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Abstract: The improved creep response of 2618Al-15vol%TiCP and 2618Al-20vol%TiCP metal matrix composites with respect to 2618Al alloy is interpreted assuming a mechanism of load transfer from the aluminum alloy matrix to the TiCP (the effective stress that the matrix alloy of the composites sustains is lower than the applied stress). It is also proposed that a threshold stress is present in the composites and the alloy. The origin of this stress is not the TiCP, but the fine dispersion of alumina particles arising from the powder particles employed in the in situ procedure involved in the materials' manufacture.

Comments on “Creep behavior of *in situ* TiC_P/2618 aluminum matrix composite”

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Abstract

The improved creep response of 2618Al-15vol%TiC_P and 2618Al-20vol%TiC_P metal matrix composites with respect to 2618Al alloy is interpreted assuming a mechanism of load transfer from the aluminum alloy matrix to the TiC_P (the effective stress that the matrix alloy of the composites sustains is lower than the applied stress). It is also proposed that a threshold stress is present in the composites and the alloy. The origin of this stress is not the TiC_P, but the fine dispersion of alumina particles arising from the powder particles employed in the *in situ* procedure involved in the materials' manufacture.

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Very recently, Ji *et al* [1] have studied the compressive creep behavior of *in situ* 2618Al-15vol%TiC_p and 2618Al-20vol%TiC_p composites at the temperatures of 523, 573, and 623 K. The equiaxial TiC particles were about 0.5-1.5 μm in size and were homogeneously distributed in the aluminum alloy matrix. 2618Al alloy processed by the same route was also studied for comparative purposes at the same three temperatures. The creep results obtained are summarized in the double logarithmic plots of steady state or minimum strain rate, $\dot{\epsilon}$, vs. applied stress, σ , shown in Figure 1. In these plots, the behavior of the three materials is compared at each temperature. A power law creep behavior is obeyed in all cases. It can be seen that the creep strength of the composites is always superior to that of the un-reinforced alloy and that the composite with 20%TiC_p is slightly more creep resistant than that with 15% TiC_p.

In their work, Ji *et al* [1] interpreted the creep behavior of these composites, and in particular their better creep resistance than that of 2618Al alloy, Figure 1, on the basis of the existence of a threshold stress term, σ_0 , associated with the TiC_p and the *S* precipitates of the 2618Al alloy. Ji *et al.* [1] justify the need for a threshold stress, in the light of the “much higher” stress exponents and activation energies obtained, compared to those of pure aluminum. They calculated σ_0 using the extrapolation method [2,3]. This procedure consists, basically, in “imposing” an *n* value on the creep equation so that a straight line is obtained when $\dot{\epsilon}^{1/n}$ is represented as a function of σ/E (see figure 3 in [1]). At any temperature, the extrapolated value of σ at $\dot{\epsilon} = 0$ is taken as σ_0 . In this manner, the initially high apparent *n* value decreases down to the “expected” *n*. Ji *et al.* [1] calculated σ_0 assuming that *n*=5. This assumption was based on the fact that *n*=5 is obtained for pure aluminum [4]. The threshold stress was finally interpreted as “the resistance of TiC_p to the deformation of the matrix and the resistance and pinning of the precipitated *S* phases to the dislocation”. In other words, on the obstacle offered by TiC_p and *S* precipitates to dislocation motion. The Orowan mechanism is considered the one underlying the threshold stress. To conclude, these authors also point out the possible occurrence of a load transfer mechanism but in a rather imprecise way.

The purpose of this communication is to reanalyze the results of Ji *et al* [1] with a different perspective of the creep of discontinuously reinforced MMCs. Our view rejects the threshold stress associated with the TiC_p and the *S* precipitates. We propose here that:

A) A threshold stress term exists but only connected to the dispersion of alumina particles arising from the aluminum alloy powder used in the materials' manufacture by the *in situ* process. In this way, the threshold stress term will exist not only for the composites, but also for the 2618Al alloy. Furthermore, since the alumina particle dispersion is similar for the alloy and the composites, a similar value of σ_0 should be expected for all materials at each testing temperature.

B) A mechanism of load transfer from the continuous matrix alloy to the TiC_p explains the improved creep strength of these *in situ* composites with respect to the 2618Al alloy.

Both, the load transfer and the threshold stress contribute in an additive manner to enhance the creep behavior of these two composites, as argued in [5].

This different view of the creep of these composites is based on the following two considerations derived from the results of Ji *et al.* [1]:

1.-Regardless of the fact that the stress exponents and activation energies for creep obtained in [1] are comparably high with respect to pure aluminum, very similar n values are obtained for the composites and alloy studied¹. This observation disqualifies the need of a threshold stress in the composite (and not in the alloy) in order to “diminish” the apparent stress exponent. The similarity between n in the composites and the alloy can be seen in Figure 2 in which the stress exponent at the three temperatures is compared on a double logarithmic scale. The trend matches quite well with that observed for a number of research data analyzed in [3]: The stress exponent is virtually identical for the composites and the alloy. It was proposed in [3] that the improved composites creep strength is mainly based on the load partitioning mechanism. This assumption was verified by the analysis conducted in [3] of a significant number of experimental studies published in the open literature on the creep of composites.

2.- The reliability of the threshold stress values, calculated by the extrapolation method, is doubtful, as argued in [3,6]. The matrix alloy of the composites studied by Ji *et al.* [1] is a commercial forging alloy belonging to the age hardening 2xxx series alloys. As pointed out in [3,6], and as also raised by Requena and Deguischer in [7], the matrix microstructure of these alloys does not remain unchanged during creep deformation

¹ The n values are reported in figure 2 and Table I of [1]. Here, similar but not exact values have been calculated since the data analysis has been conducted not from the original values but from the values extracted from the plots of figure 2 of [1]. This slight difference, however, is irrelevant for the analysis and conclusions defended in this work.

since simultaneous coarsening of precipitates should occur. Furthermore, the precipitation evolution should be different at the different testing temperatures. As a consequence, the stress exponent in these age hardening alloys may vary due to the simultaneous precipitation process [6,7,8,9,10]. The significance of the σ_0 term obtained by this method is, therefore, very controversial.

The availability of creep data of *in situ* 2618Al alloy in Ji *et al*'s work makes it possible to calculate the experimental creep strengthening of the composites due to the addition of the TiC_p in the same way as in [3]. This strengthening is readily taken as the difference between the flow stress of each composite and the alloy at any given strain rate. The resulting difference or creep strength increment, $\Delta\sigma = \sigma_{comp} - \sigma_{alloy}$, is represented as a function of σ in the plots of Figure 3. Figure 3a) is for the 2618Al-15vol% TiC_p composite and Figure 3b) for the 2618Al-20vol% TiC_p one. The linear dependence resulting for both materials is remarkable. It is also noteworthy that this tendency is independent of the test temperature and that the slope of the linear trend is higher for the 20% TiC_p composite than for the 15% TiC_p one. These findings suggest, as mentioned in [3], that load partitioning is the relevant mechanism that explains the creep strengthening of these two composites and, hence, that $\Delta\sigma$ is related to the stress carried by the reinforcement.

To confirm the above conclusion, the Shear Lag model proposed by Ryu *et al* [11] will be used to calculate the load transferred to the reinforcement, σ_T . This model has been used previously to calculate σ_T in other composites, and has compared successfully with the experimental $\Delta\sigma$ which these composites undergo, as shown in [3,12]. A similar comparison can be conducted in the present case.

A detailed description of this model is in [11]. Accordingly, σ_T is calculated from the following equation,

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

where f and S_{eff} are the volume fraction and the effective aspect ratio of the reinforcing particles (the model takes into account the possible misalignment of elongated particles with the loading direction), respectively.

Equation (1) predicts that the stress borne by the reinforcement is linearly dependent on σ , with the term in parenthesis being the proportionality constant. The equation is valid, as mentioned, for the case of composites with elongated and misaligned particles. It is simplified in the case of composites with equiaxial particles, such as the present case. Furthermore, assuming that particles are equiaxial cylinders (aligned with the loading direction) or spherical particles, it would be $S_{eff} = 1$ and $S_{eff} = 1.25$, respectively [13].

The trends predicted by the model, equation (1), assuming cylinders or spheres are included in the plots of Figure 3 as dashed lines. A very reasonable correlation with the experimental values of $\Delta\sigma$ is obtained for both composites, considering the simplicity of the model and that no other contribution is taken into account. The model predicts slightly higher creep strength than the experimental data. Several factors can explain this slight discrepancy:

- 1.- It may be possible that damage or decohesion phenomena occurs at the metal-ceramic interface. These phenomena are likely to occur when composite preparation involves melting of the metallic matrix [5]. [Damage at the interface reduces the efficiency of the stress transfer mechanism with respect to the value predicted by the model, Equation 1, as shown in Figure 3.](#)
- 2.- The possible influence of the accelerated ageing process on $\Delta\sigma$ should also be taken into account. This process, which affects composites with age-hardening alloys [14] such as 2618Al, indicates that the precipitation state of the alloy and the composite matrix is not the same: The composite matrix microstructure weakens faster than the alloy under given testing condition.
- 3.- Some stress relaxation by diffusion could also be another factor which diminishes $\Delta\sigma$ [15].

Finally, a refinement of the microstructure in the composite matrix caused by the addition of the reinforcement could also affect $\Delta\sigma$ [3]. This contribution, contrary to the others above, would enhance the composites' matrix behavior, as explained in [3]. This factor is difficult to analyze since no data of grain size is reported in [1]. It is likely, however, that the dispersion of the alumina particles maintains a similar microstructure in the alloy and in the matrix of the composites. This contribution, therefore, can be omitted.

Despite all these considerations, the good correlation of experimental data and the model's predictions shown in Figure 3 is noteworthy, which indicates that these other factors are not significant.

Once the fraction of stress transferred to the TiC_p is known, it is possible to make a comparison between the composite matrix creep behavior and that of the 2618Al alloy by considering the effective stress, $\sigma_{\text{eff}} = \sigma - \sigma_T$, that the plastically deforming matrix undergoes during testing. As described in [3], the power law creep behavior which governs the steady or minimum creep rate of this behavior is,

$$\dot{\varepsilon} = A' \left(\frac{\sigma - \sigma_T}{E} \right)^n \exp(-Q_c/RT) \quad (2)$$

Where A' is a material's constant, E the Young's modulus, Q_c the activation energy for creep, R the universal gas constant ($R=8.314 \text{ kJ mol}^{-1} \text{ K}^{-1}$), and T the absolute temperature. This equation is similar to the common creep equation with the term σ_0 (see equation 1 in [1]), but replacing this term by σ_T . The different meanings of both terms and the different ways of calculating them should be emphasized. In the present view the actual stress which the metal matrix dislocations bear is $(\sigma - \sigma_T)$, whereas in the former view it is the applied stress, σ , the one that the dislocations undergo. Figure 4 shows the data of Figure 1 after σ_T correction. A double logarithmic plot of strain rate as a function of effective stress (to calculate σ_T , the most unfavorable case of $S_{\text{eff}} = 1.25$ has been used in the model, Equation 1) is shown in this figure. As expected from the prediction of Figure 3, a good correlation between the creep behavior of the 2618Al alloy and the composites matrix is obtained at each of the three temperatures, revealing the importance of the load partition process during creep of these composites.

As mentioned in previous work [3], the load transfer mechanism does not alter the stress exponent value. An explanation of why the stress exponent of these three materials is so high, Figure 1, resides in a threshold stress associated with the alumina particle dispersion: The *in situ* procedure used for the materials' preparation involved the compaction of blend powders of 2618Al, Ti, and C in a given proportion. Clearly, the alumina layer inherent to the initial 2618Al powder particles has given rise to a fine dispersion of alumina particles in both the composites and the alloy, as occurs in powder metallurgy aluminum alloys and aluminum alloy metal matrix composites [5]. This dispersion is, therefore, responsible for *true* threshold stress in these materials.

Such a term leads to a natural increase of the apparent stress exponent and also to an increase of the materials' creep strength, as described in [6] (see for example figure 2 in [6]).

Our view of the creep of MMCs and the relevance of the load partition phenomenon is very similar to that proposed by Langdon and co-workers [16,17]. A comparison of both visions is, however, beyond the scope of the present analysis and will be the subject of future work.

The influence of the ageing process on the stress exponent of this alloy is also worth mentioning. As shown in [6], the simultaneous ageing process occurring during creep in age hardening aluminum alloys causes a variation (decrease) in the apparent stress exponent because the Ostwald ripening or precipitate coarsening process differs from the low to the high applied stress test. It was proposed that the inter-particle distance, λ , increases with a decrease in σ , such that the stress exponent value, in the framework of the substructure invariant model [18], decreases from $n=8$ down to $n=6.5$ [6]. A change in the apparent stress exponent due to differences in precipitate coarsening with creep conditions has been also suggested in [7].

A comparison of creep data of *in situ* Ji *et al*'s 2618Al alloy and a similar alloy obtained by ingot metallurgy, IM, which does not bring alumina particles into the alloy, would make it possible to calculate σ_0 and, hence, its contribution to the creep strengthening of materials [6]. The absence of these data in the work of Ji *et al* [1], however, does not permit such a calculation to be made by the subtraction method [6]. This strengthening can, nevertheless, be roughly estimated by the "classical" procedure (extrapolation method). The reliability of the values obtained by this method has been criticized in the introductory part because of the stress exponent value used by Ji *et al* [1], not consistent with the occurrence of simultaneous coarsening of precipitates during creep deformation. To avoid this, a stress exponent value of $n=6.5$ will be considered for such a calculation. This n value is consistent with the expected value of n if the influence of the simultaneous precipitation process occurring during creep is taken into account [6]. On this basis, the $1/6.5$ power of $\dot{\epsilon}$ has been represented as a function of the effective stress $(\sigma-\sigma_f)$. As mentioned, this stress coincides with the applied stress for the case of the alloy since no load transfer occurs in this material. For the composites, however, $(\sigma-\sigma_f)$ is the "actual" stress that the aluminum alloy matrix undergoes during creep deformation [3]. The resulting data is shown in the plot of Figure 5. A very good

correlation for the alloy and the composites is obtained supporting the idea that the extrapolated value of σ at $\dot{\epsilon}^{1/6.5} = 0$, i.e., σ_0 , is common for the alloy and composite materials. It is also seen that σ_0 decreases with increasing testing temperature. At 623 K σ_0 is virtually null and at 573 and 523 K it is obtained, $\sigma_0=14$ MPa and $\sigma_0=87$ MPa, respectively.

Because of the absence in [1] of creep data of IM 2618Al alloy, the reliability of the above σ_0 values can be finally assessed if the experimental creep data of this *in situ* 2618Al alloy is compared with data of IM 2618Al alloy from other investigations. Bardi *et al* [19] have conducted torsion tests on 2618Al alloy to investigate its warm formability. Bardi *et al*'s alloy was processed by an IM procedure [20]. Therefore, the absence of alumina particles is guaranteed. In their work, Bardi *et al* report the converted torsion tests data into equivalent strain rate and stress such that the data can be directly compared with data from conventional uniaxial tests. A comparison of the creep data at 573 K from Ji *et al*'s *in situ* 2618Al alloy (with the presence of this alumina dispersion) and the IM 2618Al alloy studied by Bardi *et al* [19] is made in the plot of Figure 6. The apparent increasing stress exponent observed with increasing strain rate in the IM alloy is very likely to be related to a power law breakdown behavior. This behavior is usually observed in a region of high strain rate or applied stress [4]. Despite the different ranges of strain rates studied in both works [1,19] (experimental σ data under similar testing conditions only appears around 10^{-3}s^{-1}), the tendency indicates that Bardi *et al*'s alloy is, as expected, less creep resistant than Ji *et al*'s one. Furthermore, the strength difference at 10^{-3}s^{-1} is very close to the value of $\sigma_0=14$ MPa obtained above from the extrapolation method (assuming $n=6.5$). Consequently, this good accordance supports the idea that this alumina particles dispersion, through the associated threshold stress, is the principal mechanism responsible of the increased creep resistance of the *in situ* 2618 alloy with respect to the ingot metallurgy one, Figure 6, as demonstrated for other alloys [6].

In summary, a threshold stress associated with a dispersion of alumina particles arising from the original 2618Al alloy powder used in the *in situ* procedure, rather than the TiC_p , is responsible of the high stress exponent observed in these materials. The threshold stress also leads to an increase of their creep strength. The σ_0 term, common to the alloy and the composites, has been calculated from the extrapolation method

under the assumption that $n=6.5$ and after subtraction of the load transferred to the reinforcement. This threshold stress is not responsible for the better creep resistance of the composites with respect to the 2618Al alloy investigated by Ji *et al* in [1]. This improvement is only associated with a load partitioning phenomenon, as has been demonstrated.

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Figure captions

Figure 1.- Double logarithmic plots of strain rate vs. stress creep data of the 2618Al-15vol%TiC_P and 2618Al-20vol%TiC_P composites and the 2618Al alloy reported by Ji *et al* [1].

Figure 2.- Comparison between the apparent stress exponent, n , of the *in situ* 2618Al-15vol%TiC_P and 2618Al-20vol%TiC_P composites with that of the 2618Al alloy studied by Ji *et al* [1].

Figure 3.- Stress increment as a function of applied stress, σ , and prediction of Ryu *et al*'s Shear-Lag model [11] of the load transferred, σ_T , in each composite for the cases of: equiaxial cylinders, $S_{eff} = 1$, and spherical particles, $S_{eff} = 1.25$. a) 2618Al-15vol%TiC_P composite and b) 2618Al-20vol%TiC_P composite. A good correlation between the experimental strength increment and prediction is obtained.

Figure 4.- Double logarithmic plot of strain rate vs. effective stress, $\sigma_{eff} = \sigma - \sigma_T$, as predicted from Ryu *et al*'s model [11], of the 2618Al alloy and the 2618Al-15vol%TiC_P and 2618Al-20vol%TiC_P composites at the three temperatures investigated.

Figure 5.- Dependence of $\dot{\epsilon}^{1/6.5}$ with the effective stress, $\sigma - \sigma_T$, to calculate σ_0 by the extrapolation method. A common σ_0 is obtained for the alloy and the composites. A value of $n=6.5$ is considered to eliminate the effect of the simultaneous precipitation process and precipitate coarsening during creep deformation.

Figure 6.- Double logarithmic plots of strain rate vs. stress data at 573K of the *in situ* 2618Al and IM 2618Al alloys studied by Ji *et al* [1] and Bardi *et al* [19], respectively. A slightly better high temperature deformation behavior in the *in situ* 2618Al alloy is appreciated.

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Figure 1.- Double logarithmic plots of strain rate vs. stress creep data of the 2618Al-15vol%TiC_p and 2618Al-20vol%TiC_p composites and the 2618Al alloy reported by Ji *et al* [1].

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Figure 6.- Double logarithmic plot of strain rate vs. stress data at 573K of the *in situ* 2618Al and IM 2618Al alloys studied by Ji *et al* [1] and Bardi *et al* [19]. A slightly better high temperature deformation behavior in the *in situ* 2618Al alloy is appreciated.