MICROSTRUCTURE AND MECHANICAL PROPERTIES OF SIC_P/AZ91 MAGNESIUM MATRIX COMPOSITES PROCESSED BY STIR CASTING

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Magnesium alloy (AZ91) matrix composites reinforced with various volume fractions of SiC-particulates were processed. A Stir casting process has been employed under an inert atmosphere to produce magnesium matrix composites. This process leads to a complete wetting of SiC particles in the molten magnesium. Microstructural feature, such as SiC particles distribution, and the interface between the magnesium matrix and SiC was characterized by scanning electron microscopy (SEM) and energy dispersion X-ray analysis (EDX). The results show a refined matrix structure of the composite compared to the matrix alloy, a uniform SiC distribution in the matrix and that no reaction takes place during the synthesis of the magnesium alloy matrix composites between the oxidized SiC and the molten magnesium. However, AZ91 matrix composite showed a massive formation of ternary intermetallic compound of Mg17 (ZnAl)12 at the SiC interface. The mechanical properties of AZ91 and AZ91-SiC composites have been evaluated and the results are compared with the unreinforced AZ91 alloy. Results of the mechanical properties revealed an increase in hardness value, maximum tensile strength and 0.2 % yielding strength. It was proposed that the strength increase due to SiC addition to AZ91 alloy was a result of a change in the matrix strength, i.e. an increase in dislocation density and a reduction of matrix grain size. However, it is also evident that the strain to failure significantly decreased as the volume fraction of the particulate increased. Examinations of the fracture surface show good bonding between SiC/matrix, extent of fine dimples, matrix cracking and particle cracking.

INTRODUCTION

The attractive properties of magnesium and its alloys are their low density, high strength to weight ratio in cast form or wrought form [1]. Recently, as a result of general requirement for lighter weight automobiles to conserve energy, there has been a growing use of magnesium in the automobile field. The extraordinary growth in magnesium structural casting is explained by the need of car manufactures to lower fuel consumption while increasing comfort and safety of cars [2-3]. Reduction of car weight may be achieved by utilization of light weight structural magnesium castings. Therefore, the magnesium alloys attract lastly more and more interest of the manufacturers and designers, which makes them also attractive from the research point of view [4]. Despite its low density and the possibility to be economically formed into

complex net shape parts, magnesium suffers from several material inherent deficiencies like low young's modulus, insufficient strength, low wear and creep resistance that currently prevent the substitution of steel and aluminum for highly loaded structural parts [5]. Reinforcement of magnesium alloys with ceramic particulates has engineered a new family of materials that are marketed under the trade name metal-matrix composites offers one possibility to overcome these deficiencies. The excellent mechanical properties of these materials, together with weight saving (using reinforcement with lower density than the metal matrix) and relative low cost in production makes them very attractive for a variety of engineering applications [6-10]. The reinforcement used can take the form of continuous fibers [11], whiskers, short fibers [11,12] or particles [13-15]. Low density ceramics, e.g. boron, carbon, silicon carbide, and alumina are materials which have been produced in these various forms. The fiber reinforced composites offer the highest specific stiffness along the reinforcement direction [11], while particulate reinforced composites are more isotropic in the properties and are also easier to the process via powder metallurgical or casting route [13-15]. Silicon carbide particulates remain the most commonly selected reinforcement because of its cost, compatibility with magnesium matrix and high modulus [16-17]. There are different techniques that can be used for the fabrication of metal matrix composites such as powder metallurgy in the solid state [18, 19], preform infiltration [20, 21] compocasting or stir casting [22-24]. By liquid metallurgy technique, the second phase particles are introduced into the melt prior to casting and the composite melt is substantially cast either by gravity or under pressure. This technique being simple, economical and easily applicable in foundries, it gains wide popularity [17, 25]. The disadvantages of these processes, compared to solid state processes, are generally related to higher processing temperature used, which result in a greater propensity for matrix/ reinforcement chemical reactions [26]. The powder route also has some major disadvantages: It involves handling large quantities of highly reactive, potentially explosive powders and the manufacturing route is relatively complex and limited in the initial product forms it can produce. As a result, the product is expensive in comparison with the cast route [27]. Another problem, the reinforcement distribution will depend on the blending and consolidation procedures, as well as the relative size of the matrix and particles. If the matrix powder is large relative to the reinforcement, the reinforcing particles will agglomerate in the interstices of the coarse particle, and will be very inhomogeneously distributed in the final product [28]. Despite the importance of a uniform reinforcement distribution, too much exuberance for achieving this goal can cause reinforcement damage or breakage, and blending operations have all caused reinforcement breakage, especially when the reinforcement aspect ratio is high [29].

Accordingly, the present work was undertaken to synthesize magnesium based metal matrix composite (AZ91) reinforced with various volume fractions of SiC-particulates using conventional casting technique. Microstructural characterization, the distribution of the dispersoid phase in the matrix and the particle/matrix interfacial reaction between SiC and AZ91-matrix were investigated by using optical microscopy, SEM and EDX-analysis. The mechanical properties for the produced composites are also evaluated and the results obtained are compared with the unreinforced alloy.

EXPERIMENTAL WORK

Materials

The nominal composition of the matrix alloy (commercially designated as (AZ91) used in the present study was listed in Table (1). The SiC particulates with average sizes of 50 µm were selected as the reinforcement phase. The chemical analysis and (the) properties of SiC particulates were tabulated in Tables 2 and 3 respectively.

Table 1: The chemical analysis (wt.%) of AZ91, which is used as a matrix

Al	Zn	Mn	Si	Mg
8.9	0.69	0.11	0.01	bal.

Table 2: The chemical analysis of the SiC particles (wt. %), which are used in the present work.

SiC	C	Si	SiO_2	Fe	Al_2O_3,	CaO,	MgO
99.3	0.20	0.05	0.1	0.005	<0.01	<0.01	<0.01

Table 3: Some physical and mechanical properties of the SiC particles that are used in the present study.

Density g/Cm^3	Young's Modulus GPa	Tensile strength GPa	Thermal expansion coefficient ($10^{-6}K^{-1}$)
3.2	480	3	4.7

PROCESSING

The synthesis of the magnesium alloy matrix composites used in the present study were carried out according to the following procedure. The process consists of mixing preheated SiC with molten AZ91 alloy, melt stirring and composite casting. The lack of wettability was overcome by pre-treatment (oxidized) of SiC particles. Oxidation of SiC particles were carried out in an electrical furnace in air at 1100 °C for 2 hours. Also, the pretreatment of the surface of the dispersoid (SiCp) aims to remove the adsorbed gases from the surface and to make the entire surface uniformly active. About 200 g of the matrix alloy AZ91 was melted at 725 C° under an inert atmosphere in a graphite crucible. The stirrer was inserted just below the surface of the melt by using a speed of 900 rpm. This rotational speed creates a vortex of adequate depth. The silicon carbide particulate (pre-heated) was gradually added to the vortex at a rate of approximately 40 g.min^{-1}. The details of this method were discussed in a previous investigation [16]. To ensure complete assimilation of the particles, the melt was continuously stirred for 1 min and the molten metal was poured at 700 °C into the permanent mold (20 mm width, 40 mm length and 120 mm height).

MICROSTRUCTURAL CHARACTERIZATION

The microstructure examinations of AZ91 (Mg9AlZn1) and AZ91-SiC alloys were investigated, using optical microscope and scanning electron microscope (DSM 950 der Firma Zeiss) with energy dispersive spectroscopy (EDS der firma Link).

Test specimens taken from both the unreinforced magnesium alloy (AZ91) and the composite counterparts (AZ91/ SiC) were mechanically ground on a series of SiC emery papers under stream of water. The ground specimens were polished on a series of polishing cloth with diamond paste of 6, 3 and 1 µm. To reveal the intrinsic microstructural features, the specimens were etched. The etchant is a solution mixture of 1 ml HNO3 (conc.), 75 ml ethylene glycol and 24 ml water. Optical microscope was conducted on polished specimens cross-sections to determine the volume fractions, and the distribution of SiC in matrix. The volume fraction is achieved by point counting, using 10 x 10 grid on 20 different areas for each cross-section. The smallest and largest matrix volume fractions were discarded and averages were calculated from the remaining values according to the ASTM Standard E-562.

MECHANICAL PROPERTIES

The macrohardness was established by means of Vickers hardness testing using 10 Kg load. At least 10 macrohardness measurements were made for each specimen to ensure accurate results. The macrohardness values have also been checked for reproducibility with 3 specimens. The results cited are the average values from such multiple tests.

Tensile tests of unreinforced alloy and composites were performed on specimens 20 mm gauge length and 4 mm diameter with threaded ends, in accordance with the DIN 50125 D-4 (20 requirements. Tensile tests were done using a testing machine (der Marke Zweck-1475 with program software Z7005). The test was carried out at an applied strain rate of 0.5 mm/min. The reported results of the tensile tests are the average of three individual tests on three different specimens.

RESULTS AND DISCUSSION

Microstructure

The binary Mg-Al system was the basis for early magnesium casting alloys and the maximum solubility of Al is 12.7 wt. % at 437 C° (eutectic temperature) decreasing to about 2 % at room temperature [30]. The aluminum is partly dissolved in the magnesium-rich solid solution matrix and partly precipitated as the β-phase $Mg_{17}Al_{12}$ which is usually formed at the grain boundaries and is most prevalent in sand or permanent mold casting [30]. However, when zinc is added to magnesium- aluminum alloys, the magnesium aluminum eutectic takes a completely divorced form, in which massive particles of $Mg_{17}Al_{12}$ compound –or of $Mg_{32}(AlZn)_{49}$ compound if the ratio of zinc to aluminum exceeds 1:3- are surrounded by magnesium solid solution [31].

The unreinforced (AZ91) and reinforced material structures shown in Fig.1 (a-d) were solidified under the same solidification condition. The microstructure of the magnesium alloy AZ91 consisted of magnesium aluminum solid solution (α-solid solution) dendrites and intermetallic phase of $Mg_{17}Al_{12}$ which precipitate mainly along the grain boundaries, Fig. 1-a. The intermetallic phase of the $Mg_{17}Al_{12}$ type − gray

phase - was revealed, mostly at grain boundaries. The matrix structure of composites shows much smaller grain sizes, generally, the larger the volume fraction of SiC particles in the composite is, the smaller the grain size will be, Fig. 1 (b-e). The main reasons for the refinement of AZ91 alloy containing SiC-particles, may be that the sites for heterogeneous nucleation of the primary magnesium phase were numerous enough to decrease their size and the restricted growth of magnesium crystals caused by the presence of rigid SiC particles. The defects in the SiC particles—such as stacking faults, dislocations, and pits or grooves would also act as favourable sites for heterogeneous nucleation [32]. Some studies [33,34] reported significant grain refinement in the matrix in the SiC particle reinforced AZ91 magnesium alloys. On the other hand, it has also been reported that the primary magnesium phase could not heterogeneously nucleate on the surface of SiC particles, and thus the SiC particles would not refine the matrix grains [35].
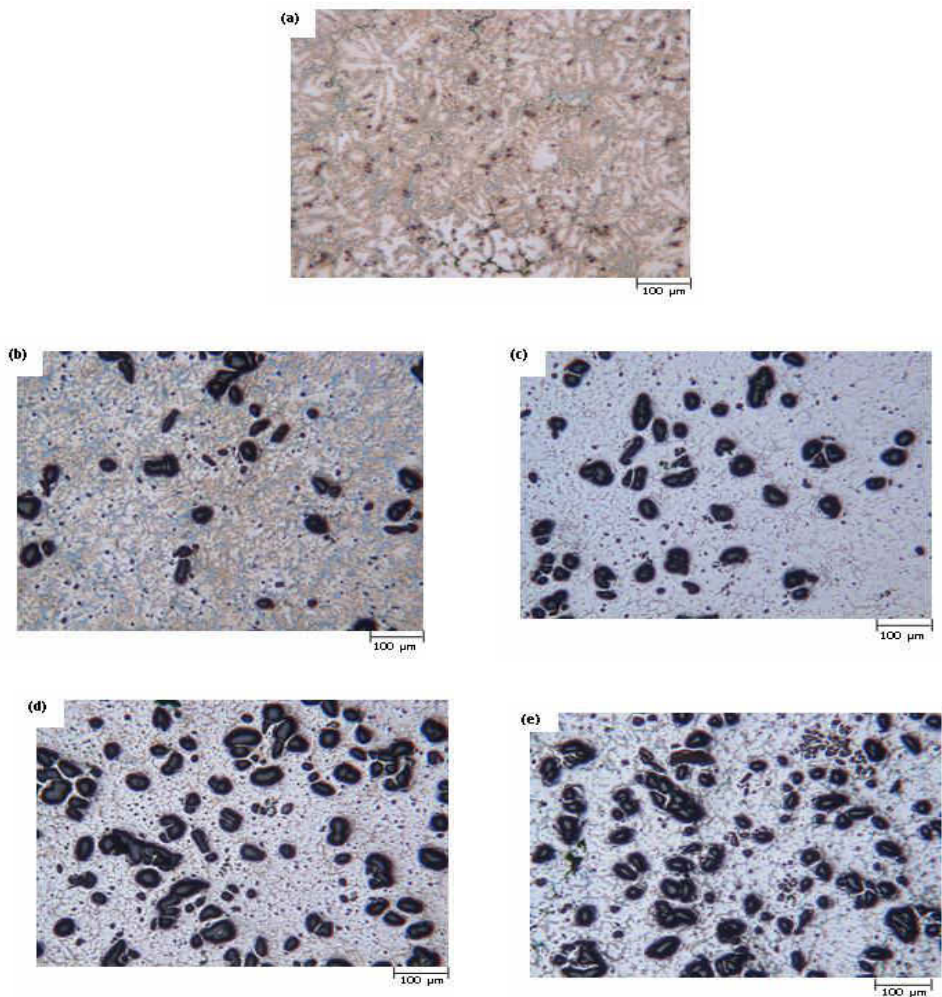
Fig. 1: Optical micrograph showing the as cast microstructure of (a) unreinforced AZ91 and AZ91 reinforced with (b) 5 vol. % SiC, (c) 10 vol. % SiC , (d) 15 vol. % SiC , (e) 20 vol. % SiC

The solidification microstructure of SiC/AZ91 composite that was investigated showed that a majority of the SiC particles were pushed by the primary magnesium phases and segregated at the grain boundaries. At the same time, some SiC particles were entrapped in the magnesium grain. Fig. 1(b-e) shows a typical microstructure of an Mg-alloy-silicon carbide particles composite, where the SiC particle is pushed by α-magnesium grains into the last freezing liquid where it is associated with intermetallic compound $Mg_{17}Al_{12}$. In such a situation, primary magnesium could not nucleate at the SiC particle surfaces, and the latter would be pushed by the solidifying primary magnesium. The conditions and mechanisms for heterogeneous nucleation of magnesium on the SiC particles were investigated in a recent research paper [36]. The nucleation of primary phases on reinforcing particles can also be influenced by the fact that these particles moves in the melt and sometimes they may reach a higher temperature zone leading to remelting of the nuclei. The temperature differences between the ceramic dispersoids and the melts may be responsible for the absence of nucleation of lower melting phases like α-magnesium and α-aluminum. Because α-magnesium does not nucleate on the   surface of the particulates reinforcement, the primary phases dendrites (grains) appear on the mold boundaries first, and push the particles in direction opposite to heat transfer, as the temperature in the mold decreases. A particle-free zone appears within an area where many grains develop in the same time.

The distribution of SiC particulates was mostly uniform and can be attributed to the ability of magnesium to wet SiC particulates, Fig.1 (b-e). The distribution of reinforcement in metal matrix rather becomes complicated, so that the reinforcement distribution is influenced by several factors such as stirring action [37], density difference between the particles and the molten metal and solidification rate [38]. In creating a homogeneous distribution of particles in molten alloy, the high shear rate caused by stirring the slurry result in a fairly uniform particle distribution in the radial direction, and also prevents particles from settling.   After mixing and before solidification, one major processing problem is that particles either sink or float, depending on the particle–to-liquid density ratio [39]. In foundry operations, segregation of the particles may occur between the time stirring has stopped and the melt has solidified. At sufficiently long holding times, top parts of the casting are completely denuded particles which settle to the lower part of the casting as a function of time [40]. Therefore, the melt must be re-stirred prior to casting if long holding times in the molten are used. According to Geiger and Zaki [41] the settling rate will also be a function of the particle density with particle shape and size possibly playing a role. At high volume fractions in MMCs, the particles interact with each other and the settling will be hindered [42]. At lower volume fraction of particles the settling time is less. So, it is apparent that a slurry with large size particles will have to be stirred all-time until casting [43]. In practice the situation is complicated by the fact that there is a range of particle shape and size. Particles of different shape and size will settle at different rates producing agglomeration. A near uniform distribution of SiC–particles was also observed for magnesium alloy metal matrix, Fig.2a. High magnification observations revealed to the absence of debonding between the reinforcing SiCp and the magnesium alloy metal matrix, Fig.2-b. This clearly suggests that the method of preparation used in this study is effective, resulting in a near uniform distribution of the reinforcing phase   in the metal matrix.
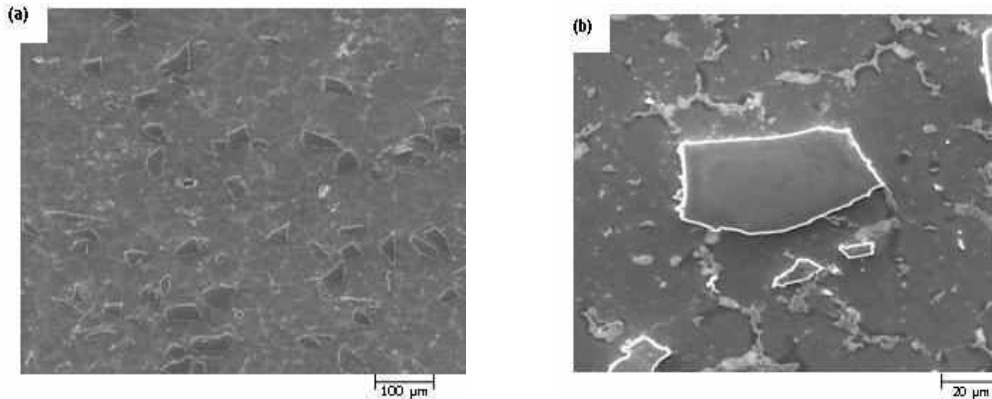
Fig. 2: Representative SEM photomicrograph showing (a) distribution of SiC in the matrix AZ91, (b) view of SiC-particle and its surrounding in an AZ91 matrix.

Composite Interface Behavior

In a composite, the nature of the interface becomes a factor to limit the contribution of the dispersed phase towards a given property of a composite. The primary role of the interface between the matrix and the reinforcement phase in a metal matrix composite is to transfer the load from the matrix to reinforcement. The interface may also be expected to serve as mechanical fuse or as a diffusion reaction barrier. The key features of the interface are the chemical reactions and the strength of bonding. Chemical interaction at a metal matrix composite interface is defined as mass transfer across the interface. Chemical reaction results in a new distribution of elements near the interface, the criterion and elimination of phases, and a new interface morphology. This occurs when components from the matrix and reinforcement dissociate and interdiffuse to form new phases. Finally the composition of the matrix can affect the reactivity of the interface. These are the major issues that motivate investigations of chemical reactions at metal matrix interface. The interaction between SiC -particulates and matrix will depend on the thermodynamic potential of the elements involved, the matrix composition, the nature of the surface reinforcement, and the fabrication method. It should be also noted that in this study, silicon carbide particles were oxidized in air at 1100 °C for 2 hours before being introduced into the magnesium melt. Then, the SiC-surface was covered with SiO_2 during pre-treatment of SiC according to the following reactions.

$$SiC + 2O_2 \ \rightarrow \ SiO_2 + CO_2$$
$$SiC + 3/2O_2 \ \rightarrow \ SiO_2 + CO$$
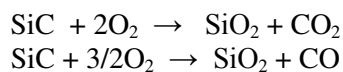
H Ribes [44] has examined by Auger spectroscopy (JEOL JAMP30) the thickness of the oxide layer formed on the SiC particles , indicating a 50 nm-thick SiO_2 layer surrounded the SiC particles. From several measurements performed on various particles, Costello and Tressler [45] demonstrated that the average value of the thickness of oxide film was evaluated as 50 nm. The oxide layer could not only be effective for prohibiting at a direct contact between SiC and the matrix but could also supply the SiO_2 required to react with Mg to form MgO or Mg_2Si. Also, SiC may react

with Mg or Al to form Mg_2Si or Al_4C_3. The formation of MgO or Mg_2Si or Al_4C_3 is possible by any of the following reactions.

The free energy values for reactions 1, 2 and 3 take place at 670 °C [46], but reaction 4 is at 700 °C [47].

$2Mg + SiO_2 \rightarrow 2MgO + Si$ $\qquad \Delta G = -318$ KJ/mol $\qquad$ (1)

$4Mg + SiO2 \rightarrow Mg_2Si + 2MgO$ $\qquad \Delta G = -397$ KJ/mol $\qquad$ (2)

$2Mg + SiC \rightarrow Mg_2Si + C$ $\qquad \Delta G = -15$ KJ/mol $\qquad$ (3)

$4Al + 3SiC \rightarrow Al_4C_3 + 3Si$ $\qquad \Delta G = -125$ KJ/mol $\qquad$ (4)

Thermodynamic calculations performed for this composite system have made it possible to determine the likelihood of the reaction of SiC with alloying elements to form carbides or oxides and silicides. Detailed identification of compounds will require the use of scanning electron microscopy.
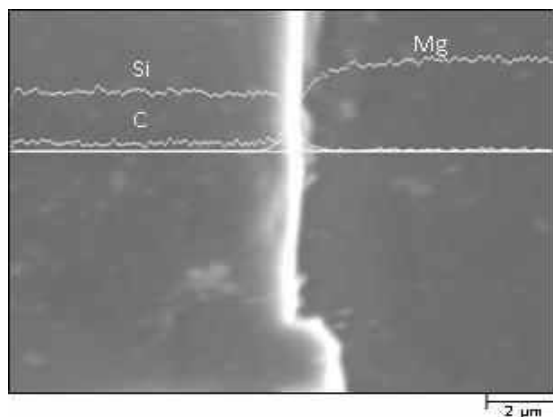

Fig. 3: SEM view of SiC boundary in AZ91 and distribution of concentration of Si, C, and Mg at SiC/AZ91 interface.

Figure 3 illustrates the SEM micrograph for intimate bonding between discontinuous SiC particles and the Mg matrix, which completely surrounds the particle with an intermediate interactions zone. The interfacial integrity between SiC-particulate and the matrix shows good stability and also no debonded at the interface was observed. Elemental scan lines at the interface indicate that there is no Si, C or Mg at the interface, Fig.3, suggesting that there is no significant diffusion of these elements from the matrix into SiC surface. This indicates that no reaction occurred between the sintered SiC and the molten magnesium alloy AZ91 (Mg9AlZn1). Therefore, Mg_2Si, Al_4C_3 and MgO cloud not be detected as a result of the previous reactions. Figure 4-a shows the interface between SiC and magnesium matrix alloy AZ91 and it reflects the angular shape of the particles. On the other hand, the Si, Mg, and Al mapping showed that there no Si, Mg, or Al precipitated in the vicinity of the interface, (Fig. 4b-e). Therefore, there appears to be no observable segregation of Al, Mg, or Si in the interface (.i.e. no reaction takes place during stir casting between the oxidized SiC particles and the molten magnesium alloy AZ91). This is a consequence of the low temperatures (700-725 °C) and the holding time (max. 2 min.) in the liquid state of the AZ91 alloy, possible with this process. By using short times in the process of mixing the suspended liquid composite, the oxide film may remain on the particles,

forming an intermediary layer at the component interface. These results agree with
the results in the literature [48]. Higher temperature normally accelerates interfacial
reactions, as governed by the Arrhenius law. Prolonged holding time at 700 °C of the
material may lead to an extensive chemical reaction between oxidized SiC and AZ91.
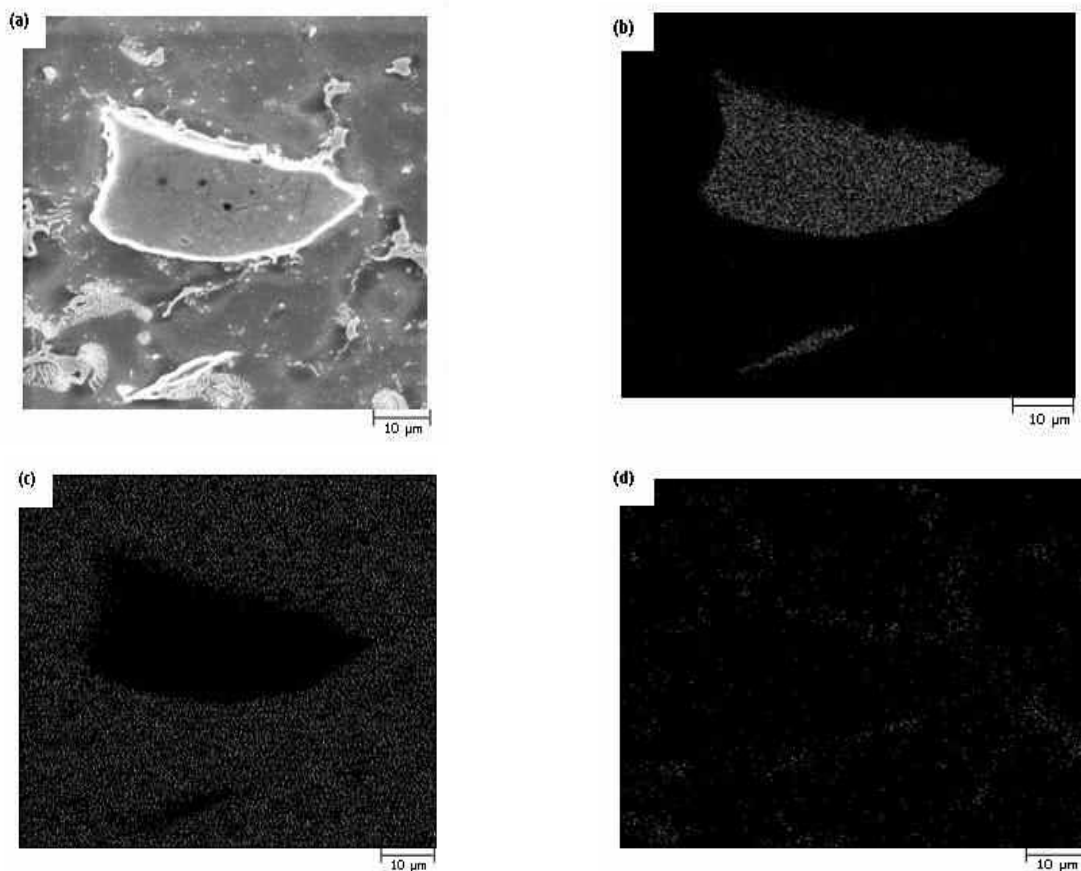

Fig.4: Typical appearance of interaction zone, (a) SEM view of SiC particle and its
surrounding in an (AZ91) alloy and (b), (c), (d) X-ray mapping showing the
distribution of alloying elements of Si, Mg and Al around SiC particle respectively.

A comparison study of the interfacial reactions in pure magnesium and AZ91
alloy based composites reinforced with SiC particles has evinced the effect of a matrix
alloy composition on the particle/matrix interfacial phenomena. In the pure Mg based
composite, SiC particles were stable and no reaction products were found at the
interface [49]. K. N. Braszczy [48] examined by transmission electron microscopy
(TEM) the interfacial reaction between SiC and alloy AZ91 type that contains chiefly
aluminum with zinc and manganese. The results show that no reaction takes place
between the SiC and the molten magnesium alloy. Although aluminum is able to react
with SiC forming very unfavorable carbide Al_4C_3, it does not exhibit this tendency in
magnesium alloys. AZ91 matrix composite showed a massive formation of ternary
intermetallic compound of $Mg_xZn_yAl_z$ at the SiC interface, (Fig. 5). Elements of Mg,

Zn and Al identified as Mg_{17} (Al Zn) $_{12}$ was analyzed by EDXA as shown in Fig. 5-b. It was reported that Zn was mainly distributed in the β-$Mg_{17}Al_{12}$ phase by substituting for a part of aluminum, with a form of Mg17(Al,Zn)12 or $Mg_{17}Al_{11.5}$ Zn $_{0.5}$ at temperatures below 437 °C [50]. In the results of F. Wu. et. al. [51] it has been shown that $Mg_{17}Al_{12}$ precipitate is formed discontinuously at the fibre/matrix interface, even if only a small amount of Al is present in the magnesium matrix [51].

To obtain composite materials with the desired microstructure and properties, the interfacial reaction should be controlled through selecting an appropriate matrix alloy, conducting an appropriate surface treatment of the reinforcement, and correctly controlling the process parameters.
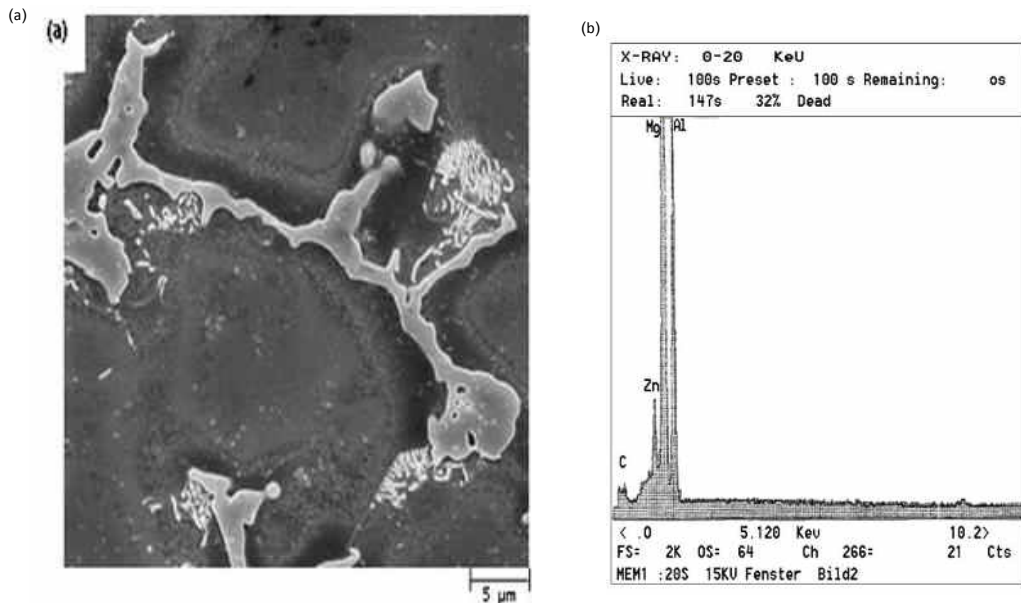

Fig. 5 : SEM view of interfacial formation of Mg_{17} (Al Zn)$_{12}$ in an AZ91 alloy (a) and its EDX-analysis (b)

MECHANICAL BEHAVIOUR

The results of hardness measurements revealed that an increase in the volume fraction percentage of SiC particulates produced an increase in the hardness value of the metallic matrix, (Fig. 6). The hardness of composites increases from HV10 = 65.3 to 124.8 with increasing the volume fraction of SiC - particles from 0 % to 20 %. This can be attributed to increasing the presence of a harder SiC particulate in the matrix, which surrounded by a softer and relatively tough matrix and can contribute to wear resistance by favoring more plastic behavior.

Tensile properties of AZ91 SiC-particulate composite are shown in Fig.7 and Table 4. The results reveal that the addition of SiC–particulates to the base alloy increases the 0.2% yielding and ultimate tensile strength, but considerably reduces the strain to failure.

Fig. 6: Variation of the hardness value of
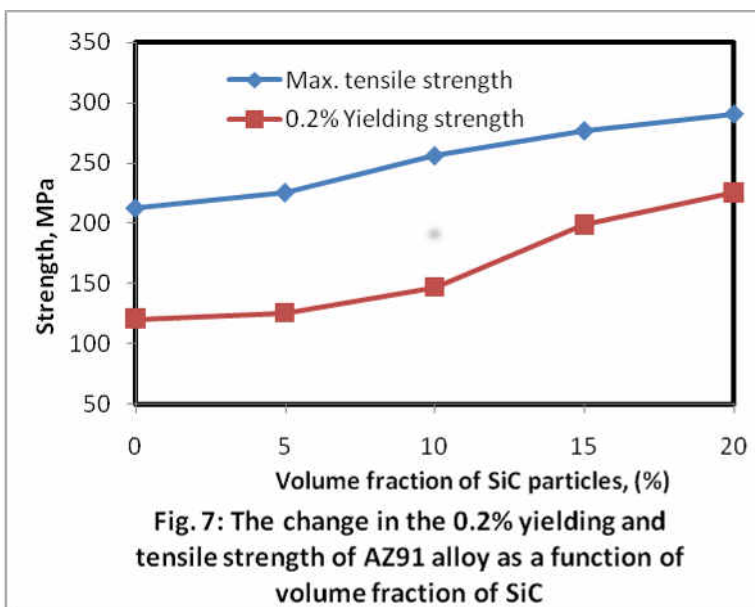AZ91 with diffrent precentage of SiC-particles
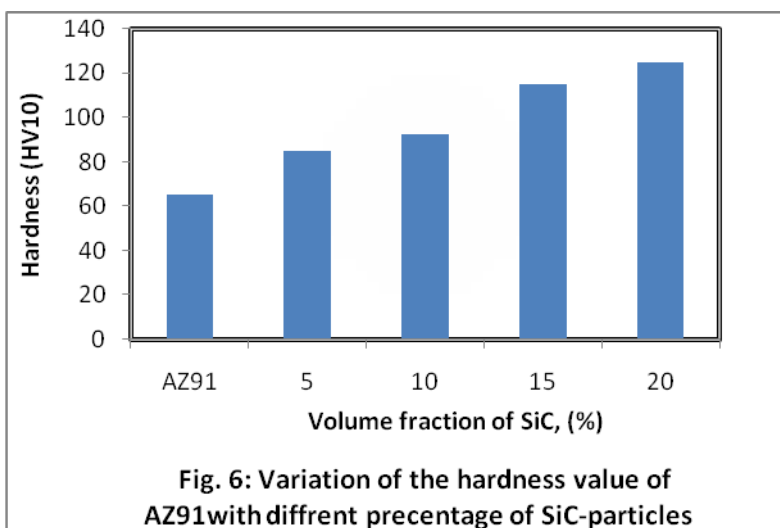
        As expected the volume fraction of the particulate has a significant effect on the yield and maximum tensile strength. In comparison with AZ91 alloy, the values of tensile strength and yielding strength of 20 vol. % SiC (50 µm) composite increased by about 36.9% and 86.8 % respectively. However, it is also evident that the strain to failure significantly decreases as the volume fraction of the particulate increases, (Fig. 8).   This means that the matrix probably does not have sufficient internal ductility to redistribute the very high localized internal stresses. It is also obvious that the resistance to the dislocation motion of the hard particles reduces the ductility of the composite materials. Moreover, the presence of the SiC-particulates impedes the plastic flow of the matrix, initiating failure at low strains by the formation of cavities in the vicinity of the particulate. Recent work has demonstrated that composite failures (is) associated with particle cracking and void formation in the matrix within clusters of particles [28]. In addition, the failure associated with particles clusters can be attributed to the higher   stress triaxiality generated in these regions. In related studies conducted on as cast AZ91-SiC composites [17] , it was proposed that the presence of SiC particulate lead to localized damage, such as interface debonding and particulate breakage, resulting in lower elongation value. In fact it has pointed out by a number of authors [52, 53] that high matrix strength results in more particle – related damages, while the material with lower matrix strength exhibited a smaller percentage of reinforcement damage at the fracture strain. Often the ductility must be considered in the design, especially when it is low as the material becomes less forgiving and already at small deformation, cracks can appear. Generally, the strengthening effects, which can occur in the composite material, can be divided into direct and indirect strengthening. Direct strengthening mechanism includes the ability to transfere stress from the matrix to the stronger reinforcing particles [28]. This, in turn, depends on achieving a strong interfacial bond between the matrix and the reinforcement. Therefore, If the bonding between the matrix and the reinforcement is strong enough, the applied stress can be transferred from the soft matrix to the hard particle phases. If the interfacial bond is weak, the interface will fail before any effective stress transfer to the particle can occur, and no strengthening is achieved [54]. Indirect strengthening
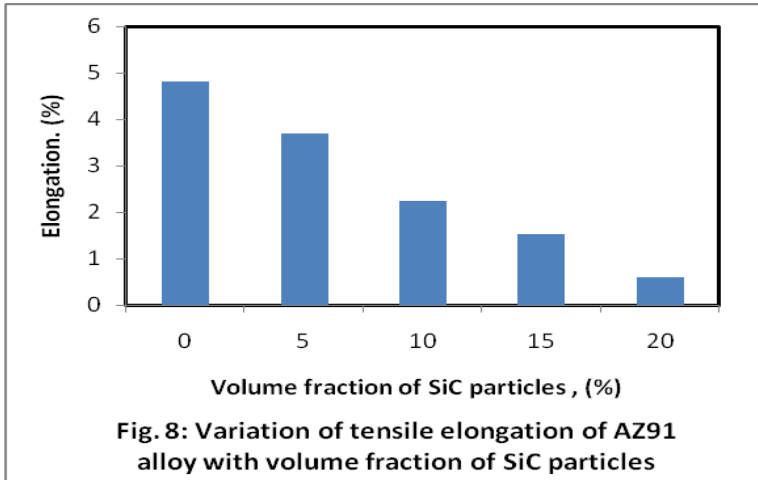
mechanisms include increased initial work hardening of the composite [55], grain refinement of the matrix alloy caused by the reinforcement phases, and the generation of a high dislocation density in the matrix as a result of difference in the coefficient of thermal expansion (CTE) of the matrix alloy AZ91(26×10^{-6} K^{-1}) and the SiC reinforcement (4.7×10^{-6} K^{-1}) which are the key strengthening mechanisms in magnesium composites [56]. Also the strain mismatch between the matrix and the reinforcement usually generates a higher density of dislocation in the matrix around the reinforcement, thus strengthening the material.

Different models describing strengthening mechanisms in two –phase alloys have been reviewed by Midling [57]. Virtually, all models predict that the yield strength of the composites is increased and coefficient of thermal expansion decreased with increasing volume fraction compared (to) that of unreinforced alloys [58].

Table 4: Mechanical properties at room temperature of AZ91 andAZ91/SiC

Alloy	Hardness (HV10)	Tensile Strength, (MPa)	0.2% Yielding Strength, (MPa)	Elongation, (%)
AZ91- 0% SiC	65.3	212.6	120.5	4.8
AZ91- 5% SiC	85.2	225.4	125.5	3.7
AZ91- 10% SiC	92.6	255.8	146.8	2.24
AZ91- 15% SiC	115.4	276.3	198.3	1.52
AZ91- 20% SiC	124.8	290.2	225.2	0.6

Fig. 7: The change in the 0.2% yielding and tensile strength of AZ91 alloy as a function of volume fraction of SiC

Fig. 8: Variation of tensile elongation of AZ91 alloy with volume fraction of SiC particles

FRACTOGRAPHY

The rupture surface of AZ91/SiC composite in the as cast condition is shown in Fig. 9. The general form of the fracture surface was found to be the same for all volume fractions. Examination of the fracture surface in the SEM shows that the SiC/matrix interface is made up of network of very fine dimples (arrowed). The extent of SiC particles cracking shows the strong bonding between the SiC and AZ91. However, regions rich in reinforcing SiCp revealed to features of locally brittle failure. The tensile fracture surface revealed (to) a combination of shallow dimples of varying sizes and failure of the reinforcing SiCp through cracking. However, close examination of the exposed silicon carbide also shows the existence of residual magnesium on the particle. The results suggested that the SiC/AZ91 interfacial bonding is reasonably good. The apparent SiC/AZ91 matrix separation is probably simple due to the fact that there exists high stress concentration at the interface during deformation. Thus, the interfacial bonding is not the main factor causing the interfacial separation. Finally, composite fracture can take place by all three modes simultaneously, i.e. within the magnesium matrix, along the matrix /reinforcement particle interface, as well as by shearing of the reinforcement particles, depending upon the strength of the particle/matrix cohesion.
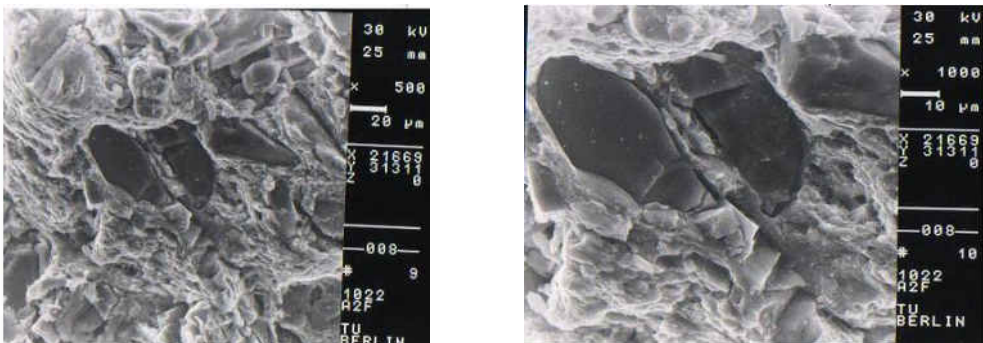

Fig. 9 : SEM micrograph of fracture surface of AZ91 alloy reinforced with SiC particles

CONCLUSION

The major conclusions from the present work are as follows:

1- Stir casting processes were used to produce SiC particle reinforced magnesium matrix composites. This process leads to a good dispersion and complete wetting of SiC particles in the molten magnesium alloy.

2- The presence of SiC particles in the matrix alloy AZ91 affects the matrix microstructure. The grain structure of the composites is refined compared to the matrix alloy.

3- Results of EDX-analysis indicate the absence of oxide, silcidie, and carbide phase at SiC-AZ91 interface, while a massive formation of ternary intermetallic compound of $Mg_{17}(ZnAl)_{12}$ at the SiC interface was detected.

4- The addition of SiC particles to the matrix increases the hardness value, the tensile strength and 0.2 yielding strength compared with the unreinforced alloy (AZ91). However the elongation is lower in the composites, decreasing with increasing the volume fraction of SiC particulate.

5- Examinations of the fracture surface show that the extent of fine dimples, matrix cracking and particle cracking and the SiC/Al interfacial bonding is reasonably good.

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التركيب والخواص الميكانيكية لمتراكبة سبيكة الماغنيسيوم و كربيد السيليكون المحضرة بطريقة التقليب والصب

تم تحضير متراكبة مكونة من سبيكة الماغنيسيوم ونسب مختلفة من كربيد السيليكون وتطبيق طريقة التقليب والصب في وجود جو من غاز خامل، وأدت هذه الطريقة للتحضير الي ضمان وجود بلل بين حبيبات كربيد السيليكون (المادة المدعمة) وبين سبيكة الماغنيسيوم (الرحم). وتم دراسة التركيب الحبيبي للمتراكبة والمتمثل في توزيع حبيبات كربيد السيليكون وعلاقة الترابط بين سبيكة الماغنيسيوم (الرحم) وحبيبات كربيد السيليكون (المدعم) عن طريق الميكروسكوب الاليكتروني والاشعة السينية . وبينت النتائج التباين بين التركيب الحبيبي لسبيكة الماغنيسيوم قبل وبعد اضافة كربيد السيليكون وبينت النتائج توزيع حبيبات كربيد السيليكون داخل سبيكة الماغنيسيوم وعدم وجود تفاعل بين حبيبات كربيد السيليكون المؤكسد وسبيكة الماغنيسيوم، وتبين ان سبيكة الماغيسيوم كونت مركب ثلاثي من الماعنيسيوم والزنك والالومنيوم علي سطح كربيد السيليكون.

وبينت النتائج تحسن في الخواص الميكانيكية للمتراكبة المحضرة نتيجة تدعيم سبيكة الماغنيسيوم بحبيبات كربيد السيليكون، كما بينت وجود زيادة في الصلادة والصلابة لهذه المتراكبة المنتجة.

ولوحظ ان الزيادة في الصلابة للمتراكبة المنتجة كانت نتيجة زيادة صلابة سبيكة الماغنيسيوم (الرحم) نتيجة وجود حبيات كربيد السيليكون أو أن الصلابة نتيجة الزيادة في كثافة اسطح التشوة وصغر حجم حبيبات سبيكة الماغنيسيوم (الرحم)، كما تم اثبات أن العلاقة بين الانفعال والانهيار يقل بزيادة كمية حبيبات كربيد السيليكون.

وتبين من دراسة مقطع سطح الكسر وجود قوة ترابط بين حبيبات كربيد السيليكون وسبيكة الماغنيسيوم (الرحم) ووجود بروزات عبارة عن حبيبات دقيقة وأن الكسر تم في حبيبات سبيكة الماغنسيوم (الرحم) وحبيبات كربيد السيليكون (المدعم).