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Short communication

The tensile behavior of two magnesium alloys reinforced with silicon carbide particulates

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Abstract

In this paper is reported the results of a study aimed at establishing an understanding the role of particulate reinforcement on tensile deformation and fracture behavior of magnesium alloys discontinuously-reinforced with silicon carbide (SiC) particulates. An increase in particulate reinforcement content was observed to decrease ultimate tensile strength and ductility of the composite when compared to the unreinforced counterpart. Cracking of the individual and clusters of reinforcing particulates present in the microstructure dominated tensile fracture of the composite, on a microscopic scale. Final fracture occurred as a result of crack propagation through the matrix between particulate clusters. The fracture behavior of the composite is discussed in light of the concurrent and mutually interactive influences of intrinsic microstructural effects, deformation characteristics of the metal matrix and the particulate reinforcement, nature of loading and local stress state.

Keywords: Silicon carbide particulates; Tensile fracture; Metal matrix composites

1. Introduction

A sustained interest arising from an increasing need to improve fuel economy, minimize vehicle emissions, enhance styling options, improve overall performance, while concurrently maintaining safety, quality, reliability, durability and also ensuring profitability, are certainly a few of the challenges that need to be addressed by both existing and future generation materials that find use in performance-critical applications. This burgeoning need has provided the desired impetus for the development, emergence and use of a “new” class of metallic materials, termed as metal-matrix composites (MMCs) [1–4]. A class of metal-matrix composite materials, which have emerged to attract attention and use in automotive components, space applications and consumer-related products, are the magnesium alloy matrices discontinuously-reinforced with particulate, whisker or short-fiber reinforcements. These composites will be referred to henceforth in this paper as DR MMCs. The presence of a discontinuous reinforcement phase in a continuous magnesium alloy metal matrix can result in properties not attainable by other means, thus enabling an extension of the potential range of possible applications [5]. The ability to achieve improvements in mechanical properties is dependent on mutually interactive influences of: (a) intrinsic properties of the composite constituents; and (b) the size, shape, orientation, volume fraction and distribution of the reinforcement phase in the metal matrix [6–10].

Selection of reinforcement type and geometry (shape) is critical to obtaining the best optimum combination of properties at a substantially low cost [11–13]. Of the available choices silicon carbide particulates (SiC_p) are the most preferred reinforcements primarily because enhanced properties can be easily achieved with little or no penalty on density [13]. From a design perspective, the attractiveness in preferring and choosing a discontinuously-reinforced magnesium alloy MMC for many applications stems from an improvement in specific modulus, that is, the density compensated increase in elastic modulus. Emerging applications for the DR magnesium alloy-based MMCs involve automotive and aerospace components such as automotive pulleys;

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gog-tooth sprockets, cylinder liners and aircraft engine castings.

Only few studies have focused in entirety on developing an understanding of the influence of reinforcement particle on matrix microstructure and tensile response of these alloys and there certainly exists a need to expand this aspect of mechanical characterization. In fact, presence of discontinuous particulate reinforcements in a ductile magnesium alloy metal matrix will tend to alter the microstructure of the material during secondary heat treatments when compared with the unreinforced monolithic counterpart [7,14]. A change in intrinsic microstructural features will exert an appreciable influence on mechanical response and fracture behavior. The objective of this paper is to record the influence of discontinuous particulate reinforcement on tensile deformation and fracture behavior of two magnesium alloys reinforced with particulates of silicon carbide. The final fracture behavior of the composites is discussed in light of concurrent and mutually interactive influences of reinforcement effects on alloy microstructure, deformation characteristics of the composite constituents, and the microscopic mechanisms governing fracture.

2. Materials and processing

2.1. Materials

Two magnesium alloys AM 60 (Mg–6Al) and AZ92 (Mg–6Al–2Zn) were selected for this study. The reinforcement used was silicon carbide particulates (henceforth referred to as SiC_p) having an average diameter of 20 μm. Test castings of both the unreinforced alloys and the composite counterpart (base alloy reinforced with silicon carbide particulates) were produced using the squeeze casting technique. For both alloys, two volume fractions of the carbide particulate phase were used (a) 10, and (b) 15%. The test castings were then subject to the following heat treatment sequence prior to characterization of mechanical properties and failure-damage analysis.

(1) *AM60 alloy*—T5 treatment, i.e.: direct quench from the casting temperature followed by aging at 288 °C for 2 h. The composites are denoted as AM60/SiC/xxp-T5, where xx denotes the volume fraction of the reinforcing phase.
(2) *AZ92 alloy*—T6 treatment: solution heat treatment at 405 °C and subsequently aged at 218 °C for 2 h. The composites are denoted as AZ92/SiC/xxp-T6, where xx denotes the volume fraction of the reinforcing phase.

2.2. Processing

The stir casting route (also known as the vortex method) is a commercially feasible, economically viable and hence a widely adopted technique for the production of particulate-reinforced metal-matrix composites based on aluminum and magnesium alloy families. In this research investigation, the stir casting equipment (designed and fabricated in situ) was used for producing the magnesium alloy-based metal-matrix composites. An electric resistance furnace was used to hold the molten metal at the desired temperature while it was being stirred.

The required amount of base alloy was melted in a large furnace. After skimming off the dross, the crucible containing the melt was transferred to a smaller holding furnace whose temperature was maintained at 700 °C. A stirrer was then introduced into the melt and the process of mechanical stirring initiated. The stirrer speed was maintained at around $600 + 20$ rpm. The rotational speed facilitates in inducing a vortex having adequate depth. The silicon carbide particulates (pre-heated to 700 °C) were then added into the vortex. Two different volume fractions of the reinforcing SiC_p, namely: 10 and 15 vol.% were used. Stirring of the resultant melt (mixture of molten metal and reinforcing SiC particulates) was continued for 3 min following the addition of the particles. The composite melt was then poured into a squeeze casting die and squeeze cast at a pressure of 70 MPa. The castings (both unreinforced magnesium alloy and magnesium-matrix composites) were cut to size using a high-speed abrasive cutting machine equipped with a diamond-cutting wheel.

3. Experimental techniques

3.1. Sample preparation

Blanks were cut from both the AM60/SiC/xxp-T5 and AZ92/SiC/xxp-T6 composite extrusions using a diamond-coated blade. Smooth cylindrical test specimens were precision machined from the blanks using diamond tooling. The specimens were machined with the stress axis parallel to the longitudinal (extrusion) direction and conformed to specifications in ASTM Standard E8-93 [15]. To minimize the effects of surface irregularities and finish, the test specimen surface was prepared by mechanically polishing the gage section using progressively finer grades of silicon carbide impregnated emery paper and then finish polished to obtain a mirror finish and free of all circumferential scratches and surface machining marks.

3.2. Mechanical testing

The tensile tests were performed on microprocessor-based computer controlled universal test machine (UTM) equipped with a 100 KN load cell. The tests were conducted in controlled laboratory air environment (relative humidity of 55%) at ambient temperature (27 °C). The test specimens were deformed in accordance with ASTM E8 at a strain rate of $1 \times 10^{-2}\ s^{-1}$.

3.3. Microstructural evaluation and fracture surface analysis

Metallographic samples were cut from the extruded composite extrusion, mounted in bakelite and mechanically ground on progressively finer grades of SiC impregnated emery paper, using copious amounts of water as lubricant. The mounted and ground samples were then mechanically polished using 1 μm size alumina-powder suspended in distilled water. Finish polishing to mirror finish was achieved using 0.5 μm alumina powder. Reinforcement morphology and its distribution in the metal matrix, and other key microstructural features were examined in an optical microscope and photographed using bright-field illumination technique.

Fracture surfaces of the deformed tensile samples were examined in a scanning electron microscope (SEM) (Model: JEOL-JSM-84A) to: (a) determine the macroscopic fracture mode; and (b) characterize the fine-scale topography and establish the microscopic mechanisms governing fracture. The distinction between macroscopic mode and microscopic fracture mechanisms is based on the magnification level at which the observations are made. Samples for SEM observation were obtained from the failed specimens by sectioning parallel to the fracture surface. Matching fracture surfaces were viewed in many cases in order to determine the presence of fractured SiC_p on both halves of the specimen.

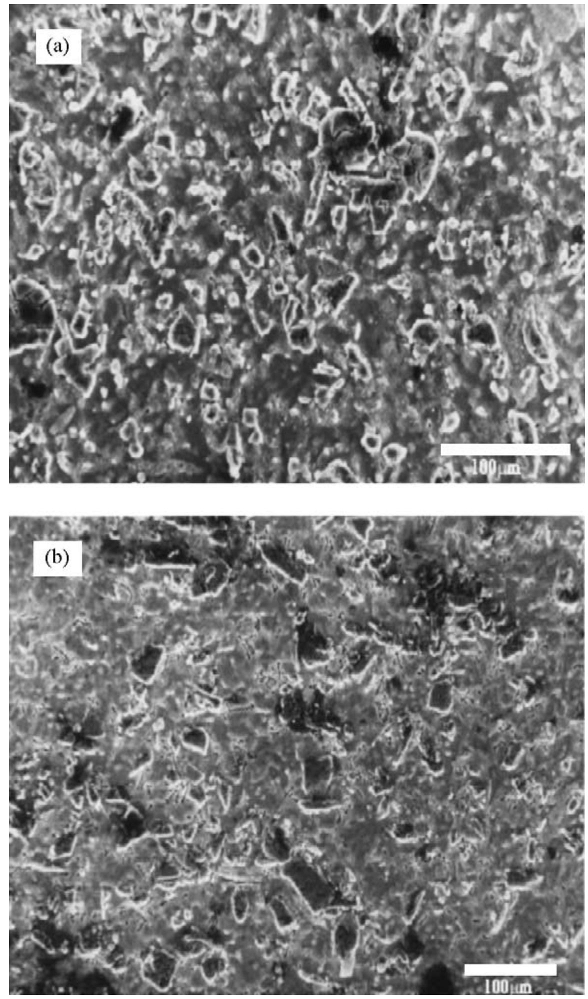


Fig. 1. Optical micrographs illustrating microstructure of the composites: (a) AM60/SiC/xxp-T5; and (b) AZ92/SiC/xxp-T6.

4. Results and discussions

4.1. Undeformed microstructure

The optical micrographs showing microstructure of the two as-processed plus heat-treated composites are shown in Fig. 1. The silicon carbide particulates (SiC_p) in both AM60 and AZ92 magnesium alloy metal matrix were non-uniform in size and near uniformly dispersed through the alloy matrix. Agglomeration or clustering of the reinforcing SiC was observed at random intervals. An agglomerated site consisted of a few larger SiC particulates intermingled with the smaller and more uniformly shaped particles. The pressure applied during squeeze casting facilitated in achieving a refined microstructure. No attempt was made in this study to determine the particle size distribution for the materials.

4.2. Tensile properties

The ambient temperature tensile properties of the unreinforced alloys and the composite counterparts are summarized in Table 1. Test results for both composites reveal a nominal increase in the tensile yield strength with addition of particulate reinforcements to the base alloy but with a concurrent decrease in ultimate tensile strength. The increase in yield strength is as high as 11% for AM60/SiC/xxp-T5 and only as high as 4% for the AZ92/SiC/xxp-T6 composite counterpart. The degradation in ultimate tensile strength with addition of particulate reinforcement to the magnesium alloy metal matrix, with respect to the unreinforced metal matrix, is as high as 10% for the AM60/SiC/xxp-T5 composite and as high as 10% for the AZ92/SiC/xxp-T6 composite. The degradation in ultimate tensile strength of the composite relative to what is theoretically expected arises

Table 1
Room temperature tensile properties of the two magnesium alloys and magnesium composites

Material	Condition	Yield strength (MPa)	Ultimate tensile strength (MPa)	Elongation (%)
AM60	T5	122	224	9.60
AM60/SiC/10p	T5	128	218	4.20
AM60/SiC/15p	T5	133	212	2.30
AZ92	T6	205	280	7.2
AZ92/SiC/10p	T6	208	274	2.2
AZ92/SiC/15p	T6	212	272	1.7

from the concurrent and mutually interactive influences of residual stress distribution interactions, and failure of the reinforcing SiC particulates by fracture.

The ductility of the magnesium alloy-based composites, quantified in terms of tensile elongation, decreased with an increase in SiC reinforcement content. The decrease in tensile elongation was as high as 6% for both the AM60/SiC/xxp-T5 and AZ92/SiC/xxp-T6 composites. For the chosen magnesium alloy–SiC composites having coefficient of thermal expansion mismatch between the magnesium alloy metal matrix and the reinforcing SiC particulates, the plastic deformation of the ductile magnesium alloy metal matrix, in the presence of discontinuous SiC particulate reinforcements, is non-uniform, i.e. heterogeneous, primarily due to the hard, brittle and essentially elastically deforming particles resisting plastic flow of the soft, ductile and plastically deforming metal matrix. Further, as a result of the elastically deforming particles resisting the plastically deforming metal matrix, an appreciable internal stress or back stress is generated, which contributes to an overall strengthening of the reinforced metal matrix with respect to the unreinforced counterpart. This is aided by the conjoint influence of: (a) small contribution from dispersion strengthening caused by the presence of reinforcing SiC_p in the magnesium alloy metal matrix; (b) dislocation–dislocation interactions; and (c) dislocation–microstructural feature interactions.

4.3. Tensile fracture

The AM60/SiC/xxp-T5 and AZ92/SiC/xxp-T6 composites exhibited limited ductility on a macroscopic scale with fracture essentially occurring normal to the tensile stress axis. However, microscopic examination of the fracture surfaces at high magnifications, revealed features reminiscent of locally ductile and brittle mechanisms. Representative fractographs of the tensile fracture surfaces are shown in Figs. 2–3.

4.3.1. AM60/SiC/xxp-T5 composite

High magnification observations of the fracture surfaces revealed bimodal failure with the presence of both ductile and brittle fracture regions. The matrix of the composite was covered with pockets of dimples coupled with microscopic cracks and cracked reinforcing SiC particulates (Fig. 2b), features reminiscent of locally brittle failure. Basically little difference was observed in the microscopic fracture surface features of the tensile samples of the two composites AM60/SiC/10p-T5 and AM60/SiC/15p-T5.

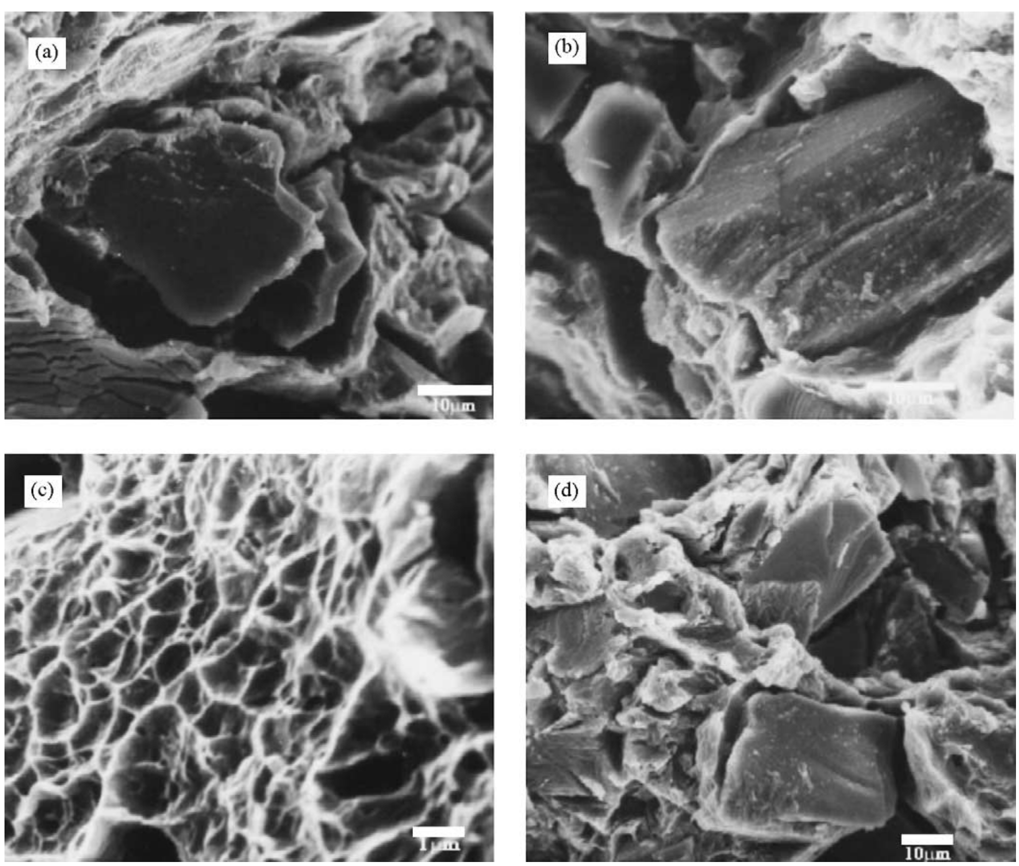
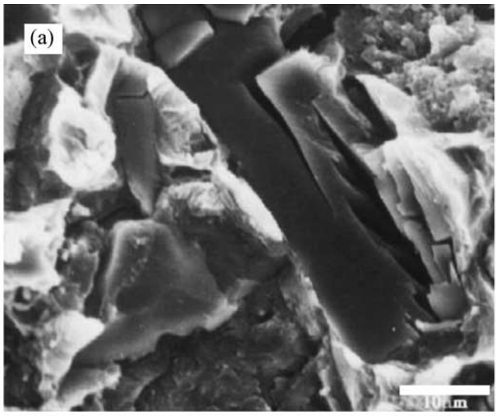


Fig. 2. Scanning electron micrographs of the tensile fracture surfaces of the AM60/SiC/10p-T5 composite showing—(a) mixed mode: ductile failure of matrix and brittle failure of reinforcement; (b) decohesion at the particulate–matrix interface; (c) presence of ductile dimples between the reinforcing SiC particulates; (d) agglomeration of particulates on tensile fracture surface.



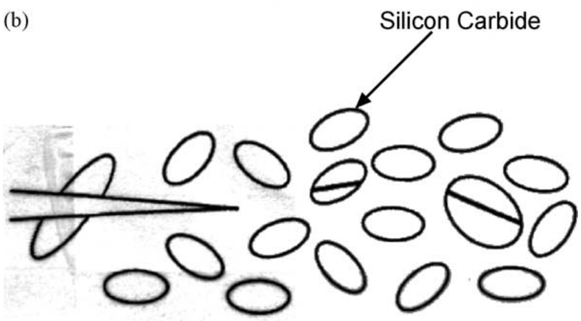


Fig. 3. (a) Scanning electron micrograph showing cracked SiC particle; (b) A schematic showing the fracture mode for this SiC reinforced magnesium alloy matrix.

A critical examination of the tensile fracture surfaces revealed damage associated with fracture to be highly localized at the discontinuous SiC particulate reinforcements with little evidence of void formation away from the fractured particles. Failure of the reinforcing SiC particle through cracking and/or debonding at its interfaces with the metal matrix is governed by the conjoint and mutually interactive influences of local plastic constraints, particle size and degree/severity of agglomeration. The local particle constraints are particularly important for the larger size particles and particle clusters during composite fracture. The characteristics that contribute to constraint are particle shape, size (r) and volume fraction (f). Contributions from particle shape are neglected since all particles are considered spherical and near uniformly dispersed through the alloy matrix. The intrinsic brittleness of the reinforcing SiC particle coupled with the propensity for it to fracture due to localized deformation, results in both second-phase and reinforcement particle cracking being the contributing fracture mode.

The limited growth of the fine microscopic voids during far-field loading coupled with lack of their coalescence, thus inhibiting the dominant fracture mode to be ductile failure, clearly indicates that the plastic deformation properties of the AM60 alloy are significantly altered by the presence of reinforcing SiC_p. Very few of the fine microscopic voids coalesce and the halves of these voids are the shallow dimples observed covering the transgranular fracture surfaces (Fig. 2c). The lack of formation of ductile dimples, as the dominant fracture mode, is essentially attributed to be due to the constraints on plastic flow in the composite matrix caused by the presence of discontinuous SiC_p reinforcements and not to limited ductility of the magnesium alloy. With a gradual increase in strain, during tensile deformation, the larger-sized SiC particles fracture first, followed by fracture of the smaller-sized SiC particles. The tendency for the larger particles to fracture at a lower stress than the smaller particles is because the flaw size in a particle is directly proportional to its size. Assuming the flaw size (d) to be a fraction (x) of the particle diameter ($2r$) then the stress that causes the reinforcing particle to fracture is expressed as:

$$\sigma_{\text{particle fracture}} = \frac{K_{\text{particle fracture}}}{(2\pi d)^{0.5}}$$

where $K_{\text{particle fracture}}$ is the fracture toughness of the reinforcing SiC particle or second-phase particle in the microstructure.

In regions of SiC particulate clustering, or agglomeration, the short interparticle distance facilitates linkage between neighboring voids and microscopic cracks as a direct result of decreased propagation distances between the cracked SiC particles. Based on observations of the fracture surfaces, the fracture plane of the cracked SiC particle is essentially perpendicular to the loading axis, suggesting the importance of normal stress in inducing particle fracture. Failure of the reinforcing SiC particles both by cracking and decohesion at its interfaces with the metal matrix is responsible for the overall inferior ductility of the composite compared to the unreinforced counterpart (Table 1).

4.3.2. AZ92/SiC/xxp-T6 composite

For both volume fractions of the particulate reinforcement phase tensile fracture surfaces of this composite revealed the damage associated with fracture to be localized at the discontinuous SiC particulate reinforcements with little evidence of microscopic void formation away from the fractured particulates. Fracture of the hard, brittle and essentially elastically deforming SiC reinforcement was observed to be greater in regions of particulate clustering. This is essentially attributed to the following:

(a) enhanced local stresses resulting from a restriction of plastic deformation; and
(b) intrinsic brittleness of the reinforcing SiC particle with the propensity for it to fracture due to localized deformation.

These two mutually competitive factors result in particulate cracking and aided by decohesion, or separation, at the matrix–particle interfaces, to being the dominant fracture modes (Fig. 4). The damage of the composite microstructure, during uniaxial loading, arising from the synergistic influences of SiC particulate cracking and decohesion at the magnesium alloy metal matrix–SiC particulate interfaces,

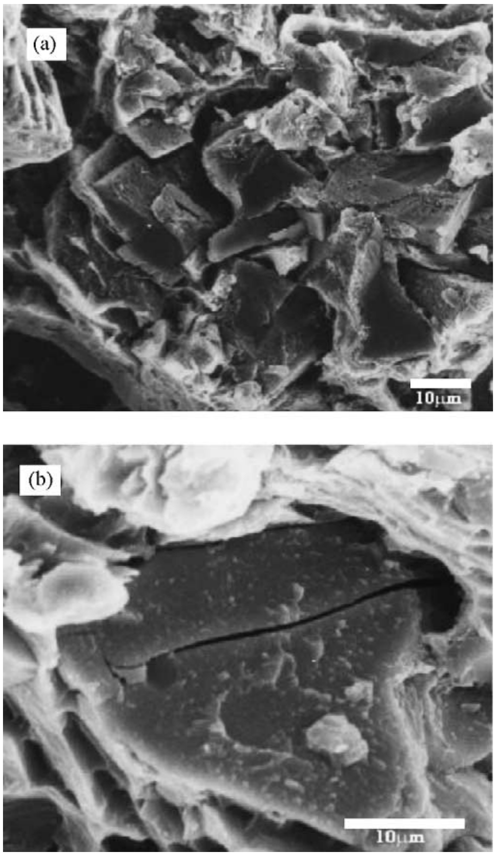


Fig. 4. Scanning electron micrographs of the tensile fracture surface of the AZ92/SiC/xxp-T6 composite showing: (a) predominantly brittle fracture morphology; (b) failure of the reinforcing SiC particulate by cracking.

results in a detrimental influence on tensile ductility. Assuming the magnesium alloy metal matrix–SiC reinforcement particle interfaces are strong, the triaxial stresses generated during far-field tensile loading favors limited growth of the microscopic voids in the matrix of the composite. The lack of extensive void growth and their coalescence, as a dominant fracture mode for this magnesium alloy-based metal-matrix composite, suggests that:

(a) The deformation properties of the magnesium alloy metal matrix are appreciably altered by the presence of discontinuous SiC particles.
(b) The overall fracture strain is critically controlled by both void nucleation strain and linkage strain. An observation of the tensile fracture surfaces revealed that less than 25% of the reinforcing SiC particles had fractured. This indicates that not all of the SiC particles were loaded to their fracture stress suggesting the importance and role of SiC particulate distribution in the metal matrix.

5. Conclusions

A study of the tensile properties of two magnesium alloy-based metal-matrix composite provides the following key observations:

1. For both volume fraction of the particulate reinforcement the yield strength of AM60 and AZ92-based particulate MMCs is only marginally higher than the unreinforced counterpart. Addition of particulate reinforcement resulted in a marginal degradation in ultimate tensile strength. The tensile strength is higher than the yield strength indicating the occurrence of work hardening beyond yield.
2. For both volume fractions of the SiC particle reinforcement the percent elongation of the composite is lower than the unreinforced counterpart.
3. The presence of hard, brittle and elastically deforming SiC particulates in a soft, ductile and plastically deforming magnesium alloy metal matrix causes fine microscopic cracks to initiate at low values of applied stress. Fractography revealed limited ductility on a macroscopic scale, but microscopically features were reminiscent of locally ductile and brittle mechanisms.
4. Constraints in mechanical deformation, induced in the metal matrix, by the hard, brittle and elastically deforming SiC_p reinforcement phase coupled with local stress concentration effects at the magnesium alloy metal matrix–reinforcement particle interfaces promotes failure through the conjoint influences of particle cracking, decohesion at its interfaces with the metal matrix and fast fracture through the matrix. Fracture of the plastically deforming metal matrix occurred through the formation, growth and coalescence of the microscopic voids.

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