Ductile-to-brittle transition in tensile deformation of particle reinforced metals

Aude Hauert[1,a], Andreas Rossoll[1,b] and Andreas Mortensen[1,c]

[1] Ecole Polytechnique Fédérale de Lausanne, Laboratory for Mechanical Metallurgy, EPFL STI IMX LMM, Station 12, Lausanne CH-1015, Switzerland.

[a]aude.hauert@epfl.ch, [b]andreas.rossoll@epfl.ch, [c]andreas.mortensen@epfl.ch

Keywords: ductile-to-brittle transition, metal matrix composite, particle fracture, aluminium, alumina, mean-field analysis, tensile failure.

Abstract. The aim of this study is to contribute to our understanding of the fracture behaviour of two-phase materials that combine brittle particles with a ductile matrix, using model metal/ceramic composites produced by infiltrating ceramic powder beds with liquid pure aluminium or aluminium alloy to produce a roughly half-ceramic/half-metal composite. Like many other materials of their class, these composites exhibit two modes of tensile failure; they fail either by tensile instability, or alternatively they break prematurely, in a brittle fashion. We explore the physics of fracture in these materials by testing samples of varying size and geometry. Furthermore, we establish a model to link damage build-up and composite fracture using micromechanical modelling, which accounts for both failure modes. A ductile-to-brittle transition is observed both in modelling and experiment when the matrix is strengthened, or the particles made weaker. An effort is made to keep the model at a level of simplicity that keeps the underlying physics clear, and makes it attractive for practical application.

Introduction

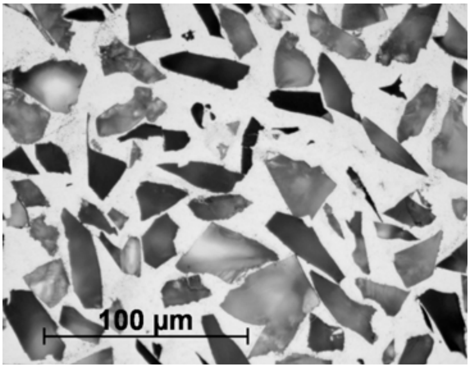
Generally speaking, materials fail under tensile loading according to one of two macroscopic failure modes: (i) a ductile failure mode, preceded by some kind of tensile instability, i.e. necking of the specimen, and (ii) a brittle failure mode that occurs prior to tensile instability. Both of these failure types are influenced by the occurrence of internal damage in the form of voids or cracks, which eventually link up to a macroscopic crack. Whether one or the other failure mode prevails may depend on several parameters but generally speaking, increasing strength goes often with increasing brittleness.

In composite-type two-phase materials, three types of internal damage can be encountered: (i) cracking of the reinforcing phase, (ii) matrix void growth, or (iii) interface decohesion [1, 2]. For continuous fibre-type composites under uniaxial loading along the fibres, internal damage development is generally dominated by fibre failure, and assessment of the stress-state is comparatively simple. This has encouraged the development of models that describe with relatively good precision and clarity how the occurrence of fibre fractures induces final failure of the material. A factor of key importance in this regard is how the load shed by a broken fibre is redistributed; as this is generally quite complex, two extreme cases of load transfer have been generally considered: Global Load Sharing (GLS) induces ductile-like behaviour, while Local Load Sharing (LLS) promotes brittle behaviour [3]. In GLS, a broken fibre sheds its load equally on all the intact remaining fibres, such that further damage develops randomly, eventually leading to failure of the sample at the onset of tensile instability. In LLS, only its (in the extreme case: nearest) neighbours take the load from a broken fibre, which leads to the development and growth of clusters of fibre breaks and favours an avalanche-like catastrophic growth of damage, leading to brittle tensile failure.

A similar transition in tensile failure mode is observed in many materials that combine a brittle discrete phase with a continuous matrix phase, including notably Particle Reinforced Metal-Matrix Composites (PRMMCs): some of these materials break at the onset of tensile instability, while others fail before [4-6]. The aim of this study is to build on our understanding of the fracture of composites with continuous fibre reinforcement in order to tackle the physics of failure in PRMMCs or similar brittle inclusion / ductile matrix two-phase materials.

Infiltrated ceramic particle reinforced metals

The starting point of our work is the exploration of model, microstructurally simple, material, that make it possible to avoid most of the peripheral complicating features encountered with commercial composite materials. This model material is of pure aluminium or aluminium alloy reinforced with a high volume fraction of Al_2O_3 particles and is produced in our laboratory by gas pressure infiltration [7, 8]. The resulting composite features a high volume fraction (about 50%) of uniformly distributed ceramic particles strongly bonded to a pore-free and metallurgically simple matrix [9] (Fig. 1).



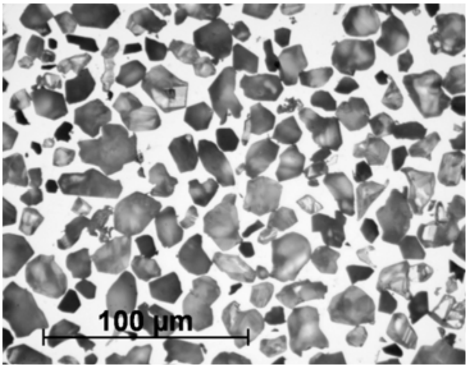


Figure 1: Optical micrographs of PRMMCs with (a) 35 µm angular and (b) 25 µm polygonal Al_2O_3 reinforcements (dark phase).

Composites made of pure aluminium reinforced with large alumina particles fail at the onset of tensile instability (following Considere's criterion) [10], whereas composites made with an alloyed matrix or with finer angular alumina particles fail in a brittle fashion, Fig. 2a. For both failure modes, tensile failure is preceded by extensive build-up of internal damage, measured by monitoring the progressive decrease of the composite material's Young's modulus, Fig. 2b. In all cases these composites fail at a tensile strain well below that of their matrix: internal damage clearly governs their tensile failure, and hence strongly influences their ductility.

These materials are fairly tough [11, 12]. To account for the brittle failure behaviour of the composite on the basis of its fracture toughness, one would need initial internal defects to be in the range of 1 to 2.5 mm. We have no indication of the presence of such large defects in the composite prior to mechanical loading. Also, there is no reason why brittle failure would not be observed, were it caused by such a pre-existing defect, in the pure aluminium matrix composite, as this does not have a very different microstructure nor a widely different fracture toughness. Brittle failure must therefore be caused by defects that grow in the material during loading and eventually link up to form a macroscopic crack of critical size.

Internal damage in these composites can take two forms: matrix voiding between the particles, or particle fracture. In some, particle fracture is very clearly dominant: we focus on those materials, and draw a parallel with continuous fibre composites with a goal of understanding the brittle-to-ductile tensile failure transition that they display.

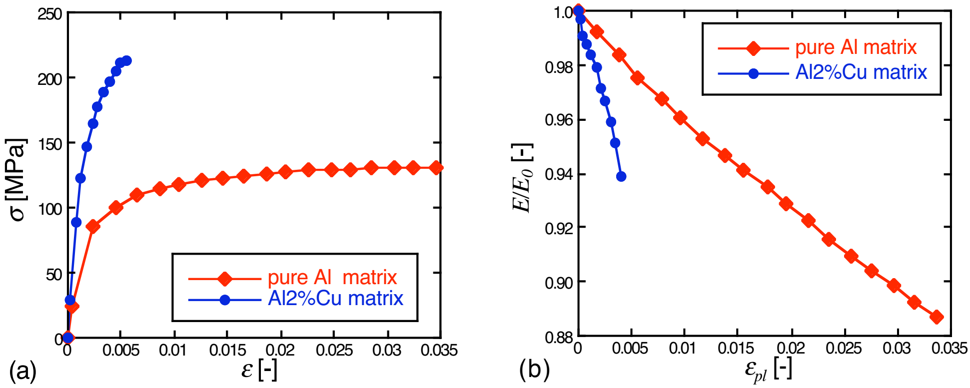


Figure 2: Pure aluminium and Al-2%Cu alloy reinforced with angular 35 micron alumina particles. (a) Typical true stress-strain curves, and (b) evolution of Young's modulus (normalized by its initial value) as a function of plastic strain. Failure of the pure aluminium matrix composite occurs following Considere's criterion, whereas the stronger alloyed materials fail before.

Experimental work and numerical analysis

Tensile tests – The properties of the constituents of our model PRMMCs need to be known for modelling. The *in-situ* stress-strain curve of the matrix can be derived from tensile tests with unload-reload cycles to monitor the evolution of Young's modulus, as described in [13]. The particle strength distribution is not known *a priori*, because it cannot, at present, be directly measured in a way comparable to the testing of individual fibres or dry fibre bundles. To assess the intrinsic particle properties, "back-calculations" from tensile tests with periodic unload-reload cycles are therefore carried out with the aid of a simplified micromechanical model, cf. further below.

Bend tests – Homothetically similar smooth bars of two different sizes were tested in four-point bending, in order to examine whether there is a volume effect on strength. Tests were conducted on composites made of 35 μm size angular alumina particles and a pure Al matrix. Small bend specimens exhibit the largest strain at failure and tensile specimens the lowest, which indicates indeed a volume-dependence of strength. Such a size effect indicates that fracture is of statistical nature, but (as indicated above) there is no indication that the critical crack is present initially in these materials—it apparently develops by internal damage build-up.

Two-dimensional finite element modelling of smooth bend bars was carried out with the code ABAQUS/Standard, Version 6.5 [14]. For these FE calculations, the composite material is modelled as homogeneous and isotropic, and is assumed to obey von Mises plasticity. A test function on the hydrostatic part of the stresses was computed at each integration point in order to differentiate whether the stress state is mainly tensile or compressive. Tensile loading is assumed to induce progressive damage and softening of the material. From Fig. 2b it follows that damage as monitored by the drop of Young's modulus evolves linearly with plastic strain in this composite.

The stress-strain properties of the composites were obtained from tensile tests, and back-calculated to an "effective" (i.e. non-damaged composite) stress used in order to describe compressive loading [13]. These flow curves (tensile and compressive) were fitted with a power law having a strength coefficient K that evolves with the equivalent plastic strain ε_p as

$$K = K_0 \cdot (1 - \alpha \cdot \varepsilon_p) \tag{1}$$

where K_0 is a constant and α a damage parameter that remains at zero for compressive loading. Damage is thus modelled as isotropic and assumed not to introduce any volume change: despite these somewhat crude approximations, the fit of simulated load-deflection curves with experimental data is very satisfying. The location of failure in the sample is also identical for simulation and experiment; for both, strain localizes opposite of the upper rollers of the specimen fixture (in the part of the beam that is under tension). The good agreement confirms that for pure Al reinforced with angular Al_2O_3 particles particle fracture depends strongly on the composite plastic strain. On the other hand, from Fig. 2a it follows that stress plays also an important role on particle cracking, since the higher stressed alloyed samples break at much lower strain than the unalloyed samples.

Double-notched bend tests – In order to gain further insight into whether damage by particle cracking is stress- or strain-dominated, double-notched bend specimens were tested in four-point bending. Loading of such a sample introduces damage simultaneously at two well-defined and confined locations, namely in each of the two ligaments. As eventually only one ligament fails, the other ligament allows observation of damage just prior to fracture [15]. The strain fields on the surface of the samples were visualized by photoelasticity and were used to control sample alignment as well as to observe strain localization.

Metallographic cuts of the unbroken ligament reveal a large stable process zone in the pure aluminium matrix composites. Fractured particles can easily be observed. Some of the cracks are wide open and aligned in a zone approximately 1 mm long, Fig. 3. Fractured particles in front of the unbroken notch are localized in the region of maximum principal stress (away from the notch root, Fig. 4b) as well as in the region of maximum strain (at the notch root, Fig. 4a). This confirms our finding that—on the level of a homogenized continuum—both stress and strain contribute to particle fracture in pure aluminium matrix composites with a high volume fraction of reinforcement, V_r.

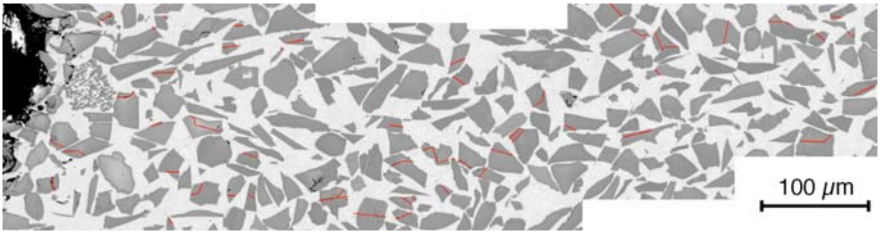
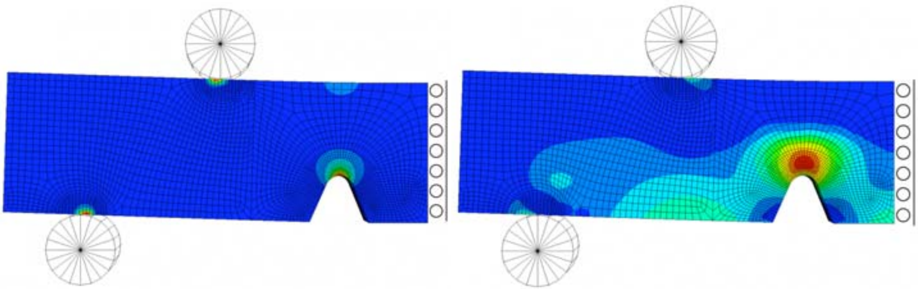


Figure 3: Metallographic cut of an unbroken notch in a double notched bend bar (in the centre plane of the specimen). Metallographically visible particle cracks are highlighted with red lines.



(a) (b)

Figure 4: FEM computed fields of (a) the largest principal strain and (b) the largest principal stress in the centre plane of a double notched bend bar (3D computation).

Analytical modelling of tensile and fracture behaviour

For analytical treatment, we consider a simple composite material made of homogeneously distributed hard spherical particles, strongly bonded to a soft matrix that undergoes strain hardening according to the Hollomon power-law. The flow curve of the non-damaging composite material is modelled using a simplified non-linear mean field approach [16] combined with the Torquato Identical Hard Sphere (TIHS) elastic scheme [17, 18]. The latter yields the best estimate out of several generally employed mean-field models for Young's modulus of such high volume fraction PRMMCs [19].

Under tensile loading, the modelled composite undergoes progressive damage in the form of particle fracture only. For a starting point, particles are assumed to fail according to a Weibull strength distribution under the action of the largest principal stress, which is the particle stress component parallel to the tensile loading direction (x_3). Particle fracture is introduced in the mean-field model using the Vanishing Cracked Particle (VCP) approach [13], i.e., by simply replacing a broken particle with an equivalent amount of matrix phase such that a composite that gradually accumulates damage is assimilated to a sequence of non-damaging composite materials of decreasing volume fractions of reinforcement.

In analogy with models for fibre reinforced composites, the load shed by particle fracture is redistributed either on the composite directly neighbouring the broken particle (Local Load Sharing mode) or to the entire intact remaining composite material (Global Load Sharing mode). These two limiting cases are used as bounds to compute the fraction of broken particles (f_b) for a given composite stress, and thus to obtain the stress-strain curve of the damaging composite.

Computation of f_b for the LLS mode – In LLS, not all particles are subjected to the same stress. We consider an extreme case of LLS where the load shed from a broken particle is transmitted only to its direct neighbours (and the surrounding matrix). To compute the fraction of broken particles in LLS, we adapt Batdorf's analysis developed for fibre composites [20]. For a given composite stress σ, the average stress on the particles in the loading direction (σ_r^{33}) causes a fraction P_1 of single particle breaks (called "singlets" in Batdorf's terminology). We assume a Weibull particle strength distribution,

$$P_1 = 1 - \exp\left[-\frac{V}{V_0} \cdot \left(\frac{\sigma_r^{33}}{\sigma_0} \right)^m \right] \tag{2}$$

with V the volume of an individual particle, V_0 a reference volume, σ_0 the stress value for which the fracture probability equals 63% for a particle of volume V_0, and m the Weibull modulus. Particles neighbouring these singlets in the same plane normal to the loading direction experience an increased stress

$$\sigma_n^{33} = c_1 \cdot \sigma_r^{33} \tag{3}$$

with c_1 the stress concentration factor on particles neighbouring a singlet. Due to this stress increase, some of the neighbouring particles may break, which may lead to the formation of clusters of two broken particles. Further stress redistribution occurs, which may in turn lead to the formation of clusters of i broken particles, called "i-plets" by Batdorf. The probability of formation of these i-plets is given by

$$P_i = P_1 \cdot \prod_{j=1}^{i-1} 1 - \exp\left[-\frac{N_j \cdot V}{V_0} \cdot \left(\frac{c_j \cdot \sigma_r^{33}}{\sigma_0} \right)^m \right] \tag{4}$$

with N_i the number of particles neighbouring an i-plet and c_i the stress concentration factor on particles neighbouring an i-plet. The total fraction of broken particle can be computed as

$$f_b = \sum_i i \cdot (P_i - P_{i+1}) = \sum_i P_i \cdot [i - (i-1)] = \sum_i P_i \quad . \tag{5}$$

The difficulty in this calculation is the computation of the c_i factors. This is done by assimilating an i-plet with an ellipsoidal inclusion of matrix phase (but with a lower Poisson's ratio) embedded in the composite material. Calculations give (relatively close) lower and upper bounds for c_i and are detailed in [21].

Computation of f_b for the GLS mode – Computation of f_b under GLS is done iteratively. Indeed, if we take the applied composite stress, σ, as the control parameter (i.e., if we assume load control), the stress on the intact particles σ_r^{33} and the fraction of broken particles are interdependent parameters that must be calculated together. This is done iteratively, starting with a first estimation of the increase in the fraction of broken particles caused by an increment in the composite stress σ. The remaining intact particles experience a further increased stress σ_{r1} (index $r1$ stands for intact reinforcement) and this causes additional particle breaks. Stress redistribution and further particle cracking thus continue until convergence is reached, i.e. until all remaining particles are strong enough to carry the increased applied macroscopic load.

Failure – Two modes of failure are simulated, either failure by the onset of tensile instability or abrupt failure. For both LLS and GLS, failure by tensile instability is predicted when Considere's criterion is reached

$$\sigma = \frac{d\sigma}{d\varepsilon} \tag{6}$$

with ε the composite strain. Sudden failure is possible only in LLS and is supposed to occur when a critical cluster of broken particles is created that leads to a fatal avalanche of damage. This brittle failure criterion is an adaptation of Batdorf's model and is defined as the stress for which an i-plet transforms into an i+1-plet under a vanishing small load increase.

Model predictions – Under LLS, the model captures the ductile-to-brittle transition that occurs with an increase of matrix strength, Fig. 5. Such a behaviour is known from experiment, e.g., Fig. 2 in this reference, or Fig. 2 of [11]). The used LLS rule provides a lower bound for strength, since only the nearest neighbours of a broken particle take part in the stress redistribution. A GLS model is used to provide an upper bound for the failure stress when failure is governed by tensile instability.

The model predicts a transition in failure mode with matrix strength also with other parameter sets (for example with $V_r = 20\%$). Perhaps the most important implication of this model is that increasing particle strength does eventually lead to a transition to ductile failure. This suggests that strong and ductile MMCs can be produced by combining a strong matrix with high strength particles. The LLS model also predicts a size effect on composite strength, i.e. brittle failure occurs earlier with increasing sample size (i.e., number of particles). Reference [22], in which a simpler version of the present model was presented, discusses this particular point in more depth.

Currently work is underway to apply the model to our experimental data. First results on the Weibull parameters inferred from tensile tests with our model are already available. The Weibull parameters of 35 µm angular particles embedded in either Al4.5%Cu or Al2%Cu are $m = 3.5$ and $\sigma_0 = 660$ MPa, as derived with the GLS model, or $m = 3$ and $\sigma_0 = 850$ MPa for the LLS model. The strength of the same particles embedded in pure aluminium, can on the other hand not be described by a simple stress-dominated Weibull distribution: as mentioned above, in these composites particle fracture is also governed by the average macroscopic strain. Physical reasons for this are that on a more local scale, the stress state is very inhomogeneous, not only from one particle to the other (inter-particle inhomogeneity), but also within each particle (intra-particle). Indeed, in these composites local stress concentrations are induced by particle-to-particle contacts within the composite, which can increase with plastic strain, as observed by Kouzeli [23]. This can explain how strain on the level of the homogenized continuum translates into stress on the level of individual particles.

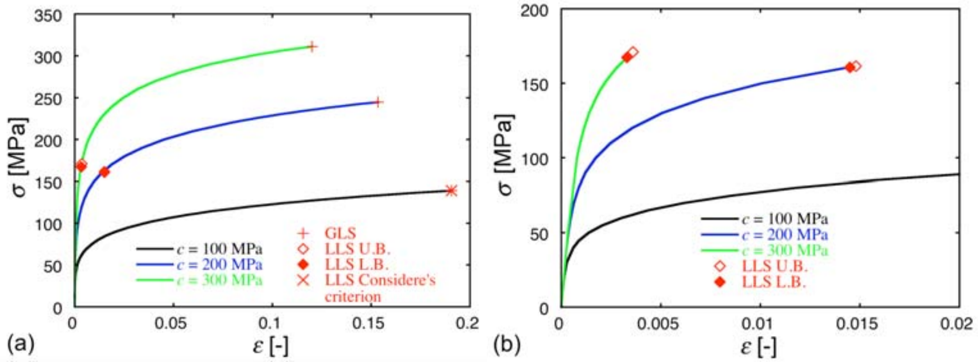


Figure 5: (a) Simulated composite stress-strain curves for different matrix strength coefficients, and failure stress as predicted for LLS and GLS. (b) Close-up at small strains for the LLS model. The symbols ◇,◆ denote abrupt failure, while + and ✕ denote failure by tensile instability. Power law matrix behaviour $\sigma_m = c\ \varepsilon_m^{\ n}$ ($n = 0.2$, $V_r = 0.5$, $N = 10^7$, $m = 3$, $\sigma_0 = 700$MPa).

Overall, despite its simplifications, the model summarized in what precedes captures essential features of the deformation, damage and fracture of matrix-inclusion-type materials that damage by particle cracking, and the basic characteristics of ductile as well as brittle behaviour are well described. The price to pay for the relative simplicity of the present analysis is that it is limited to uniaxial loading. The consideration of non-uniaxial loading would, however, induce major difficulties, which would be further enhanced for non-radial loading.

Conclusions

An analytical model has been developed that predicts the tensile curve and strength of particulate composite materials that undergo damage in the form of particle fracture. Two extreme modes of load sharing, a fully local and a global load sharing mode, are accounted for in the model, which yield a lower and an upper bound for the failure stress. Under LLS, the model predicts either of two different failure modes, *i.e.* failure by the onset of tensile instability or abrupt failure.

The model explains the experimentally observed ductile-to-brittle transition that is seen with an increase of matrix strength. Local load redistribution upon particle fracture can thus explain why some composites fail in an abrupt manner, via an avalanche-like growth of damage. The model also evidences the potential offered by high strength particles for producing strong composites.

Current work aims at the determination of the Weibull parameters of the particles from experimental tensile data. Model and experimental data also show that for a soft matrix, namely pure aluminium, the role of plastic strain on particle fracture cannot be ignored in densely packed composites on the level of the homogenized continuum. This finding is explained by particle-particle interactions, which are strongly enhanced by the close packing of particles. The role of strain has thus to be accounted for in failure modelling of these composites.

Acknowledgements

This research was supported by the Swiss National Science Foundation, Project No. 200020-107556.

References

[1] N. Chawla and K. K. Chawla: *Metal Matrix Composites* (Springer, United States of America 2006).

[2] T. W. Clyne and P. J. Whiters: *An Introduction to Metal Matrix Composites* (Cambridge University Press, United Kingdom 1993).

[3] S. L. Phoenix and I. J. Beyerlein, in: Comprehensive Composite Materials, edited by A. Kelly and C. Zweben, volume 1 Fibre Reinforcements and General Theory of Composites, chapter, 1.19, Pergamon (2000).

[4] W. J. Poole and N. Charras: Mater. Sci. Eng., A Vol. 406 (2005), p. 300

[5] W. J. Poole and E. J. Dowdle: Scr. Mater. Vol. 39 (1998), p. 1281

[6] D. J. Lahaie, J. D. Embury and F. W. Zok: Compos. Sci. Technol. Vol. 64 (2004), p. 1539

[7] A. Mortensen, in: Comprehensive Composite Materials, edited by A. Kelly and C. Zweben, volume 3 Metal Matrix Composites, chapter, 3.20, Pergamon (2000).

[8] A. Miserez, R. Muller, A. Rossoll, L. Weber and A. Mortensen: Mater. Sci. Eng., A Vol. 387-89 (2004), p. 822

[9] A. Miserez, S. Stuecklin and A. Mortensen: Mater. Sci. Technol. Vol. 18 (2002), p. 1461

[10] M. Kouzeli, L. Weber, C. San Marchi and A. Mortensen: Acta Mater. Vol. 49 (2001), p. 3699

[11] A. Miserez, A. Rossoll and A. Mortensen: Acta Mater. Vol. 52 (2004), p. 1337

[12] A. Miserez, A. Rossoll and A. Mortensen: Eng. Fract. Mech. Vol. 71 (2004), p. 2385

[13] R. Mueller, A. Rossoll, L. Weber, M. Bourke, D. Dunand and A. Mortensen: accepted for publication in Acta Materialia (2008)

[14] *ABAQUS Users' Manual 6.5* (Hibbit, Karlsson & Sorensen, Inc., USA 2005).

[15] C. Liu, S. Pape and J. J. Lewandowski, in: *Proceedings ICCI-II*, edited by H. Ishida, Elsevier, Cleveland (1988)

[16] R. Mueller and A. Mortensen: Acta Mater. Vol. 54 (2006), p. 2145

[17] S. Torquato: J. Mech. Phys. Solids Vol. 45 (1997), p. 1421

[18] S. Torquato: J. Mech. Phys. Solids Vol. 46 (1998), p. 1411

[19] A. Hauert, A. Rossoll and A. Mortensen: to be published

[20] S. B. Batdorf: J. Reinf. Plast. Compos. Vol. 1 (1982), p. 153

[21] A. Hauert, A. Rossoll and A. Mortensen: to be published

[22] A. Hauert, A. Rossoll and A. Mortensen, in: *Proceedings 16th Int. Conf. on Composite Materials (ICCM16)*, edited by K. Kageyama, paper on CD-ROM, Kyoto, Japan (2007)

[23] M. Kouzeli, L. Weber, C. San Marchi and A. Mortensen: Acta Mater. Vol. 49 (2001), p. 497