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Local and global measures of the fracture toughness of metal matrix composites

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ABSTRACT

The fracture behavior of two metal matrix composites (MMCs) at different aging conditions is investigated both at the global and local level. The study is focused on the processes of void initiation near the crack tip by either particle fracture or particle/matrix decohesion, and on fracture initiation. The global fracture properties are determined by conventional fracture mechanics tests. The local conditions for void and fracture initiation are determined using fractographic analyses and subsequent analytical computations, based on the HRR-field equations, mean-field theory, and a simple analytical model from literature. The correlation between the global and local fracture properties is studied. The applicability of existing models to predict the fracture toughness of MMCs is discussed. The results of the analyses suggest that the maximum principal stresses in the particles at the moment of void initiation are not constant, but exhibit a dependency on the composite yield strength.

1. Introduction

1.1. Global deformation behavior of MMCs

Despite of the attractiveness of metal matrix composites (MMCs) because of the increased strength and stiffness, their low fracture toughness in comparison to the fracture toughness of the unreinforced matrix material is a primary obstacle against their widespread application in engineering. There is a significant body of research into the deformation behavior and microstructure–properties relationships of MMCs [1–6]. Due to the absence of appropriate experimental tools, however, both the material properties and the microstructural parameters have been considered mostly at the global level. Various relationships have been worked out to relate strength and fracture toughness to global microstructural parameters (particle volume fraction, average particle size, particle shape, particle orientation, etc.) [7–10]. These experimental results can be roughly generalized as follows [11]: stronger matrix alloys produce stronger composites, see e.g. [12,13]; the increase in strength due to the reinforcement decreases with increasing matrix strength; in the case of high-strength alloys, the reinforcement may even lead to a reduction in strength [14]; yield strength and ultimate tensile strength improve with increasing particle volume fraction, however, this leads to lower ductility and fracture toughness [15]; decreasing particle size yields to increasing tensile strength [16,17].

There is also a significant body of theoretical research into the mechanical behavior of composite materials, see e.g. [18–31]. Whereas the prediction of thermo-elastic properties has reached a very high standard, the extension of these methods to the plastic regime is still somewhat problematic. For example, the deformation behavior of a model composite, which consists of elastic ellipsoids embedded in an elastic–plastic matrix, in uniaxial tension has been simulated recently by means of first- and second-order homogenization models and comparative finite element (FE) simulations [22]. It has been found that accurate predictions of the global stress–strain behavior do not guarantee accurate predictions of the behavior at the microscopic level, i.e. in the different phases. In comparison to FE simulations, the homogenization models become less accurate when the aspect ratio of the ellipsoids increases. In another example [26], the mechanical response of metal–ceramic composites was analyzed. It was modeled by considering a von Mises plasticity model for the metal phase and a Drucker–Prager associative elastic–plastic material model for the ceramic phase. FE modeling of the stress–strain response of the axisymmetric unit cell was carried out to validate the iterative homogenization method. In [27], ductile failure of a dual phase steel, consisting of a hard martensite phase and a soft ferrite matrix, was studied using elastic–plastic finite element analyses. Failure was predicted as a natural outcome of plastic instability during the deformation process in form of localized plastic straining in the representative volume element.

The FE modeling of heterogeneous materials relies on the concepts of a representative volume element and a repeating unit cell. A critical review of these two concepts has been presented recently in [28]. The importance of interface properties (interface strength

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and toughness) was outlined in [23]. It was shown in this parametric study that lower values of interface strength and toughness result in earlier interface decohesion and a moderate reduction of the composite strength. The influence of the interface strength on the ductility shows a more complex character with a minimum of the ductility at a certain value of interface strength. In a numerical study [29], the ratio between interface debonding and particle cracking was investigated for certain MMC-configurations. It was shown that by increasing or decreasing the strength of an interface layer, the onset of fracture initiation can be affected. The effect of particle clustering on mechanical properties of composite materials was studied in [24]. It was shown that an inhomogeneous particle distribution results in a reduction of the composite flow stress and ductility.

1.2. Global fracture behavior of MMCs

Different models have been proposed to predict the fracture initiation toughness of MMCs [2,30–32]. All models present fracture toughness as a function of global mechanical properties (yield strength, Young's modulus, etc.) and averaged microstructural parameters (particle size, particle volume fraction, etc.). In this paper, we will focus our attention on two models. Hahn and Rosenfield's ductile fracture model [30], is the most commonly used model for MMCs. Assuming that crack growth will occur if the extent of the heavily deformed zone ahead of the crack tip, i.e. the process zone, becomes comparable to the width of the unbroken ligaments separating cracked particles (or the interparticle distance), Hahn and Rosenfield calculated the fracture toughness as

$$K_{IC} = \left[2\sigma_y E\left(\frac{\pi}{6}\right)^{1/3} d\right]^{1/2} f^{-(1/6)}. \quad (1)$$

In Eq. (1), σ_y is the yield strength of the material, E the Young modulus, d the particle size, and f the particle volume fraction. The model gave a good prediction for some types of MMC, with accuracy of 30% [30]. However, the data calculated by Eq. (1) were far from the experimental results for other types of MMC, especially for Al7075–10%SiC MMCs tested in [31,32]. Thus, Majumdar and Pandey [31] modified this model, first by postulating that failure occurs when a critical strain is reached at a characteristic microstructural distance ahead of the crack tip, which is identical to the interparticle spacing. Then they replaced the assumption of a certain magnitude of the critical strain by the condition that the length of the process zone of the blunted crack equals the interparticle spacing [33]. This leads to the relation [31],

$$K_{IC} = 0.77\left[\frac{\beta\sigma_y E d}{d_N(1-\nu^2)}\right]^{1/2} f^{-(1/6)}. \quad (2)$$

In Eq. (2), the dimensionless constant d_N depends on the strain hardening coefficient, $n = 1/N$, and on the reference strain, $\varepsilon_0 = \sigma_y/E$ [34]; the constant $\beta = 0.5$, and ν is the Poisson's ratio. In comparison to Eq. (1), the modified model gave a better fit to the experimental fracture toughness values for the MMCs studied in [31].

1.3. Local fracture behavior of MMCs

Experimental tools have been developed recently that allow us to study the deformation and fracture behavior of materials at the local level [35–38]. These tools are in situ tensile, compression, or fracture mechanics tests performed in the scanning electron microscope (SEM), a local deformation analysis system and an automatic fracture surface analysis system. Applying these tools, it has been demonstrated that the local deformation and fracture behavior may vary significantly within a single MMC specimen, depending on the local microstructure of the composite material. For instance, the effects of heat treatment and homogeneity of the particle distribution on the inhomogeneity of plastic deformation in Al6061–10%SiC MMCs were investigated by in situ tensile tests in the SEM and automatic local deformation analyses [39]. It was found that even the unreinforced matrix material deforms inhomogeneously, exhibiting a deformation band pattern which is nearly independent of the loading stage. MMCs with small particles (10 μm) exhibit a similar deformation band pattern, which is controlled by the particle arrangement. The fracture of small particles does not induce far-reaching deformation bands and, therefore, these MMCs have a high strength and ductility. MMCs with coarse reinforcements (100 μm) are characterized by early particle fracture or decohesion. This induces damage-controlled deformation bands in the matrix. For small and large particles, clusters lead to enhanced damage localization and lower ductility. The same material was investigated in [40], but the particle volume fraction was also varied. The deformation band pattern observed were classified into three different types: (1) particle-arrangement controlled, or (2) damage-controlled deformation bands, both oriented about 45° to the loading axis, and (3) damage-induced deformation bands perpendicular to the loading axis. The global fracture strain was related qualitatively to the deformation band pattern. Materials with pattern (1) have the highest ductility; pattern (3) appears, in general, in materials with very low ductility. The critical crack tip opening displacement, where particle cracks proceed into the matrix material, was in the MMCs with 100 μm particles one order of magnitude larger than in the MMCs with 10 μm particles. This finding, which is important for numerical modeling aspects, was ascribed mainly to the different pre-deformation of the matrix material at the moment where particle fracture occurs [40].[1]

In [41], the effect of the local architecture in front of the crack tip (the distance between particle and crack tip, r, and the angle of particle location with respect to crack plane, θ) on the local conditions for void and fracture initiation was studied in an Al6061–10%Al_2O_3 MMC. It was shown that the angle of particle location, θ, can significantly influence the local fracture behavior, whereas no noticeable effect of the distance, r, between particle and crack tip on the local fracture properties was found. The effect of the particle size on the local conditions for particle/matrix debonding was studied in [42] for a mild steel St37 with MnS-inclusions. An increasing diameter of MnS-inclusions from 2.5 μm to 7 μm leads to a decrease of the maximum interfacial stress by 50%, from 1480 MPa to 960 MPa. A strong effect of the homogeneity of the particle distribution on the local and global fracture properties of an Al6061–20%Al_2O_3 MMC with a particle size between 1 μm and 5 μm was reported in [43]. With increasing uniformity of particle distribution, the global fracture initiation toughness increased from 1.5 kJ/m^2 to 2.7 kJ/m^2. It was found that the local fracture initiation toughness in diffuse particle clusters can be a factor 10 lower than in regions with a homogeneous particle distribution. The relation between the local and the total work of fracture in an Al–50%Al_2O_3 composite material was investigated in [44]. The local work of fracture in the process zone was estimated using detailed fractographic examinations and sim-

[1] To understand the relevance of this finding, assume two specimens of unreinforced, microductile matrix material with a rather low crack growth resistance. Assume in one specimen a 10 μm long pre-crack, in the second one a 100 μm long crack. In a fracture mechanics test, both specimens should exhibit the same fracture initiation toughness, i.e., the same COD_i. In the MMCs with small and large broken particles, the cracks are surrounded by hard particle material on the upper and lower crack flanks. The matrix material will feel a constraint effect due to the interface, but the conditions when the crack starts to propagate into the matrix should not be so different from the cases of the pre-cracked unreinforced matrix materials. Note that, if COD_i differs by a factor 10, the local strain in the matrix material just in front of the crack tip will also differ by a factor 10.