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Influence of the elastic stress relaxation on the microstructures and mechanical properties of metal–matrix composites

J. Douin[a,*], P. Donnadieu[b], A. Finel[a], G.F. Dirras[c], J.F. Silvain[d]

[a]LEM (UMR 104), CNRS/ONERA, BP 72, 29, Avenue De La, Division Leclerc, 92322 Châtillon cedex, France
[b]LTPCM-ENSEEG-INPG, Domaine Universitaire BP 75, 38402 Saint Martin d'Hères cedex, France
[c]LPMTM, Institut Galilée, Université Paris-Nord, 93430 Villetanneuse cedex, France
[d]ICMCB, 87 Av. Dr A. Schweitzer, 33608 Pessac cedex, France

Abstract

Role of the residual stresses on the mechanical properties of metal–matrix composites is studied. It is shown that the stress relaxation can be responsible for the morphologies and spatial distribution of precipitates. Direct measurements of the residual stress is also emphasized and the influence of dislocations in the accommodation process and during interface crossing is exemplified.

Keywords: A. Metal-matrix composites (MMCs); B. Residual stress; B. Mechanical properties

1. Introduction

The mechanical performance of materials is of great importance for numerous applications ranging from the structures of aircraft and automobile to the connections of microelectronic circuits. Among all the structural materials, composite materials with a metallic matrix have a large importance due to their improved mechanical properties.

The mechanical behavior of metal–matrix composites (MMC) results from the existence of inclusions within the matrix and many of the properties of these multiphase materials depend mostly on the morphology and spatial distribution of minority inclusion phases within the majority metallic matrix phase. Very often, it is the influence of the interfaces between the inclusions and the matrix that governs the mechanical properties of the composite materials. Thus the mechanical and especially plastic properties of MMC are ascribed to the mechanical behavior of the matrix and of the interfacial zone.

The existence and the response of defects to internal and external fields dictate in many ways the mechanical behavior of the matrix. The matrix behavior itself is governed by the behavior of the dislocations in the vicinity of the inclusions. As all interfaces form obstacles to the propagation of plastic glide and generate internal stresses during deformation, dislocations/interfaces interactions appear to control a large number of the mechanical properties of MMCs, and this paper will focus on a few examples of such interactions.

2. Influence of the elastic field on composite morphology

The morphology and spatial distribution of inclusions in a matrix are in some cases governed by the dynamics of a phase transformation process. For example, cooling a single-phase alloy can result in the formation of small nuclei of another phase within the high-temperature phase. Therefore, the growth associated with the structure of the reinforcement and the subsequent coarsening process frequently determine the morphology and spatial distribution of the inclusions. Nucleation and growth result from a complex interaction between chemical and elastic relaxations, that involve a non-linear kinetic process.

One of the most popular techniques used to investigate this problem is the phase field method. In the present context (phase transformation in solids), the phase field method [1] associates a description of the chemical properties, through the stabilities of the different phases represented by a Ginsburg–Landau free energy, with a modelization of the elastic strain, according to the model developed by Khachaturyan [2]. The time evolution of the concentration field is calculated on discrete values in a fixed grid following a Cahn–Hilliard-type equation [3], whereas the

* Corresponding author. Tel.: +33-1-46-734442; fax: +33-1-46-734155.
E-mail address: douin@onera.fr (J. Douin).

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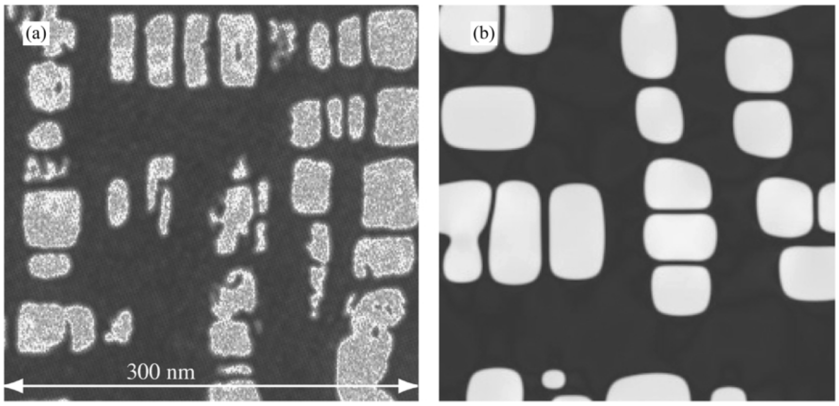

Fig. 1. Influence of the elastic stresses on the morphology of γ' precipitates within a γ matrix. (a) Ni–11.84%Ti alloy. The size of the micrograph is 300 nm ([5], courtesy of A.J. Ardell); (b) corresponding phase field simulation from an initially disordered alloy. The calculated shape, size and relative positions of the precipitates are in good agreement with the observation. Notice the alignment (rafting) of the precipitates minimizing the residual strain.

long-term order fields, which describe the state of local order, follow a Allen–Cahn dynamical equation [4]. According to the coarse graining approximation, fluctuations smaller than the grid step size are averaged.

An example of such a modelisation is given in Fig. 1, in the case of a Ni–Ti alloy. In this Ni–Ti alloy, two phases can coexist: the disordered face-centered cubic γ phase, and the ordered face-centered cubic $L1_2$ γ' phase. The Ginsburg–Landau free energy has been calibrated to reproduce the observed γ/γ' two-phase regime at 720 °C and the interfacial free-energy between the precipitates and the matrix (13 mJ/m^2). As for the elastic energy, we used the elastic constants of pure Ni and the experimental lattice mismatch $\varepsilon = 0.0085$ between precipitates and the matrix. In good accordance with the observations, the simulations show that in order to decrease the residual stresses in the γ matrix, the γ' precipitates adopt a cuboidal shape with their habit planes perpendicular to the elastically soft directions, i.e. here the $\langle 001 \rangle$ directions. Moreover, after a long enough heat treatment the cuboids are found aligned together, a mechanism called rafting, usually observed in γ/γ' alloys, such as superalloys after a long period of use.

3. Stress field accommodation during the composite elaboration

In many systems, ranging from transformation-toughened ceramics to high-temperature superalloys, the minority-phase inclusions possess physical properties that highly differ from those of the matrix phase. These differences engender elastic stress fields in both the matrix and the inclusions that can strongly influence the mechanical properties of the material. It has to be mentioned that the residual stress is not usually completely removed by an adequate choice of shapes and positions of the inclusions in the composite. This is especially true when the shape of the inclusions is defined prior to composite elaboration, such as in the case of fiber-reinforced composites. For example during the cooling procedure induced by the elaboration of MMCs, residual thermal stresses are formed, either from temperature gradients within the specimen or from a difference of coefficient of thermal expansion (CTE) between the reinforcement and the matrix.

Therefore, unless adaptative coatings are deposited on fibers, metal–matrix processing will always give rise to interfacial induced stresses. However, in certain circumstances, while the thermal induced stresses are very high, no crack in the composite is observed, even for a fragile matrix. In such cases, the induced deformation is accommodated by the formation of dislocations within the matrix.

Not all dislocations would correctly accommodate the residual stress in the vicinity of the interfaces, but the matrix usually reacts by creating the dislocations that lead to crack-free composites. In most cases of a MMC reinforced by fibers, the CTE of the fiber is smaller than the CTE of the matrix. Thus as a general trend, after cooling the fiber is axially stressed in compression and the matrix in tension, while the hoop stress is positive in the matrix and negative in the fiber (Fig. 2). Observations of the dislocations microstructure in the matrix show that the stress is best

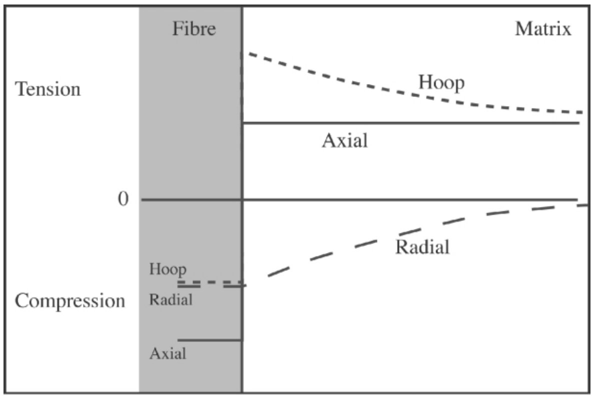

Fig. 2. Schematic representation of calculated thermal induced elastic stress of a fiber with a CTE smaller than that of the matrix.

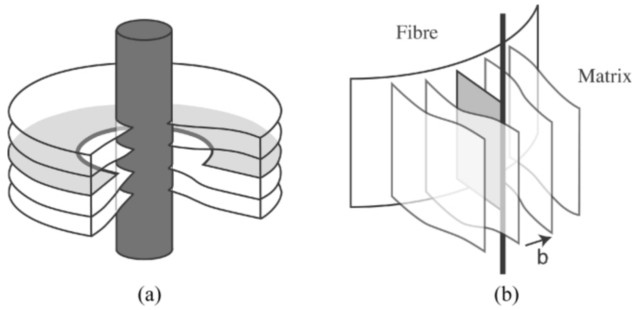

Fig. 3. Accommodation of the residual stress resulting from cooling during composite elaboration in the case of a fiber's CTE smaller than the matrix's CTE. (a) Accommodation of the axial stress by insertion of edge dislocations around the fiber; (b) Accommodation of the hoop stress by insertion of edge dislocations parallel to the fiber.

accommodated by edge or near-edge dislocations, lying in a plane surrounding the fiber for the axial stress, or parallel or close to the fiber axis for the hoop stress (Fig. 3). This was directly confirmed by observations of dislocations created in the matrix during composites formation, for example, in $NiAl/Al_2O_3$ [6], Al_2Cu/Al_2O_3 [7] and Al/Al_2O_3 [8].

4. Direct stress field measurement

Residual stress fields resulting from composite formation can be estimated using elasticity theory for simpler cases, for example when the inclusion takes the form of a very long polycrystalline fiber. However, no general theory can predict stress fields, which are created during the processing of the given (and even simple) inclusions within a given material. Also, accommodation of the stress by creation of dislocations is difficult to include in the calculations. Another way to study the residual stress in the vicinity of the interface inclusion/matrix would be to obtain measurements of the stress fields from direct observations. This can be done by the processing of images from transmission electron microscopy.

In order to measure the strain field around a nanometric precipitate, we first have to image the matrix using high resolution transmission electron microscopy (HRTEM) in the close vicinity of the inclusion. From these HRTEM images of the matrix close to inclusion seen end-on, we extract atomic displacement maps, from which we deduce strain and stress maps. The extraction of the displacement maps from HRTEM images is based on the calculation of the information related to the phase of the image parallel to a chosen diffraction vector. The image of a crystalline material can be numerically decomposed in a set of periodic Fourier components **g**. In a perfect lattice, the phase of each component **g** is constant, while in a distorded lattice, the phase of each component is proportional to **g**·**u**, where **u** describes the displacement field. The determination of the phase is made by centering an aperture around a chosen reflection in the Fourier transform of the HRTEM image,

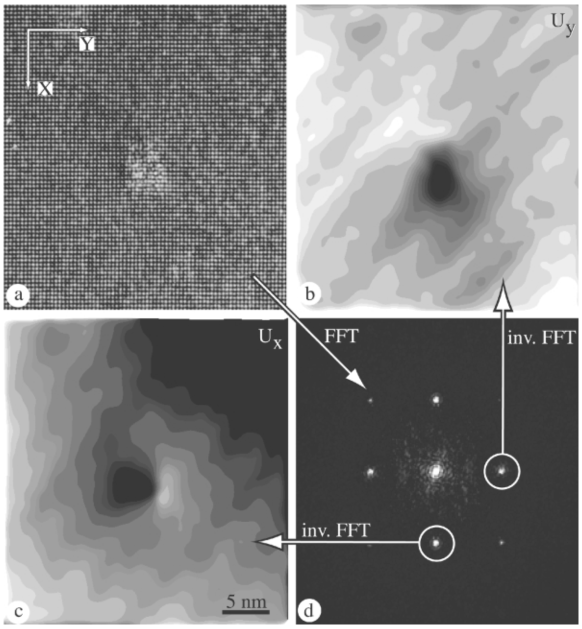

Fig. 4. Schematic representation of a displacement field determination by way of the phase image method. (a) Experimental HRTEM micrograph of a needle-shaped precipitate in a 6056 Al alloy in the T6 state; (b) and (c) components of the displacement field along the X and Y directions, respectively; (d) power spectrum of the Fourier transform of (a).

then performing an inverse Fourier transform. Details of this procedure can be found in Ref. [9].

The result of such an image processing is exemplified in Fig. 4, in the case of a 6056 aluminum alloy. In this material, the precipitates are usually found parallel to the ⟨001⟩ directions. Choosing the (200) and (020) diffraction vectors for the processing allows the extraction of components of the displacement fields u_x and u_y, respectively parallel to the [100] and [010] directions, and the strain and stress fields can simply be derived from the elastic theory.

5. Simulation of dislocations in the vicinity or at the interface

Even when the stress field around inclusions is known, the plastic behavior of the material is difficult to access, essentially because the plasticity of the composite is governed by the movement of dislocations within the matrix in the vicinity of inclusions. As the dislocations interact with the inclusions first, and often only, through the stress field created within the matrix by the inclusions, it is of prime importance to understand the relation between such stress fields and the moving dislocations. This can only be reached by simulating the movements of dislocations within these stress fields and by correlating the obtained predictions with direct observations.

Such an approach has been used in the case of the age-hardened Al–Mg–Si alloy (Fig. 5). The dislocations are described by a set of point linked by straight segments. Each

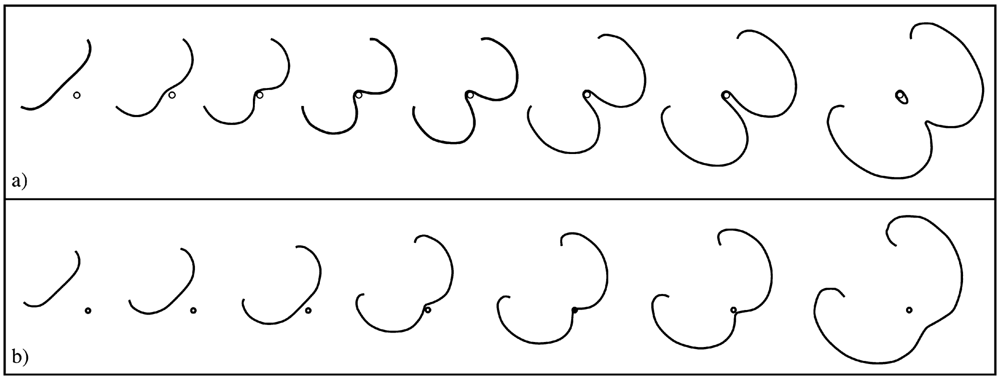

Fig. 5. Simulation of the behavior of a dislocation in the vicinity of a precipitate. (a) Strong precipitate bypassed by the dislocation (Orowan process), observed during 2% deformation; (b) soft precipitate sheared by the dislocation.

point of the dislocation is subjected to the usual dynamics of dislocations and the short and long-range interactions are included [10]. In this aluminum alloy, the simulation shows that the dislocation motion is strongly affected by the presence of nanometric needle-shaped precipitates. The correlation of direct measurement of the stress field around the inclusions together with the simulation of the behavior of dislocations in the close vicinity of the precipitates predicted the activation of the Orowan process, which is bypassing of the precipitates. This was latter confirmed by observations of the microstructure after 2% deformation [10].

6. Dislocations transmission through an interface

Finally, during the deformation of the material the dislocations may interact with the interface by trying to go through. In polycrystalline materials, the difficulty of going through the interface results in the well-known Hall–Petch effect, but in composite materials as inclusions can be often bypassed, such an effect is not always directly observed. However, the transmission of the deformation through the movement of dislocations across the interfaces is still possible and can be observed, provided that the process is not energetically prohibitive, that is, there is an adequate orientation relationship between the matrix and the inclusion or second phase (Fig. 6). Amongst the three general possible cases, only one does not lead to stress accumulation at the interface (Fig. 6a). It results from an exact correspondence of the Burgers vectors of the dislocation before ($\mathbf{b}_1$) and after ($\mathbf{b}_2$) the interface. When the difference is small, the transmission can occur, but implies the formation at the interface of a dislocation with a Burgers vector equal to the difference $\mathbf{b}_1 - \mathbf{b}_2$ (Fig. 6b). When the transmission is not easy, dislocations accumulate in piles-up at the interface, resulting in local high stress accumulation (Fig. 6c). This accumulation can result in nucleation of different dislocations on the other side or on the same side of the interface but moving to the inverse direction, or simply and quite often in cracks nucleation.

The easy transmission is the most preferable case for durability of the composite but it is also the most infrequent. It can however be observed even for an interface between two coexisting but very different phases, as in the Ti_2AlNb alloy [11,12] where the orthorhombic O phase is surrounded by a matrix with a B2 structure (Fig. 7). Even for these two quite different structures, there is a special orientation relationship for which continuity from the (110) plane of the O structure to a (112) plane of the B2 structure exists. In this

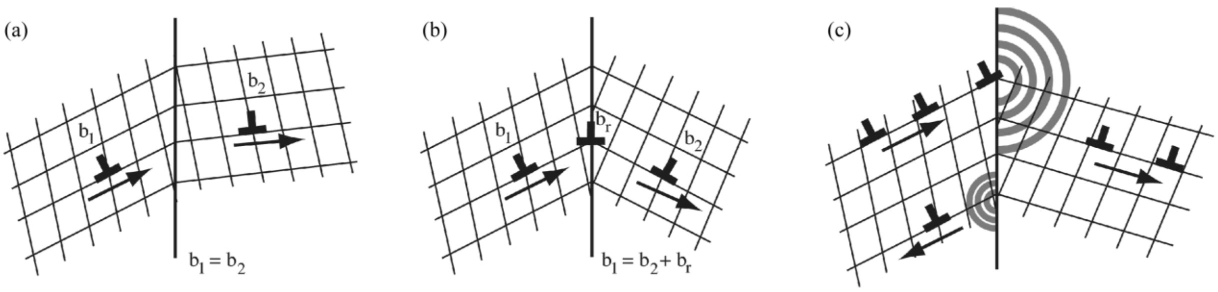

Fig. 6. Different schemes of transmission of dislocations through an interface. (a) Direct transmission resulting from a perfect correspondence of the Burgers vectors before and after the interface; (b) transmission with creation of a product dislocation at the interface; (c) indirect transmission: dislocations cannot cross the interface and new dislocations are created beyond the interface due to the stress accumulation. Notice that the accumulated stress can also be released by the formation of dislocations moving backward relative to the others.

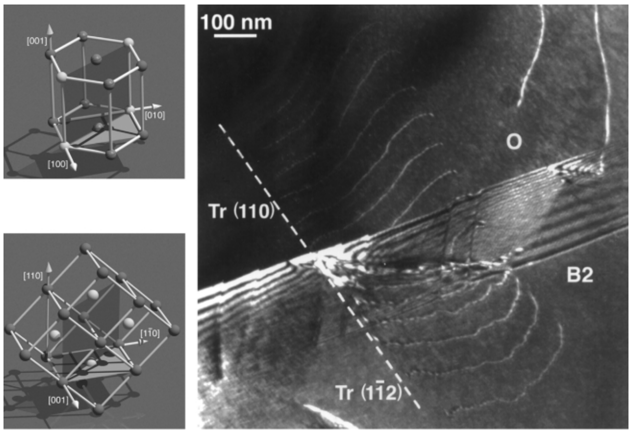

Fig. 7. Example of an easy transmission of the deformation through an interface between two phases with very different structures: ordered orthorhombic O and ordered cubic centered B2. According to the observed lattice correspondence $[001]_O \approx [110]_{B2}$ and $1/2[100]_O \approx [001]_{B2}$, the (110) glide plane of the O phase is parallel to the $(1\bar{1}2)$ glide plane of the B2 phase. In this orientation relationship, the dislocations with a $1/2[1\bar{1}0]$ Burgers vectors in the O phase transform into $[\bar{1}11]$ dislocations in the B2 phase with only a very small residue left at the interface (TEM Jeol 200 CX operating at 200 kV).

orientation relationship, dislocations gliding in the O phase with a 1/2[110] Burgers vector transmit to the B2 phase into dislocations with a $\langle 111 \rangle$ Burgers vector leading to almost no residual stress at the interface.

7. Conclusions

Residual stresses play a significant role on the mechanical properties of MMC. This residual stress can be partly relaxed during composite elaboration by an adequate choice of the microstructure or by dislocations nucleation. Except in simple cases, the stress cannot be simply modelized accurately in the close vicinity of the interfaces and must be measured.

As the interfaces between the inclusions and the matrix play the role of (strong) obstacles for slip transfer, the influence of these interfaces on the plasticity of the composite must also be taken into account very precisely. In addition, the dislocations behavior must be simulated, in order to relate the local stress field resulting from the existence of the interfaces and the mechanical properties of the composite.

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