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Threshold stress and load partitioning during creep of MMCs

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Abstract

The threshold stress, σ_o , included in the creep equation to explain the high temperature behavior of discontinuously reinforced metal matrix composites (MMCs) is criticized on the basis of microstructural considerations and a new creep data analysis. An alternative interpretation, based on changes in the composite matrix microstructure and, in particular, a load transfer mechanism, is proposed. The resulting creep equation is similar to that in which σ_o is used: in essence, σ_o is simply replaced by the stress carried by the reinforcement (referred to as σ_T). New creep data on 6061Al-15vol%SiC_w composite and the corresponding un-reinforced alloy, allowing direct experimental assessment of composite creep strengthening, $\Delta\sigma$, are analyzed. The linear dependence found of $\Delta\sigma$ with the applied stress, σ , $\Delta\sigma(\sigma)$, correlates reasonably well with Shear-Lag and Eshelby model predictions of σ_T , transferred during composite creep deformation. The possible occurrence of damage mechanisms and the complexity of modeling these mechanisms to predict the overall composite creep behavior are also discussed.

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Introduction

It is agreed that the creep of discontinuously reinforced metal matrix composites, MMCs, such as aluminum alloys with ceramic reinforcement (particles, whiskers, platelets), must be understood in terms of the creep behavior of the metallic matrix alloy since only this phase undergoes plastic deformation (due to the stiffer and stronger nature of the ceramic reinforcement). To achieve this understanding, a threshold stress term, σ_o , was incorporated in the power law creep equation,

$$\dot{\epsilon} = k' \sigma^n \quad (1)$$

(where $\dot{\epsilon}$ is the strain rate, k' a temperature and material's dependent constant, σ the applied stress, and n the stress exponent) such that the net stress ($\sigma - \sigma_o$), instead of σ , is the one included in equation (1). In this way, a rationalization of the composites creep data to that of metallic alloys results: *i.e.*, typically high values of n , and the activation energy for creep, Q_c , decrease down to values commonly obtained for the alloys [1-8]. While the above approach has been assumed by many researchers, a satisfactory interpretation of σ_o in terms of micro-mechanisms and microstructural parameters, however, has not yet been found. In fact, it is relevant that very recent publications on the creep of MMCs do not discuss the possible origins of this stress [8-12].

In many of these studies the creep of MMCs are investigated without comparing their behavior to that of the corresponding un-reinforced matrix alloys [5]. When this comparison is made, it has been shown that besides σ_o , a load transfer coefficient was needed to correlate the creep data of composites and un-reinforced alloys into a common behavior [5,13]. While the load transfer from matrix to reinforcement is a well known mechanism, the meaning of σ_o even in this new context still remains obscure [13].

In recent work [14], a different analysis of the creep of composites has been made. Ryu *et al.* [14] have interpreted the creep of MMCs considering *only* a load transfer mechanism and hence, on the basis of the effective stress that the matrix should undergo during loading. By using a generalized Shear-Lag model [15], they were able to rationalize the creep data of whisker and particle reinforced composites extruded under different conditions (*i.e.*, with different degree of whisker alignment) to that of the un-reinforced alloy. The possible role of σ_o was not considered.

So, it can be seen that the inherent mechanisms that govern the improved creep behavior of discontinuously reinforced MMCs are not yet well understood. Given the present weak predictive capacity, it is essential to analyze the creep behavior of MMCs on the basis of that of the corresponding un-reinforced alloys. This is important because sometimes, the addition of the reinforcement is detrimental, rather than beneficial, for this behavior [16,17]. If such a comparison

is not made, one could be analyzing reinforcing mechanisms when other processes and/or severe damage are occurring. As we will be able to show, such a comparison not only makes it possible for us to deepen our understanding of the strengthening mechanisms in these composites, but also to envisage the possible contribution of damage processes to their overall creep behavior.

In this work, composite creep strengthening of 6061Al-15vol%SiC_w prepared by powder metallurgy, PM, was studied in terms of the creep behavior of the corresponding PM 6061Al alloy. Shear-Lag and Eshelby models of the stress transferred have been used to interpret the strength differences found. Furthermore, a thorough analysis of the creep data of aluminum and aluminum alloy MMCs (whenever data of corresponding un-reinforced alloy is reported) of published investigations [2,8,11-14,18-39] has been conducted.

Materials and experimental procedure

Powder of 6061Al alloy, of average particle size less than 45 μm [40], was mixed in a ball-mill with SiC whiskers 20-40 μm in length and average diameter of 0.4 μm [41]. The specific details of the PM process are found elsewhere [41,42]. The 6061Al bulk alloy was also prepared by the same PM route. Table I summarizes the main processing parameters employed. Conventional metallographic techniques, including optical, OM, and scanning electron microscopy, SEM, were used to study materials microstructure. Longitudinal metallographical sections were especially useful to determine the average whisker aspect ratio, S .

Texture measurements were carried out by X-Ray diffraction in a Siemens Kristalloflex diffractometer equipped with a D5000 goniometer and a closed Eulerian cradle using the Schulz reflection method. Pole figures of the (111), (200), (220) and (311) reflections were determined from which the orientation distribution function, ODF, of the Euler angles ϕ_1 , Φ and ϕ_2 was calculated. Details of ODF determination and the volume fraction of crystallites associated with specific texture components are given elsewhere [42,43]. Texture measurements were also employed to study the alignment of SiC_w with the extrusion direction. This was possible because the whiskers were single crystals in which the long axis coincided with $\langle 111 \rangle$ [14,42].

The dislocation density was estimated on the basis of the accelerated ageing phenomenon in accordance with previous studies by hardness and calorimetric measurements [41,44,45].

Cylindrical tensile creep samples, 10 mm in gauge length, and 3 mm in diameter, machined parallel to the extrusion axis and with threaded heads were used. Tests were carried out at temperatures in the range of 573-723 K (300-450°C) under constant stress (4-50 MPa) provided by an Andrade's cam. A load cell, inserted in the loading system, allowed monitoring of the applied

load. The sample elongation was measured through LVDTs suitably attached to the sample. Test data was stored into a computer through appropriate data acquisition boards.

Results

The microstructure and the texture of similar composites and un-reinforced 6061Al alloys prepared by the PM procedure have been described in [42,46-48]. In summary, the materials microstructure is shown in the SEM micrographs of Figure 1. A tendency of the grains to elongate with the extrusion axis direction (horizontal) is evident. A trend of the SiC reinforcement to align with the extrusion axis direction can also be seen, Figure 1b). The grain size is about 3.5 μm and 1.5 μm in the alloy and the composite, respectively, Table I.

The increased dislocation density of the composite was studied previously [41,44]. It was demonstrated from the accelerated ageing phenomenon [41] and detailed calorimetric studies [44] that the dislocation density of the composite is about double that of the alloy, Table I.

The extrusion of materials into cylindrical bars leads to microstructures with axial symmetry, as confirmed by the texture measurements. Both, the un-reinforced alloy and the composite matrix develop a typical $\langle 111 \rangle + \langle 100 \rangle$ fiber texture (with the fiber axis parallel to the extrusion axis direction) of extruded aluminum [42]. The texture in the alloy is, however, more intense than in the composite due to the more severe process of particle stimulated nucleation, PSN, for recrystallization during extrusion [42]. Table I summarizes the volume fraction of crystallites associated with each of the texture components of both materials.

After extrusion, part of the SiC remains at random, but a fraction, f_a , becomes aligned with the extrusion axis. The orientation distribution of this fraction of whiskers is also axially symmetric. The plot of Figure 2 is a one dimensional (111) pole figure showing this alignment. In this figure, the diffracted intensity, I , of the Al-111 reflection (averaged over the azimuth angle, from 0° to 360°) is plotted as a function of the misorientation angle, θ (with $0^\circ \leq \theta \leq 90^\circ$), $I=I(\theta)$. As in [42], two contributions corresponding to the random and oriented whiskers are obtained, $I(\theta)=I_{rd}+I_a(\theta)$, where the oriented population, $I_a(\theta)$, obeys a Gaussian distribution, $I_a(\theta)=a \exp(-b\theta^2)$. Table I summarizes I_{rd} , a , and b values. Following [42], f_a is obtained from these values, Table I.

The results of the creep tests are shown in Figs. 3 and 4. Figure 3 shows typical true strain rate vs. strain curves showing a short primary creep and a minimum creep rate, rather than a steady creep rate in both the un-reinforced alloy and the composite under any of the conditions investigated. It can also be seen that similar creep curves, at equivalent applied stress, are obtained for the composite and the alloy. [This suggests that potential damage mechanisms associated to the metal-ceramic interface in this composite are not relevant for the creep behavior.](#) Finally, it can also be

appreciated that the minimum creep rate is reached at lower strain in the composite than in the alloy. In Figure 4, the minimum creep rate, $\dot{\epsilon}$ is represented as a function of σ , in a double logarithmic scale, for both materials at the different testing temperatures. The better creep resistance of the composite can be seen. In both materials and at any temperature, the power law dependence, equation (1), is obeyed. Except for the highest temperature, similar high n values are found for the alloy and the composite. This similarity is suspicious, and is a hint to argue against the need to include σ_o term in the creep equation to understand the high temperature behavior of MMCs.

The threshold stress concept: a criticism of its meaning in MMCs

The term σ_o was proposed originally to understand the creep of oxide dispersion strengthened, ODS, alloys [49]. These alloys show remarkably high values of the apparent n and Q_c . Specific mechanisms of dislocation-particle interaction have been considered as responsible for σ_o [2,50]. In these alloys, the size of the particles is of the order of the dislocation core. Hence, the applicability of models describing the interaction of individual dislocations with particles is feasible. These “micro-mechanisms” have been also applied to explain σ_o in MMCs. For these materials, however, the scale of the reinforcing particles (particle size and inter-particle distance) is much larger (at least three orders of magnitude) than in ODS alloys. As a consequence, models accounting for the interaction of individual dislocations with the large reinforcing particles of MMCs may be called into question [35]. This is, probably, the reason why the above mechanisms have failed in the attempt to explain the meaning of σ_o in MMCs: in particular its dependence on temperature $\sigma_o(T)$ [3,51].

Furthermore, the reliability of the σ_o values obtained from the extrapolation method is, in some cases, doubtful. This method is well described elsewhere [1,3-5,52]. It consists of “imposing” an n value to the creep equation so that, at a given temperature, a straight line should be obtained when the $1/n$ power of $\dot{\epsilon}$ is plotted as a function of σ/E . The extrapolated value of σ at $\dot{\epsilon}=0$ is taken as σ_o . Typically, the fits to obtain σ_o are made using n values of $n=3$, $n=5$, and $n=8$, which are identified with specific deformation mechanisms [53]. As done in [52], the n value which best fits a straight line and the corresponding σ_o are the data included in the creep equation. However, the reliability of these σ_o values, in particular for composites in which the matrix alloy is age hardenable [2,8,13,14,18,19,21,24,25,28,31], should be carefully considered. As has been suggested [8,17,36,40,54], n in these alloys may be affected by the simultaneous precipitation process which occurs during creep. Therefore, the significance of the σ_o values obtained by this method is, at least, questionable.

Importance of load partitioning during high temperature deformation of MMCs

Whereas the importance of load partitioning to explain the high stiffness of MMCs at room temperature is well established, its relevance in the high temperature plastic regime has not yet been stated so clearly. It explains the increased elastic properties (Young's modulus) of MMCs in agreement with the theoretical predictions such as those provided by Eshelby [55], Shear-Lag [15,56] and Variational methods [57]. In fact, Young's modulus of metallic alloys is highly insensitive to microstructural changes [58]. Understanding the improved yield strength (plastic properties), also requires a load transfer mechanism together, now, with other microstructural factors (*e.g.*, the increased dislocation density [41,59]). Going further, when high temperature properties (creep) are investigated, the role of load transfer is rather a subject of serious controversy. As mentioned above, this mechanism has also been considered in some investigations as a way to explain the creep of MMCs, together with σ_o [5,8,12,13,35,60,] or alone [14,18].

The relevance of this mechanism at high temperature can be well appreciated when thermal cycling and isothermal creep data of whisker reinforced MMCs, with well aligned whiskers, tested along different directions, are compared. It has been shown [61-63] that: a) the n value differs dramatically ($n=1$ in the thermal cycling and $n=8$ isothermal tests); b) and most important, the isothermal creep behavior is anisotropic. The longitudinal samples (whiskers parallel to the loading direction) are stronger than the transverse ones. On the contrary, the thermal cycling creep strength is irrelevant to the testing direction: the behavior is isotropic. These differences have been attributed to different mechanisms, in particular to the dramatically different role played by the Al-SiC interface [64]. In the isothermal tests, the anisotropic behavior is easily understood if a load transfer mechanism, which is more effective in the longitudinal samples, is taken into account. In the thermal cycling tests, however, load transfer is dominated by other interfacial processes. The thermal cycling causes the generation of large internal stresses as a consequence of the difference in the thermal expansion coefficient of the aluminum and the SiC. Limited stress relaxation by relieving dislocations occurs. A σ driven motion of this large fraction of dislocations "pumped" at the Al-SiC interface in every thermal cycle dictates the isotropic creep behavior under these conditions [64].

Further evidence is given by the difference in creep strength of whisker *vs.* particle reinforced composites. As reported in [4,18,65], the whisker reinforced composites have better creep resistance than the corresponding particle reinforced composite tested under similar conditions. Finally, the load transfer during creep of MMC has been also studied by neutron diffraction [66], a phase sensitive technique that enables the stress state of each phase to be measured separately. The experimental data obtained have been assessed by Eshelby model predictions.

Due to the “macroscopic” scale of the reinforcing phase, specific physical properties should be considered in models accounting for the creep of these composite. In particular, the reinforcing particles behave like perfectly elastic solid bodies. Similarly, the metallic-ceramic interface should also play a crucial role in such a creep model, consistent with a load partitioning phenomenon.

Analysis of creep results

In the light of the failed efforts to find a rational meaning of σ_o and the relevance of load transfer, the creep behavior of MMCs will be explained considering this mechanism. For this purpose, it is important first, to include the stress partition process into the creep equation. This is readily done, if no other interfacial process takes place, by means of the rule of mixtures,

$$\bar{\sigma}_M (1 - f) + \bar{\sigma}_R f = \sigma \quad (2)$$

where $\bar{\sigma}_M$ is the average stress on the matrix, $\bar{\sigma}_R$ is the average stress on the reinforcing whisker, and f the volume fraction of reinforcement ($f = 0.15$). Then, $\bar{\sigma}_M$ is,

$$\bar{\sigma}_M = \sigma - (\bar{\sigma}_R - \bar{\sigma}_M) f = \sigma - \sigma_T \quad (3)$$

where $(\bar{\sigma}_R - \bar{\sigma}_M) f$ is σ_T , *i.e.*, the average stress on the matrix is the difference between σ and the extra stress borne by the reinforcement [56]. The creep of MMCs should then be understood on the basis of the following assumptions:

- 1.- Only the matrix deforms under the power law creep equation, $\dot{\epsilon} = k' \bar{\sigma}_M^n$, equation (1).
- 2.- In a first approximation, no damage mechanisms or other interfacial processes occur.

The steady state creep rate behavior of the composite should then be determined by the equation,

$$\dot{\epsilon}_{ss} = A' \left(\frac{\sigma - \sigma_T}{E} \right)^n \exp(-Q_c/RT) \quad (4)$$

Where A' is a material's constant, E the Young's modulus, R the universal gas constant ($R = 8.314$ kJ/mol K), and T the absolute temperature. An equation similar to the common creep equation (with the term σ_o) is obtained. In the present view, however, any matrix-dislocation-based plasticity model used to account for the creep of MMCs should consider that the actual stress which the dislocations bear is $(\sigma - \sigma_T)$. This is contrary to the former view of equation (4) with σ_o instead of σ_T where the stress σ is acting on the dislocations [2].

It is now essential to determine the precise stress partition between matrix and reinforcement. This depends on the specific microstructural parameters of the reinforcement. Stress partitioning can be estimated experimentally (from the comparison of composite and un-reinforced alloy creep data) and theoretically (from models such as the Shear-Lag or the Eshelby [15,55,56,67]).

a) Strengthening due to load partitioning

The experimental creep data at 673K of both materials are shown in the log-log plot of Figure 5a). The creep strength of the composite is, as expected, higher than that of the alloy. Also, similar stress exponents, n_{220} and n_{219} , are obtained, revealing a “similar” deformation mechanism in both materials. The load transferred at the different strain rates, σ_T , can at a first approach be related to the difference between the flow stress of the composite and the un-reinforced alloy, $\Delta'\sigma = \sigma_{comp} - \sigma_{alloy}$, as done in [40]. The variation of $\Delta'\sigma$ with σ , shown in the plot of figure 5b), reveals that the higher the σ value, the higher the extra stress needed for the composite to deform at a given $\dot{\epsilon}$. The dependence is approximately linear. This dependence correlates well with the idea that load partitioning is a relevant mechanism for the creep strengthening of MMCs and therefore, that $\Delta'\sigma$ is related to σ_T , the stress carried by the reinforcement.

This linear dependence has prompted us to analyze data from other authors on the creep of discontinuously reinforced MMCs, Table II [2,8,11-14,18-39] (creep data of the corresponding un-reinforced alloy of the composite studied in [37] are reported in [39]). In the research papers of Table II data from the corresponding un-reinforced Al alloys are also reported. The resulting analysis is summarized in the plots of Figure 6, in which the stress increment, $\Delta'\sigma$, is, again, plotted as a function of σ for composites prepared by PM, Figure 6a), and, IM Figure 6b).

From these figures, several important features can be observed. It can be seen, first, that a linear tendency of $\Delta'\sigma$ to increase with σ , in particular for the PM composites, very similar to that obtained in figure 5b), is found. This important finding gives strong support again to the significance of load partitioning during creep of MMCs. In the PM composites, figure 6a), the $\Delta'\sigma$ values reached are considerably higher than in the IM materials. Furthermore, the linearity of $\Delta'\sigma$ with σ is more evident. In figure 6a) the common behavior manifested by the data of several composites (data of refs. [2,14,19,38]), which reveal a rapid increase of $\Delta'\sigma$ with σ , should be noted. These composites are reinforced with whisker or fibers except for one reinforced by bulky particles [2]. This composite has a high fraction of reinforcement: 30% vol. The $\Delta'\sigma$ vs. σ data of this composite are located in a lower range of applied stress, accounting for the lower range of stresses that can be borne in comparison to the whisker-reinforced composites. The rapid increase of $\Delta'\sigma$ with σ is an indication that the fraction of transferred stress is very high: roughly, about half that of the applied stress. The high stress that the material from Kuchařová *et al.* [38] can sustain is remarkable. A value as high as 400 MPa at 648 K is reported. In the framework of the present analysis and considering the fraction of reinforcing fibers, 15% in vol., the average stress that these fibers bear goes up to almost 1.5 GPa.

There is a second group of PM composites [9,11,22,27,30,37], for which the slope of $\Delta'\sigma$ vs. σ data is lower and more scattered. The increase of $\Delta'\sigma$ with σ occurs in a more erratic manner. This can be well understood on the basis of underlying damage processes occurring during creep deformation, which must reduce the effectiveness of load transfer mechanism to a certain extent. The data scatter, associated with the influence of irregular damage processes at metal-matrix interfaces, is an indication of the complexity of quantifying this “mechanism” in the framework of a predictive model of composites creep behavior.

The picture revealed by the IM materials, figure 6b), is significantly different from that of the PM ones. The data of these composites is mostly grouped in a region of lower stress values. It is to be noted the anomalous trend shown by two of these materials reinforced by whiskers [18] and saffil fibers [21]. For these materials the data is located in a region of high stress. Here, an initially high $\Delta'\sigma$ decreases with increasing σ . This is very likely an indication that the effectiveness of the load transfer mechanisms, in principle very significant, is increasingly inhibited by the action of local stresses responsible for severe damage mechanisms at matrix-reinforcing interface as σ increases. The linear trend of $\Delta'\sigma$ with σ of the remaining IM composites is, somewhat, more scattered than can be appreciated in the PM composites, particularly when compared to the first group of these PM composites. In some cases, negative values of $\Delta'\sigma$ are obtained (the composite is less creep resistant than the corresponding un-reinforced alloy). This again can be well interpreted on the basis of severe damage mechanisms occurring during creep deformation.

All these observations together not only reveal the remarkable significance of load partitioning mechanism to understanding the creep of MMCs, but also that these mechanisms are more effective in the PM composites than in the IM ones. This is consistent with the fact that PM is a better procedure to create a “cleaner” metal-ceramic interface than IM and, hence, that in the PM composites it is easier to transmit the applied load to the reinforcement. Again, IM is more prone to generate artifacts, resulting from chemical reactions, at the metal-ceramic interface during materials preparation which may inhibit the ideal load transfer process. As mentioned above, some composites are weaker than the corresponding un-reinforced alloys [16,17,29], and it is telltale sign that in all these cases the composites were prepared by IM.

b) Matrix strengthening

The strength increase $\Delta'\sigma$ cannot be strictly identified with σ_T . A matrix strengthening factor should be taken into account (constant A' in equation (4)). As summarized in Table I, the microstructure of the composite matrix and the un-reinforced alloy differ in grain size, d , dislocation density, ρ , texture, as well as in the precipitation state. Matrix strengthening can be

evaluated from the behavior of the un-reinforced alloy, Fig. 4a), and by assuming that the creep behavior is governed by Sherby's equation for constant substructure [49], as proposed in [40].

$$\dot{\epsilon} = K \left(\frac{D_L}{b^2} \right) \left(\frac{\lambda}{b} \right)^3 \left(\frac{\sigma'}{E} \right)^8 \quad (5)$$

where K is a material constant (equal to about 10^9 for high stacking fault energy materials), $D_L = D_o \exp(-Q_L/RT)$ is the lattice diffusion coefficient of aluminum, ($D_o = 1.7 \times 10^{-4} \text{ m}^2/\text{s}$, $Q_L = 142 \text{ kJ/mol}$ [68]), b is de Burgers vector, equal to: $b = 2.86 \times 10^{-10} \text{ m}$ in aluminum, λ is the inter-particle distance between incoherent, β , particles, E is the Young's modulus, and σ' is the effective stress resulting from the presence of a "true" threshold stress associated with the Al_2O_3 particles in PM Al alloys [3,40]. The high n values obtained, Figure 4, in comparison to that predicted by equation (5), are in fact attributed to this threshold stress, as it was found to be independent of σ [40].

The strengthening due to the microstructure is, hence, associated with different λ values in the alloy and the composite. Consequently, the creep rate of the composite matrix, $\dot{\epsilon}_{comp}$ in terms of that of the un-reinforced alloy, $\dot{\epsilon}_{alloy}$, should be given, according to equation (5), by,

$$\dot{\epsilon}_{comp} = \frac{1}{a^3} \dot{\epsilon}_{alloy} \quad (6)$$

where $a = \lambda_{alloy} / \lambda_{comp}$. It is difficult to calculate appropriate values of λ_{alloy} and λ_{comp} because precipitation coarsening during creep alters the value of λ between low and high stress tests [17,36,40]. Due to the higher ρ in the composite matrix, Table I, an accelerated ageing phenomenon occurs [41]. This implies that the density of nuclei for precipitation is larger than in the un-reinforced alloy because dislocations are easy places for nuclei formation. A ratio of $\lambda_{alloy} / \lambda_{comp}$ can be estimated assuming that the inter-particle distance between precipitation nuclei is inversely proportional to ρ , i.e., $\lambda_{alloy} / \lambda_{comp} = \rho_{comp} / \rho_{alloy} = 3.9$, (data of Table I). The predicted creep behavior of the composite matrix at 673 K, according to equation (6), is plotted in Figure 5a) (dotted lines). There is no variation in n , in accordance with the idea that the deformation mechanism is the same in the alloy and the composite matrix.

Now, an appropriate comparison between composite and composite matrix strength, $\Delta\sigma$, to analyze the effect of load partitioning *only* can be made. Figure 5b) shows the strength increase at 673 K (dark points) after matrix strengthening subtraction and, hence, due to load partitioning. Figure 7

shows the resulting dependence at all temperatures, $\Delta\sigma(\sigma)$. Despite some scatter, the good fit of all the data into a reasonable common linear dependence, except at 573K, is remarkable.

A similar analysis of the composites in Table II would reveal a more rigorous view of the load partitioning in these composites. This analysis goes beyond the present scope of this research.

Model predictions

The linear dependence $\Delta\sigma(\sigma)$ found after subtraction of matrix strengthening, Figure 7, can be compared to Shear-lag (scalar) and Eshelby (tensorial) model predictions of the load transferred to the reinforcement, $\sigma_r(\sigma)$. A rigorous prediction is obtained when the misalignment of the reinforcing whiskers with the extrusion axis, Figure 2, is considered [47].

a) Shear-Lag

The Shear-Lag models propose that a fraction of an external stress applied to a discontinuously reinforced composite is transferred to the reinforcement by means of a shear stress at the matrix-reinforcement interface. No stress transfer to the end surfaces of the fibers is considered. In the original shear-lag model [69], the reinforcement is modeled as perfectly aligned cylinders of aspect ratio S , homogeneously dispersed in the matrix. Other authors suggest a more complete description of the load transfer phenomenon considering both the transfer of shear stress at the sides of the fibers and normal stress at the end surfaces [15,56,67,70]. Other complex microstructural considerations, such as texture evolution close to the reinforcement during plastic strain or plastic relaxation effects, should be included if a reliable description is to be aimed at. Although these characteristics are not included at the moment in any Shear-Lag model, it is worth discussing the model predictions because of its direct insight into the load transfer phenomenon and its strong theoretical foundation.

From the existing Shear-Lag models [15,56,67,70], the one developed by Ryu *et al.* [15] will be employed for simplicity. This model, based on the one proposed in [56], takes into account the effective aspect ratio of individual whiskers, S_{eff}^I , misaligned θ with the loading (extrusion axis) direction,

$$S_{eff}^I = S \cos^2 \theta + \left(\frac{3\pi - 4}{3\pi} \right) \left(1 + \frac{1}{S} \right) \sin^2 \theta \quad (7)$$

Then, the average effective aspect ratio of the reinforcement, S_{eff} , can be obtained from,

$$S_{eff} = \int_0^{\pi/2} S_{eff}^I(\theta) \gamma(\theta) d\theta = \int_0^{\pi/2} S_{eff}^I I(\theta) 2\pi \sin \theta d\theta \quad (8)$$

where $\chi(\theta)$ is the density function, $\chi(\theta) = I(\theta) 2\pi \sin(\theta) = [I_{rd} + I_d(\theta)] 2\pi \sin(\theta)$, which defines the degree of alignment of the reinforcement with the loading direction, Figure 2. Equation (8) is solved numerically by the Simpson method [42] leading to $S_{eff} = 1.70$.

Now, the effective stress on the matrix, σ_{eff} , can be calculated from the model [15] according to,

$$\sigma_{eff} = \sigma \left[1 - \left(\frac{f(S_{eff}/2 + 1)}{f(S_{eff}/2 + 1) + (1 - f)} \right) \right] = \sigma - \sigma \left(\frac{f(S_{eff}/2 + 1)}{f(S_{eff}/2 + 1) + (1 - f)} \right) = \sigma - \sigma_T \quad (9)$$

As can be seen, the predicted stress borne by the reinforcement is linearly dependent on σ with the term in parenthesis the proportionality constant (equal to 0.246). The σ dependence of σ_T predicted by equation (9) is the dotted line shown in the plot of Figure 7.

b) Eshelby

The Eshelby model [55] is a straightforward and widely used method for calculating the stress state distribution in both phases of a two phase material when an external stress is applied or a mismatch due to a change in temperature occurs. The key feature of this model is to carry out an imaginary substitution of the reinforcement by a homogeneous inclusion made of the matrix material. This concept is named “equivalent homogeneous inclusion” and makes it possible to establish and solve an elastic problem equivalent to the real one. As the Eshelby model is based on the continuum theory, the elastic behavior of an MMC is described in this model without any microstructural considerations, apart from reinforcement geometry and sometimes orientation [47]. Therefore, microstructural contributions that may influence the stress borne by both phases, such as grain size effects, texture evolution or plastic relaxation at high temperatures, are missing. However, it is worth considering that predictions of such an Eshelby model must give a good approximation of the load transferred to the reinforcement excluding these other considerations.

The Eshelby model, commonly used to predict composites properties at room temperature, is employed here to calculate the stress borne by the reinforcement under creep loading conditions. The mean residual stress tensor, $\bar{\sigma}^M = (\bar{\sigma}_1^M, \bar{\sigma}_2^M, \bar{\sigma}_3^M)$ (subscripts 1, 2, and 3 refer to the loading direction and the perpendicular ones, respectively) due to σ for well aligned whiskers is given by,

$$\bar{\sigma}^M = f C_M (B - I) \{ (C_M - C_R) [B - f(S - I)] - C_M \}^{-1} (C_R - C_M) C_M^{-1} \sigma \quad (10)$$

where, B is the Eshelby tensor, I the unity tensor, and C_M and C_R the elastic constant of the matrix and reinforcement, respectively [71]. The C_R values have been taken at room temperature because of their small variation in the range under consideration (573 – 723K). The C_M values have been obtained from Young’s modulus and Poisson’s ratio data [72].

As done in the Shear-Lag model prediction from the previous section, the Eshelby analysis has also taken into consideration the misalignment of the reinforcement [47]. Some temperature variation of load transfer is expected due to temperature variation of C_M . The difference is, however, minimal and an average predicted dependence will be used.

Once the mean matrix stress due to σ is known, and applying Tresca's yield criterion, the effective stress along the testing direction can be obtained as,

$$\sigma_{eff} = (\bar{\sigma}_{3M} - \bar{\sigma}_{1M}) = P\sigma \quad (11)$$

where P is a dimensionless constant ($P \geq 1$) which quantifies the increase in load that a particular composite material can bear due to the load transfer mechanism. This implies that the actual stress responsible for composite matrix deformation by creep is only a fraction of the external applied load. Then, the identification of the term, σ_T , from equation (11) with the load borne by the reinforcement is straightforward; *i.e.*,

$$\sigma_T = (P^{-1} - 1)\sigma \quad (12)$$

The value obtained of P is $P=0.85$, which leads to the σ dependence of σ_T shown in figure 7 as a solid line. The difference between the Shear-Lag and Eshelby model predictions, figure 7, should be attributed to the different reinforcement shape assumed in each case (cylinders *vs.* ellipsoids) and hence, to a geometric effect of the predicted stress transfer.

As can be seen, the fit of the experimental data to the prediction obtained from the models is reasonably good, given the simplicity assumed in these two models. Other factors, such as stress relaxation by diffusion [73] and tensile damage mechanisms, not here considered, may also have contributed to the creep behavior of the composite, in particular in the low stress regime where the experimental $\Delta\sigma$ data fall below the predictions' trends. In this range, the experimental data corresponds to tests conducted at the two highest temperatures investigated: it is likely that diffusional processes have led to stress relaxation and, hence, to composite weakening. The possible influence of stress relaxation can be envisaged from the trend observed: it is $\Delta\sigma=0$ for a positive σ . In the high stress regime, however, the models predict higher values of σ_T than those of $\Delta\sigma$ obtained experimentally. The data in this regime corresponds to tests conducted at the two lowest temperatures investigated. At present, there is as yet, no satisfactory explanation for this divergence, but clearly other sources for strengthening, not considered here, would also influence the increased composite creep behavior. It is possible that strengthening mechanisms associated with low temperature behavior such as the increased dislocation density and the interaction of geometrically necessary dislocations, GNDs, not affected by dynamic annealing processes at these low testing temperatures, could play a role in composite creep strengthening and the divergence

from the model predictions [41]. The scarcity of experimental data in this stress regime, however, suggests that further experimental and theoretical work is needed to account for this difference.

Discussion

The microstructural considerations, the experimental data analysis conducted, and the reasonable good fit with model predictions of σ_T support the relevance of load partitioning to explain the improved creep resistance of MMCs. As already mentioned, damage mechanisms have not been considered in the models, due to the complexity of including such processes in a common pattern for discontinuously reinforced MMCs. Damage processes could be treated as a separate “softening” mechanism, or also, as another relaxation process of the transferred stress. The possibility of including these processes in an overall predictive model of the composite creep behavior, however, is still weak. As shown by several investigations [74,75], only descriptive and phenomenological pictures of the specific damage mechanisms, occurring in composites, are conducted, with little effort focused on developing a predictive capacity on the overall creep process. Furthermore, damage mechanisms may differ from one composite to another or under different test conditions [74,76], whereas the load transfer mechanisms, however, are common to all these materials and under different conditions. In summary, it is possible to study, in great detail, the specific damage mechanisms occurring during composite deformation, but it is very difficult to predict which mechanism will operate and its importance before the test is conducted. Such an analysis is essential for a thorough assessment of composite creep behavior.

A final analysis which can be done with the data of Table II is to compare the apparent n of the aluminum alloy matrix composites reported with that of the corresponding un-reinforced alloy [2,8,11-14,18-39]. This is shown in the double logarithmic plot of Figure 8 in which data from the IM and PM materials are compared. From this figure, clear differences between the IM and PM composites can be seen as follows: Firstly, the n -values of the IM materials are lower than those of the PM ones: n ranges from about 3-10 in the alloys and 4-20 in the composites. In the PM materials, however, the n values, range from about $n = 7$ to about $n = 100$ for the alloy and $n = 9$ to about $n = 130$ for the composites. And secondly, the n -values of the PM materials follow more closely the line $n_{comp} = n_{alloy}$ than the IM materials. In other words, a wider scatter of data in the IM than for the PM materials is evident. Specifically, about half of the data points of the IM materials closely follow the linear fit $n_{comp} = n_{alloy}$, but for the remaining points, it is $n_{comp} > n_{alloy}$. The data of the present research, also included in this plot, fit as well, as expected, with the PM materials' data. The wider scatter in the n value in the composite material than in the un-reinforced alloy is worth mentioning.

The evidence that the apparent n of the PM composites does not increase by the addition of the reinforcement confirms the idea that the deformation mechanism in the composite matrix is the same as in the un-reinforced alloy. This reveals that σ_o associated with the addition of the reinforcement, as calculated by the extrapolation method [1,3-5,52] to reduce the apparent n , is meaningless. The high n values in these materials should be attributed to the presence of the Al_2O_3 dispersion, inherent in the PM procedure, as in ODS alloys [2,40]. These particles are responsible for a “threshold stress” type of mechanism in agreement with previous studies [2,52] in both, the un-reinforced alloy and the composite. This mechanism has been explained [2,40,52] as an interaction of dislocation with the alumina particles (absent in the IM materials) through the detachment model proposed by Artz and coworkers [50]. The wide variation in n is because of the different temperatures investigated, but also because of the significant effect of the precipitation phenomena occurring in the age-hardenable alloys of Table II. The precipitation process which occurs in these types of alloys as creep progresses influences the value of n [17,40].

On the other hand, the wider scatter of the IM materials in comparison to the PM ones (it is $n_{comp} > n_{alloy}$ for many of these materials) should be also attributed to the specific processing route employed. It is likely that the possible interface reactions occurring in the IM composites, absent in the PM ones, provide a different load bearing capacity in the IM composites and, hence, to a different creep behavior and n value. In the IM composites, further mechanisms, besides microstructure strengthening and load partitioning, should be taken into account, but not a σ_o “mechanism”.

The temperature dependence of σ_o found in several investigations should be carefully considered. The σ_o values analyzed have been obtained by the extrapolation method, and their reliability has been criticized above. Instead, the temperature dependence of σ_T obtained by the present method is negligible, and is in reasonable agreement with the predictions of the two models employed.

Summary

The σ_o term included in the creep equation of discontinuously reinforced metal matrix composites, MMC, to rationalize the apparent high n and Q_c values obtained has been criticized. It has been proposed that the improved creep strength relies mainly on a microstructural factor (smaller interparticle distance) and a load transfer mechanism. The data comparison of the creep behavior of a PM 6061Al-15vol%SiC_w composite and the corresponding 6061Al alloy reveals that the additional stress needed by the composite, $\Delta\sigma$, to creep at a given strain rate (after matrix strengthening subtraction) is linearly dependent on σ . In this new frame, a similar creep equation to that in which the threshold stress was considered is obtained. Now, however, $\Delta\sigma$ is interpreted

as a load transfer term. The values obtained agree reasonably well with the σ_T values predicted by the Shear-Lag and Eshelby model predictions.

Also, a similar linear dependence $\Delta' \sigma(\sigma)$, as that obtained in the present composite, is found from a thorough literature review of creep data of aluminum alloy MMCs and corresponding alloys, in particular, when the processing route employed is the PM one. Such a result, common to a representative number of investigations, gives strong support to the relevance of load partitioning during high temperature deformation of MMCs materials. Finally, (Table II) the relevance of possible damage mechanisms to the overall composite creep behavior has been shown from the comparison between composite and un-reinforced alloy creep data in the above investigations. It can be clearly seen that these mechanisms are more prone to occur in the IM composites than in the PM ones. This is most likely to be attributed to the presence of artifacts at the metal-matrix interface arising from reactions during composite processing. The unpredictable occurrence of these processes makes a rigorous mathematical treatment of damage into a predictive composite model of composites creep behavior very difficult.

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References

- [1] Mishra RS, Pandey AB. Metall Trans A 1990;21:2089-90.
- [2] Park K-T, Lavernia EJ, Mohamed FA. Acta Metall Mater 1990;38:2149-59.
- [3] Mohamed FA, Park K-T, Lavernia EJ. Mater Sci Eng A 1992;150:21-35.
- [4] González-Doncel G, Sherby OD. Acta Metall Mater 1993;41:2797-805.
- [5] Park K-T, Mohamed FA. Metall Mater Trans 1995;26:3119-29.
- [6] Ma ZY, Tjong SC. Comp Sci Technol 2001;61:771-86.
- [7] Čadek J, Kuchařová K, Zhu SJ. Mater Sci Eng A 2000;281:162-8.
- [8] Spigarelli S, Cabibbo M, Evangelista E, Langdon TG. Mater Sci Eng A 2002;328:39-47.
- [9] Peng LM, Zhu SJ, Ma ZY, Bi J, Wang FG, Chen HR, Northwood DO. Mater Sci Eng A 1999;265:63-70.
- [10] Huang M, Li X, Yi H, Ma N, Wang H. J. J Alloys Compd 2005;389:275-80.
- [11] Desmukh SP, Mishra RS, Keding KL. Mater Sci Eng A 2005;410-411:53-7.
- [12] Moreno MF, González Oliver CJR. Mater Sci Eng A 2006;418:172-81.
- [13] Li Y, Langdon TG. Metall Mater Trans A 1998;29:2523-31.
- [14] Ryu HJ, Chung H, Seung IC, Hong SH. J Mater Res 2004;19:3633-40.
- [15] Ryu HJ, Cha SI, Hong SH. J Mater Res 2003;18:2851-8.
- [16] Krajewski PE, Allison JE, Jones JW. Metall Trans A 1993;24:2731-41.
- [17] Requena GC, Deguischer HP. Z Metallkd 2005;96:805-11.
- [18] Nieh TG. Metall Trans. A 1984;15:139-45.
- [19] Pickens JR, Langan TJ, England RO, Liebson M. Metall Trans A 1987;18:303-12.
- [20] N.Hansen, D.J.Jensen, Y.-L. Liu. Proc. 9th Int. Riso Symp. 1988. p.365-70.
- [21] Komenda J, Henderson PJ. Metal matrix composites – Processing, microstructure and properties. 12th Riso Int. Symp. 1991:449-54.

- [22] Zedalis MS, Bryant JD, Gilman PS, Das SK. JOM 1991;8:29-31.
- [23] Dlouhy A, Merck N, Eggeler G. Acta Metall Mater 1993;41:3245-56.
- [24] Yu D, Chandra T. Mater Sci Forum. 1993;113-115:563-8.
- [25] Furukawa M, Wang J, Horita Z, Nemoto M, Ma Y, Langdon TG. Metall Mater Trans A 1995;26:633-9.
- [26] Krajewski PE, Jones JW, Allison JE. Metall Trans A 1995;26:3107-18.
- [27] Zhu SJ, Peng LM, Ma ZY, Bi J, Wang FG, Wang ZG. Mater Sci Eng A 1996;215:120-4.
- [28] Matsuda N, Akaike J, Hongo K, Matsuura K. Mater Sci Eng A 1997;234:751-4.
- [29] Liauo CS, Huang JC. Metall Mater Trans A 1997;28:1859-69.
- [30] Bidlingmaier T, Vogt D, Wanner A. in: G. Ziegler (Ed.), Werkstoffe und Werkstoffverbunde, DGM Informations gesellschaft, Frankfurt, 1996: p.245.
- [31] Wakashima K, Moriyama T, Mori T. Acta Metall Mater 2000;48:891-901.
- [32] Winand HMA, Whitehouse AF, Withers PJ. Mater Sci Eng A 2000;284:103-13.
- [33] Kausträter G, Skrotzki, Eggeler G. Mater Sci Eng A 2001;319-21:716-20.
- [34] Zhu SJ, Peng LM, Zhou Q, Ma ZY, Kuchařová, Čadek J. Mater Sci Eng A 2002;282:273-84.
- [35] Yawny A, Eggeler G. Mater Sci Eng A 2004;387-9:905-9.
- [36] Requena GC, Deguischer HP. Mater Sci Eng A 2006;420:265-75.
- [37] Čadek J, Kuchařová K, Zhu SJ. Mater Sci Eng A 2000;283:172-80.
- [38] Kuchařová K, Čadek J, Zhu SJ. J Mater Sci 2003;38:3535-43.
- [39] Zhu SJ, Kuchařová K, Čadek J. Metall Mater Trans A 2000;31:2229-37.
- [40] Fernández R, González-Doncel G. J Alloys Compd 2007; 440:158-67.
- [41] A, Ibáñez J, López V, Lieblich M, González-Doncel G. Scripta Mater 1996;34:471-8.
- [42] Borrego A, Fernández R, Cristina MC, Ibáñez J, González-Doncel G. Comp Sci Technol 2002;62:731-42.
- [43] Pérez-Prado MT, Cristina MC, Ruano OA, González-Doncel G. Mater Sci Eng A 1998;244:216-23.
- [44] Borrego A, González-Doncel G. Mater Sci Eng A 1998;245:10-8.
- [45] Borrego A, González-Doncel G. J Mater Sci 2003;38:2761-5.
- [46] Bruno G, Fernández R, González-Doncel G. Mater Sci Eng A 2004;328:188-97.
- [47] Fernández R, Bruno G, González-Doncel G. Acta Mater 2004;34:5471-83.
- [48] Fernández R, Bruno G, González-Doncel G. Mater Sci Eng A 2005;403:260-8.
- [49] Lin J, Sherby OD. Res Mechanica 1981;2:251-93.
- [50] Artz E, Wilkinson SD. Acta Metall 1986;34:1893-8.
- [51] Ma ZY, Tjong SC, Li YL, Liang Y. Mater Sci Eng A 1997;225:125-34.
- [52] Park KT, Lavernia EJ, Mohamed FA. Acta Metall Mater 1994;42:667-78.
- [53] Sherby OD, Walser B. Scripta Metall 1982;16:213-9.
- [54] Gariboldi E, Vedani M. Advanced Eng Mater 2000;2:737-41.
- [55] Eshelby JD. Proc Roy Soc A 1957; 241:376-96.
- [56] Nardone VC, Prewo KM. Scripta Metall 1986;20:46-8.
- [57] Hashin Z, Shtrikman SJ. Mech Phys Solids 1963;11:127-40.
- [58] Ashby MF. Acta Metall Mater 1993;41:1313-35.
- [59] Arsenault RJ. Mater Sci Eng 1984;64:171-81.
- [60] Čadek J, Kuchařová K, Zhu SJ. Mater Sci Eng A 1999;272:45-56.
- [61] Hong SH, Sherby OD, Divecha AP, Karmarkar, MacDonald BA. J Comp Mater 1988; 22: 102-23.
- [62] Daehn GS, González-Doncel G. Metall Trans A 1989;20:2355-68.
- [63] González-Doncel G, Karmarkar SD, Divecha AP, Sherby OD. Comp Sci Technol 1989;35:105-20.
- [64] González-Doncel G, "Tensile ductility and fracture of aluminum-SiC composites under changing temperature conditions" in "Thermomechanical fatigue & fracture" 2002 Ed. Aliabadi MH. WIT press, Shouthampton 245-85.
- [65] Ma ZY, Tjong SC. Comp Sci Technol 1999;59:737-47.

- [66] Winand HMA, Whitehouse AF, Withers PJ. Mater Sci Eng A 2000;284:103-13.
- [67] Starink MJ, Syngellakis S. Mater Sci Eng A 1999;270:270-7.
- [68] Frost HJ, Ashby MF, in *Deformation Mechanisms Maps*, Pergamon Press, Oxford, 1982.
- [69] Cox HL. British J. Appl Phys 1952;3:72-8.
- [70] Clyne TW. Mater Sci and Eng A 1989;A122:183-92.
- [71] Clyne TW, Withers PJ. "An introduction to metal matrix Composites" Cambridge (MA): Cambridge University Press; 1993.
- [72] Simmons GS, Wang H, in *Single Crystal Elastic Constants and Calculated Aggregate Properties: a Handbook*. MIT Press, Cambridge, Mass. (1971).
- [73] Rösler J, Bao G, Evans G. Acta Metall Mater 1991;39:2733-8.
- [74] Vedani M, D'Errico F, Gariboldi E. Comp Sci Technol 2006;66:343-9.
- [75] Zong BY, Derby B. J Mater Sci 1996;31:297-03.
- [76] Barnes SJ, Prangnell PB, Roberts SN, PJ Withers PJ. Scripta Metall Mater 1995;33:323-9.

Material (code)	T_{ext} (K)	P_{ext} (MPa)	d (μm)	ρ (m^{-2})	$V_{\text{f}}\langle 111 \rangle$	$V_{\text{f}}\langle 100 \rangle$	S, S_{eff}	f_a	I_{rd}	a	b
Alloy (E220)	763	543	3.5	8×10^{12}	56	35	-	-	-	-	-
composite (E219)	753	669	1.5	3.1×10^{13}	41	17	2, 1.70	0.23	0.83	3.69	27.69

Table I. Processing conditions (extrusion temperature, T_{ext} , and extrusion pressure, P_{ext}) and microstructural parameters of the alloy and the composite. d is the grain size, ρ the dislocation density, $V_{\text{f}}\langle 111 \rangle$ and $V_{\text{f}}\langle 100 \rangle$ the volume fraction of crystallites with the crystallographic $\langle 111 \rangle$ and $\langle 100 \rangle$ oriented with the extrusion axis, respectively, S is the average whisker aspect ratio, f_a is the fraction of aligned whiskers with the extrusion axis (loading) direction, following a Gaussian distribution function (the remaining fraction, $1 - f_a$, is randomly aligned). Parameter I_{rd} refers to the population of randomly aligned whiskers and a and b define the Gaussian distribution function of the aligned whiskers.

Author, year [ref.]	Material (*)	Processing	Test mode	Test temp. K	Stress exponent (composite)	Stress exponent (unreinforced alloy)	Remarks
Nieh, 1984 [18]	* 6061Al-20vol%SiC _w * 6061Al-30vol%SiC _p	IM	Tensile	561	20.5 15.1	4.9	Increasing stress exponent with stress in the unreinforced alloy
Pickens <i>et al.</i> , 1987 [19]	* 6061Al-20vol%SiC _w	PM	Torsion	700	10.5	6.8	
Hansen <i>et al.</i> , 1988 [20]	Al-0.8%Al ₂ O ₃ -2vol%SiC _w	PM	Tensile	673	4.4	4.7	
Park <i>et al.</i> , 1990 [2]	* 6061Al-30vol%SiC _p	PM	Double shear	648, 678	9.8 11.0	13.6, 12.4	Decreasing stress exponent with stress
Komenda <i>et al.</i> , 1991 [21]	* AlCu3-20vol%- δ -Al ₂ O ₃ (saffil fibers)	IM	Tensile	423	15.7	9.2	Up to 50% strain was primary creep
Zedalis <i>et al.</i> , 1991 [22]	8009Al-11vol%SiC _p	PM	Tensile	589	20.3	28.8	No reinforcing effect above 589 K
Dlouhy <i>et al.</i> , 1993 [23] ⁺	AlSi7Cu3- 15vol%Al ₂ O ₃ (saffil fibers) AlSi7Cu3Mg- 15vol%Al ₂ O ₃ (saffil fibers)	IM	Tensile	623	10.0 7.3	6.0 8.1	Coarsening of precipitates does not influence creep rate of MMCs
Yu & Chandra, 1993 [24]	* 6061Al-15vol%Al ₂ O ₃ (particles)	**	Compression	673	8.3	7.8	Load transfer and matrix strengthening
Furukawa <i>et al.</i> , 1995 [25]	* 6061Al-20vol%Al ₂ O ₃ (microspheres)	IM	Tensile	623, 673	5.0, 5.2	4.2, 3.5	
Krajewski <i>et al.</i> , 1995 [26]	Al-15vol%TiC _p Al-1.5Mg-20vol%TiC _p	XD (IM) XD (IM)	Tensile	423, 523	8.6-21.6 8.5, 9.3	5.4-20.8, 6.6 1.7-7.6, 3.6	Load transfer and matrix strengthening
Zhu <i>et al.</i> , 1996 [27]	8009Al-15vol%SiC _w	PM	Compression	573, 623, 673, 723	14, 18, 18, 11	16.7, 17.6, 13.3, 15.4 ⁺⁺	High and similar n values for the composite and the alloy. Threshold stress analysis (with $n=5$) in the alloy and the composite.
Matsuda <i>et al.</i> , 1997 [28]	* 6061Al-10vol%Al ₂ O ₃ (particles) * 6061Al-20vol% Al ₂ O ₃ (particles)	IM	Tensile ?	573, 773	6.8-12.5, 12.3-5.4 7.5, 17.5-5.3	9.0, 3.5	No strength difference between 10vol and 20vol% reinforced composites.
Liao & Huang, 1997 [29]	AlSi12-12vol%Al ₂ O ₃ (saffil fibers)	IM	Tensile	523, 623, 723	-, 7.6, 7.0	-, 8.7, 6.0	Strength evaluation through load transfer and microstructural change. Negative n values at low temperature and high strain rates. Virtually, no increase in composite creep strength.
Li & Langdon, 1998 [13]	* 6092Al-25vol%SiC _p	PM	Tensile	623, 723	9.3, 9.8	9.3, 8.1	Analysis of results by means of a threshold stress and load transfer mechanism.
Peng <i>et al.</i> , 1999 [9]	8009Al-15vol%AlBO _w	PM	Compression	573, 623, 723	13,14,14	16.7, 17.6, 15.4 ⁺⁺	Similar n values for the composite and the alloy. Analysis with threshold stress ($n=5$) and load transfer mechanism. High n values.
Bidlingmaier <i>et al.</i> , 1996 [30]	AlSi12CuMgNi-13.9 vol%Al ₂ O ₃ (saffil fibers)	IM	Tensile	523, 623, 673	10.5, 5.7, 6.3	3.7, 4.5, 5.3	Significant variation of n with temperature
Wakashima <i>et al.</i> , 2000 [31]	* 6061Al-10vol%SiC _p	PM	Tensile	673	12.4	8.6	The incorporation of a threshold stress gives rise to $n=3$ in the alloy (viscous flow).
Winand <i>et al.</i> , 2000 [32]	Al-10vol%SiC _{p(3μm)} Al-10vol%SiC _{p(20μm)} Al-20vol%SiC _{p(3μm)} Al-20vol%SiC _{p(20μm)} Al-10vol%SiC _w Al-20vol%SiC _w	PM	Tensile	543	30.2, 17.1 31.0, 14.7 36.0, 20.4	16.7, 12.9	Direct measurements of phases stress during loading by neutron diffraction. Stress relaxation occurs during stages I and II of creep.
Kausträter <i>et al.</i> , 2001 [33]	AlMg5-15vol% Al ₂ O ₃ (saffil fibers)	IM	Tensile	623	9.2	3.1	Load transfer to fibers by work hardened zone formation, recovery and fiber breakage are relevant during creep deformation.
Zhu <i>et al.</i> , 2000 [34]	Al-SiCp/Al ₄ C ₃ (SiC/AlC1) Al-SiCp/Al ₄ C ₃ (SiC/AlC2)	PM	Tensile	623, 723	19.1, 23.3 ⁺⁺ 23.8, 29.6 ⁺⁺	31.9, 32.3 ⁺⁺ 25.7, 29.3 ⁺⁺	Oxide dispersion strengthened matrix alloys. Creep strengthening analysis through threshold stress and load transfer mechanisms.
Spigarelli <i>et al.</i> , 2002 [8]	* 2024Al-15vol%SiCp	PM	Tensile	548, 573,603	17.0, 8.8, 16.3	17.5, 12.2, 17.8	Differences in alloy elements between matrix composite and alloy. Stress dependence of n . Threshold stress dependent on stress because of continuous precipitation during test.
Yawny <i>et al.</i> , 2004 [35]	AlZn8-8vol% Al ₂ O ₃ (saffil)	IM	Compression	573	10.1	5.6	Threshold stress and load transfer together are not able to explain the creep of this composite.

Ryu <i>et al.</i> , 2004 [14]	*2124Al-20vol%SiC _w (extrusion ratio 10:1) *2124Al-20vol%SiC _w (extrusion ratio 15:1) *2124Al-20vol%SiC _w (extrusion ratio 25:1)	PM	-	573	10 10 10	10	Same <i>n</i> value (<i>n</i> ≈10) for the un-reinforced alloy and the composites. Creep strengthening analysis by load transfer mechanism. Misalignment is taken into account.
Deshmukh <i>et al.</i> , 2005 [11]	AlMg6Sc1Zr1-10vol%SiCp	PM	Tensile	423, 477, 533	11.8, 12.6, 14.6	12.3, 12.4, 8.8	None of the models used to analyze the threshold stress account for this stress in the alloy. Threshold stress in the composite leads to <i>n</i> =5 but no explanation is given of its meaning.
Requena <i>et al.</i> , 2006 [36]	AlSi12CuMgNi-10vol%Al ₂ O ₃ (fibers) AlSi12CuMgNi-15vol%Al ₂ O ₃ (fibers) AlSi12CuMgNi-20vol%Al ₂ O ₃ (fibers)	IM	Tensile	573	3.3 3.5 3.8	3	Similar <i>n</i> value (<i>n</i> ≈3-4) for the un-reinforced alloy and the composites. Overloading tests. Load transfer mechanism.
Moreno & Oliver, 2006 [12]	N5 (pure Al + 5.3vol%SiC _f) N11 (pure Al + 11.6vol%SiC _f)	PM	Compression	673	15-25 15-35	10-40	Variable <i>n</i> (decreases) with stress. Presence of a threshold stress related to oxide contamination.
Čadek <i>et al.</i> , 2000 [37]+++	8009Al-15vol% SiCp	PM	Tensile	623, 673, 723	61.7, 14.5, 87.7, 16.9, 130, 18.6	58.4, 21.8, 105, 13.1, 70.6, 13.9	Threshold stress of the composite higher than in the un-reinforced alloy. No load transfer effect. Very high <i>n</i> values at low applied stress.
Kuchařová <i>et al.</i> , 2003 [38]+++	8009Al-15vol% Al ₂ O ₃ (fibers)	PM	Tensile	648, 698, 748	49.2, 12.7, 84, 11.7, 64.2, 14.8	57.7, 14.6, 82.4, 16.8, 68.8, 17.1	Increased composite threshold stress due to load transfer effect. Very high <i>n</i> values at low applied stress.

* Ageing matrix alloy

** Not specified processing but seemingly to be IM

+ Analysis made from data of Table 2 of ref [23] as some data points in plot of fig.8 in this reference are missing.

++ Stress exponents values taken from ref. Peng&Zhu, Int.J.M.P&Tech. vol.18 (2003) 215-254.

+++ Analysis of stress exponent made on the basis of two regions of high and lower *n* values.

Table II.

Summary of the creep studies in the literature on discontinuously reinforced MMCs (aluminum alloy matrix) which include data of the corresponding un-reinforced alloys.

a)

b)

Figure 1.- Microstructure of a) the un-reinforced 6061Al reference alloy (E220) and b) the 60601Al-15vol.% SiC_w, composite (E219). Both were obtained by a powder metallurgical procedure using 6061Al from the same powder batch. In b) a same region, as observed by secondary electrons and back-scattered electrons to reveal the grain size, is shown.

Figure 2.- One dimensional pole figure showing the distribution of the long axis direction of the reinforcing whiskers. The Gaussian fit corresponds to whiskers which are aligned with the extrusion axis. The constant intensity value I_{rd} corresponds to the random population.

a)

b)

Figure 3.- Double logarithmic plot showing typical strain rate vs. strain curves at 623 K of, a) the PM 6061Al un-reinforced alloy (E220) and b) the PM 6061Al-15vol%SiC_w composite (E219).

Figure 4.- Minimum creep rate as a function of the applied stress in a double logarithmic plot of, a) the PM 6061Al alloy (E220) and, b) the PM 6061Al-15vol%SiC_w composite (E219). Numbers denote the stress exponent at each temperature.

Figure 5.- a) Creep rate as a function of the applied stress at 673 K in a double logarithmic plot of the PM 6061Al alloy (E220) and the PM 6061Al-15vol%SiC_w composite (E219). Dotted line refers to the expected composite matrix behavior (without reinforcement). b) Applied stress, σ , dependence of the composite creep strength increment, $\Delta'\sigma$ and $\Delta\sigma$ (after subtracting matrix strengthening), with respect the un-reinforced alloy at 673 K. A linear correlation of $\Delta'\sigma$ and $\Delta\sigma$ with σ is obtained.

Figure 6.- Variation of the strength increment, $\Delta'\sigma$, with σ of composites (aluminum alloy matrix) studied in the literature, Table II. a) Data from the IM materials, b) data from the PM materials. A linear correlation between $\Delta'\sigma$, and σ , as obtained for the present composite, Figure 5b), is particularly apparent for the case of the PM composites. The linear trend of the data from the material of ref. [38], Fig. 5a), extends up to σ values of 400 MPa.

Figure 7.- Dependence of $\Delta\sigma$ with σ , as obtained from the experimental data of the composite and the predicted composite matrix behavior (matrix strengthening subtracted) at all temperatures investigated. A linear correlation between $\Delta\sigma$ and σ , with small variation due to temperature, is obtained. Dashed and solid lines correspond to σ_T , the predicted σ dependence of the load carried by the reinforcement according to the Shear-Lag and Eshelby models, respectively. The agreement between $\Delta\sigma$ and σ_T is significant considering the simple assumptions of the models.

Figures captions

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Figure 8.- Variation of the composite apparent stress exponent n with that of the corresponding un-reinforced alloy for the PM and IM materials (from the data reported in Table II). For comparison, the data points of the present research are also included. The higher n values of the PM materials in comparison to those of the IM ones are evident. Numbers denote the corresponding reference number.