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Title: Understanding the creep fracture behavior of aluminum alloys and aluminum alloy metal matrix composites

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Abstract: In a previous study on the creep fracture behavior of aluminum alloy metal matrix composites and their respective un-reinforced alloys a new form of the phenomenological Monkman-Grant, MG, relationship was proposed as follows:

where t_f is time to failure, $\dot{\epsilon}_0$ the minimum or steady strain rate, $f(n')$ a scalar function of the Monkman-Grant exponent, n' , and ϵ_f the strain to failure. This equation is, in fact, very similar to the Dobeš-Milička, DM, one. In the new equation, however, the exponent n' of the minimum strain rate term, $\dot{\epsilon}_0$, is eliminated ($n'=1$). In this way, the creep rupture behavior of materials can be deepened by analyzing the physical meaning of a new unit-less parameter, ϵ_2/ϵ_f . It is readily seen that this parameter equates ϵ_2/ϵ_f , where ϵ_2 would be the rupture strain accumulated during the secondary creep regime if the entire creep test progressed under the steady state period and ϵ_f is the strain to failure. So, ϵ_2/ϵ_f quantifies the relative importance of secondary creep strain with respect to that of primary and the tertiary creep stages. For the case of the Al alloys significant variations of ϵ_2/ϵ_f with experimental creep conditions (applied stress) are found when the grain size and/or aspect ratio is large. On the contrary, ϵ_2/ϵ_f is nearly stress independent for fine microstructures. The additional influence of the metal-ceramic interface in metal matrix composites leads to a more complex evolution of this term.

Understanding the creep fracture behavior of aluminum alloys and aluminum alloy metal matrix composites

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Abstract

In a previous study on the creep fracture behavior of aluminum alloy metal matrix composites and their respective un-reinforced alloys a new form of the phenomenological Monkman–Grant, MG, relationship was proposed as follows:

$$t_f \dot{\epsilon}_{min} = \frac{f(n')}{(\dot{\epsilon}_{min}^{n'-1})} \epsilon_f$$

where t_f is time to failure, $\dot{\epsilon}_{min}$ the minimum or steady strain rate, $f(n')$ a scalar function of the Monkman–Grant exponent, n' , and ϵ_f the strain to failure. This equation is, in fact, very similar to the Dobeš-Milička, DM, one. In the new equation, however, the exponent n' of the minimum strain rate term, $\dot{\epsilon}_{min}$, is equal to one. In this way, the creep rupture behavior of materials can be deepened by analyzing the physical meaning of a new unit-less parameter, $K = \frac{f(n')}{(\dot{\epsilon}_{min}^{n'-1})}$. It is readily seen that this parameter equates

$\frac{\epsilon_2}{\epsilon_f}$, where ϵ_2 would be the rupture strain accumulated during the secondary creep

regime if the entire creep test progressed under the steady state period and ϵ_f is the strain to failure. So, K quantifies the relative importance of secondary creep strain with respect to that of primary and the tertiary creep stages. For the case of the Al alloys significant variations of K with experimental creep conditions (applied stress) are found when the grain size and/or aspect ratio is large. On the contrary, K is nearly stress independent for fine microstructures. The additional influence of the metal-ceramic interface in metal matrix composites leads to a more complex evolution of this term.

Key words: Aluminum Alloys; Metal matrix composites (MMCs); Creep failure; Powder processing; Casting.

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Introduction

In previous work [1], creep rupture data reported by these authors as well as data recorded from published work on discontinuously reinforced metal matrix composites and un-reinforced Al alloys were analyzed. The analysis, conducted in the framework of the Monkman-Grant, MG, equation [2], was done with two objectives: to understand better the specific role played by the ceramic particles in the creep rupture behavior of MMCs and to deepen the significance of the MG equation. This equation reads:

$$t_f \dot{\varepsilon}_{min}^{n'} = C \quad (1)$$

where t_f is the time for creep rupture, $\dot{\varepsilon}_{min}$ the minimum or steady state strain rate, and n' and C are constants. n' is called the MG exponent, which is close to but different from unity, and C is known as the Monkman-Grant constant. The latter one would represent the total elongation to failure in the case that $n'=1$ and $\dot{\varepsilon}_{min}$ dominate creep deformation. The importance of this equation depends on the fact that it is obeyed by most engineering materials and in its potentiality to predict, from data obtained in laboratory creep tests, rupture life times of structures designed to operate over very long time periods. Despite this important technological use, however, the underlying phenomena responsible for the trend dictated by this equation are not well known. It is agreed that “the creep deformation is the macroscopic manifestation of the damage accumulated during creep deformation” [3]. This implication is clear when $n'=1$ because only in this case, does C have a physical meaning: This term indicates a strain, which is associated with $\dot{\varepsilon}_{min}$. This meaning vanishes when $n' \neq 1$ since C is no longer a strain. In fact, the efforts to explain the MG equation on a microstructural basis have been made under the approach that $n'=1$ [4,5,6]. It was shown in [1], however, that this condition is virtually never obeyed. Furthermore, whereas this is a good approximation for relatively high values of $\dot{\varepsilon}_{min}$, the resulting value of $\dot{\varepsilon}_{min}^{m'}$ can be significantly different when $\dot{\varepsilon}_{min}$ is low (below 10^{-5} s^{-1}) for a typical value of $m'=0.85$. Therefore, further efforts to find a rational meaning of the equation in its general form are justified. In [1], a new form of equation (1) was proposed to deepen the significance of the correlation among creep rupture data. Simple data reorganization, under the assumption that constant C is $C = f(n')\varepsilon_f$ (following Krasowsky [7] for power law type of equations, where ε_f is the strain to failure), leads readily to the following relationship:

$$t_f \dot{\varepsilon}_{min} = \frac{f(n')}{(\dot{\varepsilon}_{min}^{n'-1})} \varepsilon_f \quad \text{or} \quad t_f \dot{\varepsilon}_{min} = K \varepsilon_f \quad (2)$$

In this form, the new strain rate exponent in the equation is equal to unity allowing a new *modus operandi* to understand better the meaning of the dependence among creep rupture parameters.

Equation (2) is also very similar to the equation proposed by Dobeš and Milička, DM, in a work [8] published 20 years after that of Monkman and Grant [2]. In their equation DM include ε_f as an additional term in a predictive equation which correlates creep rupture data parameters. In this case, the equation, which is in fact a refinement of the MG one, is expressed as:

$$t_f \dot{\varepsilon}_{min}^{m'} = C' \varepsilon_f \quad (3)$$

where m' and C' are new constants. A step forward of the DM equation with respect to the MG one relies in the fact that the deviation of experimental data from a common trend is less pronounced than that given by the MG equation (see figures 3 and 4 in [8]). On the other hand, the technological interest of the DM equation is somewhat lost since elongation to failure is required to “predict” time to failure, *i.e.*, one has to conduct tests to failure to apply the equation. Despite this inconvenience, however, it is of a great scientific interest to deepen the significance of this dependence and the influence of materials microstructure. As in the MG equation, the fact that $m' \neq 1$ (although close) also complicates the understanding of its physical meaning and the underlying microstructural basis that lead to its predictive capacity.

Several other modifications of the MG equation have been proposed elsewhere [9,10,11,12]. These studies concern very different alloys, ranging from nickel base superalloys [9], Zircalloys and different steels and alloys [10,11], and intermetallics [12]. Their data are not analyzed in this work as they do not include Al alloys and MMCs. A discussion on these other modifications of the MG equation is, hence, beyond the scope of this research.

As mentioned above, an analysis of the data reported from the literature, besides the authors' own data, was made in an attempt to deepen the underlying meaning of the MG equation as well as to understand better the role played by the ceramic particles in the creep rupture behavior of these MMCs [1]. In the present research, further analysis based on this form of the DM equation is made and discussed. Since the strain rate exponent of the new equation is equal to one, additional thinking to understand the

creep rupture mechanisms responsible of the dependence among creep rupture parameters can be further attempted. For this purpose, analytical derivations from the MG equation to equations (2) and (3) are developed and the required conditions to comply with the parameter involved in the DM equation are imposed. In this framework, the new term of equation (2), $\left(\frac{f(n)}{\dot{\varepsilon}_{min}^{n'-1}}\right) = K$, is discussed.

Materials and experimental part

The materials investigated are 6061Al alloys (obtained by ingot metallurgy, IM, and powder metallurgy, PM, procedures, as described in [13]) and 6061Al-15vol%SiC and 6061Al-40vol%SiC composites of previous work by these authors [1,3,13,14]. Their microstructure, the powder metallurgical procedure to obtain them, and the experimental details to conduct the creep tests are already described in [1, 13, 14, 15]. Here, the additional information recorded from the creep tests, namely, ε_f , is also reported and included in the analysis.

Besides the experimental results obtained from the above un-reinforced alloy and composites, data collected from the open literature [2,16, 17, 18, 19, 20, 21, 22, 23,24,25,26] have also been analyzed. A summary of such literature creep tests and results will be supplied in this investigation. In these studies tests were conducted either at constant stress or constant load. This difference leads inevitably to a certain scatter of results since time to rupture at constant load should decrease with respect tests carried out at constant stress.

A microstructural study of crept samples has been conducted employing conventional metallographical procedures and optical microscopy.

Results and data analysis

The results obtained from the creep experiments conducted on the four materials investigated are summarized in Table I. In this table, the time to failure, t_f , and ε_f data at the different testing conditions are summarized. As mentioned before, data reported from the literature on aluminum alloys and aluminum alloy MMCs, whenever information on ε_f is reported, have been also analyzed, Table II. So, the fitting parameters resulting from equations (1) and (3) are summarized in this table. For the case of the DM analysis, equation (3), only materials which include ε_f data have been used to obtain m' and C' parameters.

The better correlation of creep rupture data provided by the DM equation with respect that of MG, as demonstrated in [8], is also supported by the data of the aluminum alloys and aluminum alloy MMCs listed in Table II. Both correlations can be appreciated in the plots of Figure 1. As can be seen, creep rupture data of all the Al alloys faithfully follow the MG relation, Figure 1a). It is possible that the scatter is partially due to the fact that these results were obtained from test conducted, both, at constant load and constant stress. When the ε_f correction is made (DM approach), however, a very close correlation among data of the eight alloys is obtained, Figure 1b. Similarly, the creep rupture data of the aluminum alloy matrix MMCs also obey the MG relation, as revealed by the plot of Fig. 1c). All data fits follow, roughly, a similar pattern of behavior. Again, this common behavior is substantially improved by the DM equation, Fig. 1d).

In summary, the correlation between $\dot{\varepsilon}_{min}$ and t_f is better described if t_f is normalized by ε_f , Equation (3), as suggested in [8]. It is also remarkable that, contrary to the MG plots, both, alloys and composites, follow virtually the same common trend. As discussed in [8], this dependence indicates the close connection between the deformation mechanism and the process which leads to crack and cavity formation and final rupture of the material. Only data of 6061Al-15vol%SiC of the present work is slightly more scattered than the data from the literature, but the common trend dictated by all materials is also obeyed by this composite.

The correlation proposed in [7] between constants involved in power law type equations (such as the MG and DM ones) prompted us to develop equation (2) in an attempt to deepen creep rupture behavior of un-reinforced Al alloys and Al alloy MMCs [1]. A connection between C and n' is, indeed, found when experimental data reported from several investigations (see Table II) are plotted in a $\ln C$ vs. n' graph, as shown in Figure 2. As can be seen, a close linear correlation between $\ln C$ and n' , is obtained indicating that C can be expressed as a function of n' . It is worth noting that the values of the n' parameter are much more grouped for the alloys than for the composites. In the former ones, n' extends towards values which are significantly lower than those for the alloys. The equation relating C and n' in the plot of figure 2 is an exponential relation, *i.e.*, of the form $C = Ae^{-Bn'}$. This exponential relation should account for the specific dependence of $C = f(n')\varepsilon_f$ proposed in [7].

The analysis conducted in the light of Figure 2 between MG parameters can also be attempted with the DM ones for materials in which ε_f data are reported in the literature, Table II. This analysis is shown in the plot of Figure 3, in which the natural logarithm of the DM constant, C' in equation (3), is, again, represented as a function of the corresponding exponent m' for the composites and un-reinforced alloys of Table II. A linear correlation between $\ln C'$ and m' is, once more, obtained, *i.e.*,

$$C' = A'e^{-B'm'} \quad (4)$$

In this case, however, two essential differences between this fit and that of Figure 2 are to be noted: First, whereas all the n' data of Figure 2 are virtually lower than unity, the m' data in Figure 3 are more grouped around unity (and the range of variation is also lower). And second, the modification introduced by the DM equation (data displacement with respect to figure 2 is such that the fit provided by the new data constants nearly intersects the value $\ln C'=0$ and $m'=1$, designated by the dark cross in the plot of Figure 3.

Discussion

It is remarkable the improved fit provided by the DM equation (the strain to failure value is considered) with respect that of MG, Figure 1, and, in particular, the fact that the new constants m' and C' are further grouped around the value $\ln C'=0$ and $m'=1$, the dark cross in the plot of Figure 3. As mentioned, this value represents the obvious dependence $t_f \dot{\varepsilon}_{min} = \varepsilon_f$ among creep rupture parameters for the case in which the entire creep curve progresses under the secondary creep regime. Therefore, the still divergence of data around this value invites us to think that the non-stationary creep regimes (in particular the tertiary one) are of major importance in the prediction of creep rupture parameters from the DM model. Furthermore, the fact that these data (C' and m') obey the exponential equation (4) suggests that an interdependence between the secondary and tertiary creep regimes parameters exists, in a way that reminds of Wilshire's Θ projection model [27].

A step forward in the interpretation of the DM equation can be made if terms of the MG equation are reorganized such that the minimum strain rate exponent is equated to one. As a result of this reorganization, equation (2) is obtained. In other words, the purpose of equation (2) is to find a rational interpretation of these constants and the role of the non stationary creep regimes in the prediction of creep rupture parameters in aluminum

alloys and aluminum alloy MMCs. Since a correlation between MG and DM parameters, respectively, is obtained (as described by the plots of figures 2 and 3), their interpretation in terms of microstructural factors is somewhat simplified.

A very similar examination has been made previously by Dunand *et al.* [6]. These authors, however, do not reorganize the MG equation to derive a minimum strain rate exponent of one. They directly “assume” $m'=1$ in the equation. As above mentioned, this is a good approximation when $\dot{\epsilon}_{min}$ is high, but, for typical values of m' (e.g., $m'=0.85$), it can lead to considerably different values of $\dot{\epsilon}_{min}^{m'}$ when $\dot{\epsilon}_{min}$ is low.

As mentioned in [1], the term K in this equation “quantifies the relative importance of secondary creep strain with respect to that accumulated during the primary and, particularly, the tertiary creep stages”. Specifically, a physical meaning of this factor is found from the analysis of complete creep curves, like that of figure 4. In this figure, a creep curve of 6061Al PM alloy tested at 623 K, under 29 MPa, is shown. The strain terms ϵ_2 and ϵ_f are visualized in this plot. The magnitude of ϵ_2 would be identified with the strain accumulated during the secondary creep regime if the entire creep test had progressed under the steady state period. From these plots it is seen that, $\epsilon_2 = t_f \dot{\epsilon}_{min}$.

Therefore, K is readily identified with $\frac{\epsilon_2}{\epsilon_f}$, giving, hence, a physical meaning to this

term. Dunand *et al.* [6] also arrive to this conclusion (their equation 12), but, again, they make an analysis under the assumption that $m'=1$.

Once a physical meaning of K has been found, it would be now helpful understanding its evolution with the testing conditions. This evolution is obtained by plotting this term as a function of $\dot{\epsilon}_{min}$ from known C' and m' values (Table II). This exercise can be

done since it is also found that $K = \frac{f(n)}{(\dot{\epsilon}_{min}^{n'-1})} = \frac{\epsilon_2}{\epsilon_f} = \frac{C'}{\dot{\epsilon}_{min}^{m'-1}}$. In principle, only one

constant would be necessary for this analysis since C' and m' are interdependent parameters, as shown by the plot of figure 3 and equation (4). In this first report,

however, the tabulated values of Table II will be used. The dependence of K with $\dot{\epsilon}_{min}$ obtained for the different alloys of Table II is seen in the plot of figure 5. As can be seen, K increases with $\dot{\epsilon}_{min}$, although the variation (slope) can differ significantly: It can increase rapidly, such is the case of the 6061Al alloys of the present work, or the

2124Al alloy of ref [16], or be nearly flat, as for example the ALCO alloy of ref. [20]. For the case of the Al alloy MMCs, the variation of K with $\dot{\epsilon}_{min}$, Figure 6, is more complex: a decrease of K with $\dot{\epsilon}_{min}$, besides the tendencies observed in figure 5 is also appreciated in some materials.

The onset of tertiary creep has been associated with a combination of necking and cavitation phenomena by some authors [28]. Dobeš [29] also reported the importance of void formation in tertiary creep. A more detailed insight into the rupture phenomena in high temperature creep of aluminum alloys suggests that strain softening can also play an important role [30]. Other authors, however, consider only the localization of deformation or necking for triggering this stage [21]. For the case of MMCs, necking is not usually observed but cavity formation at the metal-ceramic interface is considered as the controlling mechanism [21]. In general, a combination of mechanisms can be operative during high temperature deformation. Here, cavity formation was, in fact, observed near the fracture region of alloy samples of the present investigation, as shown in Figure 7. In this figure, cavity formation near the fracture region of the 6061Al PM alloy (under the creep conditions specified in Figure 4) can be seen. As shown, material accumulates damage preferentially in a heterogeneous form (see regions inside the ovals). Therefore, and taking into account the absence of necking in MMCs, cavity accumulation/coalescence was also considered as the main responsible mechanism of fracture occurrence for materials tabulated in Table II.

Based on the above assumption, a connection of K with the microstructure and damage and cavity nucleation mechanisms is proposed for the case of the un-reinforced Al alloys, figure 5. There are several cavity nucleation mechanisms well described in the literature, see *e.g.*, [31]. It is claimed that cavities often nucleate on grain boundaries, and that those nucleated on boundaries transversely oriented with respect the tensile axes direction are vital for fracture occurrence. During secondary creep, cavities nucleate and grow in a stationary condition. Growth mechanism is related to vacancy diffusion at elevated temperatures and to plasticity at lower temperatures [31]. In the frame of MG type models, coalescence of cavities occurs as deformation progresses, leading to final failure when a certain limit is achieved. Then, it is possible that the onset of the tertiary creep occurs when accumulation/coalescence of cavities exceeds a certain value, which depends on the testing conditions and the material's microstructure. In this manner, the capacity of strain accumulation during coalescence of cavities can

also influence K : A high capacity for accumulation/coalescence of cavities (and extended tertiary creep) means large differences between ε_f and ε_2 ; *i.e.*, $K \ll 1$. On the contrary, low capacity for cavity accumulation/coalescence (and limited tertiary creep) means similar values of ε_f and ε_2 ; *i.e.*, $K \sim 1$.

Going further, an explanation of the tendencies exhibited in Figure 5, can be found in the grain size and its aspect ratio (and probably also the texture), as suggested in [31]. So, a correlation between the microstructure and the tendencies of the plot of figure 5 can be envisaged from the accompanying micrographs. Micrographs of 2124Al, 5083Al, and ALCO alloys have been taken from refs. [16], [18], and [20], respectively, and those of the 6061Al PM and 6061 IM alloys are from the present work. From these micrographs, it can be inferred that strain accumulation during coalescence of cavities in materials with large grain size and/or aspect ratio should be highly influenced by testing conditions. Thus, cavity coalescence at grain boundaries reduces the effective area that the applied load bears. This reduction is crucial for the final fracture occurrence and, hence, for the amount of strain accumulated during tertiary creep (and for the magnitude of K): In large grain size and/or aspect ratio materials, cavities coalesce more heterogeneously than in small size materials. Then, the grain boundaries (transverse) that have accumulated cavities will nucleate fracture the earlier the higher the applied stress (or the resulting $\dot{\varepsilon}_{min}$). Only if the applied load (or $\dot{\varepsilon}_{min}$) is low, the transverse boundaries will be able to sustain the stress during the tertiary creep stage. In this manner, the possibility to accumulate strain during the tertiary regime increases, leading to a low K value. Contrarily, if the applied load is high, boundaries are not able to sustain the stress and limited tertiary creep strain and a high K value is obtained. This would explain the large variation of K observed in some alloys, Figure 5.

On the contrary, the reduction in effective area due to cavity coalescence in materials with a fine and homogeneous microstructure may be expected to be much more limited. Stress concentration is, in this case, absent due to the more homogeneous distribution of cavity formation. Hence, little influence of applied stress on K is expected, leading to “flatter” tendencies with $\dot{\varepsilon}_{min}$.

A correlation of K with the microstructure and damage mechanism for the case of MMCs is a more complex task. Although these composites reveal tendencies similar to those of the un-reinforced alloys, Figure 3, other factors, very likely related to the metal-reinforcement interface, should be taken into account in the fracture process to

explain (and predict) the decrease of K with $\dot{\varepsilon}_{min}$ observed in some cases, Figure 6. The nature of this interface is strongly dependent on the processing conditions [14]. It must affect, hence, not only the creep behavior [14] but also cavity formation and coalescence during creep.

Further work is, hence, needed to establish the specific correlation between K and the microstructure and deformation/damage mechanism.

Conclusions

The rupture creep behavior of aluminum alloys and aluminum alloy metal matrix composites has been studied on the basis of authors' own data and the results recorded from published investigations. The analysis has been made in the frame of the Monkman-Grant, MG, equation and, mainly, of the Dobeš-Milička, DM, one. As a result, a new form of the DM equation, relating minimum creep rate and time to rupture,

is proposed as follows, $t_f \dot{\varepsilon}_{min} = \left(\frac{f(n')}{(\dot{\varepsilon}_{min}^{n'-1})} \right) \varepsilon_f$, where $\left(\frac{f(n')}{(\dot{\varepsilon}_{min}^{n'-1})} \right) = \frac{C'}{\dot{\varepsilon}_{min}^{m'-1}}$ and the strain rate

exponent in this new equation is eliminated.

The following are the main conclusions of this research:

1.- A good linear correlation between $\ln C$ and n' (MG parameters) for the materials investigated has been found. However, data of the Al alloys are more grouped than the data of the MMCs: The range of variation of n' is 0.64-1.00 and 0.33-1.11 for the Al alloys and Al alloy MMCs, respectively.

2.- Similarly, a good linear correlation between $\ln C'$ and m' (DM parameters) is also found for all materials. In this case, the DM exponents are much closer to unity than the MG ones. This indicates that the introduction of the strain to failure term, ε_f , rationalizes this equation so that it approaches the physically understandable case of $m'=1$.

3.- By eliminating the minimum strain rate exponent from the DM equation, a new unit-

less term, $\left(\frac{f(n')}{(\dot{\varepsilon}_{min}^{n'-1})} \right) = \frac{C'}{\dot{\varepsilon}_{min}^{m'-1}} = \frac{\varepsilon_2}{\varepsilon_f}$ (where ε_2 represents the strain accumulated during the

secondary creep regime, if the entire creep test had progressed under the steady state

period), appears in this new form of the DM equation (equation (2)). The ratio $\frac{\varepsilon_2}{\varepsilon_f} = K$

ranges from “flat” tendencies to dependencies strongly influenced by $\dot{\varepsilon}_{min}$ (testing conditions). This is related to the capacity of the material to accommodate cavity coalescence during the tertiary creep stage.

4.- It is suggested that the evolution of K revealed for Al alloy, figure 5, depends on the microstructure: In materials with non-homogeneous distribution of damage, as for example those with large grain size and/or aspect ratio, significant variations of K with experimental creep conditions would be expected. These variations must be smaller for finer microstructures. For the case of Al alloy MMCs, also the interface associated to the reinforcing particles plays a significant role. This results in more complex evolutions of this term with $\dot{\varepsilon}_{min}$.

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List of figures

Figure 1. Double logarithmic plots of time to failure as a function of minimum strain rate (MG plots) and time to failure divided by true strain to failure as a function of minimum strain rate (DM plots) for un-reinforced Al alloys (a and b) and Al alloy matrix composites (c and d). Data are from materials of the present research and from the research works of Table II.

Figure 2. Natural logarithm of C as a function of the MG exponent, n' , for materials summarized in Table II. Both, Al alloys (open symbols) and Al alloy MMCs (black symbols) are included. Numbers denote the corresponding reference number. The good linear correlation between $\ln C$ and n' is evident.

Figure 3. Natural logarithm of C' as a function of the DM exponent, m' , for materials summarized in Table II (both, Al alloys and Al alloy metal matrix composite are included). Open symbols are alloys and black symbols are composite materials. Numbers denote the corresponding reference number. The fit obtained between the natural logarithm of C and n' , Figure 2, is also included for comparison.

Figure 4. Creep curve of the 6061Al PM alloy tested at 623K and 29 MPa. The strain term ε_2 , and the strain to failure, ε_f , are represented.

Figure 5. Dependence of K with $\dot{\varepsilon}_{min}$ for the un-reinforced Al alloys of table II for which data of m' and C' are reported. Also, the microstructure of some of the alloys is shown, revealing its correlation with the K dependence with testing conditions (micrographs of 2124Al, 5083Al, and ALCO alloys have been taken from refs. [16], [18], and [20], respectively).

Figure 6. Dependence of K with $\dot{\varepsilon}_{min}$ for the Al alloy MMCs of table II for which data of m' and C' are reported. The variation is more erratic than for the alloys, Figure 5, accounting for the influence of the metal-reinforcement interface.

Figure 7. Cavity formation/distribution close to the rupture surface in 6061Al PM tested at 623K and a) 29MPa and b) 25MPa. Ovals delimit damage concentration zones.

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a)

b)

Figure 1. Double logarithmic plots of time to failure as a function of minimum strain rate (MG plots) and time to failure divided by true strain to failure as a function of minimum strain rate (DM plots) for un-reinforced Al alloys (a and b) and Al alloy matrix composites (c and d). Data are from materials of the present research and from the research works of Table II.

Figure 2. Natural logarithm of C as a function of the MG exponent, n' , for materials summarized in Table II. Both, Al alloys (open symbols) and Al alloy MMCs (black symbols) are included. Numbers denote the corresponding reference number. The good linear correlation between $\text{Ln } C$ and n' is evident.

Figure 3. Natural logarithm of C' as a function of the DM exponent, m' , for materials summarized in Table II (both, Al alloys and Al alloy metal matrix composite are included). Open symbols are alloys and black symbols are composite materials. Numbers denote the corresponding reference number. The fit obtained between the natural logarithm of C and n' , Figure 2, is also included for comparison.

Figure 4. Creep curve of the 6061Al PM alloy tested at 623K and 29 MPa. The strain term ϵ_2 , and the strain to failure, ϵ_f , are represented.

Figure 5. Dependence of K with $\dot{\epsilon}_{min}$ for the un-reinforced Al alloys of table II for which data of m' and C' are reported. Also, the microstructure of some of the alloys is shown, revealing its correlation with the K dependence with testing conditions (micrographs of 2124Al, 5083Al, and ALCO

alloys have been taken from refs. **[Error! Bookmark not defined.]**, **[Error! Bookmark not defined.]**, and **[Error! Bookmark not defined.]**, respectively).

Figure 6. Dependence of K with $\dot{\epsilon}_{min}$ for the Al alloy MMCs of table II for which data of m' and C' are reported. The variation is more erratic than for the alloys, Figure 5, accounting for the influence of the metal-reinforcement interface.

Figure 7. Cavity formation/distribution close to the rupture surface in 6061Al PM tested at 623K and a) 29MPa and b) 25MPa. Ovals delimit damage concentration zones.

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alloys have been taken from refs. **[Error! Bookmark not defined.]**, **[Error! Bookmark not defined.]**, and **[Error! Bookmark not defined.]**, respectively).

Figure 6. Dependence of K with $\dot{\epsilon}_{min}$ for the Al alloy MMCs of table II for which data of m' and C' are reported. The variation is more erratic than for the alloys, Figure 5, accounting for the influence of the metal-reinforcement interface.

Figure 7. Cavity formation/distribution close to the rupture surface in 6061Al PM tested at 623K and a) 29MPa and b) 25MPa. Ovals delimit damage concentration zones.

Temperature (K)	Appl. stress (MPa)	Time to failure (s)	Strain to failure	Minimum creep rate (s^{-1})
523	40	7.2×10^5	0.016	3.5×10^{-8}
523	48	1.6×10^5	0.20	3.0×10^{-7}
523	49	9.0×10^4	0.075	3.5×10^{-7}
523	49	2.4×10^5	0.20	9.0×10^{-8}
523	53	9.8×10^3	0.22	1.0×10^{-5}
523	61	5.6×10^2	0.22	2.8×10^{-4}
573	28	2.1×10^6	0.18	1.8×10^{-8}
573	31	1.5×10^5	0.21	5.9×10^{-7}
573	37	1.2×10^4	0.18	7.0×10^{-6}
573	40	2.1×10^4	0.21	3.0×10^{-6}
573	42	3.0×10^3	0.25	5.6×10^{-6}
573	47	9.0×10^2	0.18	1.0×10^{-4}
623	23	1.1×10^5	0.19	2.5×10^{-7}
623	29	1.9×10^4	0.20	7.8×10^{-6}
623	29	3.7×10^4	0.24	2.3×10^{-6}
623	39	7.7×10^2	0.25	2.6×10^{-4}

6061Al PM alloy

Temperature (K)	Appl. stress (MPa)	Time to failure (s)	Strain to failure	Minimum creep rate (s^{-1})
623	23	2.0×10^3	0.45	2.1×10^{-4}
673	8	2.7×10^6	0.08	9.5×10^{-9}
673	5	7.8×10^5	0.13	1.2×10^{-7}
723	6	4.1×10^5	0.10	6.0×10^{-8}
723	9	2.5×10^4	0.45	4.8×10^{-6}
723	12	5.6×10^3	0.45	4.5×10^{-5}

6061Al IM alloy

Temperature (K)	Appl. stress (MPa)	Time to failure (s)	Strain to failure	Minimum creep rate (s^{-1})
623	29	9.3×10^5	0.004	5.0×10^{-8}
623	35	2.4×10^6	0.0055	1.1×10^{-8}
623	35	1.6×10^4	0.003	1.0×10^{-7}
623	37	6.2×10^4	0.05	2.0×10^{-7}
623	50	2.7×10^2	0.18	3.0×10^{-4}
673	17	2.1×10^6	0.002	4.7×10^{-10}
673	23	3.7×10^5	0.004	5.8×10^{-8}
673	23	3.5×10^5	0.004	1.0×10^{-9}
673	23	7.6×10^4	0.016	6.7×10^{-8}
673	30	3.3×10^4	0.025	3.0×10^{-7}
673	35	1.1×10^3	0.12	3.3×10^{-5}
673	43	3.7×10^2	0.18	2.8×10^{-4}
723	12	1.8×10^5	0.007	8.0×10^{-9}
723	15	1.8×10^5	0.005	7.3×10^{-9}
723	15	4.2×10^4	0.02	1.0×10^{-7}
723	17	6.5×10^3	0.0033	1.7×10^{-7}
723	20	1.9×10^3	0.05	1.4×10^{-5}
723	30	4.2×10^2	0.02	1.8×10^{-4}
773	5	1.8×10^5	0.018	5.8×10^{-8}
773	6	7.1×10^4	0.03	3.0×10^{-7}
773	8	8.5×10^3	0.04	1.5×10^{-6}
773	8	1.4×10^5	0.027	7.0×10^{-7}
773	13	1.9×10^3	0.17	5.0×10^{-5}

6061Al-15vol%SiC composite

Temperature (K)	Appl. stress (MPa)	Time to failure (s)	Strain to failure	Minimum creep rate (s^{-1})
673	60	5300	0.012	1.4×10^{-6}
673	66	2600	0.013	3.0×10^{-6}
673	73	500	0.018	3.5×10^{-5}
723	26	5500	0.017	1.6×10^{-6}
723	30	1171	0.028	1.3×10^{-5}
723	43	180	0.026	1.0×10^{-4}

6061Al-40vol%SiC composite

Table I.- Creep conditions and creep fracture parameters for the 6061Al PM and 6061Al IM alloys and the 6061Al-15vol%SiC and 6061Al-40vol%SiC composites.

Authors [ref.]	Material	Processing	S. rate, s ⁻¹	Temp, K	$n'//C$ (MG)	$m'//C'$ (DM)	G. size (μm) //A. ratio	Reinf. Size//A. ratio
Present research ^a (constant stress)	6061Al	PM	$2.8 \times 10^{-4} - 9.7 \times 10^{-10}$	573- 723	0.80//0.90	0.89//1.53	3.5//2	-
	6061Al	IM	$2.1 \times 10^{-4} - 9.5 \times 10^{-9}$	623- 723	0.72//4.12	0.92//1.29	3.0//2	-
	6061Al-15vol%SiC	PM	$3.0 \times 10^{-4} - 1.4 \times 10^{-6}$	623- 773	0.67//1.16	1.00/0.34	1.5//4	2//2.0
	6061Al-40vol%SiC	PM	$3.5 \times 10^{-5} - 4.7 \times 10^{-10}$	673- 723	0.77//0.16	0.92//1.60	-	-//1
Monkman, Grant [Error! Bookmark not defined.] -	Al, 2S,3S and Zn, Mg and Cu Solid solutions of Al	IM	$3.5 \times 10^{-3} - 4.5 \times 10^{-8}$	288- 593	0.85//1.3	-	-	-
Dunand <i>et al.</i> [Error! Bookmark not defined.] (constant load)	Al99,9%-25% Al ₂ O ₃ p Corase Grain (CG)	IM	$2.3 \times 10^{-4} - 5.9 \times 10^{-9}$	608- 723	0.84//0.13	0.99//0.51	>1000/-	0.28//1
	Al99,9%-25% Al ₂ O ₃ p Fine Grain (FG)	PM	$2.7 \times 10^{-4} - 1.3 \times 10^{-8}$	608- 723	0.93//0.05	1.02//0.42	1.3/-	0.28//1
Taminger <i>et al.</i> [Error! Bookmark not defined.] (constant stress)	2124Al	PM	$2.5 \times 10^{-5} - 1.8 \times 10^{-7}$	394- 539	0.73//1.40	0.95/0.24	250//4.7	-
	2124Al-15% Al ₂ O ₃ w	PM	$1.5 \times 10^{-5} - 2.0 \times 10^{-8}$	367- 541	0.91//0.48	0.88//0.75	77//2.6	(8.2±2.4)//6.3
Pandey <i>et al.</i> [Error! Bookmark not defined.] (constant load)	Al-10% SiCp	PM	$1.3 \times 10^{-6} - 3.0 \times 10^{-9}$	623	0.74//0.85	1.10//0.05	5//-	1.7//1
	Al-20% SiCp	PM	$8.5 \times 10^{-7} - 7.5 \times 10^{-10}$	623	0.77//0.16	0.90/2.44	5//-	1.7//1
	Al-30% SiCp	PM	$1.7 \times 10^{-6} - 2.2 \times 10^{-8}$	623	0.85//0.04	0.96//1.13	5//-	1.7//1
Ishikawa <i>et al.</i> [Error! Bookmark not defined.] ^b (constant stress)	5083Al	IM	$8 \times 10^{-5} - 4 \times 10^{-8}$	623- 773	0.83//0.606	0.83//1.68	100/-	-
Ma <i>et al.</i> [Error! Bookmark not defined.] (constant load)	Al-TiB ₂ p	Reactive hot press. (not PM)	$1.3 \times 10^{-4} - 9.9 \times 10^{-8}$	573-673	0.91//0.04	1.03//0.27	-	(0.5-5)//1
Rosler <i>et al.</i> [Error! Bookmark not defined.] ^c (constant load)	ALCO (Al-O)	PM	$7.4 \times 10^{-4} - 4.4 \times 10^{-10}$	573- 773	0.74//0.20	0.96//0.32	1.9/-	-
	ALCOR		$1.7 \times 10^{-4} - 3.6 \times 10^{-9}$	773	0.64//0.54	0.93//0.80	330/-	-
Requena <i>et al.</i> [Error!]	6061Al 6061Al-22vol% Al ₂ O ₃ p	Stir casting (not PM)	$1.1 \times 10^{-9} - 2.5 \times 10^{-7}$ $5.8 \times 10^{-9} - 2.2 \times 10^{-6}$	573	0.85//0.08 1.11//0.003	0.70//7.53 0.88//4.27	200/- 22±10/-	- 10-15//1-1.5

Bookmark not defined.] (constant load)								
Kim et al [Error! Bookmark not defined.] (constant load)	5083Al	-	$9.1 \times 10^{-8} - 3.1 \times 10^{-5}$	548	0.90//0.37	0.98//0.54	205//1	-
Nam & Han [Error! Bookmark not defined.] (constant stress)	AC2B (Al-Fe-V-Si-Mn) AC2B-15% Al ₂ O ₃ fibers	Squeeze Cast (not PM)	$1.5 \times 10^{-7} - 1.7 \times 10^{-8}$ $1.0 \times 10^{-6} - 2.0 \times 10^{-10}$	523 523	1.00//0.002 0.55//5.97	-	-	- -
Gariboldi et al. [Error! Bookmark not defined.] (constant load)	6061Al-14vol% Al ₂ O ₃ p 6061Al-21vol% Al ₂ O ₃ p	IM IM	$4.0 \times 10^{-7} - 1.1 \times 10^{-10}$ $1.4 \times 10^{-5} - 7.0 \times 10^{-10}$	423- 523 423- 523	0.40//263.98 0.69//3.20	-	7// 7//	9.9//1 20.6//1
Hung et al. [Error! Bookmark not defined.] -	A359 A359-10% SiCp A359-20% SiCp	IM IM IM	$1.8 \times 10^{-5} - 1.1 \times 10^{-6}$ $2.3 \times 10^{-4} - 7.4 \times 10^{-6}$ $2.1 \times 10^{-9} - 7.3 \times 10^{-7}$	448- 573 448- 573 448- 573	0.42//283.45 0.54//179.28 0.33//3531.9	-	- - -	- (9.3±1.0)//1.5 (12.8±1.0)//1.5
Whitehouse et al. [Error! Bookmark not defined.] (constant load)	Pure aluminum: unreinforced -20% safil powder -10% safil infiltr. -20% carbon fibers -10% whiskers	PM PM PM PM PM	$8.0 \times 10^{-5} - 1.1 \times 10^{-8}$ $1.8 \times 10^{-5} - 2.2 \times 10^{-7}$ $1.3 \times 10^{-5} - 4.9 \times 10^{-7}$ $2.3 \times 10^{-5} - 5.1 \times 10^{-8}$ $2.1 \times 10^{-9} - 7.3 \times 10^{-7}$	543 543 543 543 543	0.60//362.09 0.68//1.19 0.98//0.07 0.76//0.90 0.91//0.05	-	(75x7)//11 (21x5)//5 (12x4)//3 (13x4)//3 (60x24)//30	- (11-13)//1 3//5 3//5 1//5

^a Elongated grains in the PM alloy are grouped in bands (see figure 1a) in ref. **[Error! Bookmark not defined.]**.

^b Fits for DM analysis have been conducted assuming $\mathcal{E}_f = 0.36$ and independent of stress and temperature, as reported in **[Error! Bookmark not defined.]**.

^c Alloy ALCO is an ODS (Oxide Dispersion Strengthened) alloy with Al₄C₃ and Al₂O₃ fine particles and alloy ALCOR is ALCO alloy after recrystallization treatment to increase grain size. Grains in this alloy are large and elongated.

Table II. Summary of the studies on creep fracture behavior of monolithic aluminum alloys and discontinuously reinforced aluminum alloy matrix composites, MMCs. The resulting fitting parameters using equations (1), MG, and (3), DM (whenever strain to failure data is reported), are included.

Temperature (K)	Appl. stress (MPa)	Time to failure (s)	Strain to failure	Minimum creep rate (s ⁻¹)
523	40	7.2 x 10 ⁵	0.016	3.5 x 10 ⁻⁸
523	48	1.6 x 10 ⁵	0.20	3.0 x 10 ⁻⁷
523	49	9.0 x 10 ⁴	0.075	3.5 x 10 ⁻⁷
523	49	2.4 x 10 ⁵	0.20	9.0 x 10 ⁻⁸
523	53	9.8 x 10 ³	0.22	1.0 x 10 ⁻⁵
523	61	5.6 x 10 ²	0.22	2.8 x 10 ⁻⁴
573	28	2.1 x 10 ⁶	0.18	1.8 x 10 ⁻⁸
573	31	1.5 x 10 ⁵	0.21	5.9 x 10 ⁻⁷
573	37	1.2 x 10 ⁴	0.18	7.0 x 10 ⁻⁶
573	40	2.1 x 10 ⁴	0.21	3.0 x 10 ⁻⁶
573	42	3.0 x 10 ³	0.25	5.6 x 10 ⁻⁶
573	47	9.0 x 10 ²	0.18	1.0 x 10 ⁻⁴
623	23	1.1 x 10 ⁵	0.19	2.5 x 10 ⁻⁷
623	29	1.9 x 10 ⁴	0.20	7.8 x 10 ⁻⁶
623	29	3.7 x 10 ⁴	0.24	2.3 x 10 ⁻⁶
623	39	7.7 x 10 ²	0.25	2.6 x 10 ⁻⁴

6061Al PM alloy

Temperature (K)	Appl. stress (MPa)	Time to failure (s)	Strain to failure	Minimum creep rate (s ⁻¹)
623	23	2.0 x 10 ³	0.45	2.1 x 10 ⁻⁴
673	5	7.8 x 10 ⁵	0.13	1.2 x 10 ⁻⁷
673	8	2.7 x 10 ⁶	0.08	9.5 x 10 ⁻⁹
723	6	4.1 x 10 ⁵	0.10	6.0 x 10 ⁻⁸
723	9	2.5 x 10 ⁴	0.45	4.8 x 10 ⁻⁶
723	12	5.6 x 10 ³	0.45	4.5 x 10 ⁻⁵

6061Al IM alloy

Temperature (K)	Appl. stress (MPa)	Time to failure (s)	Strain to failure	Minimum creep rate (s ⁻¹)
623	29	9.3 x 10 ⁵	0.004	5.0 x 10 ⁻⁸
623	35	2.4 x 10 ⁶	0.0055	1.1 x 10 ⁻⁸
623	35	1.6 x 10 ⁴	0.003	1.0 x 10 ⁻⁷
623	37	6.2 x 10 ⁴	0.05	2.0 x 10 ⁻⁷
623	50	2.7 x 10 ²	0.18	3.0 x 10 ⁻⁴
673	17	2.1 x 10 ⁶	0.002	4.7 x 10 ⁻¹⁰
673	23	3.7 x 10 ⁵	0.004	5.8 x 10 ⁻⁸
673	23	3.5 x 10 ⁵	0.004	1.0 x 10 ⁻⁹
673	23	7.6 x 10 ⁴	0.016	6.7 x 10 ⁻⁸
673	30	3.3 x 10 ⁴	0.025	3.0 x 10 ⁻⁷
673	35	1.1 x 10 ³	0.12	3.3 x 10 ⁻⁵
673	43	3.7 x 10 ²	0.18	2.8 x 10 ⁻⁴
723	12	1.8 x 10 ⁵	0.007	8.0 x 10 ⁻⁹
723	15	1.8 x 10 ⁵	0.005	7.3 x 10 ⁻⁹
723	15	4.2 x 10 ⁴	0.02	1.0 x 10 ⁻⁷
723	17	6.5 x 10 ³	0.0033	1.7 x 10 ⁻⁷
723	20	2.3 x 10 ³	0.05	1.4 x 10 ⁻⁵
723	30	4.2 x 10 ²	0.15	1.8 x 10 ⁻⁴
773	5	1.8 x 10 ⁵	0.018	5.8 x 10 ⁻⁸
773	6	7.1 x 10 ⁴	0.03	3.0 x 10 ⁻⁷
773	8	8.5 x 10 ³	0.04	1.5 x 10 ⁻⁶
773	8	1.4 x 10 ⁵	0.027	7.0 x 10 ⁻⁷
773	13	1.9 x 10 ³	0.17	5.0 x 10 ⁻⁵

6061Al-15vol%SiC composite

Temperature (K)	Appl. stress (MPa)	Time to failure (s)	Strain to failure	Minimum creep rate (s ⁻¹)
673	60	5.3 x 10 ³	0.012	1.4 x 10 ⁻⁶
673	66	2.6 x 10 ³	0.013	3.0 x 10 ⁻⁶
673	73	5.0 x 10 ²	0.018	3.5 x 10 ⁻⁵
723	26	5.5 x 10 ³	0.017	1.6 x 10 ⁻⁶
723	30	1.1 x 10 ³	0.028	1.3 x 10 ⁻⁵
723	43	1.8 x 10 ²	0.026	1.0 x 10 ⁻⁴

6061Al-40vol%SiC composite

Table I.- Creep conditions and creep fracture parameters for the 6061Al PM and 6061Al IM alloys and the 6061Al-15vol%SiC and 6061Al-40vol%SiC composites.

Authors [ref.]	Material	Processing	S. rate, s ⁻¹	Temp, K	n''/C (MG)	m''/C' (DM)	G. size (μm) //A. ratio	Reinf. Size//A. ratio
Present research^a (constant stress)	6061Al	PM	2.8x10 ⁻⁴ – 9.7x10 ⁻¹⁰	573- 723	0.80//0.90	0.89//1.53	3.5//2	-
	6061Al	IM	2.1x10 ⁻⁴ – 9.5x10 ⁻⁹	623- 723	0.72//4.12	0.92//1.29	3.0//2	-
	6061Al-15vol%SiC	PM	1.8x10 ⁻⁴ – 4.7x10 ⁻¹⁰	623- 773	0.65//1.98	1.01/0.47	1.5//4	2//2.0
	6061Al-40vol%SiC	PM	1.0x10 ⁻⁴ – 1.4x10 ⁻⁶	673- 723	0.77//0.16	0.92//1.60	-	-//1
Monkman, Grant [2] -	Al, 2S,3S and Zn, Mg and Cu Solid solutions of Al	IM	3.5x10 ⁻³ – 4.5x10 ⁻⁸	288- 593	0.85//1.3	-	-	-
Dunand <i>et al.</i> [6] (constant load)	Al99,9%-25% Al ₂ O ₃ p Corase Grain (CG)	IM	2.3x10 ⁻⁴ – 5.9x10 ⁻⁹	608- 723	0.84//0.13	0.99//0.51	>1000//-	0.28//1
	Al99,9%-25% Al ₂ O ₃ p Fine Grain (FG)	PM	2.7x10 ⁻⁴ – 1.3x10 ⁻⁸	608- 723	0.93//0.05	1.02//0.42	1.3//-	0.28//1
Taminger <i>et al.</i> [16] (constant stress)	2124Al	PM	2.5x10 ⁻⁵ – 1.8x10 ⁻⁷	394- 539	0.73//1.40	0.95/0.24	250//4.7	-
	2124Al-15% Al ₂ O ₃ w	PM	1.5x10 ⁻⁵ – 2.0x10 ⁻⁸	367- 541	0.91//0.48	0.88//0.75	77//2.6	(8.2±2.4)//6.3
Pandey <i>et al.</i> [17] (constant load)	Al-10%SiCp	PM	1.3x10 ⁻⁶ – 3.0x10 ⁻⁹	623	0.74//0.85	1.10//0.05	5//-	1.7//1
	Al-20%SiCp	PM	8.5x10 ⁻⁷ – 7.5x10 ⁻¹⁰	623	0.77//0.16	0.90/2.44	5//-	1.7//1
	Al-30%SiCp	PM	1.7x10 ⁻⁶ – 2.2x10 ⁻⁸	623	0.85//0.04	0.96//1.13	5//-	1.7//1
Ishikawa <i>et al.</i> [18]^b (constant stress)	5083Al	IM	8x10 ⁻⁵ – 4x10 ⁻⁸	623- 773	0.83//0.606	0.83//1.68	100//-	-
Ma <i>et al.</i> [19] (constant load)	Al-TiB ₂ p	Reactive hot press. (not PM)	1.3x10 ⁻⁴ – 9.9x10 ⁻⁸	573-673	0.91//0.04	1.03//0.27	-	(0.5-5)//1
Rosler <i>et al.</i> [20]^c (constant load)	ALCO (Al-O)	PM	7.4x10 ⁻⁴ – 4.4x10 ⁻¹⁰	573- 773	0.74//0.20	0.96//0.32	1.9//-	-
	ALCOR		1.7x10 ⁻⁴ – 3.6x10 ⁻⁹	773	0.64//0.54	0.93//0.80	330//-	-
Requena <i>et al.</i> [21] (constant load)	6061Al	Stir casting (not PM)	1.1x10 ⁻⁹ – 2.5x10 ⁻⁷	573	0.85//0.08	0.70//7.53	200//-	-
	6061Al-22vol% Al ₂ O ₃ p		5.8x10 ⁻⁹ – 2.2x10 ⁻⁶		1.11//0.003	0.88//4.27	22±10//-	10-15//1-1.5
Kim <i>et al.</i> [22] (constant load)	5083Al	-	9.1x10 ⁻⁸ – 3.1x10 ⁻⁵	548	0.90//0.37	0.98//0.54	205//1	-
Nam & Han [23] (constant stress)	AC2B (Al-Fe-V-Si-Mn)	Squeeze Cast (not PM)	1.5x10 ⁻⁷ – 1.7x10 ⁻⁸	523	1.00//0.002	-	-	-
	AC2B-15%Al ₂ O ₃ fibers		1.0x10 ⁻⁶ – 2.0x10 ⁻¹⁰	523	0.55//5.97			-
Gariboldi <i>et al.</i> [24] (constant load)	6061Al-14vol% Al ₂ O ₃ p	IM	4.0x10 ⁻⁷ – 1.1x10 ⁻¹⁰	423- 523	0.40//263.98	-	7//-	9.9//1
	6061Al-21vol% Al ₂ O ₃ p	IM	1.4x10 ⁻⁵ – 7.0x10 ⁻¹⁰	423- 523	0.69//3.20		7//-	20.6//1
Hung <i>et al.</i> [25] -	A359	IM	1.8x10 ⁻⁵ – 1.1x10 ⁻⁶	448- 573	0.42//283.45	-	-	-
	A359-10%SiCp	IM	2.3x10 ⁻⁴ – 7.4x10 ⁻⁶	448- 573	0.54//179.28		-	(9.3±1.0)//1.5
	A359-20%SiCp	IM	2.1x10 ⁻⁹ – 7.3x10 ⁻⁷	448- 573	0.33//3531.9		-	(12.8±1.0)//1.5
Whitehouse <i>et al.</i> [26] (constant load)	Pure aluminum: unreinforced	PM	8.0x10 ⁻⁵ – 1.1x10 ⁻⁸	543	0.60//362.09	-	(75x7)//11	-
	-20% safil powder	PM	1.8x10 ⁻⁵ – 2.2x10 ⁻⁷	543	0.68//1.19		(21x5)//5	(11-13)//1
	-10% safil infiltr.	PM	1.3x10 ⁻⁵ – 4.9x10 ⁻⁷	543	0.98//0.07		(12x4)//3	3//5
	-20% carbon fibers	PM	2.3x10 ⁻⁵ – 5.1x10 ⁻⁸	543	0.76//0.90		(13x4)//3	3//5
	-10% whiskers	PM	2.1x10 ⁻⁹ – 7.3x10 ⁻⁷	543	0.91//0.05		(60x24)//30	1//5

^a Elongated grains in the PM alloy are grouped in bands (see figure 1a) in ref. [14]).

^b Fits for DM analysis have been conducted assuming $\mathcal{E}_f = 0.36$ and independent of stress and temperature, as reported in [18].

^c Alloy ALCO is an ODS (Oxide Dispersion Strengthened) alloy with Al_4C_3 and Al_2O_3 fine particles and alloy ALCOR is ALCO alloy after recrystallization treatment to increase grain size. Grains in this alloy are large and elongated.

Table II. Summary of the studies on creep fracture behavior of monolithic aluminum alloys and discontinuously reinforced aluminum alloy matrix composites, MMCs. The resulting fitting parameters using equations (1), MG, and (3), DM (whenever strain to failure data is reported), are included.