpodobnější mechanismus zahrnující atermické odpoutávání dislokací od takových částic je diskutován se zřetelem k experimentálnímu poznatku, že skutečné prahové napětí klesá se vzrůstající teplotou mnohem rychleji nežli smykový modul matrice. V tomto kontextu je vymizení skutečného prahového napětí ilustrováno některými experimentálními výsledky a interpretováno jako důsledek přechodu od atermického k termicky aktivovanému odpoutávání dislokací od interagujících částic. Dále jsou diskutovány otázky relace mezi skutečným napěťovým exponentem a skutečným prahovým napětím, úloha simultánně působícího efektu přenosu zatížení na výztuhu a prahové napětí, a zejména možný zdroj teplotní závislosti relaxačního faktoru charakterizujícího „pevnost“ interakce jemných nekoherentních částic v matrici s dislokacemi.

1. Introduction

It is well known that discontinuous reinforcement of aluminium with silicon carbide particulates (SiC_p) or short fibres (SiC_f) enhances the Young modulus significantly even at high temperatures [1]. On the other hand, it is equally well known that such a reinforcement in itself increases the creep strength relatively slightly. In fact, particulates or short fibres do not represent effective obstacles to dislocation motion owing to their large size and large mean spacing even at volume fractions of reinforcement as high as 30 % [2]. The discontinuous reinforcement is supposed to introduce a load transfer effect [3, 4]. Due to this effect, the stress acting in the composite matrix is lower than the external applied stress and, consequently, the creep strain rate of the composite is lower than that of the matrix metal at the same applied stress. The concept of load transfer thus predicts a close correlation between the creep behaviour (creep mechanism) of the matrix metal and the creep behaviour of the composite. However, as a rule, such a correlation is not observed [2, 5–7]. Specifically, the applied stress and temperature dependence of the minimum creep strain rate of the discontinuous aluminium matrix composites, processed by powder metallurgy, is generally much stronger than that of the matrix metal – aluminium. At the present time, it is already well recognized (e. g. refs. [8, 9]) that the above difference between the creep behaviour of aluminium and that of discontinuously reinforced aluminium is attributable to the presence of fine alumina particles necessarily introduced into the composite matrix during their processing based on the techniques of powder metallurgy (PM). This was clearly demonstrated correlating the creep behaviour in Al-based solid solution alloys and powder metallurgy Al-alloys [9]. For PM Al-alloys, namely PM 6061Al and PM 2124Al alloys, the apparent stress exponent m_c defined as $m_c = (\partial \ln \dot{\varepsilon}_m / \partial \ln \sigma)_T$, was found to decrease with increasing applied stress σ (in the definition equation for m_c, $\dot{\varepsilon}_m$ means the minimum creep strain rate and T the temperature). At the same time, the apparent activation energy of creep, $Q_c = [\partial \ln \dot{\varepsilon}_m / \partial(-1/RT)]_\sigma$, (where R is the gas constant) was found to decrease with increasing stress. At low applied stresses, m_c for PM Al alloys is generally much higher than that for aluminium, i. e. ~ 5, and the value of Q_c is much higher than that for aluminium, i. e. ~ 140 kJ mol^{-1}. At the same time, m_c as well as Q_c decrease with increasing temperature. Qualitatively similar relations between m_c and Q_c and applied stress as well as temperature were found for PM 6061Al-30SiC$_p$ [2], PM Al-30SiC$_p$ [6], PM 2124Al-20SiC$_p$ [7], ODS Al-30SiC$_p$ [10], ODS Al-5Mg-30SiC$_p$ [11], and some other composites with aluminium or aluminium alloy matrices (ODS, as usually, means oxide dispersion strengthening). Creep behaviour similar to that of ODS Al-30SiC$_p$ composite was observed investigating unreinforced ODS aluminium [12].

The "anomalous" creep behaviour, as characterized by high values of both the apparent stress exponent m_c and the apparent activation energy of creep Q_c, was interpreted in terms of true threshold stress concept (e.g. refs. [2, 6, 11]). The true threshold stress is defined as the applied stress below which creep does not occur at all or does occur by a mechanism different from that acting above it. The true threshold stress σ_{TH} generally decreases with increasing temperature. However, by definition, it does not depend on applied stress.

The above-mentioned observations on creep behaviour of PM Al alloys including ODS Al alloy clearly demonstrate that the true threshold stress in discontinuous PM aluminium and aluminium alloy matrix composites is not associated with the presence of particulate and/or short fibre reinforcement, but with the presence of fine alumina particles in the composite matrix.

2. The aim of the present paper

The aim of the present overview is to analyze possible mechanisms of dislocation motion in the field of fine alumina particles (see also ref. [13]) generally present in any aluminium and/or aluminium alloy matrix of any discontinuous composite processed by powder metallurgy techniques. In the analysis, a special attention is paid to the fact that the temperature dependence of the true threshold stress, determined from the experimental $\dot{\varepsilon}_m(\sigma, T)$ or $\dot{\gamma}_m(\tau, T)$ creep data, is strong, if compared to the temperature dependence of the shear modulus of the matrix metal or alloy. The goal is to explain the σ_{TH}/G or τ_{TH}/G ratio decreasing rather strongly with increasing temperature, since just owing to this temperature dependence of the true threshold stress, the discontinuous PM aluminium and aluminium alloy matrix composites exhibit "anomalous" creep behaviour of the temperature and applied-stress dependence of the minimum creep strain rate. An example of such a behaviour is given first.

3. An example: Creep behaviour of discontinuous ODS aluminium reinforced by silicon carbide particulates, an ODS Al-30SiC$_p$ composite

As mentioned in the introductory section of the present paper, the discontinuous reinforcement of aluminium and/or aluminium solid solution alloys does not increase significantly their creep strength (usually characterized by the minimum creep strain rate as a function of applied stress and temperature). The creep strength of PM composites under consideration may be increased rather dramatically when the matrix is dispersion strengthened, e.g. by fine alumina particles. To demonstrate such a creep strengthening some results obtained investigating creep behaviour of a PM ODS Al-30SiC$_p$ composite [10] will be shown in the present section. In the PM ODS Al-30SiC$_p$ composite, the volume fraction of SiC particulates of the nominal mean size of ~ 4.5 μm was ~ 30 % and the volume fraction of Al_2O_3 particles in the *composite matrix* was nominally 1.85 %. The mean alumina particle size was estimated to 25 nm.

Constant-tensile-stress creep tests were performed at three testing temperatures ranging from 623 to 723 K, the measured minimum creep strain rates covered nearly seven orders of magnitude, the lowest of them were well below 10^{-9} s^{-1}.

Relations between the minimum creep strain rate $\dot{\varepsilon}_m$ and the applied stress σ in double logarithmic co-ordinates are shown in Fig. 1. For comparison, also the relation for PM Al-30SiC$_p$ composite at 673 K [6] is shown in the figure. It can be seen that the minimum creep strain rates of ODS Al-matrix composite are up to eight orders of magnitude lower than those of the Al-matrix composite. The creep strengthening due to introduction of additional fine alumina particles into Al-30SiC$_p$ composite matrix is thus really dramatic.

From Fig. 1 it can be seen that the apparent stress exponent m_c defined in Section 1 decreases with increasing applied stress – at a given applied stress it decreases with increasing temperature. The increase of the apparent stress exponent with decreasing temperature strongly suggests the creep behaviour associated with the true threshold stress σ_{TH}.

In Fig. 2, values of $\dot{\varepsilon}_m^{1/n}$ are plotted against σ in double linear co-ordinates for $n = 5$. For this true stress exponent, the $\dot{\varepsilon}_m^{1/n}$ vs. σ relations are linear. Extrapolating them to $\dot{\varepsilon}_m = 0$, the true threshold stress values are obtained. These

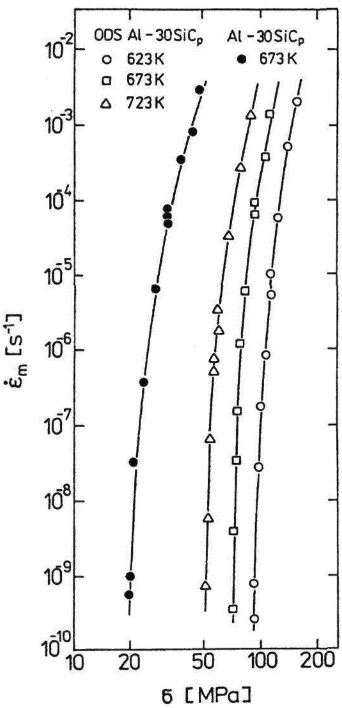

Fig. 1. ODS Al-30SiC$_p$ composite. Relations between minimum creep strain rate $\dot{\varepsilon}_m$ and applied stress σ in double logarithmic co-ordinates. The relation for Al-30SiC$_p$ composite at 673 K is shown for comparison.

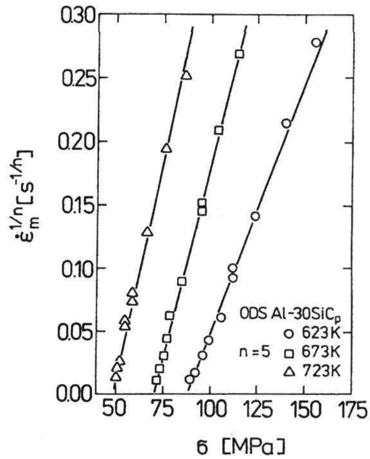

Fig. 2. ODS Al-30SiC$_p$ composite. Values of $\dot{\varepsilon}_m^{1/n}$ plotted against σ in double linear coordinates for $n = 5$.

values are plotted against temperature in Fig. 3(a). The true threshold stress decreases with increasing temperature approximately linearly. In the same figure, the σ_{TH} vs. T relation for PM Al-30SiC$_p$ composite is shown for comparison. The values of the threshold stress in ODS Al-30SiC$_p$ composite are higher by a factor of about 3.5 than those in the Al-30SiC$_p$ composite. Just owing to this fact, the creep strength of the former composite is much higher than the creep strength of the latter one.

However, the result which is most important from the point of view of the "anomalous" creep behaviour mentioned above is shown in Fig. 3(b). From this figure it follows that the true threshold stress decreases with increasing temperature more strongly than the shear modulus G; the σ_{TH}/G ratio *decreases* with increasing temperature.

It should be pointed out that the values of the threshold stress for creep in ODS Al-30SiC$_p$ composite were obtained applying three different procedures. Leaving aside the procedure [14] estimating the true threshold stress by extrapolation $\dot{\varepsilon}_m$ vs. σ relations in double logarithmic co-ordinates to $\dot{\varepsilon}_m = 10^{-10}$ s^{-1}, the results obtained applying other two procedures (Procedures B and C) are listed in Table 1. With regard to the footnotes to this table the procedures do not require any comment. The values of threshold stress values provided by both these procedures are in very good mutual agreement and the same holds for the values of the true stress exponent n.

Using the values of σ_{TH} obtained by means of Procedure B and listed in Table 1, the normalized creep strain rates $\dot{\varepsilon}_m b^2/D_L$ are plotted against $(\sigma - \sigma_{\mathrm{TH}})/G$ in double logarithmic co-ordinates in Fig. 4; b is the Burgers vector. The coefficient of matrix lattice diffusion, D_L was identified with the coefficient of lattice self-diffusion in aluminium [15] and the shear modulus G with the shear modulus of aluminium [16]. From Fig. 4 it can be seen that all the data points fit a single straight line with the slope of $\sim$ 5, which strongly suggests the matrix lattice diffusion as the creep-strain-rate controlling process and the true stress exponent n close to 5.

In Fig. 4, the relation between $\dot{\varepsilon}_m b^2/D_L$ and $(\sigma - \sigma_{\mathrm{TH}})/G$ for the Al-30SiC$_p$ composite [6] is shown for comparison. At any given normalized effective stress $(\sigma - \sigma_{\mathrm{TH}})/G$ the normalized creep strain rate $\dot{\varepsilon}_m b^2/D_L$ of the ODS Al-30SiC$_p$ composite is slightly more than an order of magnitude lower than that of the Al-30SiC$_p$ composite. Of course, this difference has nothing common with the difference of true threshold stress, Fig. 3, and has been discussed in detail in ref. [10].

Table 1. ODS Al-30SiC$_p$ composite. Comparison of the values of n and σ_{TH} as estimated by two different procedures

T [K]	Procedure B		Procedure C	
	n	σ_{TH} [MPa]	n	σ_{TH} [MPa]
623	5.0	88.9 ± 2.2	5.1	85.3 ± 0.3
673	5.0	69.7 ± 1.6	5.1	69.3 ± 0.3
723	5.0	47.6 ± 1.1	5.1	49.1 ± 0.3

Procedure B: σ_{TH} estimated assuming linear relations between $\dot{\varepsilon}_m^{1/n}$ and σ in double linear co-ordinates for $n = 5$ and extrapolating these relations to $\dot{\varepsilon}_m = 0$.
Procedure C: σ_{TH} and n estimated optimizing these parameters as well as the dimensionless constant A_o in the equation $\dot{\varepsilon}_m b^2/D_L = A_o[(\sigma - \sigma_{\mathrm{TH}})/G]^n$ under the assumption that $\dot{\varepsilon}_m$ is matrix lattice diffusion controlled. For A_o a value of $(3.477 \pm 0.501) \times 10^{21}$ was obtained.

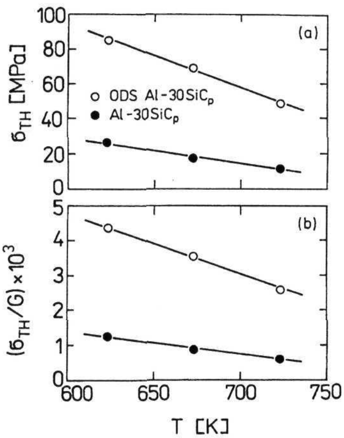

Fig. 3. ODS Al-30SiC$_p$ composite. Values of true threshold stress σ_{TH} (a) and values of the σ_{TH}/G (b) plotted against temperature.

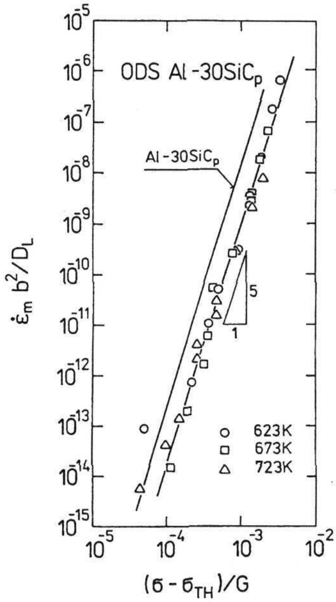

Fig. 4. ODS Al-30SiC$_p$ composite. Normalized strain rates $\dot{\varepsilon}_m b^2/D_L$ plotted against normalized effective stresses $(\sigma - \sigma_{TH})/G$ in double logarithmic co-ordinates.

The results shown in Fig. 4 suggest the following equation describing temperature and applied stress dependence of the minimum creep strain rate $\dot{\varepsilon}_m$:

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

where A is a dimensionless constant and the true stress exponent $n = 5$.

Combining Eq. (1) with the equations defining the apparent activation energy of creep, Q_c, and the apparent stress exponent m_c of minimum creep strain rate (see Section 1), the following expressions for Q_c and m_c, respectively, are obtained:

$$Q_c = \left[\frac{\partial \ln \dot{\varepsilon}_m}{\partial(-1/RT)}\right]_\sigma = \Delta H_L - \frac{nRT^2}{G}\left(\frac{G}{\sigma - \sigma_{TH}}\frac{d\sigma_{TH}}{dT} + \frac{n-1}{n}\frac{dG}{dT}\right) \tag{2}$$

and

$$m_c = \left(\frac{\partial \ln \dot{\varepsilon}_m}{\partial \ln \sigma}\right)_T = \frac{n\sigma}{\sigma - \sigma_{TH}}. \tag{3}$$

From these expressions – Eqs. (2) and (3) – it follows that both the apparent activation energy Q_c and the apparent stress exponent m_c depend not only on applied stress, but also on temperature, since the threshold stress depends on temperature. As shown in a previous paper [6] concerning creep behaviour of an Al-30SiC$_p$ composite, Eqs. (2) and (3) can be used to estimate values of $Q_c(\sigma, T)$ and $m_c(\sigma, T)$ if $\sigma_{TH}(T)$ is known. In the mentioned paper [6], the calculated values of Q_c and m_c were compared with the values obtained directly from the experimental $\dot{\varepsilon}_m(\sigma, T)$ creep data. The agreement was found very good. This result encouraged Čadek et al. [10] to calculate and present values of $Q_c = Q_c^{calc}$ and $m_c = m_c^{calc}$ for various applied stresses and a temperature of 673 K (at which σ_{TH} = 69.3 MPa and $d\sigma_{TH}/dT = 0.412$ MPa K^{-1}). The results are listed in Table 2. It can be seen that at an applied stress only slightly higher than the true threshold stress, Q_c^{calc} as well as m_c^{calc} attain extremely high values. However, even at an unrealistically high applied stress of 300 MPa, Q_c^{calc} is significantly higher than the activation enthalpy of lattice self-diffusion in aluminium, ΔH_L = 142 kJ mol^{-1} [15], and m_c^{calc} is higher than the value of the true stress exponent $n = 5$, typical for aluminium [17].

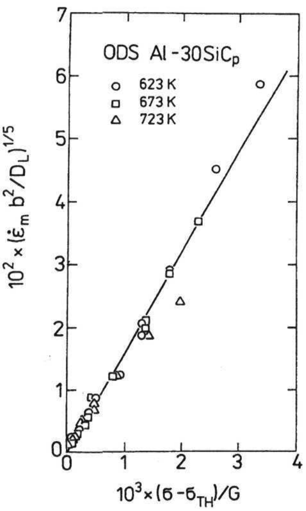

Fig. 5. ODS Al-30SiC$_p$ composite. Values of $(\dot{\varepsilon}_m b^2/D_L)^{1/n}$ plotted against $(\sigma - \sigma_{TH})/G$ in double linear co-ordinates.

Table 2. ODS Al-30SiC$_p$ composite. Values of the apparent activation energy of creep, Q_c^{calc}, and the apparent stress exponent of minimum creep strain rate, m_c^{calc}, calculated by means of Eqs. (2) and (3), respectively, for 673 K (σ_{TH} = 69.3 MPa, $d\sigma_{TH}/dT = 0.412$ MPa K^{-1}, see Table 1, Procedure B)

Parameter[a)]	σ [MPa]				
	73.5	85	114[b)]	150	300
Q_c^{calc} [kJ mol^{-1}]	2089	624	325	244	238
m_c^{calc}	99.3	28.0	13.5	9.3	6.5

a) Both Q_c and m_c depend not only on applied stress but also on temperature
b) The highest σ at which $\dot{\varepsilon}_m$ was measured at 673 K

The phenomenological creep equation (1) can be rewritten as

$$\left(\frac{\dot{\varepsilon}_{\mathrm{m}} b^2}{D_{\mathrm{L}}}\right)^{\frac{1}{n}} = A^{\frac{1}{n}} \frac{\sigma - \sigma_{\mathrm{TH}}}{G}. \tag{4}$$

Values of $(\dot{\varepsilon}_{\mathrm{m}} b^2/D_{\mathrm{L}})^{1/n}$ are plotted against $(\sigma - \sigma_{\mathrm{TH}})/G$ in double linear co-ordinates for the true stress exponent n equal to 5 in Fig. 5. As expected, the data points fit a straight line passing through the origin. This way of presentation of creep data is an alternative to their presentation as $\dot{\varepsilon}_{\mathrm{m}} b^2/D_{\mathrm{L}}$ vs. $(\sigma - \sigma_{\mathrm{TH}})/G$ in bilogarithmic co-ordinates. The former of these two presentations of creep data involving the true threshold stress is used most frequently in the literature.

4. Dislocation motion in the field of fine "interacting" particles

Alumina particles appearing during processing discontinuous aluminium and aluminium alloy matrix composites by PM route are incoherent with the matrix. Because of particle size (typically 20 nm) and interparticle spacing (typically 150 nm), they represent effective obstacles to dislocation motion. Of course, the density of alumina particles can be increased introducing similar particles in the course of composite processing. The composite matrix is then represented by an ODS aluminium alloy.

Four mechanisms to surmount the dispersed particles have been considered [13] to be associated with a true threshold stress:

(i) Dislocation bowing out between particles is associated with the stress τ_{OB} known as the Orowan bowing stress, which is given by a well-known simple equation [18]

$$\tau_{\mathrm{OB}} = 0.8 \frac{Gb}{\lambda - d}, \tag{5}$$

where λ is the interparticle spacing and d is the particle diameter. The Orowan bowing is the athermal mechanism, the bowing stress τ_{OB} represents a true threshold stress.

(ii) When a dislocation surmounts a particle by localized climb, a stress τ_{b} is needed to create an additional length of the dislocation. This stress is expressed as [19, 20]

$$\tau_{\mathrm{b}} = 0.3 \frac{Gb}{\lambda}. \tag{6}$$

If the localized climb really occurs, the "back" stress τ_{b} represents a true threshold stress.

(iii) When a particle is incoherent with the matrix, it may show an attractive interaction for a dislocation at high temperatures. Then, a dislocation surmounts an "interacting" particle by localized climb which is relatively easy as compared to the process of dislocation detachment from the departure side of the particle. The stress needed to detach the dislocation from the particle – the detachment stress τ_d – then represents the true threshold stress given by [21, 22]

$$\tau_d = \frac{Gb}{\lambda}\sqrt{1 - k_R^2} \cong \tau_{OB}\sqrt{1 - k_R^2}. \tag{7}$$

In Eq. (7), k_R is the relaxation factor characterizing the strength of the attractive dislocation/particle interaction and τ_{OB} is the Orowan bowing stress (Eq. (5)).

(iv) Mishra et al. [23] suggested that a lattice dislocation cannot enter the particle/matrix interface and surmount the interacting particle by localized climb, unless it dissociates into interface dislocations. The stress needed for such dissociation is expressed as

$$\tau_{diss} = 10^{-3}\left(\frac{2Gb}{d}\right)\exp\left(\frac{10d}{\lambda}\right). \tag{8}$$

Of course, the interface dislocations must react to form a lattice dislocation at the departure side of the particle.

Using the proper values of d and λ, Li and Langdon [13] concluded that the experimentally determined normalized threshold stress τ_{TH}/G lies within the range $\tau_{diss}/G < \tau_{TH}/G < \tau_b/G$. However, just the stresses τ_b and τ_{diss} are questionable. First, the formula for τ_b developed by Arzt and Ashby [20] was latter shown by Rösler and Arzt [24] to be hardly acceptable. These authors have demonstrated that the localized climb past a particle does not take place when this particle is "non-interacting", which is certainly not the case of alumina particles in PM 6061 Al alloy as well as in the matrix of PM 6061Al-30SiC$_p$ composite, to give two examples at least.

The model proposed by Mishra et al. [23] is still a subject of dispute. The present attitude to this model seems rather reserved for the reasons discussed e. g. by Li et al. [8].

The crucial problem of the considered models and/or processes (i) to (iv) consists in the fact that all of them predict the true threshold stress proportional to the shear modulus, while the experimental true threshold stress normalized to the shear modulus, τ_{TH}/G, decreases with increasing temperature rather strongly. This problem is still waiting for its satisfactory solution. The present authors believe that such a solution should be sought accepting the "detachment concept" described under (iii) above. This is because Eq. (7) contains the relaxation factor k_R which may be temperature dependent; just a temperature dependence of k_R might, in principle, explain the observed temperature dependence of τ_{TH}/G.

Therefore, in the following section of the present paper, only the detachment concept will be considered. Mohamed and collaborators [2, 8, 25] have speculated about the τ_{TH}/G ratio decreasing with increasing temperature just accepting this concept. Of course, the considerations of these authors will be taken into account.

5. The detachment concept and the relaxation factor k_R

Interaction of dislocations with incoherent particles was treated in several papers by Srolovitz et al. [26, 27]. These authors have shown that at high temperatures (the temperatures at which diffusion is fast enough) the interaction may be attractive, since the dislocation can relax a part of its energy when entering the particle/matrix interface.

Arzt and Wilkinson [21] modelled, in a simple way, the attractive dislocation/particle interaction introducing the relaxation factor k_R equal to Γ'/Γ, where Γ is the line energy of a dislocation in the matrix and Γ' the line energy of the dislocation in the particle/matrix interface.

The temperature dependence of the line energy Γ of a dislocation in the matrix lattice is the same as that of the shear modulus G. However, the temperature dependence of the line energy in the particle/matrix interface, Γ', may be significantly different. In fact, this energy may be affected by segregation of impurities to incoherent particle/matrix interface. The concentration of an impurity in the particle/matrix interface is proportional to $\exp[E_s/kT]$, where E_s is the interaction energy of an impurity atom with the interface and k is the Boltzmann constant. Through the energy term E_s/kT, the relaxation factor k_R may be influenced. Besides, formation of impurity atmospheres around dislocations at the departure side of a particle may affect not only the detachment stress, Eq. (7), but also its temperature dependence [8], just through the relaxation factor k_R.

Of course, the considerations of the possible effects of impurities are on the level of speculation [8]. At the present time, the relaxation factor k_R cannot be calculated from the first principles even for the simplest possible (idealized) case of a fine interacting particle embedded in a pure metal. However, an idea on possible values of k_R and on the temperature dependence of this factor can be obtained analyzing proper experimental $\dot{\gamma}_m(\tau, T)$ or $\dot{\varepsilon}_m(\sigma, T)$ creep data. This can be illustrated, e.g. by a set of $\dot{\varepsilon}_m(\sigma, T)$ creep data for an Al-30SiC$_p$ [6], ODS Al-30SiC$_p$ [10], and ODS Al-5Mg-30SiC$_p$ [11] composite. Only the last of these composites will be considered in the following. (The basic $\dot{\varepsilon}_m(\sigma, T)$ and some other creep data for the ODS Al-30SiC$_p$ composite have been presented in Section 3).

The minimum creep strain rates of the Al-5Mg-30SiC$_p$ composite with the matrix strengthened by 1.85 vol.% fine alumina particles, an ODS Al-5Mg-30SiC$_p$ composite, were measured at three temperatures ranging from 623 to 723 K; the measured creep strain rates covered seven orders of magnitude. Relations between

$\dot{\varepsilon}_m$ and σ for all three testing temperatures are presented in Fig. 6 in double logarithmic co-ordinates. From Fig. 7 it can be seen that the apparent stress exponent of minimum creep strain rate, $m_c = (\partial \ln \dot{\varepsilon}_m / \partial \ln \sigma)_T$ decreases with increasing stress σ; at low stresses it reaches values as high as 85 (see Table 3 in ref. [11]), while at high stresses it is lower than 10. It should be mentioned that a power-law breakdown sets in at $\sigma \cong 160$ MPa. Consequently, m_c increases with increasing σ above this value. The apparent activation energy of creep, Q_c, decreases strongly with increasing applied stress at any given temperature and with increasing temperature at any given applied stress (Fig. 8). The true threshold stress σ_{TH} was determined extrapolating $\dot{\varepsilon}_m^{1/n}$ vs. σ relations in double linear co-ordinates to $\dot{\varepsilon}_m = 0$ (Procedures B). The relations between $\dot{\varepsilon}_m^{1/n}$ vs. σ were linear for all the temperatures under consideration for the true stress exponent $n = 5$ [11]. The values of the true threshold stress σ_{TH} are plotted against temperature in Fig. 9. In this figure, also the relations between σ_{TH} normalized to the shear modulus G, σ_{TH}/G, vs. T are shown.

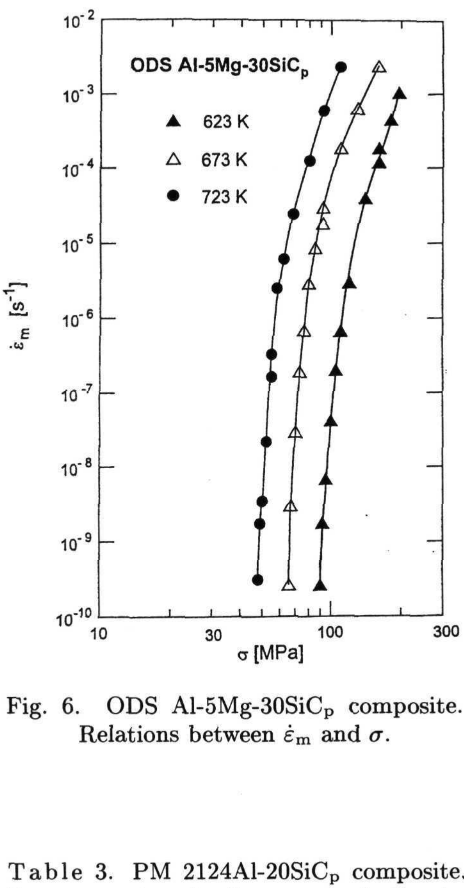

Fig. 6. ODS Al-5Mg-30SiC$_p$ composite. Relations between $\dot{\varepsilon}_m$ and σ.

Table 3. PM 2124Al-20SiC$_p$ composite. Values of the apparent stress exponent m_c, the apparent activation energy (estimated from $\dot{\gamma}_m(T, \tau)$ creep data and corrected for the temperature dependence of the shear modulus) Q_c, the detachment stress τ_d and the relaxation factor k_R for temperatures 723 and 748 K and various values of the applied shear stress τ

τ [MPa]	$T = 723$ K				$T = 748$ K			
	m_c	Q_c [kJ mol^{-1}]	τ_d [MPa]	k_R	m_c	Q_c [kJ mol^{-1}]	τ_d [MPa]	k_R
3.0	5.0	259	13.8	0.967	5.8	254	11.4	0.966
5.0	7.8	265	17.2	0.963	8.3	261	15.4	0.962
7.0	10.9	494	41.7	0.937	13.6	479	32.1	0.935
10.0	10.9	708	90.1	0.919	13.6	693	68.9	0.917

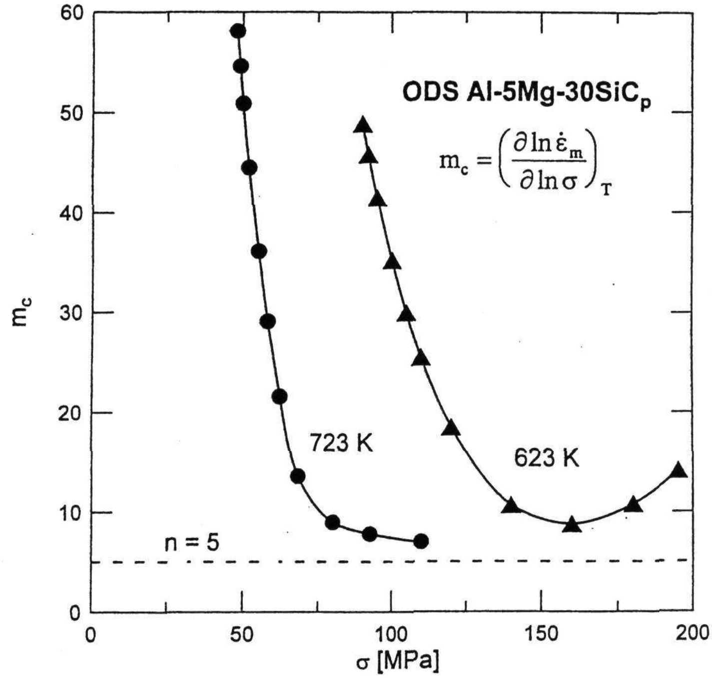

Fig. 7. ODS Al-5Mg-30SiC$_p$ composite. Relations between the apparent stress exponent m_c and applied stress σ for 623 and 723 K. At $\sigma \cong 160$ MPa, power-law breakdown sets in and, consequently, m_c starts to increase with increasing σ.

Again, σ_{TH} is not proportional to the shear modulus; it decreases with increasing temperature more strongly than this modulus.

In Fig. 10, the minimum creep strain rates normalized to the lattice self-diffusion coefficient in aluminium, D_L [15] and the squared Burgers vector b are plotted against the effective stress $\sigma - \sigma_{TH}$ normalized to the shear modulus of aluminium [16] in double logarithmic co-ordinates. All the data points fit a single straight line, which suggests, as expected, the minimum creep strain rate controlled by lattice diffusion in the matrix. The true stress exponent represented by the slope of the straight line is very close to 5. Hence, the temperature and applied stress dependence of the minimum creep strain rate can be expressed by Eq. (1) in which the true stress exponent $n = 5$. Thus, the true activation energy (activation enthalpy of creep) ΔH_c is equal to the activation enthalpy of lattice self-diffusion in the composite matrix, while the apparent activation energy of creep

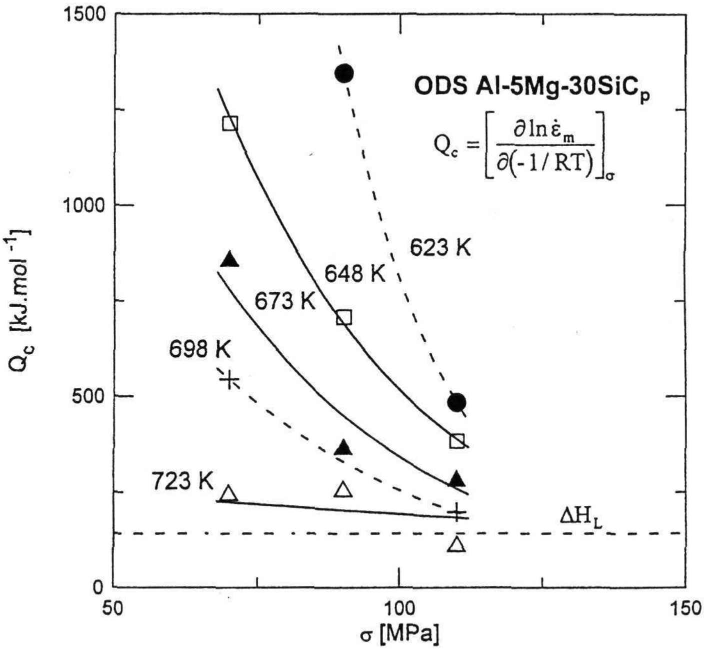

Fig. 8. ODS Al-5Mg-30SiC$_p$ composite. Relations between the apparent activation energy Q_c and applied stress σ for various temperatures.

$Q_c = [\partial \ln \dot{\varepsilon}_m / \partial(-1/RT)]_\sigma$ reaches values as high as $\sim$ 1380 kJ mol^{-1} at 623 K and the stresses only slightly higher than the true threshold stress (see Table 3 in ref. [11]). Again, the difference between Q_c and ΔH_c is due to the strong temperature dependence of the normalized threshold stress, σ_{TH}/G.

Since the threshold stress, experimentally determined from $\dot{\varepsilon}_m(\sigma, T)$ creep data, is not proportional to the shear modulus G, it cannot be identified with the stresses σ_{OB}, σ_b and σ_{diss} defined above (Eqs. (5), (6), and (8)). Neither, it can be identified with the detachment stress σ_d, Eq. (7), unless the relaxation factor k_R decreases with increasing temperature. To get an idea on a *possible* variation of k_R with temperature, the normalized threshold stress, σ_{TH}/G, will be identified with σ_d/G assuming $k_R = 0.85$ for 673 K. This value of k_R seems to be realistic, see refs. [8, 13]. Thus, from Eq. (7)

$$\frac{\sigma_d}{G} = K\sqrt{1 - k_R^2}, \tag{9}$$

where $K = \sigma_{OB}/G$ is a temperature independent constant equal to $0.84Mb/(\lambda - d)$

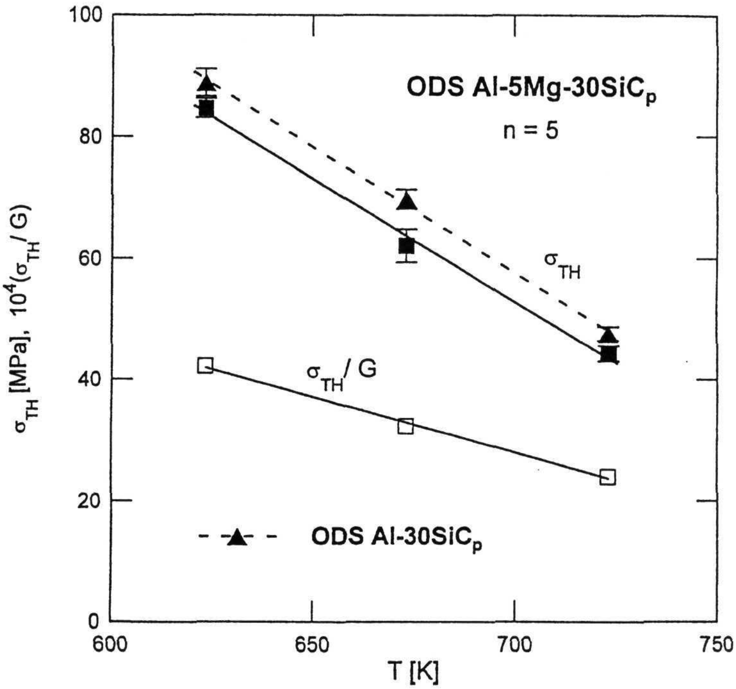

Fig. 9. ODS Al-5Mg-30SiC$_p$ composite. Relations between true threshold stress σ_{TH} and σ_{TH}/G ratio and temperature. The relation between σ_{TH} and T for an ODS Al-30SiC$_p$ composite [12] is shown for comparison.

(since $\sigma_{OB} \propto G$, see Eq. (5)), where M is the Taylor factor. Setting $k_R = 0.85, M = 3$ and $\sigma_d/G = \sigma_{TH}/G = 3.23 \times 10^{-3}$ for 673 K, this constant is found to be equal to 6.125×10^{-3}. Using this value of K, the factor k_R was estimated to 0.92 and 0.72 for 723 and 623 K, respectively. These values of k_R, and, consequently, also the temperature dependence of the relaxation factor k_R seem reasonable.

At the present time, an assumption on the temperature dependence of the relaxation factor k_R seems to be the only, though perhaps a somewhat formal, way out of difficulty regarding the observed temperature dependence of the normalized true threshold stress, σ_{TH}/G, as obtained analyzing experimental $\dot{\varepsilon}_m(\sigma, T)$ creep data. This holds not only for discontinuous aluminium and aluminium alloy matrix composites, but also for unreinforced ODS aluminium alloys [12]. This conclusion is supported by the above considerations on a possible role of impurities in dislocation-passing "interacting" particles by localized climb and final detachment.

Any further development of the concept of athermal detachment of dislocations from interacting particles as the creep strain rate controlling mechanism can be hardly contributed to by further experimental $\dot{\varepsilon}_{\mathrm{m}}(\sigma, T)$ creep data and their analysis along the contemporary conventional line. However, since the true threshold stress decreases rather quickly with increasing temperature, it may be expected to disappear at high testing temperatures. Of course, even at the temperatures too high for the *true threshold* stress to be defined, the attractive dislocation/particle interaction acts. At these temperatures, the detachment of dislocations from interacting particles can be expected to be thermally activated [28]. Therefore, the present authors believe that careful investigations of the true threshold stress disappearance at high creep testing temperatures and/or investigations of athermal to thermally activated detachment of dislocations from interacting particles as manifested by the $\dot{\varepsilon}_{\mathrm{m}}(\sigma, T)$ creep behaviour may contribute to the understanding of the strongly temperature dependent true threshold stress significantly.

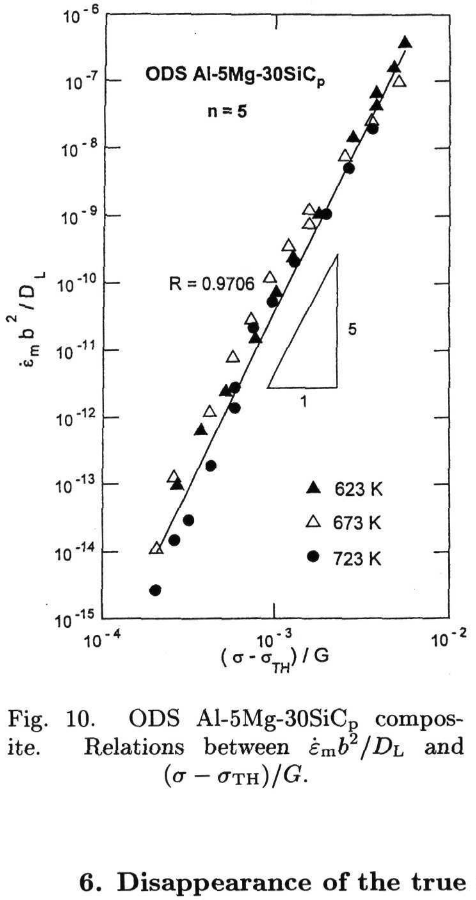

Fig. 10. ODS Al-5Mg-30SiC$_{\mathrm{p}}$ composite. Relations between $\dot{\varepsilon}_{\mathrm{m}} b^2/D_{\mathrm{L}}$ and $(\sigma - \sigma_{\mathrm{TH}})/G$.

6. Disappearance of the true threshold stress at high testing temperatures. Transition from athermal to thermally activated detachment of dislocations from fine interacting particles

Several years ago, Gonzáles-Doncel and Sherby [29] reported a disappearance of the true threshold stress associated with creep behaviour of discontinuous aluminium and aluminium alloy matrix composites at a temperature close to 735 K. These authors analyzed most of the creep data for these composites published till 1992. Although their analysis is based on the assumption of validity of the substructure invariant model of creep [30] predicting the threshold stress exponent $n = 8$, their conclusion about the disappearance of the true threshold stress at high temperatures can be taken as an important impetus for further attempts to identify the origin of the strong temperature dependence of the true threshold stress in creep of PM discontinuous aluminium and aluminium alloy matrix composites as well as PM DS and ODS aluminium matrix alloys.

Here, the explanation of creep threshold disappearance given by Gonzáles-Doncel and Sherby [29] is left aside, since it is briefly discussed elsewhere [31].

More recently, the disappearance of the true threshold stress at high testing temperatures was observed by Čadek et al. [7] in a PM 2124Al-20SiC$_p$ composite and by Zhu et al. [32] in a DS Al-Al$_4$C$_3$-10SiC$_p$ composite.

6.1 PM 2124Al-20SiC$_p$ composite

The disappearance of the true threshold stress in creep of a commercially produced PM 2124Al-20SiC$_p$ composite at temperatures higher than 698 K is demonstrated in Fig. 11. At 648 K, the apparent stress exponent $m_c = (\partial \ln \dot{\gamma}_m / \partial \ln \tau)_T$ *decreases* with increasing applied shear stress τ, which suggests an existence of the true threshold stress τ_{TH}. On the other hand, at 748 K, m_c *increases* with increasing applied shear stress, which strongly suggests *an absence* of τ_{TH}. Since, at this temperature, the apparent stress exponent m_c attains values significantly higher than $\sim$ 5 typical for the matrix alloy, one can speak of a threshold behaviour (c.f. ref. [28]), although the *true* threshold stress is not defined; instead, a back stress depending on applied stress (in contrast to the true threshold stress), can be defined. The same holds for the temperature 723 K. The apparent activation energy of creep, $Q_c = [\partial \ln \dot{\gamma}_m / \partial(-1/RT)]_\tau$ estimated from the difference between the minimum creep strain rates and corrected for the temperature dependence of shear modulus [16] at 748 K and 723 K for various applied shear stresses was found [33] to increase from $\sim$ 250 kJ mol^{-1} at $\tau = 3$ MPa to $\sim$ 700 kJ mol^{-1} at $\tau = 10$ MPa (Table 3). Thus, the apparent activation energy increases with increasing stress, which might seem absurd. An explanation of this striking experimental result is given later (see also ref. [33]).

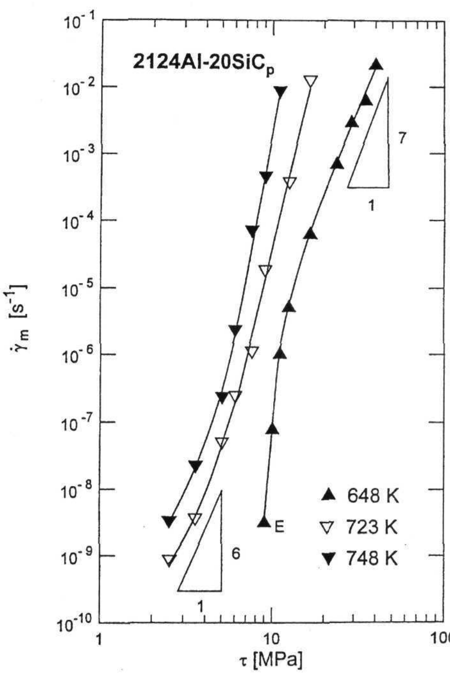

Fig. 11. 2124Al-20SiC$_p$ composite. Relations between minimum creep shear strain rate $\dot{\gamma}_m$ and applied shear stress τ for 648, 723 and 748 K.

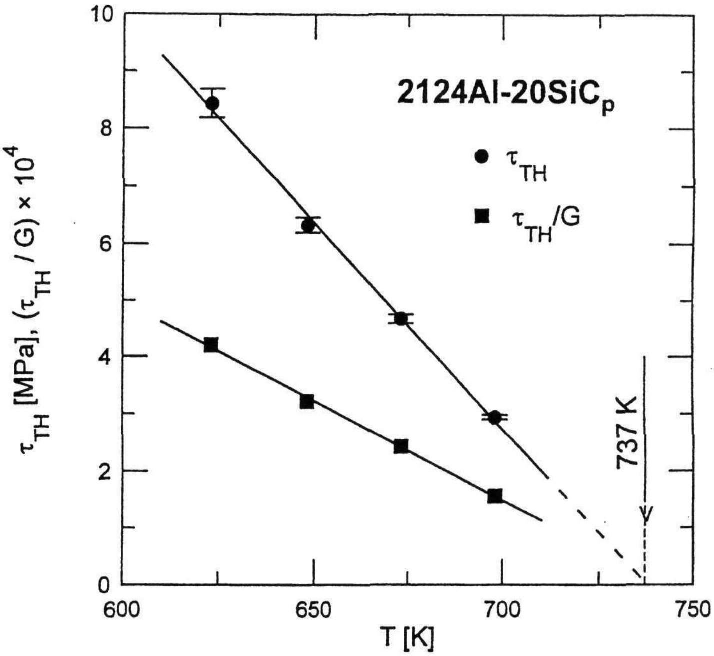

Fig. 12. 2124Al-20SiC$_p$ composite. Relations between the true threshold shear stress τ_{TH} and the τ_{TH}/G ratio and temperature. The true threshold stress is well defined up to a temperature of 698 K.

Values of the true threshold shear stress for various temperatures were determined using the linear extrapolation technique (Procedure B). At the same time, the true stress exponent n was found [7] to be close to 5. The true threshold stress values are plotted against temperature in Fig. 12. The relation between τ_{TH} and T is approximately linear; if extrapolated to temperatures above 698 K – the highest temperature at which the true threshold stress is reliably defined – it is found that this stress disappears at a temperature of $\sim$ 737 K. According to this extrapolation, the true threshold stress should be defined not up to 698 K, but up to 723 K. However, this is not the case as it follows from Fig. 13(a), in which the relation between $\dot{\gamma}_m^{1/n}$ and τ is shown for 723 K and $n = 5$. This relation clearly demonstrates that the *true* threshold stress is absent at 723 K. The curve fitted to the $(\dot{\gamma}_m^{1/n}, \tau)$ data points for this temperature apparently passes through the origin. This is not the case of the temperature 698 K as shown in Fig. 13(b). At this temperature, the true threshold stress is clearly defined.

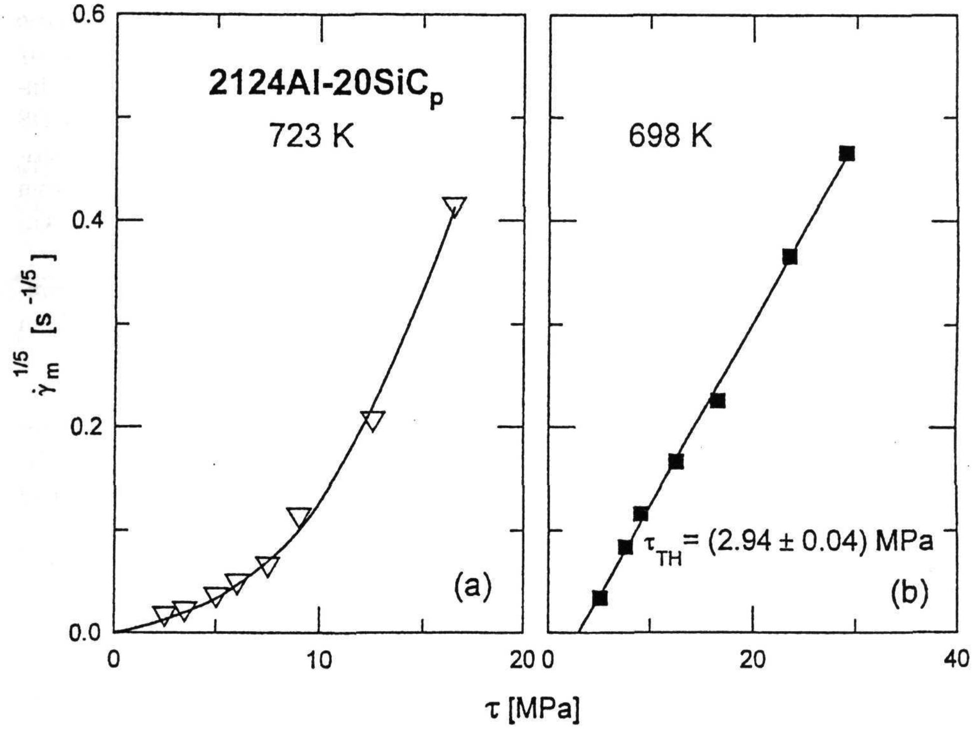

Fig. 13. 2124Al-20SiC$_p$ composite. $\dot{\gamma}_m^{1/5}$ plotted against τ for 723 K (a) and 698 K (b).

The relations between $\dot{\gamma}_m$ and τ for 723 and 748 K in double logarithmic coordinates, Fig. 11, diverge with increasing stress. This clearly indicates the apparent activation energy of creep, Q_c, increasing with the stress, which, as already pointed out, may seem absurd. However, the fact that the true threshold disappears only above $\sim$ 737 K and, thus, the athermal detachment of dislocations from fine alumina particles most probably operates in parallel with the thermally activated detachment of dislocations from these particles at 723 K, may affect the $\dot{\gamma}_m$ vs. τ relation at the latter temperature and, consequently, the apparent activation energy. Of course, to confirm this explanation of Q_c increasing with τ, and/or the anomalous $Q_c(\tau)$ at 723 and 748 K, a knowledge of the $\dot{\gamma}_m(\tau)$ creep data for at least one another testing temperature higher than 748 K, perhaps for 773 K (i.e. a temperature distinctly higher than that at which the true threshold stress apparently disappears, i.e. $\sim$ 737 K, see Fig. 12) is required.

The relations between $\dot{\gamma}_m b^2/D_L$ and τ/G for 723 K and 748 K, Fig. 14, remind, as to their shapes, those presented by Rösler and Arzt [28] for a MA 6000 nickel base superalloy strengthened in addition by fine yttrium particles. The authors interpreted the creep data for this ODS superalloy in terms of thermally activated detachment of dislocations from the interacting yttrium particles. On the other hand, the $\dot{\gamma}_m b^2/D_L$ vs. τ/G relations for PM 2124Al-20SiC$_p$ composite at temperatures ranging from 623 to 698 K are, according to their shape shown by 648 K plot in Fig. 11, typical for athermal detachment of dislocations from interacting particles [21, 22] as the creep strain rate controlling mechanism. This makes it possible to realistically assume (as it, in fact, has been already done above) that a transition from athermal to thermally activated detachment of dislocations from fine interacting alumina particles occurs at a temperature above 698 K, or, more exactly, in the temperature interval 698 K < T < 723 K. Consequently, an interpretation of the creep behaviour at the temperatures 723 and 748 K can be attempted accepting the model of creep controlled by thermally activated detachment of dislocations from fine interacting particles developed by Rösler and Arzt [28]. Such an attempt requires the knowledge of the detachment stress τ_d and the relaxation factor k_R. To estimate these quantities on the basis of the available creep data, the procedure proposed by Rösler and Arzt [28] can be applied. This procedure requires a knowledge of one structure parameter only, namely the radius r of the interacting particle. For a PM 2124Al-20SiC$_p$ composite this radius is close to 20 nm; the mean interparticle spacing $\lambda \cong 160$ nm [13].

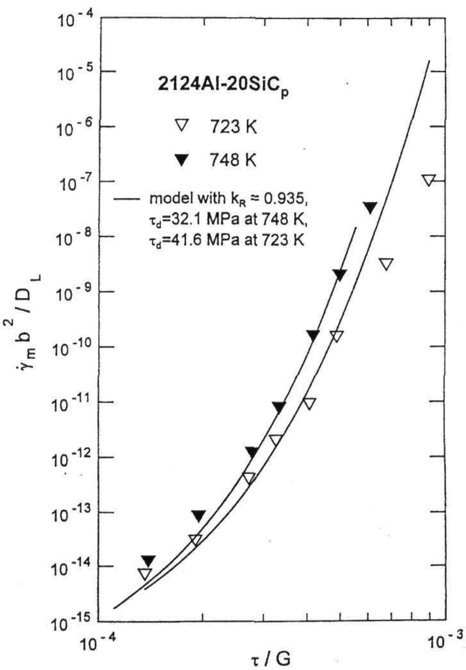

Fig. 14. 2124Al-20SiC$_p$ composite. Comparison of ($\dot{\gamma}_m b^2/D_L, \tau/G$) creep data for 723 and 748 K with the relations predicted by the model of Rösler and Arzt [28].

To apply the procedure under consideration, the τ/τ_d ratio expressed as

$$\frac{\tau}{\tau_d} = \left[\frac{3(Q_c - \Delta H_L)}{2RTm_c(1 - T/G)(dG/dT)} + 1\right]^{-1} \tag{10}$$

must be estimated first. Knowing τ/τ_d ratio, values of the relaxation factor k_R expressed as

$$k_R = 1 - \left[\frac{2kT}{3Gb^2r}\frac{m_c}{(1-\tau/\tau_d)^{\frac{1}{2}}(\tau/\tau_d)}\right]^{\frac{2}{3}}, \tag{11}$$

were calculated using $r = 20$ nm. Despite of a relatively low accuracy of Q_c estimate, quite reasonable values of k_R were obtained (Table 3), although for the applied stresses of 3 and 5 MPa they are slightly higher than the theoretical "critical" value of 0.94 [21, 22].

The model creep equation [28] expressing the normalized minimum creep strain rate as a function of temperature and applied stress can be written as

$$\frac{\dot{\gamma}_m b^2}{D_L} = C \exp\left[-\frac{E_d}{kT}\right], \tag{12}$$

where

$$C = 6\lambda\rho b \tag{13}$$

is the structure factor and

$$E_d = Gb^2r[(1-k_R)(1-\tau_d/\tau)]^{\frac{3}{2}} \tag{14}$$

is the energy of thermally activated detachment of dislocations from fine interacting particles. In Eq. (13), ρ is the density of mobile dislocations and λ is the interparticle spacing.

A comparison of $(\dot{\gamma}_m b^2/D_L, \tau/G)$ creep data for 723 and 748 K with the relations predicted by the Rösler-Arzt model [28] is shown in Fig. 14. In the comparison, the value of k_R equal to 0.935, (Table 3) was accepted [32]. The resulting fit seems quite satisfactory. However, the value of the structure factor $C = 8.22 \times 10^3$ following from this comparison for 748 K is much higher than that calculated from the structure data, Eq. (13), accepting $\rho = 1 \times 10^{13}$ m^{-2}, equal to 2.74×10^{-3}. The difference between the two values of C amounts to almost six orders of magnitude. The value of C for 723 K, 2.55×10^5, is still nearly two orders of magnitude higher than that for 748 K. Such an uncapability to predict a value of the factor C close to that following from the structure data (Eq. (13)) seems to be a shortcoming of the model under consideration. Possible reasons of this shortcoming were discussed by Orlová and Čadek [34]. Nevertheless, the analysis outlined above (see ref. [33]) supports the assumption that the transition from the athermal to the thermally activated detachment of dislocations from fine alumina particles present in a PM 2124Al-20SiC$_p$ composite matrix occurs at temperatures higher than 698 K but lower than 723 K.

6.2 DS Al-Al_4C_3-10SiC_p composite

The above idea of transition from athermal to thermally activated detachment of dislocations from fine interacting particles in the composite matrix can be, as to the present authors' belief, further supported by the results of an investigation of creep behaviour of a DS Al-Al_4C_3-10SiC_p composite denoted SiC/AlC1 in the following (again, DS means dispersion strengthening). The results for 623 and 723 K were presented by Zhu et al. [32]. These authors have shown that, at these temperatures, the creep behaviour is associated with a true threshold stress, the true stress exponent of minimum creep strain rate has been found equal to 8 rather than to 5 (see Figs. 3 and 4(a) in ref. [32]). On the other hand, at 823 K, the apparent stress exponent m_c increases with increasing applied stress from a value of $\sim$ 10 at low, to a value of $\sim$ 25 at high applied stresses, Fig. 15. Relations between $\dot{\varepsilon}_m^{1/n}$ and σ are not linear for neither value of n usually considered, i. e. 3, 5, and 8, nor even for n = 11. The $\dot{\varepsilon}_m^{1/n}$ vs. σ relations for n = 5 and n = 11 are shown in Fig. 16, from which it can be seen that the $(\dot{\varepsilon}_m^{1/n}, \sigma)$ data points can be fitted by curves approximately passing through the origin.

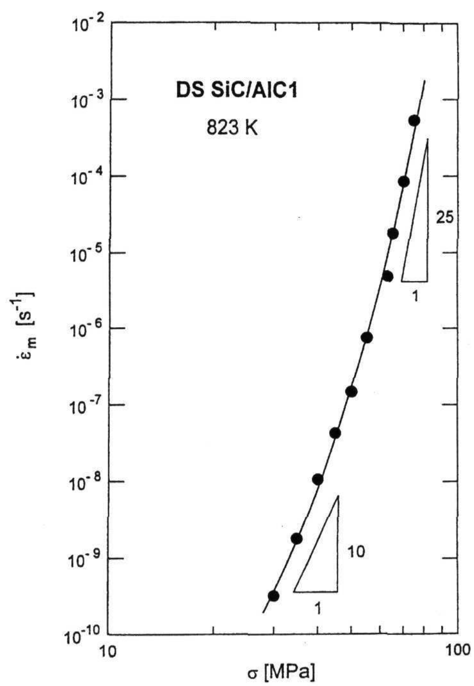

Fig. 15. DS Al-Al_4C_3-10SiC_p composite. Relations between $\dot{\varepsilon}_m$ and σ for 823 K.

The shape of the $\dot{\varepsilon}_m$ vs. σ relation (Fig. 15) again suggests the thermally activated detachment of dislocations from incoherent dispersed particles of aluminium carbide (together with a much smaller volume fraction of aluminium oxide particles). Of course, the $\dot{\varepsilon}_m(\sigma)$ creep data for a single temperature do not make it possible to analyze them similarly as those for PM 2124Al-20SiC_p composite. Nevertheless, it is possible to get an idea on the magnitude of the detachment energy E_d and the value of the apparent stress exponent m_c predicted by the Rösler-Arzt model equation [28], Eqs. (12) and (14). The value of E_d will be estimated accepting k_R = 0.94 and considering the applied stress σ = 70 MPa. For these conditions, the σ/σ_d ratio cannot be calculated by means of Eq. (10) since the apparent activation energy Q_c appearing in this equation is not known. However, the value of σ/σ_d ratio equal to

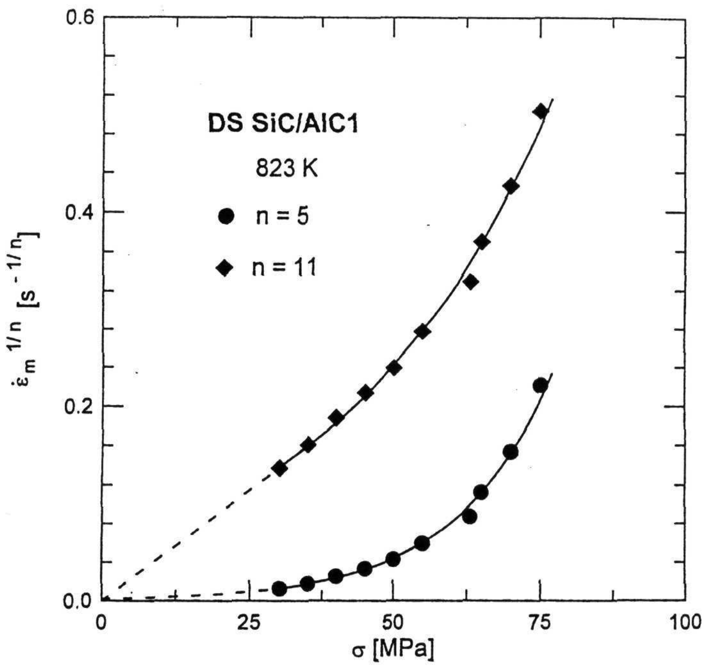

Fig. 16. DS Al-Al_4C_3-10SiC_p composite. Relations between $\dot{\varepsilon}_m^{1/n}$ and σ for $n = 5$ and $n = 11$. The $(\dot{\varepsilon}_m^{1/n}, \sigma)$ data points can be fitted by curves passing through the origin.

0.80 can be assumed as a reasonable estimate with respect to the data presented by Zhu et al. [32] in their Table 4. Accepting this value of σ/σ_d and all the above mentioned conditions, the value of E_d was estimated to 24 kJ mol^{-1}, which represents a contribution to the true activation energy (activation enthalpy) of creep, ΔH_c, for SiC_p/AlC1 composite under consideration at 823 K. The true activation energy of creep ΔH_c is thus close to 166 kJ mol^{-1}, since $\Delta H_L = 142$ kJ mol^{-1} [15]. This value of ΔH_c is relatively low, since also the $(\sigma_{TH}/G)(623\text{ K})/(\sigma_{TH}/G)(723\text{ K})$ ratio is low (Table 4 in ref. [32]), in fact, close to one. This fact was discussed in detail by Zhu et al. [32].

From the creep equation (Eqs. (12) and (14)) combined with the definition equation of the apparent stress exponent of minimum creep strain rate m_c following expression for the apparent stress exponent of the minimum creep strain rate is obtained:

$$m_c = \left(\frac{\partial \ln \dot{\varepsilon}_m}{\partial \ln \sigma}\right)_T = \frac{3Gb^2r}{2kT}(1-k_R)^{\frac{3}{2}}\left(1-\frac{\sigma}{\sigma_d}\right)^{\frac{1}{2}}\left(\frac{\sigma}{\sigma_d}\right), \tag{15}$$

where, again, r is the particle radius (in the present case the equivalent particle radius $r_c = 22$ nm was set for r in Eq. (11)). For 823 K and $\sigma = 70$ MPa, the σ/σ_d was estimated (see above) to 0.80. Using this value of σ/σ_d and $k_R = 0.94$, a value of m_c close to 21 is obtained by means of Eq. (15) for $\sigma = 70$ MPa. This value of m_c is in good agreement with experiment (see Fig. 15).

It should be emphasized once again that the creep data for a single temperature are not sufficient for a reliable enough analysis of any $\dot{\varepsilon}_m(\sigma)$ creep data. Nevertheless, the above rough estimates suggest that at the temperature of 823 K the minimum creep strain rate in SiC/AlC1 composite is controlled by the thermally activated detachment of dislocations from fine aluminium carbide (and also aluminium oxide) particles, which is certainly not the case at the temperature of 723 K [32]. Of course, it would be most interesting to estimate the transition temperature of the athermal/thermally activated detachment by further creep experiments.

7. Discussion and concluding remarks

In the present paper, no attempt was made to review all the recent results presented in the literature. Such a review can be found elsewhere [31]. Here, some contemporary topical problems are accentuated, such as the origin and temperature dependence of the true threshold stress in discontinuous PM aluminium and aluminium alloy matrix composites, disappearance of the true threshold stress and/or the transition from athermal to thermally activated detachment of dislocations from fine interacting particles in the composite matrix at high testing temperatures. Also accentuated are some problems of analysis of experimental $\dot{\gamma}_m(\tau, T)$ or $\dot{\varepsilon}_m(\sigma, T)$ creep data concerning, in particular, the true stress exponent of the minimum creep strain rate and load transfer effect.

7.1 The true stress exponent and the true threshold stress

There is an interrelation between the true stress exponent n and the true threshold stress σ_{TH}. Such an interrelation is illustrated in Fig. 2. To estimate the true threshold stress, the exponent n must be properly chosen or determined by means of a proper optimization procedure (e. g. Procedure C, Table 1). Usually, three values of n, namely 3, 5 and 8, are considered, since these values are predicted by three different models of creep widely accepted at the present time. The exponent $n = 3$ is characteristic for Alloy Class creep behaviour which is due to dislocation glide controlled by drag of atmospheres of solute atoms, e.g. Mg atoms, in aluminium. The exponent $n = 5$ is characteristic for Metal Class creep behaviour; the creep strain rate is controlled by dislocation climb and annihilation. Such a behaviour is typical for pure metals as well as for some solid solution alloys, under certain external conditions at least. The stress exponent $n = 8$ is predicted by the substructure invariant model of creep developed by Sherby et al. [30]. According to this model, the creep strain rate is controlled by recovery, occurring by dislocation climb and annihilation, dependent on diffusion, but the subgrain structure is stress invariant.

When the measured minimum creep strain rates cover at least five orders of magnitude, it is usually possible to decide reliably between the above mentioned values of the exponent n, provided of course, the scatter of data is not too large. However, rather frequently the measured creep strain rates cover less than four or even three orders of magnitude and show considerable scatter (e. g. refs. [35, 36]). Then, the guess of the value of n may be rather arbitrary. The choice of n affects the value of the threshold stress as estimated using the linear extrapolation technique (Procedure B). This is illustrated in Fig. 17, in which values of σ_{TH} estimated assuming $n = 5$ and $n = 8$ are plotted against temperature for an Al-5Mg-26$(Al_2O_3)_f$ composite processed by ingot metallurgy technique, originally investigated by Dragone and Nix [37]. For $n = 8$ the threshold stress values are much lower and the temperature dependence of τ_{TH} is much weaker. An acceptance of the value of 8 for n may eventually result in disappearance of temperature

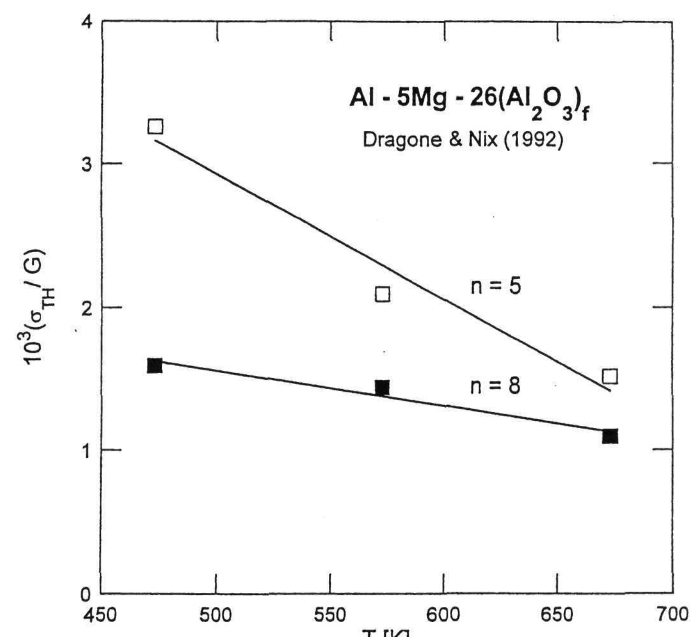

Fig. 17. Al-5Mg-26$(Al_2O_3)_f$ composite. Values of σ_{TH} estimated assuming $n = 5$ and $n = 8$, plotted against temperature ($\dot{\varepsilon}_m(\sigma, T)$ creep data of Dragone and Nix [37]).

dependence of the threshold stress. This is the case of PM Al-XSiC_p composites (X = 10, 20 and 30 vol.%) investigated by Pandey et al. [35] (see also ref. [6]) and of a PM Al-4Mg-10SiC_p composite investigated quite recently by the same authors [36]. A temperature independent threshold stress σ_{TH} (or, more exactly normalized stress σ_{TH}/G) may be in conflict with high values of m_c and Q_c observed experimentally. However, just such a conflict may serve as a qualitative check of the adequacy of the assumed value of the true stress exponent n.

It should be pointed out that, most frequently, the value of n close to or slightly lower than 5 is reported in the literature (e.g. refs. [2, 6, 7]). However, to the above considerations, two notes should be added.

(i) In some discontinuous composites, the minimum creep strain rate can be really characterized by the true stress exponent $n = 8$. At least two examples of such a creep behaviour can be found in the recent papers by Zhu et al. [32] and Čadek and Kuchařová [31]. The value of 8 should be preferred to 5 for the true stress exponent n in creep of DS Al-Al_4C_3-10SiC_p composite at 623 and 723 K [32]. Surprisingly enough, $n = 8$ should be preferred to $n = 5$ in creep of the Al--5Mg-26$(Al_2O_3)_f$ composite [31] even at external conditions, at which the Al-5Mg solid solution alloy exhibits Alloy Class creep behaviour [38, 39]. However, the true stress exponent close to 8 in itself cannot be accepted as a sufficient proof of validity of the substructure invariant model of creep of the composites mentioned as discussed in detail elsewhere [31, 32].

(ii) As already mentioned in Section 6, Gonzáles-Doncel and Sherby [29] found the temperature, at which the true threshold stress attains zero value, to be approximately the same for all the composites for which the creep data were available till 1992. To evaluate the creep data, the authors accepted the substructure invariant model of creep, predicting the value of the true stress exponent equal to 8. However, this result may be questioned because for $n = 8$ the true threshold stress was found to be temperature independent for some composites (see the results by Pandey et al. [35, 36], mentioned above), which, of course, necessarily affected the result of the analysis by Gonzáles-Doncel and Sherby.

The disappearance of true threshold creep behaviour can be quite satisfactorily interpreted in terms of the athermal to thermally activated detachment of dislocations from fine interacting, mostly alumina particles in the composite matrix. Accepting this interpretation, there does not seem any reason to expect that the true threshold stress disappears at a single, (though eventually not well-defined) temperature for any discontinuous aluminium and aluminium alloy matrix composite. Such a temperature should be expected to be affected by the creep strength of the composite in question, since this strength is determined primarily by the threshold stress level which can be dramatically enhanced by oxide dispersion or dispersion strengthening of any discontinuous composite matrix (see e.g. Fig. 3). For $n = 5$, the temperature dependence of the threshold stress not too different for different composites was reported. In fact, the temperature dependence of σ_{TH} is frequently expressed as

$$\frac{\sigma_{TH}}{G} = B\exp\left[\frac{Q_o}{RT}\right], \tag{16}$$

where B is a dimensionless constant and Q_o/RT is an energy term. Reported values of Q_c range from 19 to 25 kJ mol^{-1} [13]. For ODS Al-5Mg-30SiC$_p$ composite the σ_{TH}/G vs. $1/T$ relation in semilogarithmic co-ordinates is illustrated in Fig. 18; the energy $Q_o \cong 21.1$ kJ mol^{-1}.

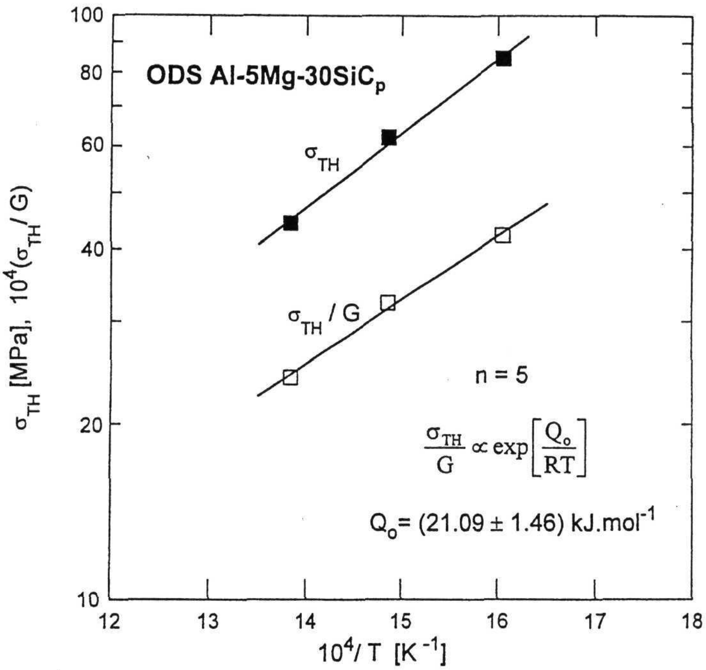

Fig. 18. ODS Al-5Mg-30SiC$_p$ composite. Threshold stress σ_{TH} and normalized threshold stress σ_{TH}/G plotted against reciprocal temperature $1/T$ in semilogarithmic co-ordinates. Same (σ_{TH}, T) data as in Fig. 9. The energy $Q_o \cong 21.1$ kJ mol^{-1} (see Eq. (16)).

7.2 The load transfer effect and the threshold effect

To take into account both the threshold effect and the load transfer effect in the description of applied stress and temperature dependence of the minimum creep strain rate, Park and Mohamed [40] suggested to introduce the load transfer coefficient β into the phenomenological creep equation (1), setting $(1-\beta)\sigma - \sigma_{TH}$ or $(1-\beta)(\sigma - \sigma_{TH})$ instead of $\sigma - \sigma_{TH}$ for the effective stress. The expression $(1-\beta)\sigma - \sigma_{TH}$ for the effective stress σ_e implies that the load transfer is developed *before* any creep deformation in the composite matrix takes place. Such a situation is illustrated in Fig. 19. On the other hand, the expression $\sigma_e = (1-\beta)(\sigma - \sigma_{TH})$ implies that creep deformation in the matrix starts when σ exceeds σ_{TH} and only then the reinforcement occurs as a result of the interaction between the creeping matrix and the rigid reinforcement.

From Fig. 19 it can be seen that the load transfer effect dominates at high applied stresses, while the threshold effect at the low ones and/or minimum creep strain rates, which may be important from the point of view of some high temperature applications of the composites under consideration. In this connection, it must be emphasized once again that the main role of discontinuous reinforcement of aluminium and aluminium alloys consists in increasing the elasticity modulus.

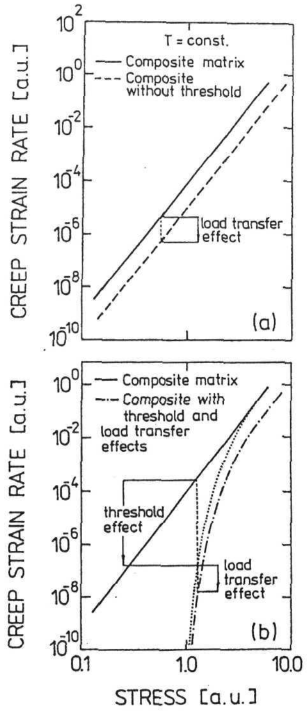

Fig. 19. Schematic representation of the load transfer effect in a composite without threshold (a) and of the load transfer effect and threshold effect in a composite with a threshold and load transfer (b). The load transfer is assumed to be developed before any creep deformation in the composite matrix takes place (after Park and Mohamed [40]). The abbreviation a.u. means arbitrary units.

Park and Mohamed [40] decided that the expression $(1-\beta)(\sigma - \sigma_{TH})$ was more appropriate primarily because they found similar values for the threshold stress in the PM 6061Al-30SiC$_p$ composite and the PM 6061Al matrix alloy. On the other hand, Li and Langdon [41] believe that it may be more appropriate to accept the expression $(1-\beta)\sigma - \sigma_{TH}$ for the effective stress. However, accepting this expression, a new quantity has to be introduced, namely the "apparent" threshold stress σ_{TH}^{app}, which is related to the true threshold stress σ_{TH} as $\sigma_{TH}^{app} = \sigma_{TH}/(1-\beta)$ [41]. The present authors prefer to use the expression $\sigma_e = (1-\beta)(\sigma - \sigma_{TH})$.

From Eq. (1), in which $(\sigma - \sigma_{TH})$ is replaced by $(1-\beta)(\sigma - \sigma_{TH})$ the following expression for Q_c is obtained:

$$Q_c = \left[\frac{\partial \ln \dot{\varepsilon}_m}{\partial(-1/RT)}\right]_\sigma = \Delta H_L - \frac{nRT^2}{G}\left(\frac{G}{\sigma - \sigma_{TH}}\frac{d\sigma_{TH}}{dT} + \frac{n-1}{n}\frac{dG}{dT}\right) - \frac{nRT^2}{1-\beta}\frac{d\beta}{dT}. \quad (17)$$

According to the theoretical models [42, 43] developed for short fibre composites, the load transfer coefficient β should be stress independent.

Park and Mohamed [40] proposed a procedure making it possible to estimate the load transfer coefficient from experimental $\dot{\varepsilon}_m(\sigma, T)$ creep data similarly as the true threshold stress σ_{TH} and the true stress exponent n. The procedure, which will not be outlined here, is based on two assumptions, namely (i) the microstructure of the reinforced aluminium or aluminium alloy is the same as that of non-reinforced aluminium or aluminium alloy and (ii) the creep behaviour of the composite is dominated by the behaviour of the matrix. At least the first of these assumptions can be a subject of discussion (see ref. [31]). Besides, the procedure to estimate β can be criticized on account of a mere application of the experimental $\dot{\varepsilon}_m(\sigma, T)$ creep data. Of course, it would be very useful to have a theoretical model for spherical particulates to estimate the load transfer effect in an independent way. Such a model could be perhaps developed for "pure" Al-SiC_p composite, i.e. for spherical silicon carbide particulates embedded in pure aluminium. It can be hardly expected that in its first version, such a model will adequately describe the situation in the matrix containing relatively large volume fraction of fine aluminium particles. However, the application of the Park-Mohamed procedure to estimate β is certainly more reliable than the application of the available models of Kelly and Street [42] and/or Wu and Lavernia [43].

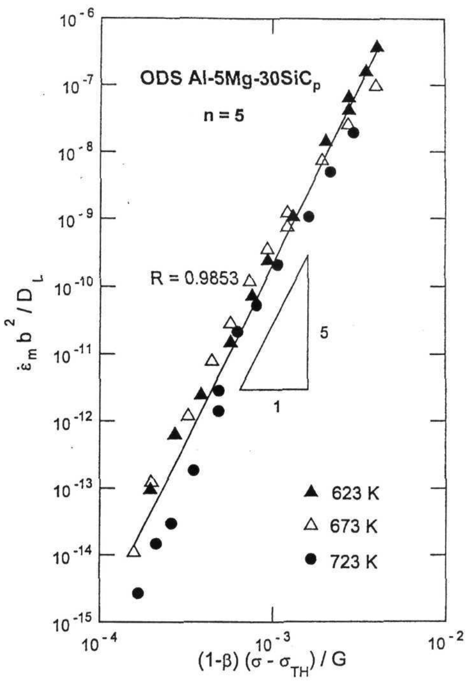

Fig. 20. ODS Al-5Mg-30SiC_p composite. Normalized creep strain rates $\dot{\varepsilon}_m b^2/D_L$ plotted against normalized effective stress compensated for the load transfer coefficient β, i.e. $(1-\beta)(\sigma - \sigma_{TH})$. For the correlation coefficient c.f. Fig. 10.

For an ODS Al-5Mg-30SiC_p composite, the load transfer coefficient was estimated to 0.26, 0.22 and 0.17 for temperatures 623, 673 and 723 K, respectively, using the Park-Mohamed procedure and recently published $\dot{\varepsilon}_m(\sigma, T)$ creep data for ODS Al [12]. Thus, for 673 K the contribution $[nRT^2/(1-\beta)]\mathrm{d}\beta/\mathrm{d}T$ amounting to 21.7 kJ mol^{-1} was obtained. This contribution is practically negligible in comparison with the value of the contribution following from the temperature dependence of threshold stress, $[nRT^2/(\sigma - \sigma_{TH})]\mathrm{d}\sigma_{TH}/\mathrm{d}T$, especially at low applied stresses (as already mentioned, β does not depend on applied stress). For instance, at $\sigma = 80$ MPa and 673 K this contribution amounts to ~ 425 kJ mol^{-1}. Thus the load transfer effect does not significantly affect the apparent activation energy of creep, Q_c, and/or the temperature dependence of the minimum creep strain rate.

In general, the load transfer effect should be taken into account, even when the matrix of a discontinuous composite is strengthened with fine interacting (e.g. alumina or aluminium carbide) particles. This is illustrated in Fig. 20, in which values of $\dot{\varepsilon}_m b^2/D_L$ are plotted against $(1-\beta)(\sigma - \sigma_{TH})/G$ for the ODS Al-5Mg-30SiC$_p$ composite in double logarithmic co-ordinates. The correlation coefficient is higher than that for $\dot{\varepsilon}_m b^2/D_L$ vs. $(\sigma - \sigma_{TH})/G$ relation, Fig. 10.

7.3 Temperature dependence of the relaxation factor k_R

From the short review of the mechanisms by which dislocations can bypass or surmount fine interacting particles, i.e. mechanisms, associated with a true threshold stress (Section 4) on one side and the experimental fact that the true threshold stress normalized to the shear modulus, τ_{TH}/G, decreases rather strongly with increasing temperature on the other one, it has been concluded that the only potential one of these mechanisms is the athermal detachment of dislocations from particles. In fact, in principle at least, the relaxation factor k_R in Eq. (7) may depend on temperature and the temperature dependence of k_R can account for the temperature dependence of τ_{TH}/G. However, the relaxation factor cannot be calculated from the first principles at the present time.

The value of the factor k_R should not depend on whether the detachment of a dislocation from an interacting particle is athermal or is aided by thermal activation. For creep controlled by the thermally activated detachment, Rösler and Arzt [28] developed the procedure of k_R estimation. This procedure follows from the model creep equation presented by these authors. The expression for k_R following from this equation is given by Eq. (11). A model simulation [11] led to the conclusion that k_R may really reasonably decrease with decreasing temperature. This supports the "detachment" concept.

A further support for the detachment concept provides the disappearance of the true threshold stress at high testing temperatures. In fact, it seems quite natural to expect that at high enough temperatures, the detachment process will be aided significantly by thermal activation in contrast with relatively low testing temperatures. However, an extrapolation of k_R obtained for temperatures, at which the thermal activation dominates, to temperatures, at which it is negligible, and thus, at which the true threshold stress is defined, is not possible at the present time, because of lack of creep data for temperatures at which the thermally activated detachment of dislocations from interacting particles is dominant.

8. Summary

In the present paper, some experimental results of recent investigations of creep behaviour of discontinuous PM aluminium and aluminium alloy matrix composites are overviewed with special reference to the contemporary, not yet well understood features of this behaviour. Among such features in the first place is an existence of true threshold stress and its temperature dependence is significantly stronger than that of the shear modulus of the matrix metal. Just owing the strong temperature dependence of the true threshold stress to the shear modulus ratio, the apparent activation energy of creep is much higher than the activation enthalpy of the matrix lattice diffusion in aluminium and the apparent stress exponent of minimum creep strain rate much higher than that for aluminium. Nevertheless, it is shown once again that the creep strain rate in the composites under consideration is matrix-lattice-diffusion controlled. The load transfer effect associated with discontinuous reinforcement is temperature dependent but it does not contribute to the apparent activation energy significantly.

A detachment concept, according to which the true threshold stress is associated with the athermal detachment of dislocations from fine interacting alumina particles in the composite matrix is accepted. The possibility for the true threshold stress to decrease with temperature more strongly than the shear modulus is accounted for assuming a proper temperature dependence of the relaxation factor characterizing the strength of the attractive dislocation/particle interaction.

Disappearance of the true threshold stress at high testing temperatures is interpreted in terms of the transition from the athermal to the thermally activated detachment of dislocations from "interacting" particles.

Also some other, not yet fully clear topics are discussed, namely the relation of the true stress exponent and the true threshold stress and the relation between load transfer effect and the threshold effect.

Acknowledgements

This work was financed by the Institute of Physics of Materials, Academy of Sciences of the Czech Republic (Project No. 7/96 K). The authors thank to Mrs. J. Duchoňová and Miss E. Najvarová for assistance in manuscript preparation.

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Received: 24.3.1999