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## Microplasticity of MMC. Experimental results and modelling

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### ABSTRACT

The microplastic behavior of several MMC is investigated by means of tension and compression tests. This behavior is assymetric : the proportionnal limit is higher in tension than in compression but the work hardening rate is higher in compression. These differences are analysed in terms of maximum of the Tresca's shear stress at the interface (proportionnal limit) and of the emission of dislocation loops during the cooling (work hardening rate). On another hand, a model is proposed to calculate the value of the yield stress, describing the composite as a material composed of three phases : inclusion, unaffected matrix and matrix surrounding the inclusion having a gradient in the density of the thermally induced dislocations.

### INTRODUCTION

The hardening of MMC compared to the unreinforced matrix is due to mechanisms including effects of microstructural parameters and internal stresses resulting from thermic, plastic and elastic . This hardening can be characterised by the value of the elastic limit. This mechanical property can be defined as the proportionnal limit (difficult to determine experimentally) or as the conventionnal yield stress (corresponding to 0.2 % of plastic strain). The behavior of the materials between these two values of the stress is defined as the microplastic behavior. Modelling of the yield stress is generally done [1] in terms of mean internal stress field and hardening due to the emission of dislocations (during the cooling) that are assumed to be randomly distributed in all the matrix volume. In order to obtain a better understanding of phenomena governing the elastic limit value, measurements of microplastic strains have been performed on discontinuous reinforced MMC at different temperatures. Results of mechanical tests are presented and analysed. Finally, a model taking the realistic microstructure of MMC into account is proposed for the prediction of the conventionnal yield stress value.

### EXPERIMENTAL

**1). Conditions :** Three kinds of MMC have been elaborated by the OSPREY route [2] and then extruded : 2618, 6061 and 7075 aluminium base alloy matrices. Each is reinforced with a 15 % volume fraction of SiC particles exhibiting an average length of about 10  $\mu\text{m}$  and an aspect ratio near 2. The composites have been heat treated under the peak aged conditions that are classically used for the unreinforced alloy. Tension and compression tests have been performed (long direction) according to the experimental procedure defined in reference [3]. Tension and compression stresses are reported versus microplastic strains measured in a plastic strain range rising from  $5 \cdot 10^{-6}$  to  $5 \cdot 10^{-3}$ . 2618 and 6061 materials have been tested at room temperature and 7075 alloy between 240 and 350 K.

**2). Results :** The microplastic behavior of the 6061, 2618 and 7075 MMC is presented on figure 1, 2 and 3 respectively. At room temperature, an assymetrical behavior of the 6061 MMC is observed (fig. 1): the proportionnal limit is higher in tension than in compression but the work hardening rate is more important in compression. Thus the curves cross for a plastic strain of about 0.2 %. The behavior of

the 2618 MMC is also assymetrical (fig. 2) but the curve crossing occurs for a very low plastic strain, because the work hardening rate is very high in compression. At room temperature (fig. 3a), the behavior of the 7075 MMC is quite similar to that of the 6061 composite. When the temperature is lowered (fig.3b), the work hardening rate increases and the plastic strain amplitude corresponding to the curve crossing is reduced. On the contrary, at high temperature (fig. 3c), tension and compression curves are nearly parallel and the curve crossing is not observed in the strain measurement field.

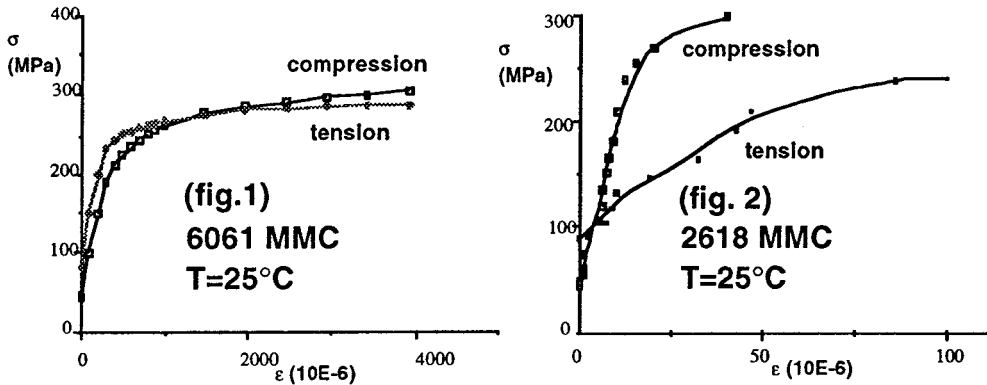


Figure 1 and 2: Experimental tension and compression stress versus microplastic strain curves (6061 and 2618 matrix composites) at room temperature.

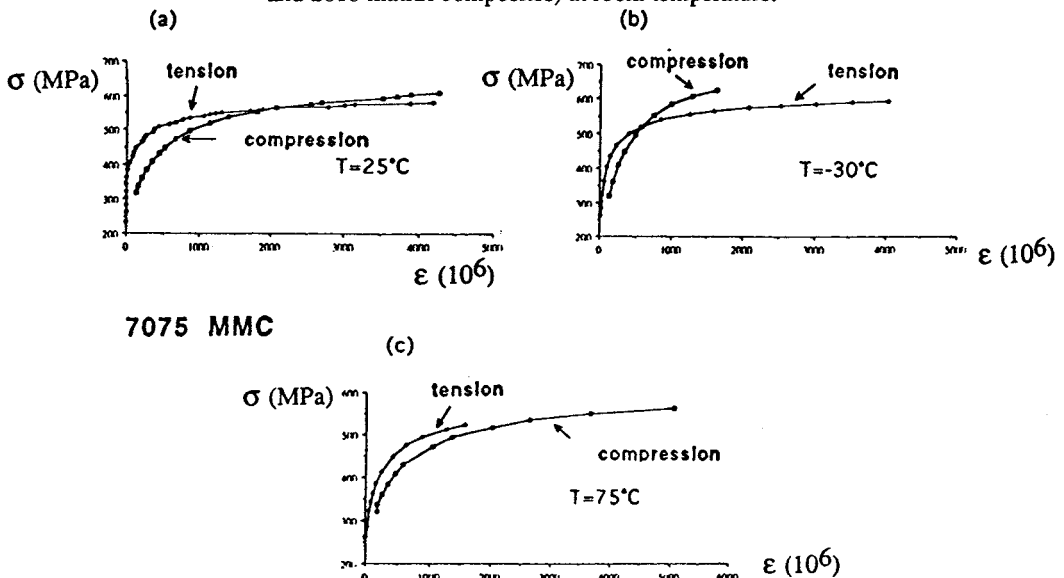


Figure 3 a,b,c : Experimental tension and compression stress versus microplastic strain curves for different testing temperatures (7075 matrix MMC).

**3). Discussion :** Proportional limit differences in tension and compression have been analysed from thermal and elastic heterogeneities. The subsequent analysis is relevant to usual MMC where particles present a higher young modulus and a lower thermal expansion coefficient than the matrix. After a cooling down responsible for the thermal stress field and with an applied mechanical stress HAMMAN et al [3] have shown that the maximum of the TRESCA's shear stress value at the interface between the matrix and the particle is higher in compression than in tension. In consequence plastic strain occurs sooner when a compression is applied. The curve crossing is due to the existence of a higher work hardening rate in compression than in tension. This phenomenon is probably due to a difference in the evolution of the dislocation density during tension or compression tests. The thermal treatment induces strain incompatibilities which are partially assumed by the emission of interstitial

dislocation loops. These loops have the same type than the loops emitted to assume incompatible strains due to an applied compression stress. Therefore their effects are cumulative. On the contrary lacunar dislocation loops created to assume incompatible strains due to an applied tension stress are annihilated with thermally emitted interstitial loops. Thus the work hardening rate is higher in compression. The effect of the testing temperature is also well understood by considering that at high temperature, recovery phenomena lead to the annihilation of the emitted dislocations, that reduces the difference between the hardening rates and therefore leads to parallel microplasticity-stress curves. At low temperature, recovery phenomena are limited and differences in work hardening rates are enhanced. It is also the case for the 2618 MMC at room temperature. Indeed the 2618 aluminium alloy has been elaborated to be used under high temperature conditions and therefore recovery phenomena are limited.

## MODELLING

**1). Yield Stress Model :** This model has been developed to calculate the value of, the yield stress, defined as the stress at which the all matrix volume is plastic. It introduces an inhomogeneous distribution of thermally enhanced dislocations (TED) in the matrix. As shown on Figure 4, the MMC is considered as a three phases material including ceramic particles, matrix volume affected by TED and finally the rest of the matrix unaffected by the presence of the particles. The difference between this approach and the existing ones is that in the thermally induced plastic zone, we consider a dislocation density gradient as it is experimentally observed [4].

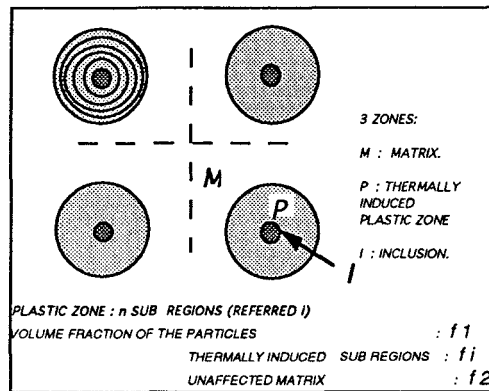


Figure 4 : yield stress model.

Considering a mixture law the yield resolved shear stress of the composite can be written as following :

$$\tau = \langle \tau \rangle_M + \tau_0 + k d^{1/2} + \frac{\alpha \mu b}{1 - f_1} \left( f_2 \sqrt{\rho_M} + \sum_{i=1}^{N} f_i \sqrt{\rho_M + \rho_i} \right)$$

We have:  $f_1 + f_2 + \sum_{i=1}^{N} f_i = 1$  N being the number of TED sub regions.

$f_1$  and  $f_2$  are the volume fractions of ceramic particles and TED unaffected matrix respectively,  $f_i$  is the volume fraction of the TED sub region  $i$ , inside which the dislocation density  $\rho_i$  is assumed to be constant,  $d$  is the matrix grain size,  $k$  the Hall and Petch constant ( $= 0.75 \text{ MPa.m}^{1/2}$ ),  $\mu$  is the shear modulus ( $= 28 \text{ GPa}$ ),  $b$  the dislocation Burgers vector ( $= 2.7 \cdot 10^{-10} \text{ m}$ ),  $\alpha$  a numerical constant ( $= 1.2$ ) and  $\rho_M$  the unaffected matrix dislocation density ( $= 10^{12} / \text{m}^2$ ),  $\langle \tau \rangle_M$  is the mean internal stress in the matrix assumed to be neglectible and  $\tau_0$  the yield stress of the unreinforced matrix. To use this relation we have to determine  $f_i$  and  $\rho_i$ . These parameters are calculated from the HAMANN's model [5] That is used as a tool in the present study.

2). **Results and analysis :** Results have been obtained in the case of 6061 MMC ( $\tau_0 = 100$  MPa) for a single SiC particle in an infinite matrix and a temperature gap of  $-500^\circ$ . This calculation is correct up to a 5 % volume fraction of SiC from which there is an overlap of the TED regions around each particle. Calculations of the yield stress variation versus the particle size obtained by the present model in three cases corresponding to different hypothesis are presented on figure 5.

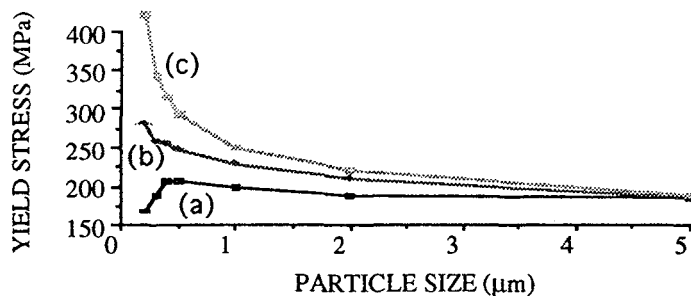


Figure 5 : Conventional yield stress variation versus the particle size (a) constant grain size, (b) grain size equal to the particle size, (c) dislocations randomly distributed in the matrix volume.

In the case a, the TED are heterogeneously distributed (gradient in the dislocation density) and the grain size is assumed to be constant. In the case b, the grain size is a function of the particle size, as it is assumed by Humphreys et al. [6]. This is a more realistic hypothesis : 6061 matrix recrystallises from the particles during the solution treatment. We tracted the case c for comparison with existent models : the same length of dislocations is emitted but they are assumed to be randomly distributed in the matrix volume (homogeneous repartition). For the very small particle size, the maximum of the yield stress, predicted in the case a is due to the fact that the only contribution that is varying is this of the plastic zone: the intrinsic yield stress of this zone becomes very high -when the particle size decreases- but its volume fraction is decreasing. It even becomes equal to zero for a particle size under which the emission is impossible. The increase in the yield stress in the case b is then mainly due to the grain boundaries effect for the smallest particle sizes. The main feature to notice is that the value calculated with the more realistic hypothesis (b) is much smaller than what could be predicted by the existent models.

#### ACKNOWLEDGMENTS

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