MICROSTRUCTURALLY INHOMOGENEOUS COMPOSITES WITH TAILORED REINFORCEMENT DISTRIBUTION

H. X. Peng and L. Yin

Advanced Composites Centre for Innovation and Science (ACCIS), Department of Aerospace Engineering, Faculty of Engineering, University of Bristol, Queens Building, University Walk, Bristol, BS8 1TR, U.K.
H.X.Peng@bristol.ac.uk

ABSTRACT

Contiguity models for multiphase composites predict that the spatial distribution of reinforcement will influence mechanical performance. Preforms consisting of spherical agglomerates of SAFFIL alumina short fibres were melt-infiltrated with aluminium to give a controlled Microstructurally Inhomogeneous Composite (denoted as MIC). The microstructure and mechanical properties were compared with those of a composite made from a conventional short fibre perform with Planar-Random arrangement of SAFFIL short fibres (denoted as PRMMC). The results indicate that the MIC possessed a higher energy absorbing capability together with modestly enhanced modulus and strength and although a relatively lower tensile elongation was observed. The contiguity model was used to relate the mechanical properties to the microstructure.

1. INTRODUCTION

There has been a persistent effort in exploiting metal matrix composites (MMCs) not only with high specific strength and stiffness but also can exhibit certain ductility and damage tolerance in applications [1-4]. An issue that has received much attention in recent years is the importance of spatial inhomogeneity [3, 5-6]. Unlike the conventional approach to seek a homogeneous reinforcement distribution in MMCs processing, some inhomogeneous MMCs with controlled distribution of reinforcements showed considerable enhancement in the overall mechanical properties [7-11]. In this context, a theoretical model based on topological transformation and phase contiguity has been derived which allows a range of physical and mechanical properties of two phase composites with any grain size, volume fraction or distribution of the second phase to be predicted from volume fraction and phase contiguity [12]. In the phase contiguity model, the mechanical and physical properties of composite materials depend more directly on the *continuous volume fraction* of reinforcement than on the overall volume fraction. The spatial distribution of reinforcement thus exerts a significant influence on the mechanical performance. This model has been shown to fit a range of published data [12] justifying the exploration of property advantages that are expected to accrue from controlled inhomogeneous structures in MMCs and of procedures for creating such structures [13].

In this paper, the principal objective is to test the conjecture that better MMC performance can be obtained by tailoring the reinforcement distribution rather than pursuing a homogeneous distribution pattern. The squeeze casting technique, one of the most economical methods of fabricating MMCs, was employed partly because the resulting composite microstructure can be controlled simply by managing the preform architecture. It will be shown in the present work that the contiguity model can be used to relate the mechanical properties to the microstructural character of the newly fabricated composites.

2. EXPERIMENTS

The as-received SAFFIL alumina short fibres (ICI, UK) were tumbled in a polyethylene container at 80 rpm for up to10 hours. Selected sizes of the resulting agglomerates (0.4 - 1mm) were packed in a steel die to make a preform with desired volume fractions. MMCs were fabricated by squeeze casting techniques with aluminium as matrix. The planar-random arrangement short fibre MMCs (denoted as PRMMCs) with the same volume fraction of alumina fibre were also fabricated to compare with the MMCs made from fibre agglomerates preform (denoted as MICs). Mechanical properties were measured by tensile and three point bending tests. SEM and optical microscopy were used to characterize the morphology of the fibre agglomerate and microstructures of the fabricated MMCs. The tensile test bars were obtained by cutting the as-cast composite cylinder parallel to the cylindrical axis to give final sample dimensions of about 55mm x 6mm x 2mm. Three point loading tests were performed on notched samples at a span of 40mm and constant cross-head speed of 0.1mm/min. The elastic modulus were determined by using the measurement of resonant frequencies Samples with approximate dimensions of 2mm x 6mm x 50mm rested on nodes at about L/4 from each end and were set in vibration by striking the sample with ceramic projectiles.

3. RESULTS AND DISCUSSIONS

3.1 Microstructures and mechanical properties

The morphology of the fibre agglomerates prepared by the tumbling technique is shown in Figure 1. The microstructure of the composite prepared from an assembly of fibre agglomerates with volume fraction 0.1 indicates that the die evacuation system provided extensive infiltration. The composite spheres are uniformly distributed in the fibre-free matrix at the macroscopic level and are in contact (Fig. 2). This material has an isotropic microstructure in contrast with the conventional MMC which reveals a planar-random fibre distribution [13]. The volume fraction of the composite spheres was approximately 61% based on the whole composite and this was obtained by measuring the area fraction of spheres. The fibre distribution within the composite sphere varied from the edge to the centre along the diameter [13]. A higher fibre volume fraction of about 30% was observed at the periphery while the local volume fraction in the centre was only a few percent. This distribution should provide a stronger outer layer. Figure 2 also indicates that the interface between the composite spheres and the fibre-free matrix was not sharp, there being a transition from the fibre-free region to the composite sphere, in contrast to the interface in conventional MMCs between two dissimilar materials.

The mechanical properties are summarized in Table 1. The microstructurally inhomogeneous composite (denoted as MIC hereafter) exhibited higher UTS and lower strain at fracture than those of the composite with the as-received perform. It is interesting to note that the elastic modulus of the fibre agglomerate reinforced composite is also higher (Table 2) despite identical fibre volume fraction.

During tensile testing, the deformation is uniformly distributed throughout the volume within the gauge length. It is evident that the composite sphere is actually acting as one reinforcing unit in the overall composite and remains nearly un-deformed while the deformation mainly occurred in the fibre-free area. The volume fraction of the ‘reinforcing phase’ is as high as 61% which is likely to be responsible for the lower overall tensile elongation observed.

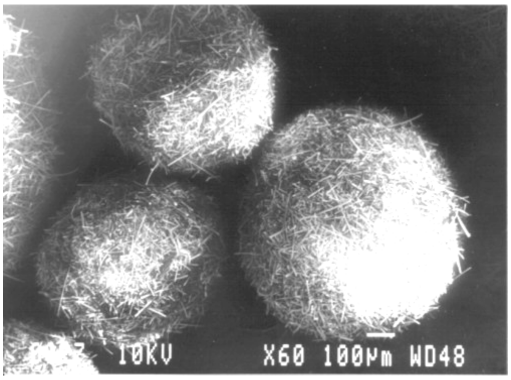


Figure1: The morphology of the fibre agglomerates prepared by the tumbling technique.

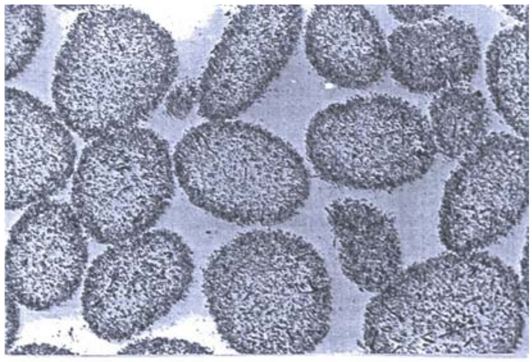


Figure 2: The typical microstructure of the composite prepared from the fibre agglomerates. (V_f = 10%)

Table 1 Comparison of the mechanical properties of composites made from the as-received preform and fibre agglomerates (V_f = 10%)

Preform condition	UTS (MPa)[a]	Strain to break (%)	Modulus (GPa)[b]
As-received (PRMMC)	188 (7)	7.1(0.5)	76 (0.5)
Fibre agglomerates (MIC)	199 (7)	4.9 (1.3)	84 (0.8)

[a] values in parentheses are the 95% confident limits from at least 6 samples.
[b]values are obtained from the resonant vibration method.

Based on three-point-bending tests on notched samples, our previous study [13] concluded that the MIC exhibited a much higher energy absorbing capability (characterized by the area below the load-displacement curve) than that with the as-received performs. Figure 3 shows the typical load-displacement curve for the PRMMC and the MIC obtained during flexural tests in which the conventional MMC was tested in the direction where the notch was parallel to the random fibre plane.

3.2 Effect of continuous volume fraction - application of the contiguity model

The higher UTS and elastic modulus compared with the homogeneous composite is consistent with investigations of aluminium matrix composites reinforced with particle agglomerates which concluded that they exhibited a significant increase in UTS and also in elastic modulus compared with a composite having a homogeneous microstructure [14]. Our results confirm that the composite performance can be altered by controlling the spatial distribution of reinforcement.

Two specific conclusions drawn from the contiguity model [12] are:

1) In discontinuously reinforced composite systems, the Young’s modulus (E) increases with increasing continuous volume fraction of the reinforcing phases. The higher the ratio of the reinforcement to matrix stiffness (E_β/E_α), the greater the

increment in stiffness with increasing continuous volume fraction of the reinforcing phase at fixed volume fraction.

2) The yield strength of two ductile phase alloys increases with increasing continuous volume fraction of the hard phase at fixed volume fraction.

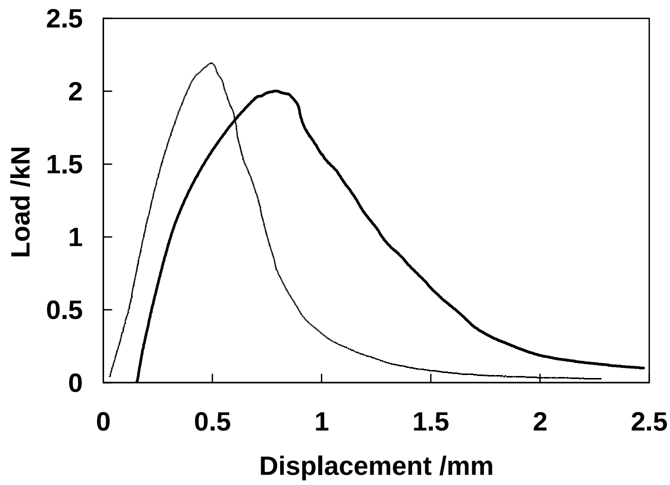
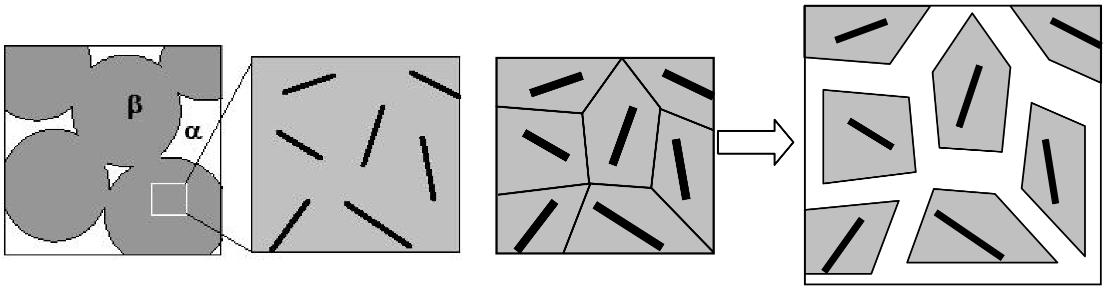


Figure 3: The typical load-displacement curve of the conventional MMC made from as-received preform (thin line to the left) and the MMC made from the fibre agglomerates (heavy line to the right) during three point loading tests.

It remains difficult to quantify the continuous volume fraction experimentally. There are also some challenges in comparing the phase contiguities of the homogeneous composite and the MIC in a qualitative way because the individual fibres are mainly surrounded by the matrix alloy, i.e., are isolated, in both composites. However, this is really a scaling problem and an equivalent microstructural transformation of the MIC can be made according to local fibre volume fraction. This enables a comparison of the continuous volume fractions between the two composites.

Following the same strategy of dealing with heterogeneous systems [15, 16], regions with the same local volume fraction can be treated as one "phase". The composite produced here effectively consists of a stiffer phase comprising the inter-connected fibre-rich spheres which is 3D continuous (denoted β -phase) and a fibre-free aluminium alloy phase in the interstitial regions (denoted α-phase) between the interconnected spheres which is also 3D continuous and interpenetrating to the fibre-rich spheres. The average local fibre volume fraction within the composite spheres ($\boldsymbol{f_\beta}$) is approximately 16.4% while the overall volume fraction of the spheres ($\boldsymbol{V_\beta}$) is 61%. The microstructural transformation proposed here includes a tessellation step and a dilution step as schematically shown in Figure 4.

Step 1- Tessellation: β-phase with its overall volume fraction $\boldsymbol{V_\beta}$ is randomly tessellated into unit-cells each containing a single fibre with a fibre volume fraction of $\boldsymbol{f_\beta}$ based on phase-I (Fig. 4a). Each cell is assigned as one monolithic β-phase "grain". These "grains" are fully interconnected in the transformed microstructure (Fig. 4a).



(a) f_I =0.164 (b) f_H =0.1

Figure 4: Schematic illustration of the microstructural transformation based on the level of local fibre volume fraction for the purpose of contiguity comparison.

Step 2- Dilution: These "grains" are diluted with matrix phase (which is also designated α-phase), to form a uniform microstructure with a fibre volume fraction of $\boldsymbol{f_H}$, where $\boldsymbol{f_H} < \boldsymbol{f_\beta}$. As shown in Figure 4b, this is the effective microstructure of a homogeneous composite with an overall fibre volume fraction of $\boldsymbol{f_H}$. As a direct consequence of lower $\boldsymbol{f_H}$, i.e., larger inter-fibre spacing in the homogeneous composite compared with that in composite spheres, this forms a completely continuous matrix (α-phase) network (Fig. 4b).

The overall volume fractions ($\boldsymbol{V}$), contiguities ($\boldsymbol{C}$) and continuous volume fractions ($\boldsymbol{f_C}$) of the α and β phase in both transformed composites can be found. For the overall volume fractions of the α and β phases in the transformed homogeneous composite, $\boldsymbol{V_\alpha^H}$ and $\boldsymbol{V_\beta^H}$, we have:

$$\boldsymbol{V_\alpha^H + V_\beta^H = 1} \quad (1)$$
$$\boldsymbol{f_\beta V_\beta^H = f_H (V_\alpha^H + V_\beta^H)} \quad (2)$$

For the transformed MIC, the overall volume fractions of α and β phases, $\boldsymbol{V_\alpha}$ and $\boldsymbol{V_\beta}$, are 0.39 and 0.61, respectively.

The α -phase network in the transformed homogeneous composite is completely continuous (Figure 4b), so the contiguity of α - phase ($\boldsymbol{C_\alpha^H}$) is unity, i.e., $\boldsymbol{C_\alpha^H}$ =1. It is obvious that the contiguity of β - phase in the homogeneous composite ($\boldsymbol{C_\beta^H}$) is zero since the β - phase "grains" are all isolated by α - phase network as shown in Figure 4b.

In the transformed MIC, it is equivocal to describe the α-phase network as *continuous* because it is formed by the interstices of the composite spheres and hence has necks. In terms of the contiguity of β - phase, the composite spheres are not completely 3-D *contiguous*, they contact each other via limited touching areas; hence the contiguity of these spheres is less than unity.

Based on the definition of the contiguity [17], the contiguities of α - and β - phase in the MIC can be calculated if simple assumptions are made. In view of the microstructural observations and the packing efficiency of the fibre agglomerates, the assumption of simple cubic packing seems reasonable. The model is a simple cubic array of packed overlapping spheres with co-ordination number of 6, with lattice parameter, *a*, and with spheres of diameter *D*, centred at lattice points (Fig. 5). The spheres touch when $D = a$. As *D* increases, the area formed by the intersection of nearest neighbour spheres defines a *contacting area* with diameter of *d*. Thus

$$O \leq d / D < 1/\sqrt{2} \quad (3)$$

The mean value of *d* was obtained by statistical measurement of the polished samples employing the stereological factor 4/π [18]. The value of *D* was obtained from examination of the bulk fibre agglomerates. In the present work, they are approximately 0.2~0.3mm and 0.75 mm, respectively.

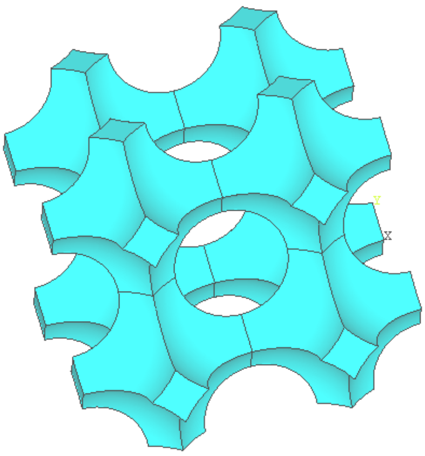


Figure 5: Schematic 3D diagram showing the contact areas (A) of the fibre-free ‘α’-phase between neighbouring unit cells. The composite spheres (β -phase) are connecting via the windows.

From the definition, the contiguity of the β phase (C_β) in an α-β binary mixture is the fraction of the total internal volume specific surface area of this phase ($S_V^{\beta\beta}$) which is shared with surface area of the same phase (S_V^{β}) and is given by;

$$C_\beta = \frac{2S_V^{\beta\beta}}{S_V^{\beta}} = \frac{2S_V^{\beta\beta}}{2S_V^{\beta\beta} + S_V^{\alpha\beta}} \quad (4)$$

where ($S_V^{\alpha\beta}$) is the area of the interfaces between α and β phase per unit volume.

In a cubic unit, the spheres make contact with each other through the contacting area with diameter *d* and with the aluminium alloy phase via sphere surfaces. For the composites made from fibre agglomerates, $S_V^{\beta\beta}$ in Equation (4) can be taken as the sum of the contact area (S_C):

$$S_V^{\beta\beta} = \sum S_C \quad (5)$$

and $S_V^{\alpha\beta}$ as the sum of the area of sphere surfaces excluding contact areas:

$$S_V^{\alpha\beta} = (\sum S_{SS} - \sum S_S) \quad (6)$$

where, S_{SS} is the surface area of the sphere and S_s is the surface area of the segment of the sphere at the contact area. Combining equations (4), (5) and (6), the contiguity of composite spheres in the composite can be expressed as:

$$C_{\beta} = \frac{3k^2}{3k^2 + 6\sqrt{1-k^2} - 4} \qquad (7)$$

where $k = d/D$

So, the contiguity of the fibre-rich spheres in the composite can be related to the ratio of the contact diameter to the sphere diameter (k).

Similarly, the contiguity of α phase which arises from the interstitial regions interpenetrating the fibre-rich spheres can be worked out by assuming that the α phase within different unit cells makes contact with itself through the area 'A' as shown in Figure 9 and makes contact with composite spheres via sphere surfaces. Then the contiguity of α phase can be expressed as:

$$C_{\alpha} = \frac{(1-k^2) - k\sqrt{1-k^2} + \arcsin k - \frac{\pi}{4}}{(1-k^2) + (\frac{\pi}{2} - k)\sqrt{1-k^2} + \arcsin k - \frac{7\pi}{12}} \qquad (8)$$

where, $k=d/D$.

Putting k=0.3 (average d=0.25mm) in Equations (7) and (8), the contiguities of α and β phases in the MIC are found to be 0.18 and 0.14, respectively.

Table 2 Comparison of the contiguities and continuous volume fractions of α and β phases

Transformed $\alpha + \beta$ structure	α - Phase		β -Phase	
	Contiguity	Continuous volume fraction	Contiguity	Continuous volume fraction
Homogeneous	1	0.39	0	0
MIC	0.18	0.07	0.14	0.08

The continuous volume fractions of the α and β phases in the homogeneous composite and in the MIC were calculated and are summarized in Table 2 from which it can be seen that the continuous volume fraction of β phase (the stiffer phase) in the MIC is larger than that in the homogeneous composite. It is a direct outcome of the contiguity model that this result in higher stiffness and strength of the MIC compared with that of the homogeneous composite as observed experimentally. It can also be seen that the continuous volume fraction of α- phase (more ductile phase) in the MIC is smaller than that in the homogeneous composite. This resulted in a lower ductility of the MIC compared with that of the homogeneous composite as obtained in experiments.

In this comparison, it is appreciated that there exists a fibre planar-random arrangement to some level in the homogeneous composite resulting from the perform preparation. As described in the experimental section, the tensile properties were obtained along directions *in this plane* and are believed to be somewhat higher than those in a composite with true randomly distributed fibres.

5. CONCLUSIONS

The fibre agglomerate reinforced composites possessed higher strength and modulus than composites reinforced with random fibre performs of the same volume faction. The shells of composite spheres behaved as a 3D continuous reinforcing phase in the composite. This composite exhibited higher energy absorption capability even though relatively lower elongation was observed.

Based on a microstructural transformation, the phase contiguities in the composites were deduced. A contiguity model was successfully employed to relate the mechanical properties to the microstructural parameter that is considered to control the property, namely *continuous volume fraction*.

This work adds support to the conjecture that tailored reinforcement architecture can enhance composite properties compared with homogeneous arrangements.

ACKNOWLEGMENTS

The authors are grateful to EPSRC for the financial support of this work under Grant No. EP/C532392 and to Prof. Fan of BCAST at Brunel University and Prof. Evans of University College of London (UCL) for their valuable inputs.

SELECTED REFERENCES

[1] Kelly A., “Composites in context ”, Comp. Sci. Tech. 1985:23: 171.

[2] Ibrahim I. A., Mohamed F. A. and Lavernia E. J., “Particulate reinforced metal matrix composites-a review”, J. Mat. Sci. 1991:26:1137-1156.

[3] Lloyd D. J., “Particle reinforced aluminium and magnesium matrix composites”, Inter. Mat. Rev. 1994:39:1-24.

[4] Sinclair I. and Gregson P.J., “Structural performance of discontinuous metal matrix composites”, Mat. Sci. Tech. 1997:13:709-726.

[5] Segurado J., Gonzalez C. and Llorca J., “A numerical investigation of the effect of particle clustering on the mechanical properties of composites”, Acta Mater., 2003:51:2355-2369.

[6] Wilkinson D.S., Pompe W. and Oeschner M., “Modeling the mechanical behaviour of heterogeneous multi-phase materials”, Prog. Mat. Sci., 2001:46:379-405.

[7] Nardone, Vincent C. ,Strife, James R., Prewo, K.M, “Microstructurally toughened particulate-reinforced aluminum matrix composites”, Metallurgical Transactions A, 1991:22A:171-182.

[8] Ellis L. Y. and Lewandowski J. J., “Effects of layer thickness on impact toughness of Al/Al-SiCp laminates”, Mat. Sci. Eng.:A,1994:183: 59-67.

[9] Pandey A. B., Majumdar B. S. and Miracle D. B., “Laminated Particulate-Reinforced Aluminum Composites with Improved Toughness”, Acta Mater. 2001:49: 405-417.

[10] Qin S. and Zhang G., “Preparation of high fracture performance SiCp-6061A1/6061A1 composite ”, Mat. Sci. Eng. :A, 2000:279:231-236.

[11] Peng H. X., Fan Z., Mudher D.S and Evans J. R. G., “Microstructures and mechanical properties of engineered short fibre reinforced aluminium matrix composites”, Mat. Sci. Eng.:A, 2002:335:207-216.

[12] Fan Z. and Miodownik A. P., "deformation behaviour of. alloys comprising two ductile phases ", Acta Metall. Mater., 1993:41: 2403-2413 and 2415-2423.
[13] Peng H. X., Fan Z. and Evans J. R. G., "Novel MMC microstructure with tailored distribution of the reinforcing phase", J. Microscopy, 2001:201: 333-338.
[14] Zhou Z., Peng H. X., Fan Z. and Li D. X., "MMCs with a controlled non-uniform distribution of submicron Al2O3 particles in a 6061 Al matrix", Mat. Sci. Technol., 2000:16:908-912.
[15] Tseng T.C., "The effect of particle clustering on the mechanical behaviour of particle reinforced composites", composites:B, 1998:29:299-308.
[16] Crobin S. F. and Wilkinson D.S., "Influence of Particle Distribution on the Mechanical Response of a PMMC", Acta Metall. Mater., 1994:42:1311-1318.
[17] Gurland J., "*The measurement of grain continuity in two phases alloys*", Trans. Met. Soc. AIME, 1958:212:452-455.
[18] E. R. Weibel, Stereological methods. London; New York, Academic Press, 1980.