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A STUDY OF INTERNAL DAMAGE OF METAL MATRIX COMPOSITES BY NEUTRON DIFFRACTION

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ABSTRACT

Using neutron diffraction, we have measured the elastic phase strains of Al/TiC and Al/SiC composites under uniaxial tensile loading. The phase strains were used to reconstruct the global elastic strain. It has been found that, above macroscopic yield, the global elastic strain response is not linear. A theoretical model shows that the nonlinearity is dictated by changes in the ratio of longitudinal phase stresses. Furthermore, the changes in this ratio resulting from matrix plasticity and reinforcement fracture are different which leads to distinct slope changes in the global elastic strain response that can be used to distinguish the onset of these two processes on the global elastic strain loading curve.

INTRODUCTION

A strain gauge measurement is often referred to as the global average strain or macrostrain and includes both elastic and, if yield has been exceeded, plastic components. For a homogeneously deforming material the gauge measurement is identical to the local strain at any location within it. This means that the elastic component of the global strain (hereafter referred to as the global elastic strain) follows the Hooke's law, even after the onset of plastic deformation. For a heterogeneous material in which the constitutive response of the constituents differs the response of the global elastic strain following plastic deformation of one or both phases is more complicated and, counterintuitively, may not obey Hooke's law. This does not mean that at any discrete point within the material that Hooke's law is not obeyed, but that the *bulk* elastic response is affected by the developing deformation pattern between phases. This provides a tool to represent and distinguish microscopic mechanical behavior during loading.

In a metal matrix composite (MMC) stiff reinforcement particles are embedded in a ductile matrix. The strain measured by a strain gauge is an average over all of the individual grains reflecting differences in the constitutive responses of each phase. However the strain gauge can not separate the elastic behavior of the indidual phases. By using diffraction based experimental techniques which implicitly measure only elastic strains it is possible, as will be shown below, to reconstitute the elastic component of the global strain during loading.

We measured elastic phase strains in two MMC systems as a function of applied stress and reconstituted the global elastic phase strain response over the whole loading regime. By comparing the experimental results with finite element and theoretical predictions we have developed a tool for representing the effects of matrix plastic flow and reinforcement fracture.

BASIS FOR RECONSTRUCTION OF GLOBAL ELASTIC STRAIN

For displacements along the boundary of a domain D, the following expression can be obtained from Gauss' theorem:

$$\int_D \partial_k u_i \delta_{jk} dD = \int_{|D|} u_i \delta_{jk} n_k dS, \quad (1)$$

where u_i is the displacement along the ith direction, and n_k is the plane normal of the boundary $|D|$. δ_{ij} is the Kronecker δ. The repeated subscript indices are summed over 1, 2 and 3. The ∂_k denotes a partial derivative along direction k. When $i = j$, the normal strains in the integrated volume D are related to the far-field displacement u_i at the boundary $|D|$ by Eq. 1.

For a two-phase composite consisting of a matrix (m) and reinforcement (r), Equation 1 reduces to:

$$(1-f)\{\varepsilon_m\} + f\{\varepsilon_r\} = \{\bar{\varepsilon}\}, \quad (2)$$

where $\{\bar{\varepsilon}\} = [u_1/L_1, u_2/L_2, u_3/L_3]^T$ is the global normal strain vector, and f is the volume fraction of the reinforcement. Vectors $\{\varepsilon_m\}$ and $\{\varepsilon_r\}$ are the average normal phase strains. Equations 1 and 2 also apply for the part of displacements corresponding to the elastic strain component. That is, the elastic component of the global strain, $\{\bar{\varepsilon}^e\}$, can be obtained by a volume-average of the elastic phase-strains.

THE GLOBAL ELASTIC STRAIN RESPONSE OF A COMPOSITE BEYOND YIELD

For a composite, each increment of an applied stress invokes a partition of the corresponding increase of internal stress between the constituents. For loading along the axes of the global coordinate system,

$$(1-f)\{\Delta\sigma_m\} + f\{\Delta\sigma_r\} = \{\Delta\bar{\sigma}\}, \quad (3)$$

where $\{\Delta\bar{\sigma}\}$ and $\{\Delta\sigma_\alpha\}$ are vectors representing the applied and the phase stress increments, respectively, consisting only of normal stresses. To describe the partitioning of stresses, we define a principle stress ratio, $R_k = (\Delta\sigma_k)_r/(\Delta\sigma_k)_m$, between phases, where σ_k, $k = 1$, 2 and 3, are the average principle stresses. This quantity will be referred to as the load sharing ratio.

We define a tensor $\boldsymbol{R}$:

$$\boldsymbol{R} = \boldsymbol{\Delta\sigma}_r \boldsymbol{\Delta\sigma}_m^{-1} = \boldsymbol{C}_r \boldsymbol{\Delta\varepsilon}_r^e (\boldsymbol{\Delta\varepsilon}_m^e)^{-1} \boldsymbol{C}_m^{-1} \quad (4)$$

where $\boldsymbol{C}$ is the stiffness tensor; $\boldsymbol{\Delta\varepsilon}_\alpha^e$ is the corresponding elastic component of the phase strain. When the principle directions of the average phase stresses correspond to the axes of the global coordinate system, only the diagonal elements of $\boldsymbol{R}$ are non-zero and correspond to the load sharing ratio, R_k. This condition holds for most MMCs, whose properties are usually, at a minimum, orthotropic.

By substituting Eq. 4 into Eqs. 3 and 2, we get:

$$\Delta\bar{\sigma} = [\boldsymbol{I} + f(\boldsymbol{R} - \boldsymbol{I})]\Delta\boldsymbol{\sigma}_m; \quad (5)$$

$$\Delta\bar{\varepsilon}^e = [\boldsymbol{I} + f(\boldsymbol{C}_r^{-1}\boldsymbol{R}\boldsymbol{C}_m - \boldsymbol{I})]\Delta\boldsymbol{\varepsilon}_m^e. \quad (6)$$

From Eqs. 5 and 6, we define a tensor $\boldsymbol{L}^t = \Delta\bar{\sigma}(\Delta\bar{\varepsilon}^e)^{-1}$. For loading along the 3-direction, the element L_{33}^t in tensor $\boldsymbol{L}^t$ is the tangent modulus (or slope) of the stress vs. elastic global strain ($\bar{\sigma} \sim \bar{\varepsilon}^e$) curve. That is,

$$L_{33}^t = \frac{(\lambda_m + 2G_m)[1 + f(R_{33} - 1)]}{1 + f\left[\frac{R_{33}}{E_r}(\lambda_m + 2G_m) - 1\right]} \quad (7)$$

where G and E are the shear and Young's moduli and λ is the Lame's constant, R_{33} is the load sharing ratio along the loading direction. Thus the slope of the global elastic strain plotted against the applied stress depends uniquely on the load sharing ratio. By taking derivative of L_{33}^t with respect to R_{33}, it can be shown that the derivative is positive when approximately $E_r/E_m > 2$, leading to slope changes in the same sense as the load sharing ratio, R_{33}.

The correspondence between L_{33}^t and R_{33} has important implications. In a ceramic reinforced metal matrix composite, the ceramic remains elastic throughout loading. Upon initial yield of the matrix, the reinforcement takes a larger share of the longitudinal internal stress, resulting in an increase in slope, L_{33}^t. However, if a deformation mode is initiated which causes a shift of stress back to the matrix (for example fracture or debonding of the ceramic particles) the slope will decrease. This opposing trend in slope change offers a possible experimental mechanism to identify the onset of the particle fracture during deformation, provided that a sufficiently large volume of material is affected.

NEUTRON DIFFRACTION VERIFICATION

To verify the relationship between L_{33}^t and R_{33} predicted by Eq. 7, and their value in describing the physical mechanisms of deformation in MMCs we measured phase strains in two MMC systems; a 2219 Al alloy reinforced with 15 volume percent (V%) TiC particles and a 2080 Al alloy reinforced with 15 V% SiC particles. The two sytems are expected to show different post yield behavior, specifically the TiC particles do not appear to fracture during deformation [1], while the SiC usually does [2]. This allows us to compare slope changes due to matrix plasticity alone with the added effects of reinforcement fracture. Cylindrical tensile bars were machined from rolled plates and heat treated to a T6 condition.

Each specimen was incrementally loaded at room temperature to a stress level sufficient to induce plasticity. Creep was not observed for either material. Phase strains were measured at each incremental load by neutron diffraction. Details of the neutron diffraction technique are described in ref. [3]. Lattice strains are measured for many individual Bragg reflections. For comparison with calculations based on a mixture of two homogeneous solids, mean phase strains were obtained by processing the experimental data using

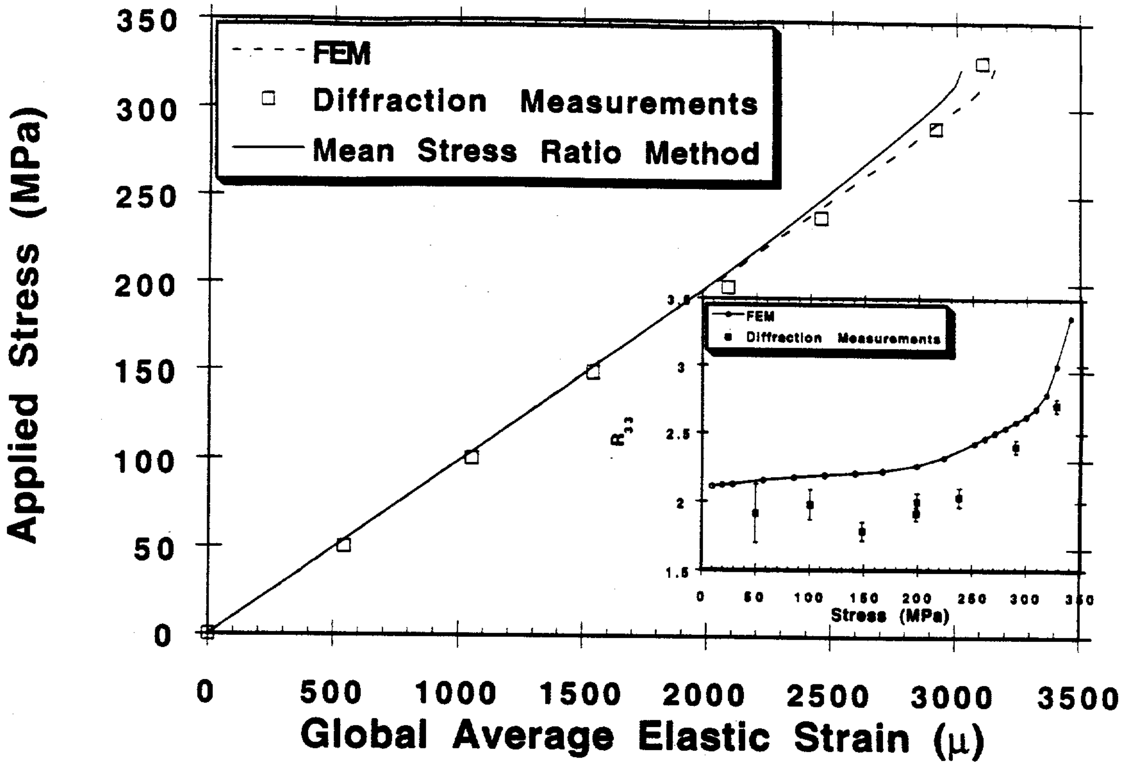
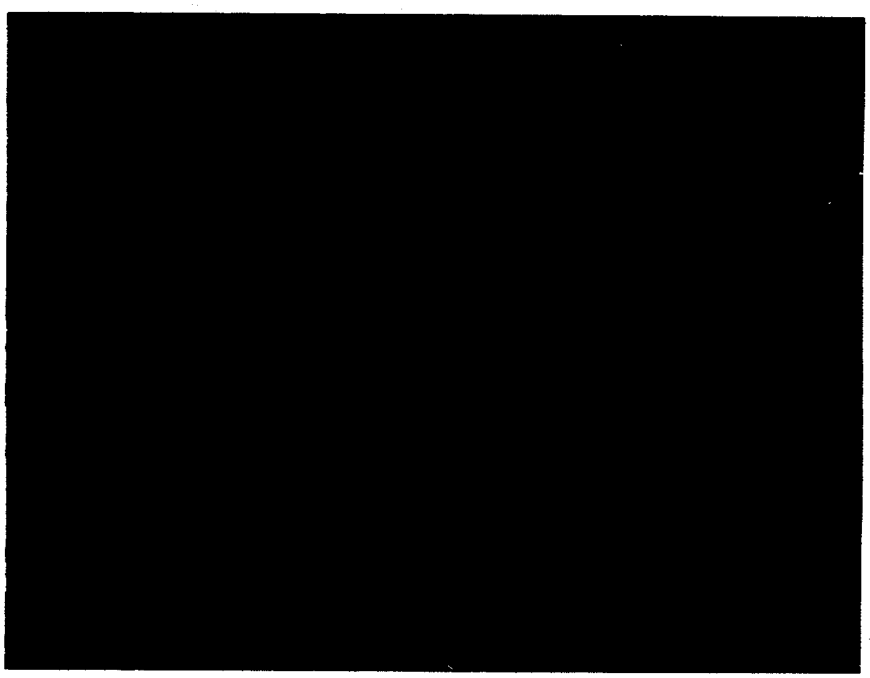


Figure1: Global elastic strain as a function of applied stress for Al/TiC MMC, and changes in R_{33} as a function of the applied stress are shown in the inset.



⇐ **Loading Direction** ⇒

Figure 2: Optical micrograph (500x) showing TiC particles in Al/TiC MMC are not fractured after loading (500x).

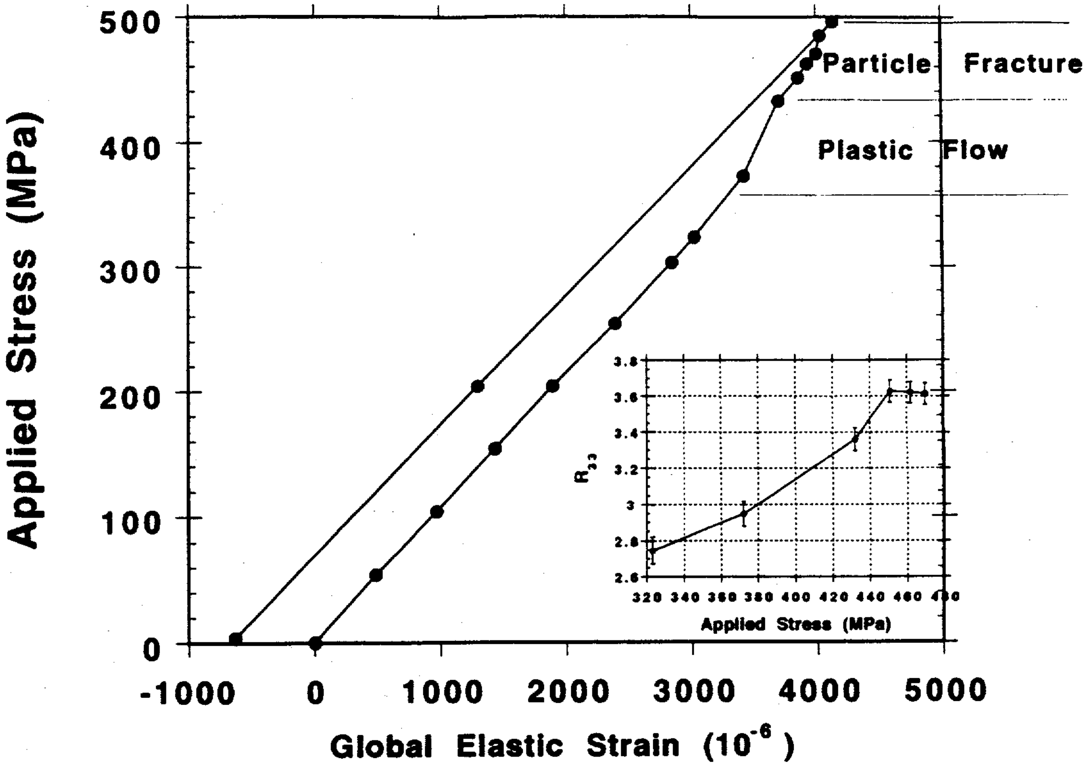
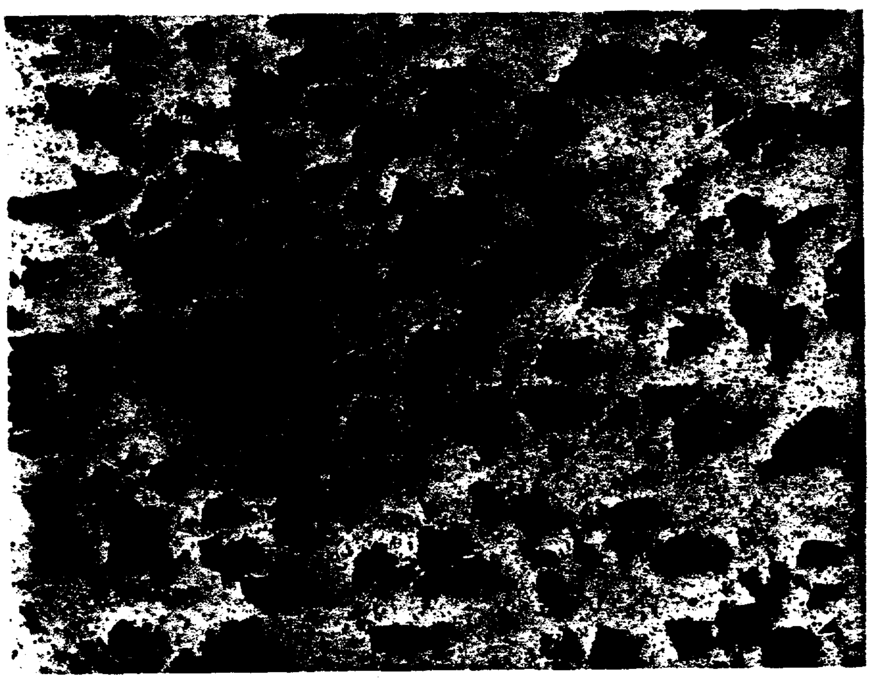


Figure 3: Global elastic strain as a function of applied stress in Al/SiC MMC, and changes in R_{33} as a function of the applied stress in the inset.



⇐ **Loading Direction** ⇒

Figure 4: Optical micrograph (500x) showing many SiC particles in Al/SiC MMC are fractured (some are marked by arrows) after loading.

Rietveld Refinement [4]. The global elastic strains were reconstructed from the individual phase strains through volume average.

To verify the $L_{33}^{t} \sim R_{33}$ relationship, an elasto-plastic three dimensional finite element (FE) model was employed, in which a periodic array of cubic particles is embedded in an infinite matrix. Details of the modeling are given elsewhere [5].

Figure 1 displays $\bar{\sigma} \sim \bar{\varepsilon}^{e}$ curve for the Al/TiC composite. The squares in the figure are deduced from diffraction phase strain measurements using Eq. 2. The dashed line shows the FE results. As the load sharing ratio (R_{33}) increases (see inset of Fig. 1) due to matrix plastic flow, the slope of the $\bar{\sigma} \sim \bar{\varepsilon}^{e}$ curve increases. Taking the R_{33} calculated by FE and substituting into Eq. 7, good agreement is obtained with the experimental data (solid line in Fig. 2). Figure 2 shows the microstructure of the gauge section after unloading. No particle fracture can be detected.

By contrast with Fig. 2, Figure 3 shows the slope changes in the $\bar{\sigma} \sim \bar{\varepsilon}^{e}$ curve of the Al/SiC composite. The slope of the loading curve displays a two-tone behavior. At about $\bar{\sigma}$ = 360 MPa, corresponding to initial yield of the matrix the slope increases, then in excess of 440 MPa, the slope decrease suggesting internal stresses are being transferred back to the reinforcement according to Eq. 7, i.e. due to decreases in R_{33} as shown in the inset of Fig. 3. The most probable cause for this decrease in load sharing ratio is reinforcement fracture. Figure 4 displays the microstructure from the gauge section after unloading. Particle fractures can be identified in the gauge section (Fig. 4), while particles are mostly intact in the grip section of the specimen (not shown) which experienced a much smaller load.

CONCLUSIONS

(1) For MMCs, the elastic component of the global strain is not linearly related to the applied load when the phase yield strengths are different.

(2) The changes in slope of $\bar{\sigma} \sim \bar{\varepsilon}^{e}$ curve is uniquely correlated to the ratio of the longitudinal phase stresses.

(3) Changes in the slope of $\bar{\sigma} \sim \bar{\varepsilon}^{e}$ curve due to matrix plastic flow and reinforcement fracture can be used to identify the onset of the reinforcement fracture in MMCs.

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