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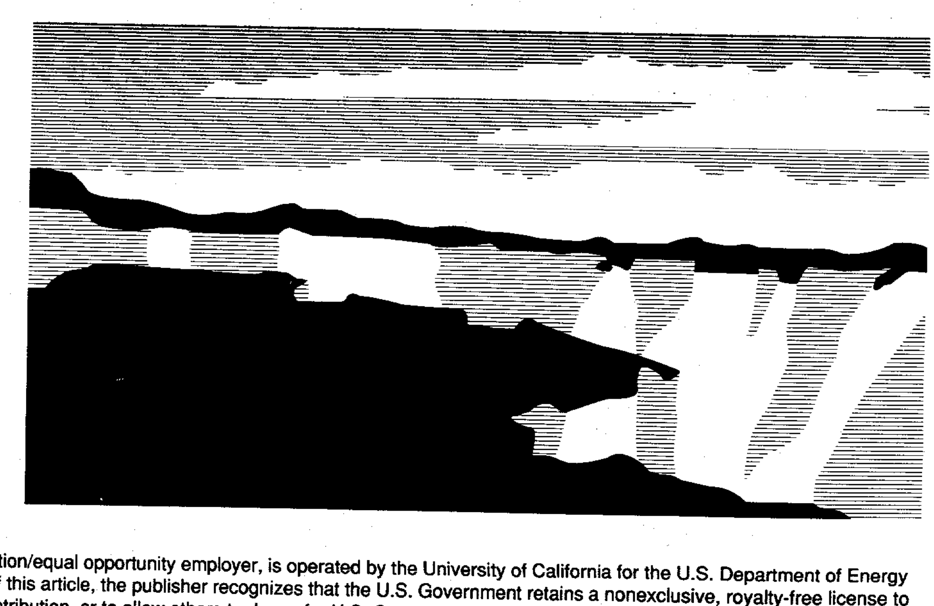

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COMPOSITE DUCTILITY-THE ROLE OF REINFORCEMENT AND MATRIX

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Abstract

The total ductility, i.e., the energy required to rupture a ductile solid is controlled by two factors: (1) the ability to withstand strain hardening before initiation of critical nuclei for final rupture (microcracks or voids); (2) the resistance to crack extension. The first group is related to the reinforcement; particle or whisker fracture, debonding and void nucleation and growth at the reinforcement-matrix interface. The second group is related to the matrix; triaxial stress state in matrix, the work hardened state of the matrix due to the ΔCTE effect, and localized matrix plastic flow. The experimental data and theoretical analysis show that the lack of strain hardening capacity of the matrix is the major contributing factor causing the low ductility of DMMCs. This may be attributed intrinsically to the exhaustion of the strain hardening capacity from ΔCTE-induced work hardening and/or extrinsically to the high matrix triaxial stresses which induces rapid void growth.

Introduction

The relatively low ductility is one of the major limitations of discontinuous metal matrix composites (DMMCs). The total ductility, i.e., the energy required to rupture a ductile solid is controlled by two factors: (1) the ability to withstand strain hardening before initiation of critical nuclei for final rupture (microcracks or voids); (2) the resistance to crack extension. The experimental and theoretical efforts that have been undertaken to understand the low ductility of DMMCs can be divided into two main categories. The first group is related to the reinforcement; particle or whisker fracture, debonding and void nucleation and growth at the reinforcement-matrix interface. The second group is related to the matrix; triaxials tress state in matrix, the work hardened state of the matrix due to the ΔCTE effect, and localized matrix plastic flow.

We will concentrate on the ΔCTE effect, i.e., exhaustion of work hardening capacity, and the localized matrix plastic flow aspects, and discuss their contributions to the ductility of DMMCs. However, we shall also discuss other aspects of ductility of DMMCs.

Discussion

The failure of SiC/Al DMMCs as revealed by fractography may be categorized into three modes; fracture of reinforcement and/or large intermetallics in the matrix, particle-matrix interfacial debonding, i.e., voids, and ductile failure of matrix. The discussion will be divided into three parts.

Fracture of Reinforcement

Clegg et al. [1] and Llorca et al.[2], while studying Al_2O_3 fiber- and SiC particulate-reinforced Al composite, respectively, observed that reinforcement fracture was the main failure mode for these composites. The damage process was found to follow Weibull statistics [2,3]. To identify critical factors that dictated the fracture of reinforcement particles in a composite, Llorca et al. [2] studied the evolution of particle fracture in the composites with different heat treatment. They found that the amount of fracturing particles was correlated with the matrix strength. The higher the matrix strength, the higher proportions of fractured particles they could detect. From FEM modeling, they suggested that particle fracture was dominated by load transfer to the particle. With an increasing load-carrying ability for the matrix, the corresponding particle stress would also increase, which then increased the tendencies for particle fracture. The particle-stress model [2] is also consistent with experimental results by Lloyd [4] who has observed in SiC/6061 Al composites more fractured particles along the fracture surface under a top-aged T6 temper than those in an as-solution-treated T4 temper.

Flom and Arsenault [5] studied the effect of particle size on the fracture of DMMCs. They found that reinforcement fracture was rarely observed when the particle size was small (< 20μm). For larger sizes there was larger than average fraction of particles at the fracture surface. They then argued that for small particles the composite rupture process was dominated by matrix ductile failure. However, in another study, Wu and Arsenault [6] made an in-situ SEM observation of crack propagation in a SiC/Al composite, and found that cracks propagated through existing "broken" particles which had already existed prior to loading. This accounts for the higher average density of SiC particles on the fracture surface, and even the presence of matching pairs of SiC particles. Christman et al. [7] studied fractography of SiC_w/2124 Al composite. They noted that the fracture surfaces of both the reinforced and unreinforced alloys exhibited similar fine dimple structures. They suggested that the failure of composite was predominantly through the matrix not along the matrix-reinforcement interfaces.

In an investigation to determine the bond strength, the following experiment was undertaken. A sphere of SiC was produced by electrical discharge machining (EDM) a large single crystal of SiC. The sphere was ~ 3 mm in diameter. The tensile sample was made by hot compaction of wrought 1100 Al alloy, so as to allow significant plastic deformation at the SiC-Al interface. This procedure result in a strong bond between SiC and Al. The resulting composite was machine into tensile samples as shown in Fig.1. The samples were tested at room temperature and Fig.2 is schematic of the stress-strain curve and tested

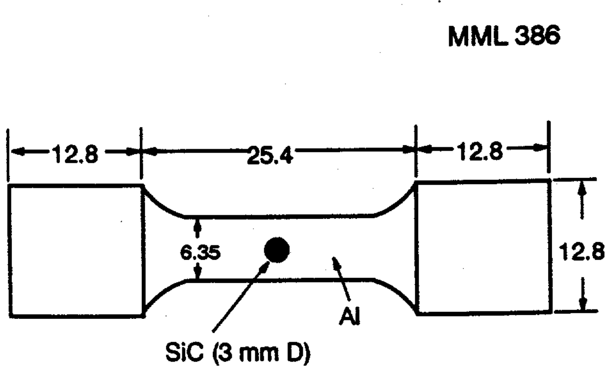

Figure 1: A schematic of the tensile sample with SiC sphere, all dimensions are in mm.

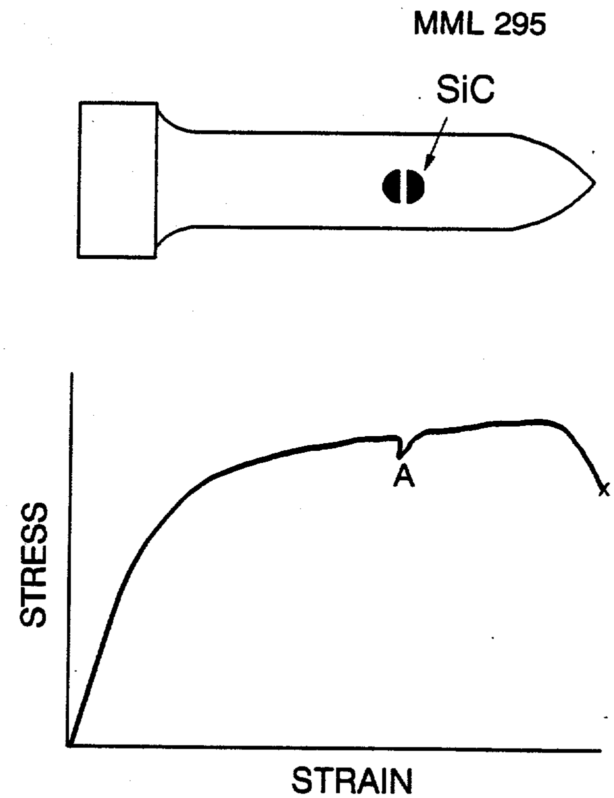

Figure 2: Schematic diagrams of the tensile sample after testing and the stress-strain curve.

sample. At point A you will note there is a downward blip in the curve, and at point A there was an audile sound, that is something had cracked. The sample continued to deform and finally necked down and fracture at a place some distance from the original SiC sphere. The sample was then electrical discharged machined longitudinally and as suspected the SiC sphere had cracked as shown in Fig.2. This result is very significant for several reasons. First, it certainly disproves any load transfer mechanism of particulate composite strengthening. Once the SiC cracks, there is no possible load transfer and the cross sectional area of the Al matrix is reduced by 25% as compared to the remainder of the Al in the sample gage length. In other words, the effective diameter of the Al is reduced to 5.50 mm were as the diameter of the reminder of the samples is 6.35 mm. These means that work hardening of the matrix due to the ΔCTE effect is greater than the work hardening due to tensile deformation. Secondly, in terms of fracture, even though the SiC sphere fractured, the crack did **NOT** propagate into the matrix. If we now consider the stress intensity factor at the tip of the crack, just as it forms and the stress intensity factor at the tip of the crack in a fractured 10 μm diameter particle, the difference is more than an order of magnitude. All other parameters being equal, if the crack does not propagate into the matrix when a 3.17 mm diameter particle fractures why should is propagate when a 10 μm particle fractures.

Debonding and Void Formation at Reinforcement/Matrix Interface

A failure mechanism by void nucleation and growth at the particles drawn directly parallel from dispersion hardened alloys is often debatable due to the particle size scale difference. In DMMCs with well bonded reinforcements, it may require a substantial amount of energy to form voids of size comparable to that of the reinforcements, and other failure mechanisms may be more favorable, including void nucleation and growth away from the interfaces [8]. Arsenault and Pande [9] studied by Auger microprobe the surfaces of whiskers on the composite fracture surface, and found evidence that the surface of the whiskers was coated with Al. Therefore, they suggested that what appeared to be debonding was actually near-interface matrix failure.

Li et al. [10] using a quantum chemical approach studied the Al--SiC interfaces, and concluded that the bonding strength was higher than the theoretical strength of Al (in the absence of interfacial chemical reactions).

The extent of void nucleation and growth (at the particles and/or in the matrix) can often be revealed through material uniaxial tensile behavior under a superimposed pressure [11]. Vasudevan et al. [12] studied the effect of superimposed pressure on particulate- and whisker-reinforced composites. They found that pressure had marked effect on both the composite flow behavior and ductility. The flow stress and ductility of the composites increased as a function of superimposed pressure while no difference was observed for the unreinforced alloy. Mahon et al. [13] noted that a superimposed pressure did not enhance the ductility of the composite. This result suggested that the composite failed by ductile tearing or shear localization (not by void nucleation and growth). Fractography by Davidson [14] revealed that in the composite he tested extensive "tearing ridges" appeared on the fracture surface, which he suggested as an indication for ductile tearing.

The presence of a high matrix triaxial stress can also decrease the matrix in-situ ductility of DMMCs. Embury [15] suggested that void nucleation might dominate the ductility since the subsequent stages of void growth and coalescence might be extremely rapid due to high matrix triaxiality Rice and Tracey [16] predicted a spherical void dilatation under triaxial stresses. This could be a possible explanation of the results found by Flom and Arsenault [5]; that is no additional voids were observed immediately below the fracture surface of DMMCs. Although the concept of rapid void growth originated from void nucleation at the particles in dispersion-strengthened alloys, it could also apply to other situations of void nucleation and growth under high triaxial stresses, for example, within the matrix. However, there are other experimental evidence that dispute the theory of rapid void nucleation-controlled failure. Whitehouse and Clyne [17] measured the evolution of voids in DMMCs by monitoring the composite density. They showed a gradual accumulation of cavitations. This suggests that void nucleation may not be a catastrophic process.

Llorca et al. studied the void nucleation and growth process using FEM modeling. By assuming that the processes of void-nucleation and void-growth were controlled separately by the effective plastic strain and the dilatational stresses respectively [18], they showed that for whisker reinforced composites void nucleation at the tip of the whisker was dominating due to high matrix triaxial stresses; whereas for spherical composites substantial void growth could exist. This suggests that the void content should be higher in a spherical composite. However, this prediction was contrary to the experimental observations by Whitehouse and Clyne [17] who showed that more voids were detected at all strain levels in a whisker composite than those in a comparable spherical composite.

Exhaustion of Work Hardening Capacity: Ductile Failure of Matrix

Arsenault [19] correlated the in-situ matrix fracture toughness and ductility in a SiC/Al DMMC with those of an cold-worked unreinforced alloy. He argued that the matrix is in a highly cold-worked state due to ΔCTE. Therefore, the matrix in-situ ductility and toughness should correspond to those in a cold-worked state. It is a general observation that ductility of a metal or alloy decreases with an increase in cold work. Since the microstructure of the matrix in 20V% SiC/Al composite is similar to that of an unreinforced alloy with 90% cold work [20], he suggested that the in-situ matrix ductility of the composite should be comparable to the cold-worked unreinforced alloy. In a related experiment, it was shown (Table 1) that cold-rolling 6061 Al alloy to 69% resulted in nearly a factor of two reduction in K_{IC}, and in the case of 139% cold-rolling there is further reduction in K_{IC}. At this level the K_{IC} value of the cold-rolled matrix alloy is comparable to the 20V% SiC_p/6061 Al alloy composite [19].

In another investigation [22] in which correlations between K_{IC} and yield stress were determined, Hahn and Rosenfield showed that for 2000 and 7000 series Al alloys there was a substantial loss of fracture toughness with increasing yield strength (for the same volume fraction of inclusions).

Table 1 Fracture toughness vs Cold Work

Material	Unreinforced Al [19]		20V% SiC/Al [21]
	Cold work (%)	K_{IC} ($MPa \cdot m^{1/2}$)	K_{IC} ($MPa \cdot m^{1/2}$)
6061 Al T6	0	43	22.4
6061 Al T6	15	31	
6061 Al T6	69	27	
6061 Al T6	139	~ 20	

They further suggested that this was due to a reduction in the work needed to link the voids in the higher strength alloys. In the case of cold-worked Al alloy and the Al matrix in a SiC/Al DMMC, this is related to the loss of strain hardening capacity. Jagannadham and Wilsdorf [23] demonstrated that dislocation-cell walls in metals and alloys could serve as microcrack nucleation sites and degraded the ability to resist crack propagation. Jones et al.[24] calculated the fracture toughness of a SiC/Al DMMC by accounting for the strength increase from work hardening due to ΔCTE [25] and for the influence of the strength on the fracture toughness, i.e., $K_{IC} \propto (\sigma_y E)^{1/2}$ [21]. They obtained a reasonable agreement with the existing experimental data.

Secondary processing is shown to affect the composite fracture process.

The change in ductility due to under- and over-aging provides some insights into the role of the matrix in affecting the ductility of DMMCs as listed in Table 2. While the changes from matrix- to reinforcement-fracture-dominated failure double the composite fracture resistance J_{1c}, only a fractional increase in strain-to-failure is obtained (see Table 1) [26]. This points to the potential importance of the matrix strain hardening capacity in determining the ductility of DMMCs. The key role of the matrix in composite failure resistance is further demonstrated by the ability to recover most of the ductility after prestraining by re-solutionizing the composite [4]. That is, particle-fracture induced by prestraining has a minimal role as compared with the matrix conditions in affecting the composite ductility. The matrix can affect the ductility of DMMCs in two ways: (1) the matrix in-situ ductility is degraded due to the addition of reinforcements; (2) the matrix plastic flow is localized.

Table 2 Tensile Ductility of Underaged (UA) and Overaged (OA) Composites

Material	Elongation (%)		J_{1c} (KJ/m^2)
	SiC/ 7xxxAl [26]	SiC/ 2124 Al [13]	SiC/ 7xxx Al [26]
UA-0V% (unreinforced)	20	~ 20	31.0
UA-13.5V% SiC_w	-	~ 5	-
UA-15V% SiC_p	4.9	-	16.3
UA-20V% SiC_p	4.3	-	11.7
OA-0V% (unreinforced)	19	~ 18	31.5
OA-13.5V% SiC_w	-	~ 3	-
OA-15V% SiC_p	3.5	-	7.4
OA-20V% SiC_p	3.4	-	5.5

Localization of plastic deformation can further degrade the macroscopic ductility of DMMCs. The inherent origin for deformation localization in DMMCs attracts attention due to its importance in improving composite ductility. To assess the degree of matrix deformation localization, Arsenault et al. [27] examined the dislocation density near the failure surface of a tensile sample. They found that the dislocation density was very high near the fracture surface, and decreased at a greater rate as a function of distance from the fracture surface in a composite with a higher SiC content. This suggested that plastic flow in a particle reinforced metal was localized due to addition of SiC. Using stereoimaging, Davidson [14] noted that the maximum local strain near fracture path is significantly in excess of the average strain-to-failure obtained by tensile tests.

Particle distribution is another factor which limits the ductility of the composites. With few exceptions, cracks in the matrix are always initiated in the particle-rich region [14,28], and follow a more random path when particle distribution is more uniform [29]. Arsenault et al. [27] studied localized plastic flow in a composite by in-situ monitoring the development of slip lines in the matrix via an optical microscope. They found that slip lines were more prominent in the particle-rich region. A companion FEM analysis [27] on periodic clustering of 4 neighboring particles showed that inhomogeneous particle distribution led to a higher rate of effective plastic strain accumulation within/near reinforcement clusters. Therefore, aside from the deformation localization led by the additions of the reinforcement [30], particle clustering induces additional plastic flow localization.

Wang et al. [31] performed another in-situ slip line observation followed by FEM modeling. They found that an idealized periodic clustering model as in [32,27] could qualitatively describe the evolution of field quantities in an actual geometry in which the FEM mesh was mapped directly from the optical images. It should be noted that the result of lower matrix triaxial stresses induced by particle clustering [32] is different from the perceptions that the matrix hydrostatic stresses are more intense in a particle cluster [4,33,34].

The periodic clustering approach [32] was also implemented into ductility predictions [35]. By assuming the same criteria that void nucleation and its subsequent growth were governed by the effective plastic strain and the triaxial stresses [18], respectively, it was shown that the void growth stage was extended in a clustered composite due to a lower matrix triaxiality, and therefore the composite ductility was enhanced. This result is contrary to the suggestions by many others that homogeneous reinforcement distributions would lead to improved composite ductility (e.g. [36,4,14]).

The periodic clustering model [32,35] only accounted for the short range interactions between particles in the cluster, the long range interactions between the clusters and the relatively homogeneous region could not be considered. To study this long range interaction (i.e. the long range fluctuation of the internal stresses caused by variations of local reinforcement concentration), Shi et al. [37] constructed an imaginary "composite" as shown in Fig.3 in which C_1 and C_2 represented the cluster and the uniform region, respectively. Because of a long range

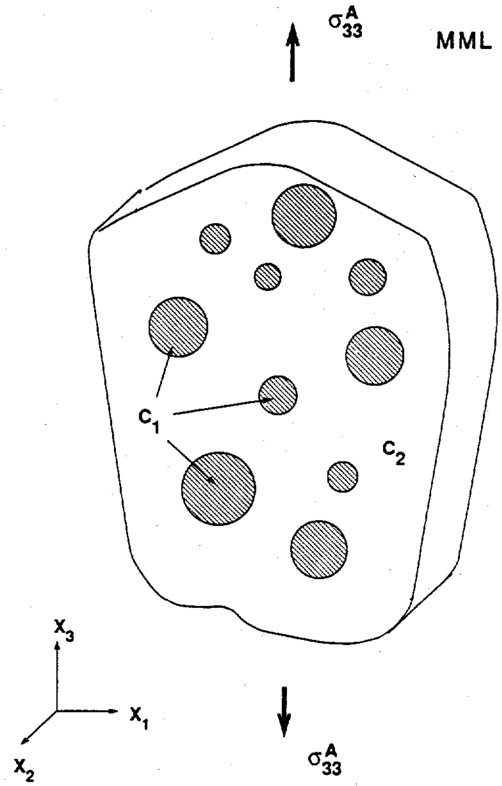

Figure 3: Schematics of mean field approximation of a composite with inhomogeneous particle distribution, where shaded spheres stand for reinforcement clusters, "C_1", embedded in a more "uniform" composite, "C_2".

interaction, they approximated this "inhomogeneous" composite by considering the smeared properties for C_1 and C_2, as if they were monolithic materials with equivalent properties from composites with predefined reinforcement volume fractions. From this approach, the short range interactions between particles in C_i's were approximated phenomenologically when the "smeared" material properties were assigned to each "phase". Within the framework of mean field interactions, they obtained the following criterion [7]:

$$\left(\frac{\sigma_y^{c2}}{\sigma_y^{c1}} - K_{12}^{-1}\frac{\mu^{c2}}{\mu^{c1}}\right)\begin{cases} >0 & C_1 \text{ yields first;} \\ =0 & C_1 \text{ and } C_2 \text{ yield simultaneously;} \\ <0 & C_2 \text{ yields first.} \end{cases} \qquad (1)$$

In this expression, the σ_y is the yield stress for the constituents, the K_{12} is the ratio between the average elastic strains in C_1 and C_2 and, the μ is the stiffness that resists shear.

Equation 1 is only applicable at the onset of global composite yielding. Since, experimentally, strain localization in a reinforcement cluster generally initiates within an early stage of deformation [14], comparison with experiments may lead to some insights into the process of plastic flow localization. Lloyd [4] recently showed that the strain distribution might be controlled by heat treatment. In a SiC reinforced 6061 Al DMMC, a T4 temper led to a rather uniform deformation until final fracture, whereas strain localization initiated at a relatively early stage of deformation in T6 condition. If the degree of deformation localization induced from different heat treatment in [4] is dictated by the changes of matrix strength, then Eq. 1 predicts the same trend as the experimental observation [4]. That is, compared with T6 condition, reduction in matrix strength from T4 has more effect on C_2 than that on C_1, i.e. a larger reduction in σ_y^{c2} while no changes in K_{12} and μ. This promotes yielding in C_1, i.e. localized plastic flow in the clusters. The fact that this variation cannot be predicted by the periodic clustering model [32,35] shows that the long range internal stress fluctuations caused by clustering is dominating the event. While clustering of particles tends to increase tendency for plastic flow [32,35,27,31], fluctuation of local particle concentration may cause long range perturbations of internal mean stresses which may either promote or suppress plastic flow in the reinforcement-rich region (Eq.1).

Results from [37] and [35] suggest that, although particle clusters may exhibit a ductile nature [35], its contribution to the global composite ductility may be offset by an early onset of local plastic flow, as noted by Davidson [14] who has estimated 50% local strain within clusters and yet the composite still suffers a low ductility (1.6 to 2.4% strain-to-failure) due to excessive local plastic flow in the clusters. In addition, increase in the loss of strain hardening capacity due to ΔCTE [19] may also offset the predicted increase in local ductility in the particle cluster [35].

To demonstrate conclusively the effect of clustering on ductility an experiment was conducted by Mirchandani and Heckel [38] in which measured elongation to failure as a function of degree of clustering (contiguity ratio). The elongation to fracture decreased significantly from ~ 14% to 1-2% with an increase in contiguity ratio from 0.1 to 0.68, as shown in Fig.4. Mirchandani and Heckel [38] also determined the tensile strength as a function of clustering and as to be expected, and contrary to Corbin and Wilkinson [39] they found that the tensile strength decreased as the degree of clustering increased, clustering increases as the contiguity ratio increases, as shown in Fig.5.

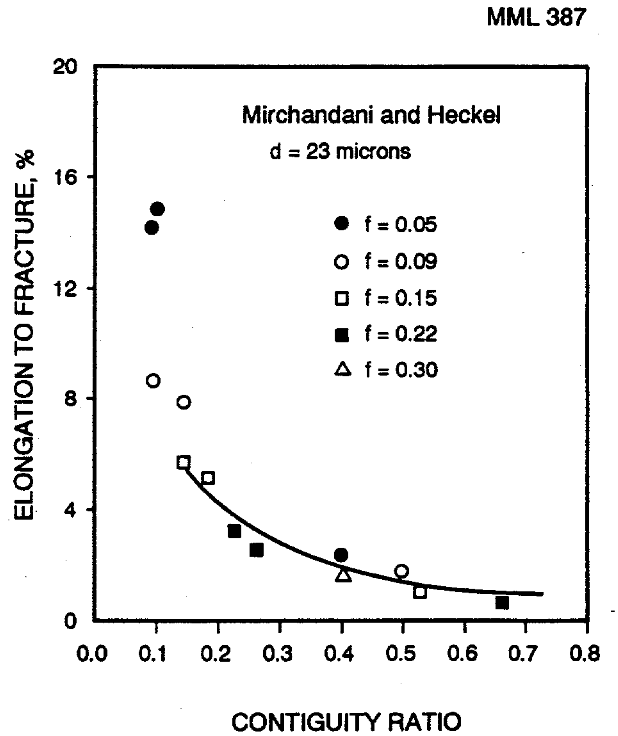

Figure 4: Correlation between ductility and contiguity ratio.

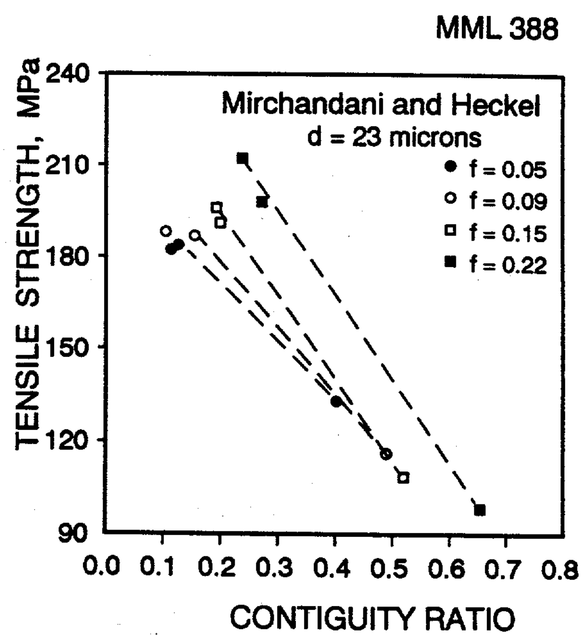

Figure 5: Variation in composite tensile strength as a function of contiguity ratio and SiC volume fraction.

Conclusion

The final rupture of DMMCs may be characterized as particle-, interface-,or matrix-controlled processes. The failure mode is dominated by the bulk matrix and reinforcement properties, and the interfacial precipitation conditions. However, no matter what the final failure mode is, the ductility of the composite seems to be controlled by the flow pattern and the in-situ ductility of the matrix. There are two factors that limit the ductility of the composite: (1) degraded in-situ matrix ductility due to thermally induced work hardening and/or the high matrix triaxial stresses; and (2) localized plastic flow. A major factor which affects the ductility of DMMCs is the fact that matrix is in a work hardened condition prior to the actual testing. The exhaustion of matrix in-situ ductility prior to loading may affect the micromechanisms of failure by, for example, limiting the ability of strain hardening. High triaxial stresses, at the same time, have similar effects. They increase the driving force for void growth and degrade microductility. With localized flow, much of the local ductility is exhausted with little contribution to the far-field displacement.

Particle clustering usually further enhances the localization of plastic flow. Although the clustering of particles is shown to elevate ductility of the cluster, in an actual composite the long range fluctuation of the internal stresses promotes highly localized flow in the cluster which reduces global ductility. In addition, work hardening induced by ΔCTE which is expected to be more severe in a particle-rich region may also counterbalance the enhanced ductility in the cluster.

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