HIGH TEMPERATURE DEFORMATION AND FAILURE IN ALUMINUM-
ALUMINA PARTICULATE METAL MATRIX COMPOSITES

by

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Abstract

This work was primarily concerned with investigating the role of the reinforcing particles on plastic flow and fracture of metal matrix composites (MMCs) in the high temperature, moderately high strain rate domain. Uniaxial tensile, uniaxial compressive, and collar compressive tests were conducted on as-cast and extruded MMCs (the uniaxial tests were also conducted on unreinforced AA6061), and the flow and fracture behavior of the materials examined. The micro-failure characteristics were ascertained through metallographic examination of the fractured samples - both perpendicular and parallel to the fracture surface, and through numerical simulations of the collar compression test on the as-cast MMC.

The reinforcing particles were found to enhance the flow stress and lower the failure strain in these materials. The detrimental effect of the particles on failure strain was higher for the as-cast MMC - and this was deemed to be due to the presence of particle rich clusters in its microstructure. In both the extruded and as-cast MMCs, particle cracking and interfacial decohesion were the primary void nucleation mechanisms. The dominance of one mechanism over another was found to be temperature dependent - the former dominating at lower and the latter at higher temperatures. The MMC ductility peaked in the mid-temperature region, where the occurrence of the two mechanisms was comparable.
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1.0 Introduction

Particle reinforced metal matrix composites (MMCs), which incorporate hard and brittle ceramic particles into a ductile metal matrix, offer a large number of incentives over monolithic metallic materials, e.g. improved stiffness/density ratio, enhanced wear resistance and better high temperature properties. At the same time, some of these MMCs (e.g. cast Duralcan MMCs) are only fractionally more expensive than metals - thanks mainly to the advances in their primary processing. As a result, in the last decade, the technological significance of these materials has increased manifold. Today, the actual and potential applications of MMCs include automotive components like driveshafts, connecting rods, brake rotors and cylinder pistons, as well as bicycle frames, structural components of advanced fighter aircrafts, gas turbines, and electronic packaging materials.

Another advantage of (particulate reinforced) MMCs is the ability to process the material using traditional metal forming processes such as extrusion. However, these materials have reduced formability compared to monolithic metallic alloys. Duralcan Corporation, for example, reported significant surface cracking of as-cast AA6061-20%Al₂O₃ composites when hot rolling operations were attempted [Lloyd 1994]. Although a tremendous amount of experimental and theoretical work has been conducted on the room temperature mechanical properties of these materials, work on properties under industrial forming conditions (high temperature and moderately high strain rates), has been limited. This has, to a large extent, prompted the present study - investigating
the plastic flow and fracture phenomena in Duralcan MMCs under these conditions is the primary objective.

Furthermore, we also seek to address the following issues:

(1) The last two decades has witnessed giant strides in our understanding of different issues related to the strength of metal matrix composites. Today, the physics of strengthening in these materials (especially at low temperature and low strain rate) is relatively well understood - the onus is on finding effective approximations to these strengthening mechanisms (making the physics more amenable to numerical treatment). Our understanding of fracture, on the other hand, is much more limited. Although the general mechanism of failure in MMCs, viz. void nucleation at second phase particles, and subsequent growth and coalescence of these voids, has been successfully identified, many specific details of these sub-processes still elude us. Consequently, microscopic examination of the MMC failure mechanism is an important part of the present study.

(2) In MMCs, damage events are primarily concentrated at second phase particles. Hence, a study on the effect of distribution characteristics of these particles on MMC failure is important. Two Duralcan AA6061-20%Al₂O₃ composites were used for this purpose - one in as-cast, and the other in extruded condition. The particle distribution was anticipated to be more uniform in extruded MMC [Chen 1994], and a comparison of the failure strains and mechanisms in the two materials was expected to bring out the importance of this parameter.

In the present study, a series of mechanical tests were conducted on MMCs (and the unreinforced matrix), where the temperature, strain rate and loading characteristics
were varied, and the corresponding strain to failure and flow properties were measured. To understand the micro-mechanisms of MMC failure under these conditions, sample sections, both parallel and perpendicular to the fracture surface, were metallographically prepared and examined. Some of the tests (collar compression tests on the as-cast MMC) were also numerically simulated in order to estimate the local stress and strains responsible for fracture in these materials.

The thesis is organized in the following manner: Chapter 2 reviews some of the available literature on the flow and fracture characteristics of metal matrix composites. Chapter 3 describes in detail the experimental and numerical techniques used in this work. The results are presented and analyzed in Chapter 4, and discussed in Chapter 5. Finally, Chapter 6 is a summary of the important conclusions that can be drawn from the present work and suggestions for future work in this field.


2.0 Background Information

This chapter briefly reviews the relevant issues related to the mechanical response of metal matrix composites. In section 2.1, the processing and microstructural aspects, as well as the application of MMCs are discussed. The deformation and failure behavior of these materials are subsequently considered in sections 2.2 and 2.3.

2.1 Metal Matrix Composites

2.1.1 Introduction

The microstructure of a metal matrix composite primarily contains two phases, the metallic matrix and the ceramic reinforcement. The choice of these two components is dependent on many factors, including the processibility, cost and applicability of the materials. Al and Ti are the most widely used matrix materials; Al for low temperature (up to 300°C) and Ti for elevated temperature usage. Al₂O₃ and SiC are the most common reinforcements, primarily owing to their suitable properties and low cost. Both are actually side products of other industries: SiC from grinding grit and Al₂O₃ from the high temperature refractory industry. Of the two, SiC enjoys relatively better mechanical properties; but suffers from its relatively higher cost and from the fact that it reacts with Al in absence of sufficient Si in the matrix. Al₂O₃, on the other hand, is widely available, chemically inert to the matrix, and forms a strong interfacial bond with Al [Clyne and Withers 1993].
The metallic matrix may be continuously or discontinuously reinforced. While the continuous ceramic reinforcements provide the greatest gain in the axial mechanical properties (moduli and strengths), these MMCs are seriously handicapped by high cost, anisotropic properties, and limitations on forming through conventional large strain deformation processes, e.g., extrusion or wire-drawing. In discontinuously reinforced MMCs, the reinforcements may be in the form of whiskers, fibers or particulates. Typically, the fiber reinforcements have an aspect ratio of 5, while that for the particulates is much smaller (~1-2, see Figure 2.1). The particulate reinforcements are, in general, more economical than the corresponding short fibers. Some of the short fibers are also reported to be environmentally unfriendly, causing asbestos-like health problems [Ibrahim et. al. 1991]. This leaves particulate reinforcements as a popular choice for the MMCs.

2.1.2 Processing Route

The MMC fabrication processes include a host of liquid and solid state processing routes, e.g., squeeze casting, slurry casting, direct chill casting, and powder metallurgy. The direct chill (DC) casting route is possibly the most economic of them all. Al₂O₃ (particulate) reinforced aluminum composites have been produced by Duralcan since 1988 by direct chill casting. The casting route, however, suffers from a relative lack of microstructural control, mainly in the distribution of the second phase (Al₂O₃) particles. The as-cast microstructure contains both particle-free and particle-clustered regions
Figure 2.1  Schematic depiction of the three types of MMCs, classified according to type of reinforcement [Clyne and Withers 1993].

Figure 2.2  Particle distribution in as-cast microstructure (A356-15%SiC) [Michaud 1993].
(see Figure 2.2), due to a variety of causes, e.g., particle agglomeration and sedimentation in the melt, gas bubble entrapment, and particle pushing by the advancing solidification front. This reinforcement segregation in the cast MMCs has a detrimental effect on the consistency and homogeneity of their mechanical response, and is believed to severely affect their ductility [Llorca et al. 1991]. In fact, the ductility of the Duralcan as-cast composites is so low that extrusion (using an overall compressive stress state) of these billets is the only feasible industrial secondary processing route.

2.1.3 The Age-hardenable Matrix

For composites with Al matrices, age-hardening alloys are usually chosen (2XXX, 6XXX and 7XXX alloys) in order to increase the strength of the composite. Of particular interest in this study is AA6061 - containing 1%Mg, 0.6%Si, 0.5%Cu and 0.2%Cr. It is generally believed [Lutts 1961, Kovacs et al. 1972] that the sequence of precipitation in this alloy is:

vacancy-Si clusters - vacancy-rich coherent Al-Mg-Si GP-I zones (spheres) - ordered, partially-coherent GP-II needles (β") - semicoherent, hexagonal rods (β') - incoherent, equilibrium β Mg₂Si plates.

These products will be present in the microstructure in different proportions, depending on time and temperature, their typical size being 5-20 nanometers. Thus, a typical MMC with an age-hardenable matrix will be reinforced by particles of two distinctly different sizes: the ceramic particles in micron scale range and the age-
hardening precipitates in the nanometer scale range. The matrix age-hardening is actually more effective than the ceramic reinforcement in elevating the low temperature material strength. Typically, the addition of 20% SiC to an annealed AA6061 matrix (yield strength: 55 MPa at room temperature [Al Standards 1978]) will raise the yield strength by around 100 MPa [Arsenault 1986], while the T6 treatment of the same alloy can raise the yield strength by 220 MPa [Al Standards 1978]. However, the major advantage of ceramic reinforcements is the large volume fraction which can be used and the resulting change in material stiffness, wear resistance and coefficient of thermal expansion.

2.1.4 Applications

Over the last one and half decades, discontinuously reinforced metal matrix composites have slowly matured into a viable advanced material. One of the earliest production usage of these MMCs was in Toyota cars (1983), when the Ni-resist iron piston ring inserts were replaced with those made of Al-5%Al₂O₃ [Kubo et. al. 1988]. Production of these pistons in Japan now run into millions of units annually.

Today, MMCs are being used in driveshafts for trucks and larger passenger cars (Corvette), engine block liners (Honda Prelude), bicycle frames (Stumpjumper M2), as electronic packaging material and as heat sinks in microprocessor based appliances [Zweben 1992]. Potential applications include brake rotors (in trial stage by both Ford and Toyota), brake calipers and connecting rods in automobiles, structural components of advanced fighter aircraft and gas turbines [Koczej et. al. 1993].
2.2 Deformation of Metal Matrix Composites

2.2.1 Strengthening in Multi-phase Materials

Prior investigations (e.g. [Fleck et. al. 1994]) have shown that the strengthening due to reinforcements larger than 10μm can be safely modeled through continuum mechanics. Continuum mechanics assumes that a body is composed of a finite number of points, and the field-variables (stress, strain etc. in case of deformation) are uniform in the immediate vicinity of these points. The application of continuum mechanics to the numerical prediction of micro-deformation has given rise to an exciting branch of study called micro-mechanics. Being intrinsically scale-independent, the micro-mechanical models cannot predict the effect of the reinforcement size on the strengthening of the composite, but several studies [Christman et. al. 1989, Llorca et. al. 1991] have examined the effects of shape, volume fraction and the intrinsic properties of the reinforcements on the composite deformation and failure characteristics. They show that the presence of the second-phase particles induce deformation constraints, and perturb the deformation fields locally. Stress concentration is observed in the stiffer second phase particles, while strain concentration occurs in the adjoining matrix.

An outstanding problem in this field is to predict the stress and strain partitioning between the two phases as a function of reinforcement characteristics (shape, volume fraction) and the imposed strain. The solution strategies and their effectiveness vary, depending on many things, viz. nature of deformation (elastic/ plastic), scale of the response of interest (macro/ micro) and the number of particles present in the domain (one-particle/ many-particles).
2.2.1.1 Effect of the Particle Shape

The effects of particle shape on the macro and micro-deformation fields have been studied, both experimentally and numerically, by several authors [Christman et. al. 1989, Bao et. al. 1991, Shen et. al. 1995]. They found that the strengthening efficiency of the reinforcement particles increased with angularity and aspect ratio. Thus, long whiskers (aspect ratio ~ 5) strengthened the matrix much more than spherical particles (see Figure 2.3). However, the higher strengthening of the longer, more angular particles came at a price. The stress and strain concentrations were more acute in these cases, and invariably, these particles failed earlier than the less angular ones (see Figure 2.4).

2.2.1.2 Effect of the Particle Volume Fraction

The composite steady-state flow stress increases with the volume fraction of the reinforcing particles (see Figure 2.5) [Christman et. al. 1989]. However, to a limited extent, the strengthening effect of the particles will depend on their distribution in the matrix. The reinforcements, as noted previously, will elevate the local stress/strain fields. The possibility of overlap of these locally elevated stress (or strain) fields depends on the inter-particle spacing, and thus, in an average sense on the reinforcement volume fraction. It has been shown [Poole 1993] that if the reinforcement volume fraction is less than 30% and the particle distribution is uniform, the overlap of the stress (or strain) concentration fields is minimal. However, in particle-reinforced MMCs, the particle distribution is hardly uniform, with the existence of high-density (particle clusters) and low-density (particle depleted) zones. In the particle clusters, the volume fraction of the reinforcements may be much higher than 30% (locally), thus giving rise to significant
Figure 2.3. Strengthening ratio (ratio of flow stress: composite to unreinforced matrix) for an elastic-perfectly plastic matrix reinforced by aligned ellipsoidal particles - numerical simulation [Bao et al. 1991].
Figure 2.4 Contour plots for void volume fraction ($f$), effective plastic strain ($\varepsilon$) and hydrostatic stress ($\sigma_m/\sigma_0$) for 2124Al-12.5%SiC composites (numerical simulation). (a) Whisker reinforcement, $\varepsilon_{\text{imposed}} = 0.013$. (b) Particle reinforcement, $\varepsilon_{\text{imposed}} = 0.038$. (c) Sphere reinforcement, $\varepsilon_{\text{imposed}} = 0.052$ [Llorca et al. 1992].
overlap of the concentration fields. In these regions, the particles bear a disproportionate share of the load, and effectively shield the matrix material trapped within. The initiation of composite failure has been sometimes attributed to this clustering phenomena [Llorca et. al. 1991].

Taking account of these clusters is a prominent problem in MMC deformation. Several problems arise in the characterization of real clusters. Since the stress/strain fields around the reinforcements are radially non-uniform, the position of the particles are important in determining the overall characteristics of the cluster deformation field. A cluster can be made of dozens of particles, which calls for a very large number of model parameters. In addition, the exact physics of the cluster formation is not very well known, and any attempt at cluster characterization can, at best, be of an empirical nature. Several investigators [Tvergaard 1990, Poole 1993] have attempted to model 'simplified clusters' in 2-dimensions by assuming a particular type of particle arrangement, usually triangular or square (see Figure 2.6). Of these two types of arrangement (square and triangular) the square arrangement of the reinforcements seems to give rise to more acute stress and strain concentrations [Poole et. al. 1994]. Other investigators have tried to characterize a cluster using 'nearest neighbor distance' parameters, and have used Dirichlet tessellation [Ghosh et. al. 1997] to characterize the cluster microstructure.

2.2.2 Temperature and Strain Rate Effects on Deformation

The flow stress of the metallic matrix decreases with increasing temperature and decreasing strain rate [Sellers and Tegart 1966]. The ceramic flow stress, on the other
Figure 2.5  Effect of reinforcement volume fraction on MMC steady state flow stress - AA1100-SiC [Christman et. al. 1989].

Figure 2.6  Schematic representation of (a) square, and (b) triangular particle arrangement [Poole 1993].
hand, is relatively insensitive to deformation conditions (at the temperature and strain rate range of interest for metallic materials). As for the overall response of the composite, the temperature and strain rate effects on it are qualitatively similar to the matrix. However, the extent to which a given change in temperature or strain rate will affect the flow stress for the composite will be different from that for the matrix [Bao et. al. 1991, Chen 1994].

The generalized (empirical) formulation incorporating the effects of temperature and strain rate on the strength of metals and metallic alloys was proposed by Sellers and Tegart [1966] and later modified by Frost and Ashby [1982]

\[ \dot{\varepsilon} = A \left( \sinh \frac{\alpha \sigma}{\mu} \right)^n \exp \left( -\frac{Q}{RT} \right) \]  \hspace{1cm} (2.1)

where, \( \dot{\varepsilon} \) is the imposed strain rate, \( \sigma \) the strength, \( \mu \) the shear modulus, and \( A, \alpha \) and \( n \) are experimentally determined constants. At low stresses (\( \alpha \sigma < 0.8\mu \)), the hyperbolic-sine function reduces to a power-law relation (creep equation)

\[ \dot{\varepsilon} = A_1 \left( \frac{\sigma}{\mu} \right)^n \exp \left( -\frac{Q}{RT} \right) \]  \hspace{1cm} (2.2)

At high stresses (\( \alpha \sigma > 1.2\mu \) or \( \sigma > 10^{-3} \mu \)), on the other hand, it reduces to an exponential relationship

\[ \dot{\varepsilon} = A_2 \exp \left( \frac{\beta \sigma}{\mu} \right) \exp \left( -\frac{Q}{RT} \right) \]  \hspace{1cm} (2.3)

where, \( \beta = \alpha n \), so that \( \alpha \) and \( n \) can be determined uniquely from high and low stress tests.

The shear modulus for Al at 300°C \( \sim 21 \) GPa [Frost and Ashby 1982], and thus, in most situations (except for creep) the stress level is high enough for the exponential law.
The MMCs generally show much higher strength levels than the matrix at lower strain or temperature. This is due to load transfer to the reinforcing phase [Corbin 1992], and the higher work hardening in the constrained matrix [McDanel 1985]. However, the matrix constraint and the load transfer mechanism is relaxed by dynamic recovery at higher strains [Taya et. al. 1991], by thermal softening at higher temperatures, or by diffusional processes at lower strain rates (and high temperatures).

At elevated temperatures and low strain rates ($10^{-4}$/s), the gain in strength due to the ceramic reinforcements is relatively minimal. Zhao et. al. [1994] observed that both the yield and the tensile strength for cast and extruded MMCs (AA2014-T6 reinforced with 10%, 15% and 20% Al$_2$O$_3$) differed little from that of the matrix above 200°C.

Bao et. al. [1991] have tried to numerically model the strengthening behavior of the MMCs at ambient temperatures and quasi-static strain rates ($-10^{-4}$/s). Two types of matrices have been investigated: an elastic-perfectly plastic and an elastic-power law hardening (Ramberg-Osgood) matrix. The MMC was found to follow a similar stress/strain response as the matrix (though at a higher stress level) and its flow stress tended to a strength value which depended on the volume fraction and the shape of the reinforcements.

The strain rate sensitivity of MMCs (AA6061-T6-20%Al$_2$O$_3$) at room temperature has been studied by Yadav et. al. [1995] over a wide range of strain rates ($10^{-4}$/s to $6\times10^5$/s). They reported that both the monolith (AA6061-T6) and the MMC were relatively strain rate insensitive over a wide strain rate range ($10^{-4}$/s - $10^2$/s). However, at elevated temperatures, the strain rate sensitivity of the materials increase. Chen [1994],
conducting a series of uniaxial compression tests on AA6061-Al2O3 (10%, 15% and 20%) in the temperature range of 400°C - 525°C and at strain rates of 0.05/s, 0.1/s, 1/s and 10/s, reported a moderate increase in flow stress with increasing strain rate.

2.3 Fracture of Particle Reinforced Metal Matrix Composites

The failure of Al based MMCs is primarily ductile. The failure mechanism consists of (a) micro-void nucleation, due to particle fracture or interfacial decohesion, (b) void growth, and (c) void coalescence into a macro-crack.

2.3.1 Void Nucleation

(a) Due to Particle Fracture

The ceramic particles reinforcing the matrix are brittle in nature, and contain microscopic flaws inherited from production or prior deformation processes. The critical stress for fracture of these reinforcements depend on the size of the flaws [Griffith 1921]

\[
\sigma_f = A\sqrt{\frac{2E\gamma}{c\pi}}
\]

(2.4)

where \( E \) is the Young’s modulus, \( \gamma \) the fracture surface energy, \( c \) is the crack length, and \( A \) is a constant. Thus, depending on the pre-existing crack length in the particles, the fracture will occur over a stress range, the distribution of which can be modeled using Weibull statistics [Caceres and Griffiths 1996]. The probability of fracture (\( p \)) in a particle of volume \( V \) under stress \( \sigma_p \) is given by
\[ p = 1 - \exp \left[ -\frac{V}{V_0} \left( \frac{\sigma_p}{\sigma_0} \right)^m \right] \quad (2.5) \]

where \( V_0, \sigma_0, \) and \( m \) are constants for particular volume of the reinforcing phase.

The probability of particle fracture increases with particle size and aspect ratio (see Figure 2.7), and in the presence of sharp corners and sides on the particle. This is primarily due to the increase in stress on the particle under these conditions [Nutt and Needleman 1987, Chen 1994]. Elongated particles with their major axis aligned in the loading direction have been found to be most prone to cracking (see Figure 2.8) [Whitehouse et. al. 1991].

(b) Due to Interfacial Decohesion

Particle cracking is the dominant micro-void nucleation mechanism at low temperatures. At elevated temperatures, with the softening of the matrix, the stress on the particles is no longer high enough for cracking, and decohesion of the metal matrix interface assumes critical importance (see Figure 2.9). Zhao et. al. [1994] found that this transition temperature depends, among other things, on the volume fraction of the reinforcements (175°C, 200°C and 250°C for 10%, 15% and 20% Al₂O₃ reinforced AA2014 respectively).

Gurland and Plateau [1963] proposed that voids would be nucleated at the interface when the local elastic strain energy is comparable to the surface energy of the newly created surfaces. Later, Tanaka et. al. [1970] showed that this energy criterion is fulfilled almost immediately upon yielding for any particle larger than \( \sim 25 \text{nm} \), and
Figure 2.7(a) Probability of particle cracking as a function of particle size $d$ at different strains [Brechet et al. 1991].

Figure 2.7(b) Probability of particle cracking as a function of particle aspect ratio $a$ at different strains [Brechet et al. 1991].
Figure 2.8  Cracking of an elongated particle aligned along the loading direction [Whitehouse et. al. 1991].

Figure 2.9  Decohesion at particle-matrix interface [Whitehouse et. al. 1991].
Brown and Stobbs [1971] pointed out that this is really a necessary rather than a sufficient condition. Now, it is generally believed that the development of a critical normal stress across the interface is essential in the decohesion mechanism [Lloyd 1991]. However, under certain circumstances, the decohesion may also be shear dominated [Needleman 1990].

Thus, the metal matrix interfacial strength, and in turn, the interface structure, assumes importance. Lee et al. [1997] have studied the interfacial structure of the Duralcan AA6061-Al₂O₃ MMCs using X-ray and convergent beam electron diffraction techniques. They conclude that the main reaction product is MgAl₂O₄ spinel (octahedral shaped crystals with average size of ~1μm). However, to date, little, if any, information is available on the interfacial strengths of the particulate MMCs under different operating conditions (temperature, strain rate etc.)

2.3.2 Void Growth

Once the micro-voids have nucleated, the stress-free surface of the voids causes a localized stress and strain concentration in the plastic field. As a result, with continuing plastic flow, the micro-void undergoes volumetric growth and shape change of a much greater magnitude than the far-field strain. If the micro-voids are nucleated far apart, thus ensuring zero interaction of the perturbation fields (i.e. perturbation of the deformation fields due to the presence of voids), it is possible to develop an adequate model in terms of a single void in an infinite plastic matrix. Rice and Tracey [1969] developed one such
model for a spherical void in a rigid, non-hardening, and an isotropic linear hardening material. They found that:

(a) under uniaxial loading, the void predominantly grows in the loading direction, with little or no lateral growth, and

(b) the growth rate increases with hydrostatic stress ($\sigma_m$), and decreases with matrix hardening.

The Rice-Tracey model investigates the growth of an isolated void. In real microstructures, however, the void (and void growth paths) are associated with other voids and second-phase particles. Neglecting this void-void and void-reinforcement interaction may, in turn, introduce serious limitations to the final failure characteristic predictions. However, the microscopically realistic simulation of ductile failure process demands unrealistic computer time and memory, and is usually solved only in an average or ‘diffuse’ manner [Gurson 1977].

In MMCs, the presence of the particles have been demonstrated to raise the hydrostatic stress levels in the matrix, especially at low temperatures [Poole et. al. 1994]. In fact, the local hydrostatic stresses may be so high that the imposition of an additional global hydrostatic stress (as in a notched tensile specimen) may not appreciably affect the failure strains. Somerday and Gangloff [1994] found that the ratio of failure strains for smooth and notched tensile specimens of AA2009-T6-20%SiC equaled unity for tests at 25-200°C, thereby rising to 3.3 at 250°C. The ratio was ~2 for the matrix material at room temperature. This high local hydrostatic stress, in turn, may be expected to elevate the nucleation and growth rates of the micro-voids.
2.3.3 Void Coalescence

Under uniaxial loading, the voids grow primarily in the loading direction. The final failure is generally brought about by the shear failure of the inter-void matrix, when the void length almost equals the inter-void spacing [Brown and Embury 1973]. Figure 2.10 schematically depicts the void growth and coalescence in metal matrix composites. The localized necking of the inter-void matrix, which marks the beginning of the void coalescence process, coincides with the attainment of the plastic limit-load for the inter-void matrix [Thomason 1968]. This limit-load criterion for the matrix failure can be represented in terms of the critical value of the mean stress $\sigma_n$, which is required to initiate localized plastic flow or necking in the inter-void matrix of the porous solid. When the micro-voids are small and widely spaced, $\sigma_n > \sigma_1$ (the maximum principal normal stress), and the localized limit load failure is prevented because

$$\frac{\sigma_n}{2\sigma_{yn}}\left(1 + \sqrt{V_f}\right)^{-1} > \frac{\sigma_1}{2\sigma_y} \quad (2.6)$$

where $\sigma_n$ and $\sigma_{yn}$ are the macroscopic and the intervoid yield stresses respectively and $V_f$ is the volume fraction of the voids. As the void length approaches the inter-void spacing, (a) $\frac{\sigma_n}{2\sigma_{yn}}$ decreases until $\sigma_n = 2\sigma_{yn}$, and (b) $V_f$ increases, both of which contribute to the attainment of the inter-void matrix limit load.

2.3.4 Experimental Observations

Over the last decade, it has been intensely debated whether void coalescence (and the 'macro-crack') follows the reinforcement particles ('the weak link'), or is primarily
Figure 2.10  Schematic representation of (a) growth and (b) coalescence of voids in particle reinforced MMCs, leading to (c) final failure [Clyne and Withers 1993].
random. Flom and Arsenault [1987] found a higher SiC particle concentration on the fracture surface than in the matrix, whereas other studies have found a lower concentration [Liu et. al. 1988], or the same concentration [You et. al. 1987]. In any case, the role of the matrix in the final failure of MMC seems to be far from insignificant. Using AA6061 based MMCs (10%, 15%, and 20% SiC reinforced) under uniaxial tension at low strain rates (0.002/s), Lloyd [1991] found that for a peak aged matrix, strain localization occurred at early stages of deformation, and made a significant contribution to the fracture strain, unlike that in a naturally aged material. Particle fracture (controlling void nucleation mechanism under those conditions) was also found to be extremely localized in case of the peak aged material, whereas it was almost uniformly distributed across the gauge length for the naturally aged material.

The effect of temperature on the failure characteristics of Al-based discontinuously reinforced MMCs is another issue of importance. Whitehouse and Clyne [1993a, b] have investigated this problem at low strain rates (~ 5x10\(^{-3}\)/s). They [1993a] reported that at elevated temperatures (220\(^{\circ}\)C and 350\(^{\circ}\)C), the onset of void nucleation was delayed, but after void nucleation, the subsequent material failure was extremely fast. They also noted that damage at elevated temperatures was highly concentrated around the fracture surface. In contrast, at room temperature [1993b], stable cavities formed well before final failure and were more uniformly dispersed throughout the gauge length.

The variation in failure strain in particulate reinforced MMCs has been studied by Syu and Ghosh [1994a]. They tested a AA2014 based MMC (15% Al\(_2\)O\(_3\) reinforced) under uniaxial tension at elevated temperatures (300\(^{\circ}\)C, 400\(^{\circ}\)C and 500\(^{\circ}\)C) at moderately
low strain rates (0.1/s). The composite ductility was found to be maximum at 400°C, whereas that of the monolith increased with temperature. Both the void nucleation strain and the failure strain were found to be higher at 400°C than that at 500°C. Similar ductility reversal in the composite has also been reported by Pickens et. al. [1987] in AA6061-20%SiC and 7090-20%SiC under hot torsion tests (at $\dot{\varepsilon} \sim 0.1/s$) in the temperature range of 260°C to 593°C. Syu and Ghosh [1994a] ascribe the composite fracture to transgranular creep at 300°C and 400°C, and to intergranular creep at 500°C.

There has been a large volume of low temperature (and low strain rate) work on deformation and fracture of metal matrix composites. On the other hand, the high temperature (and moderately high strain rate) behavior of these materials is relatively unexplored, even though this regime encompasses important large scale deformation processes like hot rolling and extrusion. The importance of the present work lies herein.
3.0 Experimental Methodology

The principal aim of the experimental work was to conduct a fundamental study on the deformation and ductility of AA6061-Al$_2$O$_3$ metal matrix composites at moderately high strain rates (0.1/s - 10/s) and high temperatures (300°C - 550°C). Three types of tests were conducted: uniaxial compression, uniaxial tension and the collar compression tests. An AA6061 based Duralcan MMC reinforced with 20% by volume (nominal) Al$_2$O$_3$ was tested at the range of temperature and strain rates. The MMC was originally manufactured through the DC casting route by Duralcan Inc., and later extruded into rods. It was tested both in the as-cast condition, as well as after extrusion. Subsequently, we shall be referring to them as as-cast and extruded MMCs, respectively.

3.1 Heat Treatment of the Samples

To ensure an uniform starting microstructure, all samples were solution heat treated for 8 hours at 530°C in an electric furnace. They were put in a glass tube, and a thermocouple was placed in proximity to the samples to monitor the local temperature during the solution heat treatment.

Further, to stabilize the microstructure at the test temperature (this may involve dissolution, or precipitation and coarsening depending on temperature), the samples were again heat treated just prior to testing. The treatment times, in this case, depended upon the test temperatures: 40 minutes at 300°C, 10 minutes at 425°C, and 5 minutes at 550°C.
3.2 MMC Characterization

A Buehler Ecomet IV automatic grinder/polisher was used for metallographic preparation of the MMC specimens. The grinding/polishing closely followed the routine suggested by Wolters [1996]. The grinding sequence was of 180, 320, 500 and 1000 grit SiC paper. A uniform pressure of ~40 psi (for a full load of 10 samples) was applied, and the grinding wheel was rotated at ~200 rpm. At least 3 papers were used at each step, with the grinding time per paper being ~1 minute. The polishing procedure used 6 μm and 1 μm diamond, and 0.05 μm silica suspensions on Texmet and Chemomet polishing cloths, respectively. During polishing, the pressure was reduced to ~25 psi and the platen rotation speed was reduced to ~150 rpm.

Previous investigations [Christman et. al. 1989, Llorca et. al. 1991, Shen et. al. 1995] indicate that the reinforcement characteristics (the size, shape and distribution of the ceramic particles) are of primary importance in determining the macro-deformation response of the MMC. However, to date, no suitable spatial distribution function exists for effectively characterizing the reinforcement arrangement in the matrix [Wilkinson et. al. 1996]. Thus, the MMC characterization was limited to determining the axial lengths, aspect ratio, equivalent diameter (= diameter of a circle with same area) and area fraction of the ceramic particles. The polished sections of the failed tensile specimens were used for the measurements, which were carried out on the C• Imaging Systems (version 5.1) and the UTHSCSA (version 1.28) image analysis software. The measurements were restricted to regions far away from the fracture surface in order to eliminate the cracked or decohesed particles from the analysis.
3.3 Mechanical Testing

The influence of reinforcement particles on the high temperature deformation (at moderately high strain rates) of the MMCs was of fundamental concern. Therefore, unreinforced AA6061 was thus used as the control material, and it was subjected to the uniaxial tensile and compression tests only. For the MMCs, both the uniaxial tests, as well as the collar compression tests were done. To study the deformation behavior of these materials over a wide temperature and strain rate range, they were tested at three different temperatures: 300°C, 425°C, and 550°C and at three nominal strain rates: 0.1/s, 1/s and 10/s (for the as-cast MMC, however, the uniaxial compression tests were restricted to only one strain rate - 1/s). At each condition, at least two tests were conducted.

All the tests were performed on an MTS/Instron 8500 servohydraulic machine. The samples rested on two stainless steel platens, which were heated using two 500 W heating cartridges in each. The platens were assembled on top of larger stainless steel platens, which were cooled by a constant flow of room temperature water (see Figure 3.1). In order to make the heating of the top platens more efficient, they were separated from the lower ones by ceramic inserts. The heating was controlled using OMEGA microprocessor based temperature controllers, one for each heating platen. The time allowed for heating from the starting to the test temperature was: 10 minutes for 300°C and 425°C, and 15 minutes for 550°C. Typically, the platens could be heated to 300°C, 425°C and 550°C in this time using 35%, 45% and 55% of the total power respectively.
Figure 3.1  Schematic diagram of the assembly used for mechanical (compressive) testing.
The specimen center-temperature was found to stabilize within 3 minutes (as measured by an independent thermocouple bored into the centre of a compression specimen) of the platen temperature stabilization. As noted in Section 3.1, all the samples were also re-heat treated (treatment involving precipitate dissolution or coarsening) just before testing. During the tests, they were covered with Saffil fibers (pure Al₂O₃ fibers) to reduce the convective and radiative heat loss to the surroundings.

During the mechanical tests, data (time, load and displacement) were acquired directly on a PC using the Instron Wavemaker software, resulting in at least 200 points per sample (except in the tensile tests at nominal strain rate of 10/s). However, since no extensometer was used for these tests, it was not possible to evaluate the initial stages of the compressive or tensile curves. The stress-strain curves were generated from the load-displacement curves.

\[ \varepsilon_{\text{eng}} = \frac{\delta l}{l_0} \]  \hspace{1cm} (3.1)  
\[ \varepsilon_{\text{true}} = \ln(1 + \varepsilon_{\text{eng}}) \]  \hspace{1cm} (3.2)  
\[ \sigma_{\text{eng}} = \frac{P}{A_0} \]  \hspace{1cm} (3.3)  
\[ \sigma_{\text{true}} = \sigma_{\text{eng}}(1 + \varepsilon_{\text{eng}}) \]  \hspace{1cm} (3.4)

where, \( \sigma_{\text{true}} \), \( \sigma_{\text{eng}} \), \( \varepsilon_{\text{true}} \) and \( \varepsilon_{\text{eng}} \) are the true and engineering stresses and strains respectively, \( \delta l \) is the change in length (measured by cross-head displacement) of the specimen of original length \( l_0 \), \( P \) is the measured load (in KN) and \( A_0 \) is the original cross-sectional area of the specimen (in mm²).
The primary error is expected to be due to the inaccuracy in measuring the extension of the sample in absence of an extensometer. No correction was made for the machine compliance, as we were primarily concerned with plastic deformation where this correction will be small.

3.3.1 Uniaxial Compression Testing

The material steady state flow stress under uniaxial loading was estimated through the uniaxial compression tests. A 50 KN load cell was used for these tests. The sample was put under load control in the MTS machine (the preload corresponding to ~3 MPa nominal stress). This preloading was critical in achieving good contact between the platen and the specimen during the heat treatment step prior to testing. The test was then shifted to displacement control (where the test was conducted) so as to achieve a constant cross-head velocity.

Though a constant cross-head velocity does not suggest a constant strain rate, the variation was minimal when compared to the widely different nominal strain rates used for the tests. As an example, let us consider a cross-head moving at a constant velocity of 1 mm/s, compressing a sample of initial length 10 mm. After 1s, the length of the specimen is 9 mm, and the true strain is

$$\varepsilon_1 = \ln(1+1/10) = 0.1.$$

At the end of the 4th second, the sample length is 6 mm. The true strain introduced during the 5th second is thus

$$\varepsilon_5 = \ln(1+1/6) = 0.15.$$
The total strain at this time is

\[ \varepsilon_{\text{total}} = \ln(1+5/10) = 0.4 \]

Thus, while introducing a strain of 0.4 at a uniform cross-head velocity of 1 mm/s, the strain rate varied from 0.1 to 0.15. The nominal strain rates used for the tests were 0.1/s, 1/s and 10/s, and in comparison, this variation was considered negligible.

The samples were machined with pocket depressions of 0.2 mm on both end faces leaving a rim of ~1 mm (see Figure 3.2 (a)). The pockets were filled with a mixture of colloidal graphite and Dow Corning vacuum grease for the 300°C and 425°C tests. This ensured uniform deformation of the compression specimen - even at large deformations (\( \varepsilon > 0.4 \)) no significant barreling was observed. For the 550°C test, lead borosilicate glass powder (softening point ~425°C) was used. The glass powder was less effective as a lubricant and the samples showed some evidence of barreling. The flow stress, however, still reached a constant value at very low strain (less than 5% strain).

During the early part of deformation of the rimmed sample, only the rim deformed, and thus, a much softer stress-strain response was obtained. This early part of the curve was neglected. Displacement was zeroed to the point when the actual sample started deforming.

3.3.2 Uniaxial Tensile Testing

The tensile tests were conducted to estimate the macro-ductility of the materials under uniaxial loading. Figure 3.2 (b) shows a typical uniaxial tensile specimen used for our tests. The specimens were held in a split-half grip that screwed into the cross-head.
Figure 3.2  Schematic representation of a typical (a) compressive, and (b) tensile sample used for the mechanical tests (dimensions in mm).

Figure 3.3  Schematic diagram showing the collar compression sample (dimensions in mm).
A displacement controlled testing procedure, as explained above, was used. All the tests were conducted with a load cell of 5 KN. Prior to the testing, the system alignment was checked at room temperature using a low carbon steel specimen (strain measured using an extensometer), whose Young's modulus (from the stress-strain curve) was calculated and matched with the literature value. In all the cases, the matching was within +/-5%.

3.3.3 Collar Compression Testing

The collar compression specimen is shown in Figure 3.3. In this test, while the central barrel is compressed axially, it expands in the radial direction. As a result, the collar experiences a tensile hoop strain.

The testing methodology in this case was very similar to that followed in uniaxial compression tests. However, in this case the sample was deformed in steps, each step introducing a hoop strain of ~0.1 in the collar (in case of the as-cast MMCs, the collar hoop strain introduced in each step was ~0.05). At the end of each step, the sample was removed and visually inspected for any macro-cracks. Since barreling of the central barrel allowed us to introduce a higher hoop (tensile) strain in the collar for a given axial (compressive) strain of the barrel, no lubricant was used for these tests.

The diameter of the central barrel (d), the diameter of the collar (d_c) and the height of the collar (t) were measured at each step to calculate the macro-strain state in the collar. The radial (ε_r), axial (ε_zz) and hoop strains (ε_θθ) were calculated from the measurements.
3.4 Experimental Analysis

3.4.1 Microscopy on Sections Perpendicular to the Fracture Surface

Sections perpendicular to the fracture surface in the collar compression and the uniaxial tensile specimens were metallographically prepared (as described in section 3.2) in order to observe the failure characteristics in and around the fracture path. Of particular interest was the micro-void volume fraction away from the fracture surface.

Back-scattered electron images at 50X were used to measure the micro-void area distribution away from the fracture surface. The micrographs were divided into areas of about 0.2 mm². Compared to the average particle size of less than 20 μm, the area was considered large enough to nullify the effect of extremely localized events such as...
isolated voids. The void area was computed through the UTHSCSA image analysis software. The void volume fraction was assumed to be equal to the void area fraction.

Another point of interest was the identification of the microstructural features associated with damage in metal matrix composites, in the temperature and strain rate range of our study. Back-scattered electron images at 150X and higher magnifications of the extruded and as-cast MMCs were used for this purpose.

3.4.2 Fracture Surface Microscopy

The fracture surfaces were microscopically examined, using a Scanning Electron Microscope (SEM) to identify the dominant mechanism of microvoid (and failure) nucleation at different temperature/strain rate. In addition, the area fraction of the reinforcement particles at the fracture surface was also measured for the extruded MMC samples fractured under uniaxial tension at 300°C, 425°C and 550°C, at a nominal strain rate of 1/s. Secondary electron images at 1000X were used for the estimation of area fraction. Each image covered an area of 0.1 mm² on the fracture surface. A 20x20 grid was set on the image and the fraction of the grid points occupied by the particles was ascertained. While converting this fraction to the area fraction of the particles at the fracture surface, a slightly different approach was taken for the cracked and decohesed particles. Parts of the cracked particle could be found on both sides of the fracture surface. On the other hand, a decohesed particle could be present on only one side of the fracture surface. Thus, for the cracked particles, the fraction of grid points occupied by the particles was interpreted as the area fraction of the particles; for the decohesed
particles, the grid fraction was multiplied by 2. The estimates were averaged over at least five fields of study.

3.5 Simulation of the Collar Compression Test

Ductile failure is essentially a localized process, and an estimate of the local strain state at failure is important. A few experimental techniques are available for the estimation of the local-fields (refer to Geers' thesis [1997] for an excellent review). However, all of them involve very elaborate and complicated procedures and are beyond the scope of the present work. In the case of MMCs, Poole [1993] showed that the finite element simulation of the deformation processes may be used to estimate the local-fields to reasonable accuracy. We simulated the collar compression tests on the as-cast MMCs for this purpose. In addition, the simulation was also expected to aid in the assessment of the stress distribution in the collar.

3.5.1 Geometry and Model

The model was created using the ABAQUS three-dimensional non-linear finite element software package. The finite element model was axi-symmetric and included only one-fourth of the sample (using radial and axial symmetry conditions). The contact between the top of the sample and the rigid, compressing platen was modeled as non-penetrating, frictional contact. An irregular mesh, with higher density of nodes at regions of expected high stress and strain gradients, was used. Figure 3.4 shows the mesh on a quarter of the sample. Mesh convergence studies were conducted to determine the ideal
meshing strategy. In the final form, the model contained more than 3000 4-noded bilinear elements (with reduced integration and hourglass control).

**Table 3.1** Input parameters for collar compression simulation of as-cast MMC

<table>
<thead>
<tr>
<th>Condition</th>
<th>Young’s Modulus (GPa)</th>
<th>Yield Strength (MPa)</th>
<th>Flow Stress (MPa)</th>
<th>Saturation Strain (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>300°C / 1/s</td>
<td>90</td>
<td>120</td>
<td>148</td>
<td>5%</td>
</tr>
<tr>
<td>550°C / 1/s</td>
<td>80</td>
<td>36</td>
<td>41</td>
<td>3%</td>
</tr>
</tbody>
</table>

**3.5.2 Material**

The material (as-cast metal matrix composite in this case) was assumed to be elasto-plastic with isotropic, linear hardening. The important elastic input parameters were the Young’s modulus, Poisson’s ratio and the yield strength of the material, whereas its plastic response was represented in terms of the strain at which the material ceased to work harden and the steady-state flow stress. Previous analyses [Clyne and Withers 1993] suggests that for the Al-Al₂O₃ system with 20% reinforcement (aspect ratio ~1), the composite stiffness is ~32% higher than the monolith stiffness. Thus, the Young’s moduli for AA6061 were obtained from literature [Frost and Ashby 1982] at different temperatures, and the MMC Young’s moduli were estimated to be 1.32 times the AA6061 moduli. The MMC yield strength, the steady-state flow stress and the strain at which the material became perfectly plastic (saturation strain) were estimated from
uniaxial compression experiments. A value of 0.34 was prescribed for the material Poisson’s ratio (elastic). The different input parameters are shown in Table 3.1.

3.5.3 Boundary Conditions and Loading

The kinematic boundary conditions were: symmetry on the axis (nodes at r=0 had \( u_r = 0 \) prescribed) and symmetry about z=0 (nodes on z=0 had \( u_z = 0 \) prescribed). The friction condition between the sample and the compressing platen was determined as follows: during the experiments with the collar compression samples, a number of measurements were taken on the samples at different stages of deformation (as explained in section 3.3.3). Simulations were subsequently undertaken using different friction coefficients and the most probable friction condition was chosen on the basis of minimum error (the error being defined as the difference between the model and experimental measurements of sample dimensions, normalized by the experimental measurements). Imposition of a sticking friction condition at the platen-sample interface provided a good fit to the experimental data (less than 5% error). Figure 3.5 schematically shows all the boundary conditions used in the model.

The loading was quasi-static, and brought about (as in the experiment) by allowing a prescribed vertical negative displacement of the rigid compression platen. The non-penetrating, no-separation contact between the sample and the platen ensured that the sample was compressed by the same amount. The rigid platen was allowed no rotational or radial displacement.
Figure 3.4  The section of sample modeled with the mesh used.

Figure 3.5  Boundary conditions used in the model.
3.6 Summary

The experimental and numerical procedures undertaken in this study have been detailed in this chapter. The next chapter presents the results obtained from these experiments and numerical simulations.
4.0 Results and Analysis

This chapter details the experimental results of this study. The deformation and damage characteristics of AA6061 and the MMCs, at both macro and micro levels are described. The microstructural studies on damage initiation and propagation, and the results from the FEM analysis of the collar deformation of as-cast MMC are also presented.

4.1 Microstructural Characteristics

4.1.1 Reinforcement Size and Shape

The particle characteristics, i.e. axial lengths, aspect ratio and equivalent diameter were obtained for the as-cast and extruded MMC using the C* Imaging Systems image analyzer (version 5.1). In order to minimize the amount of image analysis, a sensitivity study was carried out to estimate the dependence of the observed variable (i.e. particle dimensions) on the sample size (i.e. the number of particles considered). An as-cast MMC sample was used for this purpose. Table 4.1 shows the sample mean and standard deviation for the major and minor axial lengths, area and aspect ratio for sample sizes of 475, 996, 1607 and 1969 respectively.

As evident in Table 4.1, the particle characteristics do not change appreciably with sample sizes larger than 500 particles. Thus, in case of the extruded MMC, the analysis was restricted to 666 particles. The axial lengths, aspect ratio and the equivalent
Table 4.1  Sample mean and standard deviation (in brackets) for the major and minor axial lengths, area and aspect ratio of the particles, for different sample sizes in as-cast MMC (all dimension in microns)

<table>
<thead>
<tr>
<th></th>
<th>475 particles - Mean (Std. Dev.)</th>
<th>996 particles - Mean (Std. Dev.)</th>
<th>1607 particles - Mean (Std. Dev.)</th>
<th>1969 particles - Mean (Std. Dev.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Major Axis</td>
<td>15.56 (8.41)</td>
<td>15.61 (8.25)</td>
<td>15.48 (8.22)</td>
<td>15.51 (8.21)</td>
</tr>
<tr>
<td>Minor Axis</td>
<td>8.71 (5.31)</td>
<td>8.51 (5.13)</td>
<td>8.50 (5.21)</td>
<td>8.55 (5.19)</td>
</tr>
<tr>
<td>Area</td>
<td>132.76 (131.6)</td>
<td>129.72 (123.94)</td>
<td>128.83 (126.38)</td>
<td>129.55 (125.89)</td>
</tr>
<tr>
<td>Aspect Ratio</td>
<td>1.96 (0.73)</td>
<td>2.04 (0.84)</td>
<td>2.02 (0.81)</td>
<td>2.00 (0.78)</td>
</tr>
</tbody>
</table>
Table 4.2  Sample mean and standard deviation (in brackets) for the major and minor axial lengths, aspect ratio and equivalent diameter of the particles for the extruded and as-cast MMC (all dimension in microns)

<table>
<thead>
<tr>
<th></th>
<th>Extruded MMC</th>
<th>As-cast MMC</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Mean (Std. Dev.)</td>
<td>Mean (Std. Dev.)</td>
</tr>
<tr>
<td>Major Axis</td>
<td>20.76 (11.26)</td>
<td>15.51 (8.21)</td>
</tr>
<tr>
<td>Minor Axis</td>
<td>11.6 (7.29)</td>
<td>8.55 (5.19)</td>
</tr>
<tr>
<td>Aspect Ratio</td>
<td>2.04 (0.78)</td>
<td>2.0 (0.78)</td>
</tr>
<tr>
<td>Equivalent Diameter</td>
<td>18.1</td>
<td>12.84</td>
</tr>
</tbody>
</table>
diameter \( (radius = \sqrt{\frac{Area}{\pi}}) \) of the particles for the extruded (over 666 particles) and as-cast MMC (1969 particles) samples are shown in Table 4.2.

The particle characteristics, in both the as-cast and extruded MMC, show a wide scatter, resulting in high standard deviations. Nevertheless, the particles in extruded MMC samples are, on an average, larger than those in the as-cast MMC samples. This difference, however, was not chosen by design, and is related to the different particle sizes used for different castings at Duralcan Inc. (the extruded samples coming from one billet and the as-cast samples from another one).

4.1.2 Reinforcement Volume Fraction

The volume fraction of the reinforcements in the extruded and as-cast MMC were estimated from the area-fraction of the particles on the polished sections (of the broken tensile specimens). For the extruded MMC, the area fraction was estimated over 40 fields (total area \( \sim 8.7 \text{mm}^2 \), total number of particles =6721, see Figure 4.1). In case of the as-cast MMC, 30 fields (total area \( \sim 6.5 \text{mm}^2 \), total number of particles =11362, see Figure 4.2) were used for the analysis. The reinforcement volume fraction (mean) estimated in the two cases were 19.03% and 21.78% respectively. To test if the volume fraction were statistically different, a Z-test on the population means (assuming that the sample sizes, 40 and 30 in this case, were large enough for the sample means to follow a normal distribution) was conducted. The results indicated that the volume fraction of the reinforcements in the as-cast MMC samples were higher than in the extruded MMC at the 0.01 confidence level.
Figure 4.1  Variation in particle area fraction in an extruded MMC sample.

Figure 4.2  Variation in particle area fraction in an as-cast MMC sample.
The results from image analysis can be summarized as:

(a) the ceramic particles in the as-cast MMC were smaller than in the extruded ones, and
(b) the reinforcement volume fraction in the as-cast MMC material was higher (though only slightly) than in the extruded one.

The two above observations become apparent on examining sample micrographs from the extruded (see Figure 4.3) and the as-cast MMC (see Figure 4.4).

Over a smaller scale, the particle distribution in the as-cast microstructure showed both particle-clustered and particle-depleted regions (see Figure 4.5). To get an idea of the reinforcement volume fraction variation in these regions, the UTHSCSA Image Analysis software was used to estimate particle area fractions (local) in the polished as-cast tensile samples. Back scattered electron images (at 150X) from these samples were divided into several ~250μm x 250μm areas, and the particle area fraction was analyzed for each of these areas. In some of the cases, the fraction was as high as 0.4.

4.2 Material Flow Behavior

4.2.1 Compression Tests

Figures 4.6, 4.7 and 4.8 plot the flow stress curves for the extruded MMC and the unreinforced AA6061 at strain rates of 0.1/s, 1/s and 10/s respectively. Extensive serrated flow, possibly due to the Portevin-LeChatelier (PLC) effect, can be observed for both the materials at all the test temperatures when deformed at a strain rate of 0.1/s. (Note: no serrations were observed for tests under similar conditions when high purity
Figure 4.3  Extruded MMC microstructure - unetched (BSE image).

Figure 4.4  As-cast MMC microstructure - unetched (BSE image).
Figure 4.5  Particle depleted and particle clustered zones in as-cast microstructure (BSE image).

Figure 4.6  Stress-strain diagrams under uniaxial compression (at $\dot{\varepsilon} \sim 0.1$/s) for AA6061 and extruded MMC at 300°C, 425°C and 550°C.
**Figure 4.7** Stress-strain diagrams under uniaxial compression (at $\varepsilon \sim 1/s$) for AA6061 and extruded MMC at 300°C, 425°C and 550°C.

**Figure 4.8** Stress-strain diagrams under uniaxial compression (at $\varepsilon \sim 10/s$) for AA6061 and extruded MMC at 300°C, 425°C and 550°C.
copper samples were used). At 300°C, the flow curves, at all strain rates, seem to exhibit a pronounced negative work hardening effect at a true strain of ~0.1.

At the temperature and strain rate regime of this study (300°C-550°C, 0.1/s-10/s) strain hardening was extremely limited - both for the MMC and the unreinforced matrix. The materials attained the steady state flow stress at very low levels of plastic strain. The variation in the steady state flow stress (estimated here as the flow stress at a strain of 0.25) with temperature and strain rate for AA6061 and the MMCs (extruded and as-cast) is shown in Figure 4.9 and Figure 4.10, respectively. The values are the average of at least 2 tests - variation between repeat experiments was never higher than 5%. In general, the material flow stress increased with decreasing temperature and increasing strain rate. The base alloy (AA6061) and the MMC showed similar behaviour.

The ratio of the flow stress of the extruded MMC and the unreinforced AA6061 is shown in Figure 4.11. The ratio was between 1.10 and 1.30 - i.e. the flow stress for the extruded MMC was 10-30% higher than the unreinforced matrix. The as-cast material, in turn, exhibited a higher flow stress than the extruded one; at higher temperatures (425°C and 550°C), the difference was as high as 15%. However, it is to be noted that for the as-cast MMC, the compressive tests were conducted at only one strain rate (1/s).

The materials attained the steady state flow stress typically within 4-7% true strain. In case of tests at 550°C, at low strain rates (0.1/s and 1/s) the strain hardening regime was particularly short and the steady state flow stress was attained at 2-3% strain. Figures 4.12 and 4.13 show the variation in saturation strains (strains beyond which the material showed no work hardening) with temperature and strain rate for AA6061 and the
Figure 4.9  Effect of temperature and strain rate on AA6061 steady state flow stress. (Note: The points indicate the mean, and the error bars the range of the observations).

Figure 4.10  Effect of temperature and strain rate on MMC steady state flow stress. (Note: The points indicate the mean, and the error bars the range of the observations).
Figure 4.11  Effect of temperature and strain rate on the flow stress elevation in extruded MMC (over AA6061). (Note: The points indicate the mean, and the error bars the range of the observations. The tests were conducted at 300°C, 425°C and 550°C; some of the points have been slightly shifted horizontally to enhance clarity).
Figure 4.12  Effect of temperature and strain rate on AA6061 saturation strain. (Note: The points indicate the mean, and the error bars the range of the observations. The tests were conducted at 300°C, 425°C and 550°C; some of the points have been slightly shifted horizontally to enhance clarity).

Figure 4.13  Effect of temperature and strain rate on extruded MMC saturation strain. (Note: The points indicate the mean, and the error bars the range of the observations. The tests were conducted at 300°C, 425°C and 550°C; some of the points have been slightly shifted horizontally to enhance clarity).
extruded MMC respectively.

At test temperatures of 300°C and the 425°C, the compression cylinders showed minimal amounts of barreling during deformation. However, for tests at 550°C, the cylinder diameters varied considerably along the length: the difference between the maximum and minimum diameters could be as high as 10%. This was attributed to the relatively poor performance of the glass lubricant, although it was superior to no lubricant. In these cases, the final load was divided by the maximum area (corresponding to the maximum diameter) to obtain the steady state flow stress.

Estimating the material flow stress as a function of the operating conditions (temperature, strain rate) is important for industrial large scale deformation processes. For this purpose, the flow stress experimental data were fitted to a hyperbolic-sine relationship [Sellers and Tegart 1966, Frost and Ashby 1982]

\[ \dot{\varepsilon} = A \left( \sinh \frac{\alpha \sigma}{\mu} \right)^n \exp \left( -\frac{Q}{RT} \right) \]  \hspace{1cm} (4.1)

and the arbitrary constants A, \( \alpha \), n as well as the deformation activation energy Q determined from the best-fit line. The experimentally determined \( \dot{\varepsilon} \) (strain rate), T (temperature in K) and \( \sigma \) (steady-state flow stress) were used. Relevant Al shear moduli (\( \mu \)) at different temperatures were obtained from Frost and Ashby [1982]. A very fast simulated annealing (VFSA) algorithm [Routh 1997] was used for obtaining the best-fit line, primarily due to its ability to estimate the global minima in case of highly non-linear
functions. The best fit line predicted the flow stresses within ±10%. The values for A, α, n and Q for the AA6061 and the extruded MMC are listed in Table 4.3.

Table 4.3 Material constants for constitutive equations

<table>
<thead>
<tr>
<th>Material</th>
<th>A (s)</th>
<th>α</th>
<th>n</th>
<th>Q (KJ/mole-K)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Extruded MMC</td>
<td>2.16 E+10</td>
<td>0.343</td>
<td>5.033</td>
<td>149.2</td>
</tr>
<tr>
<td>AA6061</td>
<td>6.22 E+10</td>
<td>0.388</td>
<td>5.468</td>
<td>153.5</td>
</tr>
</tbody>
</table>

4.2.2 Tensile Tests

The ultimate tensile strengths were also estimated from the uniaxial tensile curves for the different materials. In case of AA6061 and the extruded MMC, the UTS was close to (within 10%) but lower than the compressive steady-state flow stress. The compressive and tensile flow curves for the extruded MMC deformed at 425°C are shown in Figure 4.14 for comparison. In this diagram, the tensile curves have been reproduced to the UTS values; the curves were then joined to the points representing the sample fracture stress (determined from load at failure and the area of the fracture surface). These UTS values were also highly reproducible (values from repeat experiments within ±5%). On the other hand, the UTS estimates from the as-cast MMC tensile specimens varied widely (the variation between repeat experiments was as high as 40%). In general, for the as-cast MMC, each tensile experiment was repeated more than twice, e.g. for the 425°C tests, four tests were done under each condition and the most common UTS
value taken as representative. The UTS values for the as-cast MMC were higher than the steady-state flow stress (~15% higher). The variation in UTS for as-cast MMC is shown in Table A1.

4.3 Material Failure - Macroscopic Observations

4.3.1 Tensile Tests

Incorporation of the ceramic particles in AA6061 significantly reduced the ductility - the failure strain (reduction in area at failure) for the AA6061 was 5-10 times higher than the extruded MMC and 25-50 times than the as-cast MMC. Figure 4.15 shows the failure strains in the three materials under tensile testing at a nominal strain rate of 1/s (each point is the average of at least 2 tests).

Failure strain measurements for the extruded MMCs were highly repeatable (variation in repeat experiments within ±5%). For AA6061, it varied by up to 15%, especially at 550°C when the reduction in area at fracture was very high, but this variation could be, in part, due to the error in measuring an extremely small area at the fracture surface. Failure strains for the as-cast MMC varied by as much as ±20% from the mean value. As noted earlier in section 4.2, at least three uniaxial tensile tests were performed on the as-cast MMC under each condition. The variations in the failure strains for the as-cast MMC, under uniaxial tension and collar compression are shown in Table A2.

The necking strain (equal to the effective strain at which there was no further increment in the macro-load required for deformation) for the AA6061 and the MMCs under uniaxial tensile tests were also measured. Figure 4.16 shows the variation in the
Figure 4.14 Compressive and tensile stress-strain diagrams for extruded MMC at 425°C and different strain rates. (Note: the ellipses on the tensile curves represent the point till which the original stress-strain curves have been reproduced. The rectangles indicate the fracture stress of the samples).

Figure 4.15 Effect of temperature (at \( \varepsilon \sim 1/s \)) on the uniaxial tensile failure strain of AA6061 and MMCs. (Note: The points indicate the mean, and the error bars the range of the observations).
necking strains for the unreinforced matrix material (AA6061) and the MMCs, tested under a nominal strain rate of 1/s. The strains were found to be almost identical for the unreinforced AA6061 and the extruded MMC - whereas their strain to failure differed by almost an order of magnitude. As with the failure strain, the necking strain in the as-cast MMC was extremely low. For all the materials, the necking strain was particularly low at 550°C.

The extremely low necking strains obtained in the 550°C tests (at lower strain rates) are clearly visible in Figure 4.17, where the strains for extruded MMC for the different temperature and strain rate tests have been plotted. Similar trends were also observed for the unreinforced and the as-cast MMC.

4.3.2 Collar Compression Tests

4.3.2.1 Failure Strains

In the collar compression tests, cracks initiated at the edge of the collar and then moved inwards (see Figure 4.18). The hoop strain ($e_{0\theta}$) in the collar (calculated as shown in section 3.3.3) was used as the failure strain in this case. Generally, for the extruded MMC, the collar compression tests yielded a lower failure strain than uniaxial tensile tests (see Figure 4.19). However, for the as-cast MMC, they were quite similar (see Figure 4.20).

In each case, the average failure strain from 2 tests was used. The variation in the failure strains between repeat experiments was similar to that in the uniaxial tensile tests.
Figure 4.16 Effect of temperature (at $\varepsilon \sim 1/s$) on the necking strain of AA6061 and MMCs. (Note: The points indicate the mean, and the error bars the range of the observations).

Figure 4.17 Effect of temperature and strain rate on extruded MMC necking strain. (Note: The points indicate the mean, and the error bars the range of the observations).
Figure 4.18  BSE image of crack propagation (in the collar) in collar compression tests.
Figure 4.19  Effect of temperature and strain rate on extruded MMC failure strain. (Note: The points indicate the mean, and the error bars the range of the observations. The tests were conducted at 300°C, 425°C and 550°C; some of the points have been slightly shifted horizontally to enhance clarity).

Figure 4.20  Effect of temperature and strain rate on as-cast MMC failure strain. (Note: The points indicate the mean, and the error bars the range of the observations. The tests were conducted at 300°C, 425°C and 550°C; some of the points have been slightly shifted horizontally to enhance clarity).
It is to be noted that during the collar compression test, the sample was inspected for failure only after each displacement step, and thus the failure strain effectively represents only an upper bound. The lower bound was dictated by the collar hoop strain in the last no-failure step, and was generally 0.05 (as-cast MMC) to 0.1 (extruded MMC) lower than the upper bound.

4.3.2.2 Strain Path

Syu and Ghosh [1994b] have investigated the strain evolution in the collar in a collar compression test. They have shown that at low imposed deformation, the strain path is quite similar to that in an uniaxial tensile case. At increased strains, however, the path shifts towards a plane strain condition. In our tests, for all temperatures and strain rates, the (collar) deformation in the as-cast MMC was in the uniaxial tensile regime (see Figure 4.21). On the other hand, for the extruded MMC, the strain path tended towards plane strain conditions at fracture in the high temperature tests (especially at 550°C, see Figure 4.22).

4.3.2.3 Test Simulation

Microscopic examination of the collar fracture indicated that the cracks were initiated at the collar edge (see Figure 4.18). Therefore, the purpose of the simulation of the collar deformation in the as-cast MMC was to estimate the prevalent micro-strain and stress in the collar. The Finite Element (ABAQUS) simulations showed:
(a) the micro-strain (hoop) at the collar edge was uniform, and was very close to (less than 1% difference at imposed hoop strain of 15%) the macro-strain (hoop) for the collar.

(b) the stress-state at the edge of the collar represented an uniaxial tensile state - the hoop stress (tensile) was more than 10 times the other principal stresses.

Figure 4.24 and Figure 4.25 show the variation in stress and strain in the collar for the as-cast MMC deformed at 300°C at a strain rate of 1/s (Figure 4.23 marks the collar in the model). As evident, the estimation of the principal stress and strain responsible for failure ($\sigma_{66}$ and $\varepsilon_{99}$ at the collar edge) is relatively simple in this case - $\sigma_{66}$ is equal to the steady state flow stress, whereas $\varepsilon_{66}$ at the edge could be estimated by the macro $\varepsilon_{66}$ at the collar.

The effect of the sample-die friction coefficient on the stress and strain states in the collar was also investigated. The micro-strains were found to be quite sensitive to the contact friction condition (changed by 50-100% when the coulomb friction coefficient was changed from 0.1 to 0.4). The micro-stresses on the other hand, were relatively insensitive. The stress state in the collar at low strains (the failure strains for as-cast MMCs were low enough) represented that of uniaxial tension. With the hoop stress saturating to the steady state flow stress value very early during the deformation, and the radial and axial stress components negligibly small near collar periphery, the stress-state in this region almost became insensitive to strain, and in turn, to the contact friction condition. However, the situation may be expected to change at higher strains (for the extruded MMC) when the collar periphery is under a plane-strain condition. This could be encountered in the simulation of the extruded MMC collar compression tests.
Figure 4.21  Collar compression strain path for as-cast MMC.

Figure 4.22  Collar compression strain path for extruded MMC.
Figure 4.23  Schematic diagram of the part of the collar compression sample modeled - the encircled region is the collar. (Note: The unit vectors 1, 2, and 3 correspond to the radial, axial and hoop unit vectors respectively).
Figure 4.24 Stress distribution in the collar (encircled region in Figure 4.23) from FEM calculations, (a) $\sigma_{11}$, (b) $\sigma_{22}$, (c) $\sigma_{33}$, and (d) hydrostatic pressure (values in Pa).
Figure 4.25 Strain distribution in the collar (encircled region in Figure 4.23) from FEM calculations, (a) $\varepsilon_{11}$, (b) $\varepsilon_{22}$, (c) $\varepsilon_{33}$, and (d) von mises strain.
However, these simulations were not done in this study, after encountering mesh problems at the later stages of deformation.

4.4 Microscopic Failure Mechanisms

4.4.1 Observations on the Fracture Surface

Figures 4.26, 4.27 and 4.28 show a typical portion of the fracture surface for the extruded MMC under uniaxial tension at 300°C, 425°C and 550°C, at a strain rate of 1/s. The fracture surface for AA6061 (at 300°C and 1/s) is shown in Figure 4.29 for comparison with the MMC fracture surfaces. The voids seem to be noticeably smaller in case of the unreinforced material. For MMCs, the primary void nucleation mechanism at 300°C was particle cracking, whereas at 550°C, decohesion at the matrix particle interface was dominant. At 425°C, both the mechanisms seem to be operative, with interfacial decohesion dominating.

As explained in section 3.4.2, area fraction of the particles at the fracture surface was measured from SEM micrographs at 1000X. The area fractions were found to vary with temperature of testing: 17% at 300°C, 51% at 425°C and 58% at 550°C.

4.4.2 Observations on Sections Perpendicular to the Fracture Surface

4.4.2.1 Extruded MMC

In extruded MMCs, high damage levels (void volume fraction ~ 0.1-0.15) were observed in low strain rate tests (0.1/s and 1/s) at 550°C. The damage was quite uniformly distributed in this case, and material even half a mm away from the fracture
Figure 4.26  Fracture surface of extruded MMC, under uniaxial tension at 300°C and 1/s.

Figure 4.27  Fracture surface of extruded MMC, under uniaxial tension at 425°C and 1/s.
Figure 4.28 Fracture surface of extruded MMC, under uniaxial tension at 550°C and 1/s.

Figure 4.29 Fracture surface of AA6061, under uniaxial tension at 300°C and 1/s.
surface was also highly damaged (see Figure 4.30). In all other conditions, the damage was extremely localized around the fracture surface. At 550°C/ 10/s, the void volume fraction was ~0.1 near the fracture surface, but it reduced to less than 0.01 within 0.3 mm from it. For the 425°C tests, near the fracture surface, the void volume fraction was ~0.05; however, it reduced to less than 0.01 within 0.2 mm from the fracture surface. In the 300°C tests, the void volume fraction never exceeded 0.01 (see Figure 4.31). This seems to suggest that for the 550°C/ 0.1/s and 1/s tests the damage events were well spread out over a large strain range, whereas for the rest of the tests (and especially in the 300°C tests) the events were highly localized. In general, the voids nucleated by interfacial decohesion seem to have grown considerably more than those by particle cracking (see Figure 4.32), i.e. particle cracking seems to be more catastrophic than interfacial decohesion.

4.4.2.2 As-cast MMC

Damage away from the fracture surface was insignificant (void volume less than 1%) under all conditions (see Figure 4.33). The as-cast microstructure revealed acute particle clustering (see Figure 4.33). Long fine cracks (~ 0.5 mm long) were observed in the particle clusters (see Figure 4.34).

4.5 Summary

Incorporation of ceramic particles in the metallic matrix significantly changed both strength and ductility - while there was 10-30% gain in the steady state flow stress in
Figure 4.30  Distribution of damage in extruded MMC sample, under uniaxial tension at 550°C and 1/s (BSE image).

Figure 4.31  Distribution of damage in extruded MMC sample, under uniaxial tension at 300°C and 1/s (BSE image).
Figure 4.32  Comparison of growth characteristics of voids nucleated by particle cracking and interfacial decohesion (BSE image).

Figure 4.33  Distribution of damage in as-cast MMC sample, under uniaxial tension at 550°C and 1/s (BSE image).
Figure 4.34  Crack propagation in a particle cluster (BSE image).
the extruded MMCs, the reduction in area at failure was 5-10 times lower. Microstructural features such as particle arrangement characteristics were found to have a large effect on the material ductility - failure strain for the as-cast MMC (with a clustered microstructure) was more than 5 times lower than the extruded MMC failure strain. Particle cracking and interfacial decohesion seem to be the dominant void nucleating mechanisms: particle cracking dominating at lower and interfacial decohesion at higher temperatures. Damage was widely and uniformly distributed at 550°C for the extruded MMC. On the other hand, damage in as-cast MMC was highly concentrated near the fracture surface. These results will be discussed in detail in the next chapter.
5.0 Discussion

This chapter presents a discussion on plastic flow and fracture of the AA6061-20%Al₂O₃ MMCs. It is divided into two parts. In the first part, the various issues regarding the elevation of flow stress in metal matrix composites are discussed. The second part deals with the ductility issues in these materials. Here, the principal mechanisms of void evolution and their effects on damage distribution and final failure are considered. The possible effects of the microstructural characteristics and experimental variables (temperature, strain path etc.) on the failure strain are also examined.

5.1 Flow Stress Elevation in MMC

5.1.1 Role of the Ceramic Particles

The presence of hard, brittle ceramics particles strengthened the matrix. The flow stress for the extruded and as-cast MMC were 10-30% and 25-35% higher than the unreinforced AA6061 respectively. As explained in section 2.2.1, the strengthening in MMCs is primarily due to the particles (stiffer than matrix) carrying a higher proportion of load - and may be expected to be affected by the particle shape and volume fraction. Under our testing conditions, however, the higher strain-rate hardening of the matrix may be expected to contribute significantly to MMC strengthening.

Under macroscopically similar deformation conditions, the MMC-matrix experiences a much higher strain (due to the non-deforming particles) than the
unreinforced AA6061. The higher strain in the matrix essentially means that the MMC-matrix is deforming at a higher strain rate than unreinforced AA6061 (even though the nominal strain rate is same for both AA6061 and MMC). Under low-temperature, quasi-static deformation, this higher strain rate in the MMC-matrix does not significantly enhance its flow stress. However, at elevated temperatures (when the matrix strain rate sensitivity is high enough) and moderately high strain rates (as in our case), the contribution of enhanced strain-rate hardening of the matrix to the overall MMC strengthening may be considerable.

Bao and Lin [1996] have numerically modeled the strengthening response of MMCs with a non-hardening, perfectly visco-plastic matrix. The matrix response has been modeled by

\[
\sigma = \sigma_0 \left\{ 1 + \left( \frac{\dot{\varepsilon}_p}{\dot{\varepsilon}_0} \right)^m \right\}
\]

(5.1)

where \(\dot{\varepsilon}_p\) is the plastic strain rate in the matrix, \(\sigma_0\) is the quasi-static uniaxial flow stress, \(\dot{\varepsilon}_0\) is the reference strain rate and \(m\) is the strain rate sensitivity. \(\sigma_0, \dot{\varepsilon}_0\) and \(m\) are unknown parameters: they could be determined on the basis of flow stress estimates under different strain rates. In our case, they were determined at each temperature (300\(^\circ\)C, 425\(^\circ\)C and 550\(^\circ\)C) from the matrix flow stress data at three different strain rates (0.1/s, 1/s and 10/s). The strain rate sensitivity obtained matched well with literature data [Fields and Backofen 1959] for aluminum at these temperatures (i.e. 0.07 at 300\(^\circ\)C, 0.15 at 425\(^\circ\)C and 0.2 at 550\(^\circ\)C). The values were then used to calculate the flow stress
elevation expected in the MMC (i.e. $\frac{\sigma_{MMC}}{\sigma_{matrix}}$), the MMC modeled as being reinforced by equi-spaced, spherical, elastic particles). The numerical estimates have been compared with the experimental observations for the extruded MMC in Figure 5.1. The numerical and experimental values match relatively well in case of 300°C and 425°C. However, at 550°C, the numerical predictions greatly overestimated the material flow stress. The reasons for this are not clear at present, however, they may be related to a change of matrix deformation behavior in the presence of the reinforcement. For example, grain boundary sliding may be much more difficult in the MMC than in the base alloy.

5.1.2 Role of Particle Distribution

The strengthening in a clustered MMC may be higher than when the particles are uniformly dispersed [Corbin 1992], primarily due to the increased flow constraints imposed on the matrix in a clustered structure. However, this is generally deemed to be only a second-order effect. In our experiments (refer to section 4.2.1) this seems to be the case, i.e. the peak flow stress of the ‘relatively uniformly dispersed’ extruded MMC differed little from that of the ‘clustered’ as-cast MMC.

5.1.3 Role of Damage

Material damage in the form of particle cracking, interfacial decohesion or matrix failure is expected to lower the elastic modulus and work hardening rate or strength of the
Figure 5.1 Temperature and strain rate dependence of flow stress elevation in extruded MMC - experimental observations (open symbols) vs. numerical estimates (filled symbols). (Note: The points indicate the mean, and the error bars the range of the observations. The tests were conducted at 300°C, 425°C and 550°C; some of the points have been slightly shifted horizontally to enhance clarity).
MMC. A number of macro-damage measurements, based on these properties of the virgin and damaged materials, have also been proposed and used by several authors (predominantly on room temperature test data) [Kachanov 1986, Lloyd 1991, Corbin 1992]. In our experiments with the extruded MMCs, strength reduction of the composites due to damage seems apparent, with the tensile flow curves lying below the compressive curves (see Figure 4.14). However, without more sophisticated strain measurements any quantitative inference in this regard may be unreliable.

5.2 Ductility Considerations

5.2.1 Void Evolution Mechanisms

Interfacial decohesion and particle cracking were the dominant void nucleation mechanisms. The particle’s tendency to crack decreased with increase in temperature. At 300°C, particle cracking dominated for all the strain rates; at 550°C, interfacial decohesion was dominant. For all the tests at 425°C, both cracked and decohesed particles could be observed. At this temperature (425°C), the number of cracked particles increased with increase in strain rate (see Figure 5.2), but even at a strain rate of 10/s, interfacial decohesion was the dominant mechanism. The effect of strain rate on the failure mechanism seems to be only secondary, possibly because, under our test conditions, strain rate change affected the material flow stress much less than a change in temperature.
Figure 5.2  Effect of strain rate on particle cracking tendency at 425°C.

(a)  $\dot{\varepsilon} \sim 0.1/s.$

(b)  $\dot{\varepsilon} \sim 10/s.$
5.2.1.1 Particle Cracking

Examination of cracked particles on the fracture surface showed evidence of flaws (e.g. see Figure 5.3). Using the dimensions of these flaws, the stress required to fracture these particles could be estimated using fracture mechanics principals. This stress could then be compared to the model predictions of the stress in the particle during MMC deformation. We will now go through this exercise for the Al$_2$O$_3$ particle shown in Figure 5.3.

The fractured particle shown in this figure contains an ellipsoidal embedded flaw. The lengths along the major and minor axes of the ellipse were measured to be 7.2 \( \mu \text{m} \) and 1.8 \( \mu \text{m} \) respectively.

For an ellipsoidal embedded flaw [Irwin 1962]

\[
K_{IC} = \frac{\sigma \sqrt{\pi a}}{\Phi \left( \sin^2 \phi + \frac{a^2}{c^2} \cos^2 \phi \right)^{1/4}}
\]  

(5.2)

where, \( \sigma \) is the stress, \( a \) and \( c \) are the semi-minor and semi-major axial lengths of the ellipse, \( \phi \) is the angular orientation of the point where the \( K_{IC} \) is being estimated. \( \Phi \) is elliptic integral, and for \( a \ll c \) (which is the case here)

\[ \Phi = 1 \]

Putting \( \phi = \pi/2 \), i.e. computing the \( K_{IC_{\text{max}}} \), we obtain

\[
K_{IC} = \sigma \sqrt{\pi a}
\]  

(5.3)

In Figure 5.3, \( a = 0.9 \mu \text{m} \), while \( c = 3.6 \mu \text{m} \). Thus,

\[
K_{IC} = 1.7 \times 10^{-3} \sigma
\]  

(5.4)
Since the crack lies at the edge of the particle (distance from the center of the crack to the particle edge ~3.5 μm), we also have to introduce edge-correction to these calculations. Using Kobayashi correction [Kobayashi 1965],

\[ K_{IC} \text{(corrected)} = 2.1 \times 10^{-3} \sigma \]  

(5.5)

It is obvious from Figure 5.3, that grain-boundary toughening of the material is minimal (i.e. each particle is essentially a single crystal). The \( K_{IC} \) in that case can be estimated from the surface energy for alumina [Broek 1991]

\[ K_{IC} = \sqrt{\frac{2E\gamma}{(1-\nu^2)}} \]  

(5.6)

where, \( E \) is the Elastic modulus, \( \gamma \) is the surface energy per unit area, and \( \nu \) is the Poisson’s ratio. In our case, \( E = 390 \) GPa [ASM Handbook 1991], \( \gamma \approx 0.5 \) J/m\(^2\) [Troczynski 1998], and \( \nu \approx 0.26 \) (elastic deformation). And, from (5.6)

\[ K_{IC} = 0.65 \text{ MPa} \sqrt{m} \]

Thus, the particle stress at which the (particle) fracture took place may be predicted to be ~300 MPa. Comparing this with the matrix flow stress of 130 MPa, the ratio of particle to matrix stress is ~2.3.

Using unit-cell FEM models, Brokenbrough and Zok [1995] have estimated the particle to matrix stress ratio for a uniform distribution of spherical particles embedded in a plastic matrix. In the fully plastic regime (\( \varepsilon \gg \varepsilon_y \), the yield strain), this ratio is shown to be independent of \( E_p / E_m \), i.e. the relative stiffness of the particle w.r.t. the matrix, and only dependent on the volume fraction of the reinforcing phase and the strain hardening exponent of the matrix (the effect of the particle shape was not considered). For a
perfectly plastic matrix, and reinforcement volume fraction of 20% (as in our case), the particle to matrix stress ratio is calculated to be ~1.8. In case of more angular particles, the ratio may be expected to be higher. Thus, the particle stress estimate (from the fractured particles) seems to be reasonably close to model predictions.

As the temperature is raised and/ or the strain rate is lowered, the matrix flow stress decreases. This in turn, decreases the particle stress. In case of the particle reinforced MMCs, the initial flaw sizes of the particles vary - and thus their strengths (i.e. the stress at which the particle may be expected to fail) vary. This variation is often represented with the help of Weibull statistics. But the strength is fairly constant with respect to temperature (or strain rate). Thus, as the temperature is raised (and the particle stress is lowered), the fraction of the particles susceptible to fracture decreases, and the importance of particle fracture as a void nucleation mechanism decreases (refer to section 5.2.3.1). Under these conditions, as mentioned previously, interfacial decohesion is the primary void nucleation mechanism.

5.2.1.2 Interfacial Decohesion

The alumina particles bore a layer of spinels at their surface - octahedral MgAl$_2$O$_4$ crystals about 1μm in size (see Figure 5.4, similar observation has been recorded by Lee et. al. [1997]). The matrix-particle decohesion was observed to be actually spinel-matrix decohesion (see Figure 5.5). Voids nucleated at the matrix interface with the spinel crystals. They grew and coalesced into a void sheet (see Figure 5.6), which ultimately
Figure 5.3  A cracked $\text{Al}_2\text{O}_3$ particle - with a flaw.

Figure 5.4  A decohesed $\text{Al}_2\text{O}_3$ particle - with the spinel crystals on surface.
Figure 5.5  A decohesed $\text{Al}_2\text{O}_3$ particle, exhibiting the voids created by spinel-matrix decohesion.

Figure 5.6  A void sheet created on the matrix due to particle decohesion.
lead to interfacial decohesion. Figure 5.4 shows a decohesed particle, with almost all the spinel crystals intact on its surface.

Decohesion primarily occurred at the top and bottom of the particles (i.e. on the faces normal to the loading direction, see Figure 5.7) and at particle corners (see Figure 5.8). The void nucleated at a point in the interface - subsequently, it experienced both longitudinal and transverse growth. The growth in the longitudinal direction was due to the dilation of the void in the loading direction, and has been partially modeled by, among others, Rice and Tracey [1969]. The transverse growth occurred as more and more of the interface (at the top or bottom of the particle) opened up - however, there was little transverse growth after the entire interface had decohesed. Figure 5.9 shows the predominant longitudinal and the limited transverse growth (after decohesion of the entire interface) of the voids.

The void coalescence occurred in two ways: (a) along the loading direction (see Figure 5.10), and (b) with the necking of the inter-void matrix when the void length was roughly equal to the inter-void spacing (see Figure 5.11), as predicted by Brown and Embury [1973]. The latter was the more important coalescence mechanism: the fact that all the fracture surfaces made large angles with the loading direction rules out any possibility of predominance of loading direction void linkage. Still, joining of voids in the loading direction increased the void length and assisted in the final coalescence.

The reduction in particle cracking with increase in temperature (or decrease in strain rate) has been explained in the previous section on the basis of the reduction in the
Figure 5.7  Void nucleation by decohesion at particle ends.

Figure 5.8  Void nucleation by decohesion at particle corners.
Figure 5.9  Void growth in extruded MMC (at 550°C, $\dot{\varepsilon} \sim 1/s$) - predominantly in the loading direction.
Figure 5.10  Void coalescence in the loading direction.

Figure 5.11  Void coalescence with necking of the inter-void matrix.
particles stress (with the lowering of the matrix flow stress under these conditions). However, the question as to why interfacial decohesion apparently becomes easier under such conditions remain unanswered. The normal stress across the particle-matrix interface (which is believed to cause interfacial decohesion) is related to the particle stress. So, as the temperature is raised, the interfacial normal stress also decreases. The increased incidence of interfacial decohesion, even though the interfacial stress was lowered, seems to indicate that the interfacial strength in this system decreases at elevated temperatures. Another possibility is the local stress elevation along shear bands at the interface (due to intense strain localization). However, to date, almost no data exists for the temperature dependence of interfacial strengths in these systems, and further research is required before anything conclusive can be said in this regard.

5.2.2 Damage Evolution Characteristics

To estimate whether the damage was highly localized or spanned a large strain range, the void volume fraction was estimated as a function of distance from the fracture surface, from micrographs of sections perpendicular to the fracture surface. In these sections, material away from the fracture surface was under lower strain (the higher the distance, the lower the strain - strain at a point in the sample being assessed from the sample diameter at that point, and the original sample diameter). Therefore, they represented the history of damage evolution: if void volume fraction was high even far away from the fracture surface, void nucleation must have occurred at low strain, and the
evolution could be considered gradual over a large strain range. On the other hand, damage concentrated around the fracture surface indicated a catastrophic failure process.

Before delving further into this discussion, it might be worthwhile to clarify what we actually mean by 'void volume' in our analysis.

5.2.2.1 Definition of 'void volume'

The evolution of damage with imposed strain (and the microstructural causes for such evolutionary traits) was of interest in this study. In other words, we were interested in damage initiation and propagation up to incipient macro-failure. It is important to remember, however, that everything was being observed post-failure. Thus, the 'void' needed to be defined in such a manner so as to minimize the damage events which took place simultaneously with final macro-failure - since they were not the cause, but really a side effect of macro-cracking.

Two definition criteria were considered: (a) any 'phase' which bears little of the imposed load, and (b) a material 'gap' or 'emptiness'. Under the first criterion, a cracked particle can be wholly considered to be a void. However, this obviously does not serve our purpose, since every particle cracked during macro-cracking would increase the local void volume considerably - and even a highly localized pre-cracking damage phenomena would seem uniform, at least in the near vicinity of the crack surface.

On the other hand, any material 'gap', once nucleated, grows as a function of far-field imposed strain [Rice and Tracey 1969]. Thus, the volume of the material 'gap' is directly related to the far-field strain imposed after its nucleation. The void volume was
thus defined to be the ‘volume of material emptiness’. This served our need exactly: a gradual change in void volume fraction over distance meant a gradual evolution of damage with strain.

5.2.2.2 Damage Characteristics vis-à-vis Nucleation Mechanism

Damage due to interfacial decohesion spanned a much larger strain range than that due to particle cracking (refer to section 4.4.2). In other words, particle cracking seemed to be much more catastrophic than interfacial decohesion. This could be due to the following:

The ceramic particles were hard and brittle - and thus cracking of a single particle occurred over a very small strain range. The local stress field was, at the same time, considerably perturbed - a full particle carried a large load, whereas a cracked particle was load-free at the fracture plane. This excess load had to be shared by the other particles and the matrix. The ‘extra load’ partitioning may be anticipated to be local in nature - the closer a particle is to the cracked particle, the higher its share of the ‘extra load’. This incremental loading now induced the neighboring particles to crack and very soon a macro-crack formed.

On the other hand, when interfacial decohesion was the dominating mechanism, the damage evolution was much slower. As evident from Figure 5.4, decohesion at the spinel-matrix interface was ductile in nature (void formation and coalescence), and thus the decohesion of even a small interface could take place over a considerable strain range. Load transfer across the interface slowly decreased as more and more of the interface
opened up. Thus, local load conditions changed slowly and were almost continuous with strain. This led to formation of small voids over a large area (diffuse rather than concentrated damage). These voids grew and coalesced (with micro-necking of the inter-void matrix) and the macro-crack was formed.

5.2.2.3 Macro-crack Path

It has been experimentally noted that a macro-crack will prefer the ‘least resistance path’ and will move through prior damage, if it exists [Arsenault 1993]. In our study, to assess whether the macro-crack propagates in a random direction through the microstructure or it follows the ‘weak link’ (i.e. moves through the particles), the area fraction of particles on the fracture surface was estimated. In case of random crack propagation, the particles could be expected to cover around 20% of the fracture surface (equal to the nominal volume fraction of the particles in the MMC). On the other hand, if the crack propagated primarily along particle-centered prior damage, the coverage could be expected to be much more.

As noted in section 3.4.2, fracture surface micrographs from 1/s tests conducted on extruded MMC were used for this purpose. In the case of 300°C tests, the area fraction of particles was around 17%, whereas, in the 425°C and 550°C tests, the fraction was higher than 50%. This drastic change in the particle area fraction at the fracture surface seems to have occurred with a change in the primary void nucleation mechanism, from particle cracking at 300°C to interfacial decohesion at 425°C and 550°C. This can be explained as follows:
In case of particle cracking, since the damage was extremely local, the macro-crack saw a virtually undamaged matrix around it (see Figure 5.12). Thus the crack path now followed the path of least distance across the sample, with no particular preference for the particle or the matrix. As a result, the presence of the ceramic particles on the fracture path was determined, in an average sense, by their volume fraction.

On the other hand, a macro-crack due to interfacial decohesion was surrounded by damaged material - damage centered around the particles (see Figure 5.13). The crack path now followed the ‘most damaged trail’, and thus, the particle distribution. Hence, the fraction of the fracture surface occupied by the particles was much more than their nominal volume fraction.

The predominant role of particle centered damage on the formation of the macro-crack can be observed in crack propagation in collar compression samples. In these cases, as shown in Figure 5.14, small cracks formed in particle-rich regions - they connected up with the failure of the intervening matrix.

5.2.3 Observations on Failure Strain

5.2.3.1 Temperature Dependence of Failure Strain

In AA6061, the failure strain increased monotonically with temperature (see Figure 4.15) in accordance with the expected behavior in monolithic metals and alloys. In this material, void nucleation, is expected to be primarily at the micron-scale dispersed particles (intermetallics). As the temperature was raised, the matrix stress was lowered and the void nucleation occurred at fewer and fewer sites, resulting in an overall
Figure 5.12  Distribution of damage around a propagating crack - 300°C, $\dot{\varepsilon} \sim 1/s$.

Figure 5.13  Distribution of damage around a propagating crack - 550°C, $\dot{\varepsilon} \sim 1/s$. 
Figure 5.14  Damage characteristics ahead of the crack tip - 300°C, $\dot{\varepsilon} \sim 10$/s.
enhancement of ductility. In our case, the failure strain for the 6061 increased by almost 5 times in the 300°C-550°C temperature range.

In extruded MMC, failure strain seemed to peak at a temperature close to 425°C (see Figure 4.19). This might be explained as follows:

As noted in section 5.2.1, the probability of particle fracture decreased with increase in temperature, while that of interfacial decohesion seemed to increase. Thus, in the domain where particle fracture was the primary void nucleation mechanism, it is understandable that the ductility increased with increase in temperature - simply, there was not enough void formation at low strains. On the other hand, at suitably elevated temperatures when interfacial decohesion was the controlling void nucleation mechanism, a fall in ductility with temperature might be expected. This is shown schematically in Figure 5.15. The highest point in the ductility curve may thus be anticipated to be at a point where the two influence domains intersect, and from microstructural studies this seem to be close to 425°C.

Another point to note here is that the strain rate sensitivity of the matrix increased with temperature (0.07 at 300°C, 0.15 at 425°C and 0.2 at 550°C, refer to section 5.1.1). This simply means that the rate at which the matrix strength increased with strain rate was higher for higher temperatures. Under normal strain rates, usually the contribution of the strain rate strengthening to the overall matrix strength is small. However, during non-uniform deformation (macro-necking), the strain rates in the necked region may be a few orders of magnitude higher than the nominal strain rate - and then the contribution of strain rate strengthening may be considerable. Since an elevation of strength increases
the stress required for failure, this in turn, will make the sample less failure-prone (in the limit, the material may exhibit superplasticity). The higher damage-tolerance exhibited by the extruded MMC at 550°C may be due to this phenomenon.

The influence of temperature on failure strain for the as-cast MMC was minimal (see Figure 4.20), possibly due to the extremely localized nature of the fracture process. However, the variation trends seem to be qualitatively quite similar to that in the extruded MMC.

5.2.3.2 Strain Path Dependence of Failure Strain

In the extruded MMC, the failure strain under uniaxial tensile loading was much higher than in collar compression tests. On the other hand, failure strains under the different loading conditions were virtually the same for the as-cast MMC.

The strain path in the collar for the extruded MMC initially was similar to that in an uniaxial tensile test. At larger strains, however, it tended to a plane strain condition, especially at higher temperatures (425°C and 550°C, see Figure 4.22). Under uniaxial tension,

\[ \sigma_1 = \sigma_{imp}, \]
\[ \sigma_2 = 0, \]
\[ \sigma_3 = 0, \]

and so, \( \sigma_{hyd} = \frac{\sigma_{imp}}{3} \)
Figure 5.15 Schematic representation of (a) temperature dependence of nucleation mechanisms, and (b) the resultant reversal in ductility.
where, $\sigma_{\text{imp}}$ is the imposed stress, 1, 2 and 3 are the three principal stress directions, and $\sigma_{\text{hyd}}$ is the hydrostatic stress component ($\sigma_{\text{hyd}} = \frac{\sigma_1 + \sigma_2 + \sigma_3}{3}$).

In plane strain deformation,

$$\sigma_1 = \sigma_{\text{imp}},$$

$$\sigma_2 = 0,$$

$$\sigma_3 = \frac{\sigma_{\text{imp}}}{2},$$

and so, $\sigma_{\text{hyd}} = \frac{\sigma_{\text{imp}}}{2}.$

Thus, under plane strain condition, the hydrostatic stress is higher compared to uniaxial deformation. Higher hydrostatic stress increases both nucleation (by increasing the maximum normal stress) and growth rates [Rice and Tracey 1969] of the voids, and thus accelerates void coalescence. The higher hydrostatic stress in a plane strain deformation may be deemed responsible for premature failure in the collar compression sample. In case of the as-cast material, the failure strains were extremely small, and the strain path essentially remained in the uniaxial-tensile domain (see Figure 4.21).

5.2.3.3 Pre-necking vs. Post-necking Strain

Under tensile loading, the pre-necking deformation in any material is macroscopically uniform - i.e. the reduction in the cross section area is constant across the gage length. However, with the onset of macro-necking, the deformation becomes highly non-uniform, with the reduction in area concentrating in and around the neck.
From Figure 4.16, we note that the necking strain for all the three materials were comparable (within 2-6%); their failure strains, on the other hand, were widely different. This seems clear in Figure 5.16 where the non-uniform strain (i.e. the post necking strain = failure strain - necking strain) has been plotted for AA6061 and the MMCs at a strain rate of 1/s. Obviously, it is this non-uniform strain that makes the difference in ductility in the three materials.

As noted several times in this thesis, the failure strain in as-cast MMC was 5-10 times lower than that in extruded MMC - even though their particle dimensions and volume fractions were comparable. The industrial significance of this is apparent - the as-cast material has to be extruded in order to achieve reasonable material ductility (or formability). A part of the answer to this puzzle may lie in the particle distribution characteristics of the two materials - the extruded MMC exhibiting a more uniform, and the as-cast MMC a more clustered particle distribution. The next section discusses this phenomenon in detail.

5.2.4 Reduction in Ductility in As-cast MMC

The as-cast MMC microstructure exhibited large particle clusters, while in the extruded MMC, the particles were more uniformly distributed. The lower ductility exhibited by the as-cast material can be, at least partially, explained on the basis of the 'clustering phenomena': easier void nucleation and coalescence in the clusters.
Figure 5.16 Non-uniform strain in AA6061 and MMCs under uniaxial tension ($\varepsilon \sim 1/s$). (Note: The points indicate the mean, and the error bars the range of the observations).
5.2.4.1 Void Nucleation

To understand why a clustered structure is more prone to void nucleation (or, why voids preferentially nucleate in clusters), the stress state in a cluster is estimated here and compared with the stress estimate for an uniform-particle-dispersion structure.

Assuming a square array of cubical particles (see Figure 5.17), the inter-particle distance can be estimated to be

\[
\lambda = \left[ \frac{1}{3\sqrt{V_f}} - 1 \right] l
\]

where, \( \lambda \) is the inter-particle distance, \( V_f \) is the reinforcement volume fraction, and \( l \) is the particle dimension. When the distribution of the particles is uniform (\( V_f = 0.2 \)) \( \lambda/l \) is equal to 0.7. In clusters however, as we have noted earlier in section 4.12, the volume fraction of the particles was \( \sim 0.4 \). The normalized inter-particle distance (\( \lambda/l \)) in a cluster was thus, \( \sim 0.36 \).

In clusters, this reduced inter-particle distance increases the flow constraint on the matrix, and this causes local stress elevation. Assuming plane strain deformation, the variation in the principal stress in the loading direction (\( p \)) with \( \Delta (=\lambda/l) \) can be approximated by [Backofen 1972]

\[
\frac{p}{\sigma_0} = 1 + \frac{1}{4\Delta}
\]

where, \( \sigma_0 \) is the far-field principal stress in the loading direction (i.e. \( p/\sigma_0 \) is the stress elevation). Our \( \lambda/l \) values of 0.7 (uniform square distribution) and 0.36 (clustered square distribution), yield \( p/\sigma_0 \) values of 1.36 and 1.7 respectively.
**Figure 5.17** 2-D schematic diagram of the particle arrangement considered - square array of cubical particles.
Thus, while particles in a uniform structure experience a stress level that is 36% higher than the far-field stress (which, in turn, is dictated by the material flow stress), in clusters the stress elevation is significantly higher: the particles experience 70% higher stress. And, since both particle cracking and interfacial decohesion is stress-dependent, increased damage nucleation may be expected in a cluster.

5.2.4.2 Void Coalescence

Void coalescence occurs when the void length is equal to the inter-void spacing (Brown and Embury's model [1973]). In a cluster, due to the higher local particle volume fraction, the inter-void spacing is smaller. Consequently, the length of the voids required for void coalescence is smaller. Thus, premature void coalescence may be expected in a cluster.

However, it must be noted here that Brown and Embury's model deals with a situation where the inter-particle spacing is larger than the particle dimensions, and thus, considerable growth of the voids is required before coalescence. In the MMCs under study, almost under no situation (except for tensile straining of extruded MMCs at 550⁰C and low strain rates) do we find void dimensions (away from the fracture surface) even comparable to the particle. Thus, the coalescence mechanism in these materials may be quantitatively different from that portrayed by the Brown-Embury model.
As noted in Section 4.1, there were a few macro-differences between the particle distribution characteristics in the as-cast and extruded samples: the as-cast samples (a) contained smaller particles, and (b) were reinforced to a higher volume fraction. We will end this section with a discussion on the possible effects of this on the observed lower ductility of the as-cast MMC.

Under the realm of continuum mechanics, the strengthening by the ceramic particles is independent of the particle size. Such seems to be true here, the peak compressive flow stress in the as-cast and extruded MMC being quite similar (refer to section 4.2.1). On the other hand, the probability of containing a critical sized flaw is higher for a larger particle, and so, the probability of fracture of a particle increases with particle size. Thus, the smaller particle size in the as-cast MMC, by itself, may be expected not to decrease, but rather increase the failure strain in the MMC.

The smaller particle size and the slightly higher volume fraction present in the as-cast MMC, however, when combined, means a lower inter-particle spacing. As for example, using the simple cubical particle, square distribution model discussed in section 5.2.4.1 (i.e. discounting any clustering in the as-cast microstructure), and using the equivalent diameter of the particles as the relevant dimension, the inter-particle spacings we obtain for the extruded and as-cast MMC are 13.4 μm and 8.5 μm respectively (when normalized by the particle sizes, the inter-particle spacings are 0.74 and 0.66 respectively). Thus, even discounting the clustering phenomena, the inter-particle spacing in as-cast MMC was smaller - however, the difference (when normalized by the particle dimension) was small. This reduced inter-particle spacing may affect the void
coalescence in these materials. However, as noted before, the exact nature of void coalescence is not very well studied for situations where the particle sizes are larger than inter-particle spacings, and further research is required before any conclusion can be drawn in this issue.

5.3 Summary

In this chapter, different issues relevant to the role of the reinforcing particles on the flow and fracture of MMCs at high temperatures and moderately high strain rates have been discussed. It is clear that cognition of the micro (e.g. void evolution) and meso-phenomena (e.g. damage distribution) is imperative in the proper understanding of the macro-process (e.g. failure strain variation).

The next section will summarize the important conclusions from this study and make some suggestions for future work.
6.0 Conclusions and Future Work

6.1 Conclusions

This work was primarily concerned with investigating the role of the reinforcing particles on the plastic flow and fracture of metal matrix composites in the high temperature, moderately high strain rate domain. The following important conclusions can be drawn based on the study:

The ceramic particles strengthened the matrix, and elevated the flow stress. The strength elevation was typically around 15-30%. Interestingly enough, the strengthening was less effective at 550°C (compared to 425°C), though the higher strain rate sensitivity of the matrix at that temperature predicts otherwise.

MMCs showed a very low strain to failure. The ductility of the extruded MMC was typically 5-10 times lower than that of unreinforced AA6061. The as-cast MMC was 25-50 times less ductile than the unreinforced matrix. In AA6061, the failure strain increased with increase in temperature and decrease in strain rate. However, the trends were less clear for MMCs. The (MMC) ductility also varied depending on the mode of loading: the extruded MMC showed a higher strain to failure under uniaxial tensile loading as compared to the collar compression test. This was deemed to be due to the higher hydrostatic stress in the latter.

For metal matrix composites, the voids nucleated at the second phase particles. Particle cracking was the most important nucleation mechanism at lower temperatures (300°C). At higher temperatures (550°C), interfacial decohesion at the spinel-matrix...
(MgAl$_2$O$_4$-AA6061) interface was the primary mode of void nucleation. Both the mechanisms were seen to be operative at an intermediate temperature (425°C). The composite ductility was found to be maximum at this temperature (425°C).

The effect of strain rate on MMC ductility was minimal. An increase in strain rate, however, was noted to favor particle cracking (as the dominant void nucleation mechanism).

For the extruded MMC, the extent of tolerable damage (damage before sample failure) increased with temperature. At the highest temperature (550°C) and lower strain rates (0.1/s and 1/s), the failed samples showed considerable damage even away from the fracture surface. It is suggested that when voids nucleated by particle cracking, the failure was more catastrophic than when interfacial decohesion was the principal nucleation mechanism.

In the as-cast MMC, under all conditions, the damage tolerance was extremely low. The coalescence of the voids to form a macro-crack was primarily concentrated in particle clusters. Particle clustering in this material may be deemed to be the principal cause for its extremely low ductility.
6.2 Future Work

Based on this study, the following are some of the topics which may be considered for future work:

(1) One of the interesting observations in this study was that the primary void nucleation mode shifted from particle cracking to interfacial decohesion as the temperature was raised (similar observations have also been reported by other authors in similar systems, see for example [Zhao et. al. 1994]). As the test temperature is raised, the material flow stress drops, until the stress on the particles is no longer high enough to crack them. So, it is understandable that the probability of particle cracking drops at higher temperatures. But at the same time, the normal stress across the interface (believed to be responsible for interfacial decohesion) also decreases. It is thus proposed that the interfacial strength in these materials drops as the temperature increases. To verify this proposition, a detailed study on the properties of metal-ceramic interfaces over a large temperature range is required - and it may be a worthy enterprise.

(2) Most of the void growth models available today (e.g. Rice and Tracey’s model [Rice and Tracey 1969]) deal with the growth of isolated voids in infinite matrices. In MMCs however, the voids are associated, not only with other voids in near vicinity, but also with the ceramic particles (in the same size range as the voids) which may aid (by raising the local stress) or deter (by creating a physical barrier to the void growth path) the void growth process. Simulation of void growth in the real microstructure of metal matrix composites may thus, be an useful undertaking.
Finally, it was noted that clustering of the second phase particles drastically reduced the ductility in as-cast MMCs. Fracture in these materials is generally catastrophic - and the influence of microstructural disorder (e.g. clustering) on such a process has been shown (in this study) to be extremely high. Experimental and numerical studies, on model and random systems, for quantitative estimation of this influence may now be undertaken: the end goal for such a study being the prediction of failure strain for a given microstructure.
Appendix

Table A1  Variation in tensile strength in as-cast MMC with temperature and strain rate

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Strain rate (/s)</th>
<th>Tensile strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>300</td>
<td>0.1</td>
<td>153, 148, 168</td>
</tr>
<tr>
<td>300</td>
<td>1</td>
<td>174, 167, 159</td>
</tr>
<tr>
<td>300</td>
<td>10</td>
<td>174, 183, 182</td>
</tr>
<tr>
<td>425</td>
<td>0.1</td>
<td>33, 62, 64, 60</td>
</tr>
<tr>
<td>425</td>
<td>1</td>
<td>54, 82, 83, 75</td>
</tr>
<tr>
<td>425</td>
<td>10</td>
<td>69, 94, 99, 101</td>
</tr>
<tr>
<td>550</td>
<td>0.1</td>
<td>29, 26</td>
</tr>
<tr>
<td>550</td>
<td>1</td>
<td>43, 41</td>
</tr>
<tr>
<td>550</td>
<td>10</td>
<td>60, 60</td>
</tr>
</tbody>
</table>
Table A2  Variation in as-cast MMC failure strain with temperature and strain rate under uniaxial tensile and collar compressive loadings

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Strain rate (/s)</th>
<th>Failure strain - Uniaxial tension</th>
<th>Failure strain - Collar compression</th>
</tr>
</thead>
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References


Poole, W. J., The Deformation of Copper-Tungsten composites, Ph. D. Dissertation, McMaster University, Canada (1993).


