MICRO-MACRO MODELLING OF THE INFLUENCE OF THE DISCONTINUOUS REINFORCED COMPOSITES HETEROGENEITIES ON DAMAGE AND FAILURE PROPERTIES

K.Derrien, D.Baptiste

ENSAM, Laboratory LM3, CNRS ESA 8006
151 Bd de l'Hôpital, 75013 PARIS, France

SUMMARY: We studied the tensile behaviour and damage of an Aluminium alloy X2080 reinforced with different volume fraction of silicon carbide particles. We use an homogenisation method in order to predict the behaviour of this metal-matrix composites. The main damage mechanism is particle failure. We study the influence of an heterogeneous dispersion of the reinforcement on the damage development and the failure strain of composites which contain locally a higher volume fraction of reinforcement. We compare experimental and theoretical results. The originality of this work lies in the modelling of the failure using a micromechanical approach.

KEYWORDS: Metal-matrix composites, Fracture, Damage, Heterogeneous repartition, Homogenisation method

INTRODUCTION

Metal-matrix Composites (MMCs) have been developed in order to value the combination of the following phases : rigidity of reinforcement, ductility and fracture toughness of the matrix. They constitute a very attractive range of materials particularly for the conception of aeronautical structures looking for performance improvement as well as structure alleviation. Our objective is to predict the tensile behaviour, the damage and the failure of these composites as a function of their microstructure. We are interesting in aluminium alloys X2080 reinforced by different volume fraction of silicon carbide particles. The materials were made by a powder blending and extruded route.

The first stage of this study has been an experimental characterisation of powder metallurgy MMCs: aluminium alloy and associated unreinforced aluminium matrix [1-3]. The results are used as input and validation data for the model. We use micromechanical approach based on Mori Tanaka's model [4] in order to connect the microstructure and the macroscopic properties of the material. The damage mechanisms are identified at the reinforcement scale, by in-situ tensile tests inside a scanning electronic microscope, they are modelled and integrated in our model, at the micro scale. This approach leads to a new law behaviour taking into account damage evolution. The final failure of the materials is then predicted using the same approach that is by introducing in the matrix a cavity growth law of Gurson's type. We study the influence of an heterogeneous dispersion of the reinforcement on the damage development and the failure strain of composites which contain locally a higher volume fraction of reinforcement. We consider that a composite which contains locally a higher volume fraction of reinforcement results on the association of two homogeneous micro-composites. Each one is modelled by Mori-Tanaka's model, we choose the homogenisation by the self consistent model for the blend of these micro-composites.

MICROMECHANICAL MODEL

A micro-mechanical model based on the stiffness prediction by the Mori and Tanaka approach is proposed for modeling the elastoplastic and damaged composite behavior. Among several well-known micro-mechanical approaches, the Mori and Tanaka's method is an alternative to find the estimates of elastic moduli and local fields in multiphase composite materials. The stress tensor is calculated in each phase as a function of the overall macroscopic stress tensor Σ, the geometrical parameters as orientation and aspect ratio , and the mechanical properties of each phase. The combination of the Eshelby equivalent inclusion method and the Mori and Tanaka's back stress analysis, leads us to the equations used in the prediction of the composite behaviour.The main assumption is contained in the strain localisation relation which defines the load sharing between the different constituents. This localisation relation is expressed by Mori and Tanaka [4] :

$$\varepsilon_r = T_r \varepsilon_0 \quad T_r = \left[L_r + L_0 \left(S_r^{-1} - I \right) \right]^{-1} L_0 S_r \quad (1)$$

ε_0, L_0 and ε_r, L_r are the average strain and stiffness tensor of the matrix and the r-th reinforcement respectively. S_r is the Eshelby's tensor.

If a uniform stress Σ is applied to the material, it was shown that the average local stress over the composite is equal to Σ, the corresponding strain in the material is noted E and we have:

$$\Sigma = c_0 \langle \sigma \rangle_0 + c_1 \langle \sigma \rangle_1 \quad E = c_0 \langle \varepsilon \rangle_0 + c_1 \langle \varepsilon \rangle_1 \quad (2)$$

$\langle \; \rangle_i$ means the volume average value in the volume of the phase i

As a result, the estimation of the composite stiffness tensor can be written as:

$$L = \left(c_0 L_0 + \sum_{r=1}^{r=n} c_r L_r T_r \right) \left(c_0 I + \sum_{r=1}^{r=n} c_r T_r \right)^{-1} \quad (3)$$

The elastic model is extended to the elastoplastic behaviour using the concept of secant moduli. The theory makes use of a linear comparison material, whose elastic moduli at every instant are chosen to coincide with the average secant moduli of the matrix to reflect its elastoplastic state. The composite is subsequently replaced by the comparison material with equivalent transformation strains. The effective stress is not defined in terms of the averaged stress in the matrix alone but in terms of the average elastic distortional energy in the matrix , see Qiu and Weng [5].

Damaged behaviour:

For modelling the damaged behaviour, we have to define a damage criterion relative to the experimental results. The particle breaking is the principal source of damage. We use a Weibull law in order to define the particle fracture. This law is in agreement with our

experimental observations, the fracture probability increases with the particle size and applied stress.
The fracture probability of each particle is a function of its volume V_r and of the maximum principal stress σ_r.

The Weibull's law can be written , see Mochida and al[6] :

$$P_r(\sigma_r, V_r) = 1 - \exp\left(-\frac{V_r}{V_u}\left(\frac{\sigma_r}{\sigma_u}\right)^m\right) \quad (4)$$

m,σ_u, are called shape and scale parameters., V_r is the volume of the r-th group and and V_u is the volume of the reference group for which the Weibull's parameters are determined.
We determine these parameters using our experimental results. We find the following parameters [2](determined in a statistical survey on a composite reinforced by 15% of SiCp).
σ_u=1500MPa, m=4, V_0 is the volume for a particle of diameter 10μm

A broken reinforcement still contributes to the global stiffness of the composite but the rigidity of a broken reinforcement is lower than the rigidity for an intact reinforcement . We choose to replace the broken particles by penny shaped cracks located perpendicular to the loading direction and which are oblate spheroids with the major axis being equal to the radius of particles and the minor axis being a function of the external load. The analytical value of the crack opening displacement h is :
$h = 2b\varepsilon^*_{33}$ where ε^* is the eigenstrain in the crack and 2b is the crack diameter (Mura [7]).
Using Mori Tanaka's method, we have :

$$\varepsilon^*_r = \left(L_0(I - S_r)\right)^{-1}(\sigma_0 - \sigma_r) \quad (5)$$

It is now possible to predict the tensile behaviour of composite material including the damage effect and the plasticity effect. At each step of the applied macroscopic load, the stress induced in the SiC particles is calculated followed by their fracture probability. At the following step the new broken particles are replaced by penny-shaped cracks.

FAILURE CRITERION:

Following the experimental observations, it appears that bands with a high plastic strain are around the end of the fissure of the broken particle. These intense strain lead to the debonding of the precipitates and it is the high growth of the micro-voids that will transform the band into a fissure. Our hypothesis is that it is the failure of the specimen that takes place by linking of the micro-cracks initiated in the matrix from the broken particles. The rule of the SiC reinforcement has an important function at all the stages of the ductile failure. The presence of SiC particles increases the number of potential sites at the beginning of the damage and the critical growth rate of cavities might be limited if the number of broken particles is sufficient and if these sites are sufficiently close (importance of the distance between two broken particles). From the theoretical and experimental studies , it seems necessary to have a precise knowledge of a stress and strain fields close to the broken particles in order to elaborate a failure criterion. Up to now we have integrated in our behavior law the porosity due to the particle failure. In order to have a more accurate description of the damage and to establish a failure criteria, we will take into account the porosity associated to the debonding precipitates.

The first stage of a failure criteria definition is to determine the range of stress and strain fields close to the SiC broken particles, in accordance with the composition of the elementary representative element and the macroscopically plastic strain. At the second stage we determine the growth rate of cavities initiated on the precipitates. The linking of the micro-cracks depends on the value of the growth rate and on the distance from the fissure where this growth rate is attained.

Stress and strain fields near the penny-shaped cracks :

The strain distribution near the crack is obtained from the Hutchinson [8], Rice and Rosengren [9] solution. This original solution is not suitable for porous materials because the yield criterion used is a Von Mises criterion. In our case we want to take into account the local porosity due to the debonding of the precipitates. So we have modified the original theory, the yield criterion is calculated using Mori Tanaka's approach and the plasticity of porous material developed by Qiu and Weng.

$$\Psi(\Sigma) = \frac{15-6c_0}{9{c_0}^2}\frac{\Sigma e^2}{\sigma y^2} + \frac{9(1-c_0)}{4{c_0}^2}\frac{\Sigma m^2}{\sigma y^2} - 1 = 0 \qquad (6)$$

Σ_m is called the macroscopic hydrostatic stress : $\Sigma_m = \sum_i \Sigma_{ii}$

σy is the yield stress, Σe is the macroscopic Von Misès equivalent stress

The yield criterion near the crack can be written as follows,

$$\sigma_{0e}^{h\;2} = 3u.\sigma_m^{h\;2} + v.\sigma_e^{h2} \qquad (7)$$

σ_{ii}^h are the components of the stress tensor centered at the tip of the crack

The advantage of such a criteria is that it is linked to the fraction of porosity, the coefficients u and v depending on this fraction.

$$3u = \frac{9(1-f_0)}{4{f_0}^2} \quad v = \frac{15-6f_0}{9{f_0}^2} \quad (8)$$

f_0=1-fp, fp is the volume fraction of porosity inside the matrix where the stress and strains fields are calculated.

This expression comes to the von Mises criterion as u=0 and v=1. In that case the hydrostatic stress has no effect and we will find the equation in connection with the HRR fields. The modified HRR fields are developed in [1].

using the yield criterion we calculate the growth of cavity in a porous material :

$$\frac{1}{\dot{\varepsilon}_e^p}\frac{dR}{R} = \frac{1}{3fp}\frac{3u}{v}\frac{\sigma_m^h}{\sigma_e^h} = \frac{9}{4(3+2fp)}\frac{\sigma_m^h}{\sigma_e^h} \quad (9)$$

The Figure 1 illustrates the modification of the HRR fields : $\frac{1}{\varepsilon e}\frac{dR}{R}$ decrease with the porosity’s volume fraction and the maximum growth goes from 60 to 45° when the porosity’s volume fraction goes from 0 to 15%.

Thus the HRR modified singularity in conjunction with the value of J completely specifies the

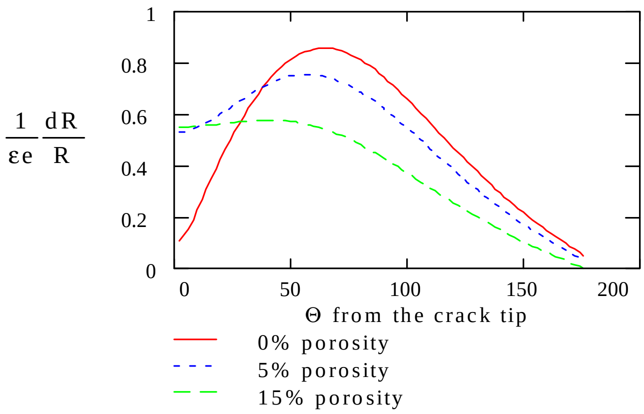


Figure 1 : ***Effect of the porosity***

near crack tip fields.

J represents the amplitude of the singular fields. We calculate J using Mori-Tanaka's model :
The energy release rate of a penny-shaped crack can be defined as Mura [7]

$$J = \frac{1}{2\pi r}\left|\frac{\partial P}{\partial r}\right| \qquad (10)$$
$$P = P_0 + E_{int} = P_0 + \frac{1}{2}\sigma\varepsilon^* V_c$$

P is the total potential energy, P_0 is the total potential energy without any inhomogeneity and E_{int} is the interaction energy between the applied stress and the inhomogeneity, V_c is the volume of the penny-shaped crack and ε* is the eigenstrain in the crack (see equation 5).

Failure criterion :

The growth rate of the cavities associated with the precipitates and therefore with their coalescence is a decreasing function of the distance to the broken particles. The ligament failure will be even more easy if the scale of the area containing porosity is large. The ligament failure arises when the growth rate becomes critical on a distance at least equal to half of the distance between two particles. The largest the volume fraction of reinforcement is, the weakest the critical distance. The volume fraction matches the average half inter-particle distance, it depends on the R ray of the particles and f the total volume fraction of reinforcement.

$$2\lambda critical = R\cdot\left(\sqrt{\frac{2\pi}{3f}}-\sqrt{\frac{8}{3}}\right) \qquad (11)$$

In order to compare the failure strain to the composites reinforced by different rates of SiC, we will calculate for one composite (here X2080 + 15% SiCp) the growth rate reached at the critical distance λcritical for a J value matching the failure strain of V.E.R (here λcritical=5.2µm (R=5µm), failure strain = 5.5%). One will use this growth rate as a reference for each composite we calculate the distance λ at which this growth rate is reached. Therefore we impose that the elementary representative element failure is taking place when λ/λcritical =1. Departing from the failure strain of a given composite, this study allows us to forecast the failure strain of composites based on their particles content.

Results :

In Figure 2 , we compare experimental and theoretical results for two kinds of materials : X2080 alloy reinforced by 15 or 20 % of particles. Our simulation gives a good agreement. Using the same method, we can calculate the failure strain of a composite as a function of the volume fraction of reinforcement, results are reported in Figure 3.

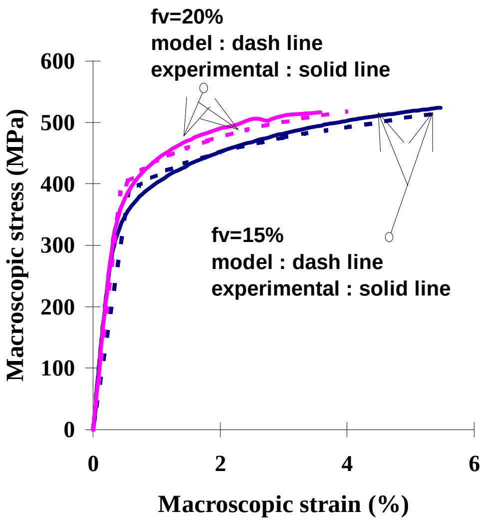


Figure 2 : Behaviour law for materials reinforced by two different volume fractions of SiCp

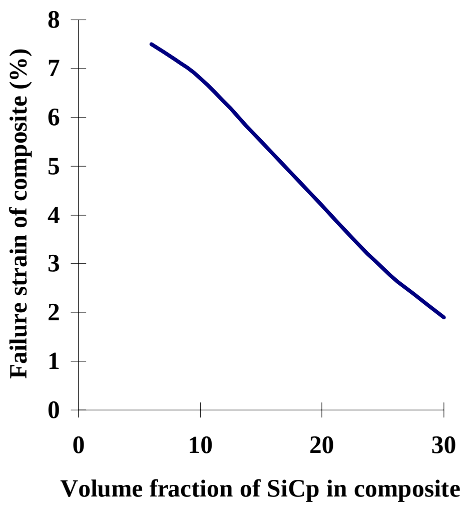


Figure 3 : Failure strain of a composite as a function of the volume fraction of SiCp contained in the matrix

THE INFLUENCE OF AN HETEROGENEOUS DISPERSION OF THE REINFORCEMENT:

Using an homogenisation method, we suppose that the distribution of the reinforcement is homogeneous. In reality, a composite is rarely homogeneous and locally we can observe a high concentration of particles that will leads to a dispersion of failure strain. We shall study the influence of a heterogeneous dispersion using two steps in our homogenisation technique.

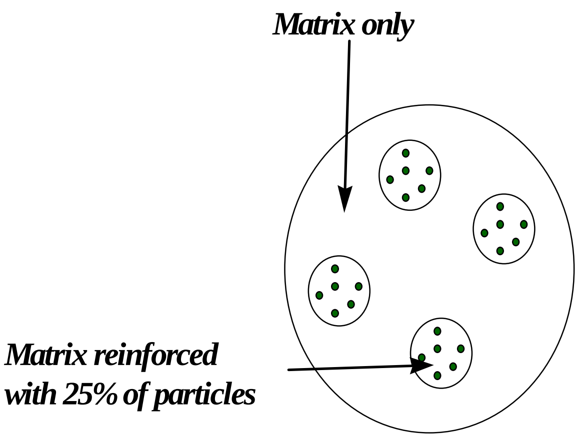
For taking into account a local concentration of particles, we shall consider two levels of heterogeneity and so, two successive homogeneisations. The first level of heterogeneity is the reinforcement, the second level represents the heterogeneity of the distribution in the composite.
If we want to take into account the interaction between composites, we naturally choose an homogenisation technique for the blend of these micro-composites. Indeed, the behaviour law of these micro-composites does not evolve independently of one another. When a fraction of material is damaged, the other comes under a consequence and has to support a higher fraction of the macroscopic load. For the first homogenisation we have chosen the Mori and Tanaka’s model, on the other hand for a morphological argument we choose the self consistent model.
With such a model it is not possible to take into account a precise spatial distribution of the heterogeneous areas. Working on that problem we have to bear in mind that the areas highly concentrated in particles are present in all the composites even though they represent a small element of it.

The details of the calculation are reported in [10].

Results :

We have studied a composite which has been manufactured for being deliberately heterogeneous by mixing an aluminium's powder with a composite's powder.

We have modelled the behaviour law of two composites including 6% of particles, the first composite is homogeneous, the second contains areas with 25% of particles and areas without particles (figure 4) .



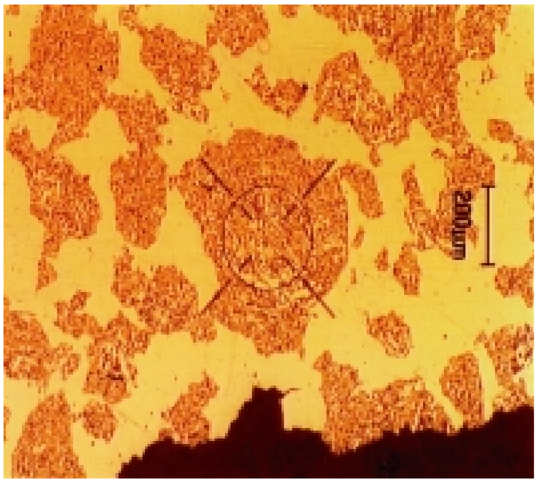


Figure 4 : heterogeneous composite

A comparison between experimental and theoretical results is plotted on figure n°5.
In the case of the heterogeneous material, the area which is highly reinforced controls the composite's failure.

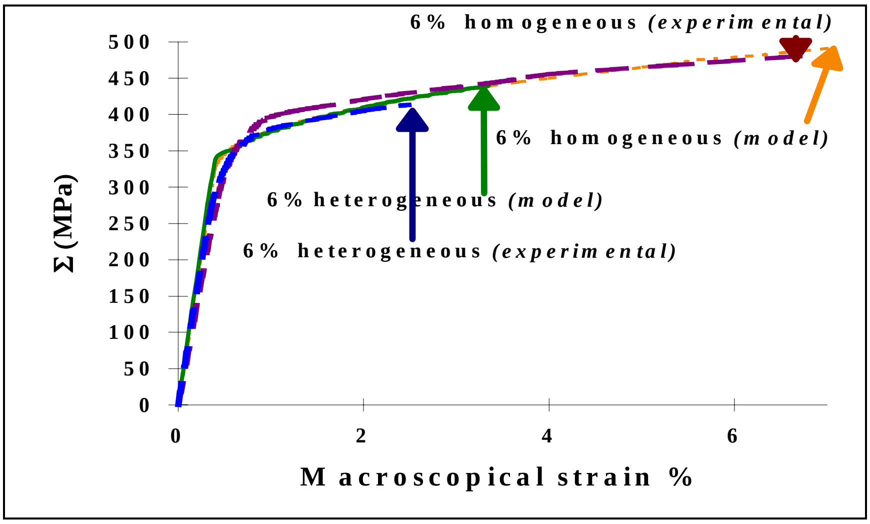


Figure 5 : The influence of an heterogeneous dispersion, comparison between experimental ant theoretical results

CONCLUSION

The use of a micromechanical model to predict the failure provides good results. It takes into account the microstructure of the material, size, aspect ratio, orientation, volume fraction of the reinforcement. It can be easily extended to other composite if one can determine mechanical properties of each phase and damage mechanisms (fiber breaking, debonding at the interface) at the origin of the macroscopic failure. The failure strength of a composite material is reactive to the defects present in the material and the heterogeneous distribution.
We have demonstrated that the particles cluster in a material could control the failure strain.
The particles breaking is less important in the clusters, however the proximity of the fissures then created is damageable for the material

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