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FLEXURAL STRENGTH AND FRACTURE STUDIES OF AL-SI/SIC$_P$ COMPOSITES

Yusof Abdullah[1], Abdul Razak Daud[1], Roslinda Shamsudin[1] and Mohd b. Harun[2]

[1]School of Applied Physics, Faculty of Science and Technology,
Universiti Kebangsaan Malaysia (UKM), Bangi 43600, Selangor, Malaysia.
[2]Malaysian Nuclear Agency, Bangi, Kajang 43000, Selangor, Malaysia.
Email: ya7747@hotmail.com

ABSTRACT

Ambient temperature mechanical properties of Al-7Si/SiC$_p$ composite prepared by stir casting process were studied. Microstructure, flexural, hardness and tensile tests of Al-7Si alloy reinforced with 10 and 20 wt.% SiC$_p$ were investigated. Samples were characterized by scanning electron microscopy (SEM) while three point bend tests were performed to study the flexural strength of the composites. Hardness and bending tests indicated that reinforcing the Al-Si matrix with SiCp improved the hardness and flexural strength. SiC$_p$ improved the flexural strength mainly by acting as barriers to cracks whereby cracks normally initiated at the debonded particle/matrix interface. The porosity content increased with the increasing number SiC$_p$ present in the matrix. However, it was believed that the high flexural strength and hardness were due to the good bonding properties of the matrix/particle interface. Fine SiC$_p$ was play a role to increase the surface area and promotes better strength of the composites.

Keywords: Al-Si/SiC$_p$ composite, flexural strength, fracture, bonding.

INTRODUCTION

The global increasing fuel price has led to a renewed urgency to address the issue of weight reduction in the automotive and aerospace industries. Since aluminium metal matrix composites (MMC) are being considered as new advanced materials due to light weight, high strength, low thermal expansion coefficient, good wear resistance and good manufacturability (Soon and Gupta, 2001; Hwu *et al.*, 1996), aluminium metal matrix composites have a good material for use in these sectors. Currently, the development of this material was successfully be used in automotive components and has increased especially for manufacture cylinder blocks, cylinder heads, pistons, piston rings and brake disc (Shorowordi *et al.,* 2003; Tekman *et al.*, 2003).

The properties of aluminium metal matrix composite mostly depend on the processing method in which capable to produce good properties to comply the industry need. Among others in the aluminium metal matrix composites, Al-Si/SiC$_p$ composites are the most candidates to be developed. Al-Si/SiC$_p$ composites can be more easily produced by the melt stir casting technique due to its good cast ability and relatively inexpensive (Shorowordi *et al.,* 2003). Tekmen et al., (2003) reported that the melt stirring method is economical, easy to apply and convenient for mass production. However, the problem encounter for this technique was low wettability and particle settling. To improve wettability and particle homogeneity during casting, various method have been used including coating or oxidizing the reinforcement particles, adding some surface active elements (magnesium and lithium) into the matrix, increasing the liquid temperature and stirring of molten matrix alloy for an adequate time period during incorporation (Tekman *et al.*, 2003).

Generally, metal matrix composites exhibit internal residual stress arising from various sources, such as plastically induced effects and thermal property mismatches between the matrix and reinforcement materials. These stress concentrations has lead to failure. It was reported that fatigue cracks initiated from porosity which in stressed region and high strain region of void at reinforcement-matrix interface (Fizpatrick *et al.*, 2002). One of the research interests in metal matrix composites is to study how the particle reinforcement affects the failure mechanisms and hence controls the fracture of metal matrix composites. Therefore it's important to study the strains and stresses caused by process and the plastic deformation of composites. The objective of this investigation was to determine the behaviour of the flexural, hardness, tensile and fracture properties of Al-7Si/SiC$_p$ composites. The microstructure and mechanical properties of the T4 heat-treated composites are compared with unreinforced materials.

EXPERIMENTAL WORKS

Commercial Al-7Si alloy (AC4C according to JIS specification) was used as the matrix and SiC particles with an average particle size of 3.0 μm were used as the reinforcement material. The chemical composition of the Al-7Si alloy was shown in Table 1.The stir casting method was used for the production of composite billets. Furnace was used to melt and hold the Al-7Si matrix alloy at 750 ºC. During melting, 1.5 wt% Mg was added to improve the wettability of the matrix and improving the interfacial bonding between the SiC$_p$ and the Al-Si matrix during casting (Henriksen and Johnsen, 1990). Heat treatment of SiC$_p$ was done at the temperature of 1000 ºC for 2 hours and then SiC$_p$ was added into Al-Si melt manually, stirring at 300 rpm and when completed the stirring process continued at 900 rpm for 8 minutes and then poured into a permanent steel mould to form ingot. Similar ingot was also produced for Al-7Si alloy. The ingots of the composites and unreinforced Al-7Si were subjected to a solution heat treatment (T4) for 5 hours at 530 ºC.

Table 1: The starting composition of Al-7Si alloy matrix material (wt %).

Si	Mg	Cu	Fe	Zn	Ti	Ni	Bi	V	Al
7.33	0.24	0.15	0.27	0.01	0.1	0.02	0.03	0.01	Balance

For three point bending test, A Universal Testing Machine was used to determine the flexure strength of the composites. Before testing, the specimens were polished up to 3 μm diamond paste. Six specimens with size of 5x5x60 mm^3 were performed on testing with a loading rate of 0.5 mm/min. The average flexural strength of composites and their standard error were also calculated. The distribution of the SiC_p in Al-Si/SiC_p and the fracture path of specimens were examined by using a JEOL scanning electron microscope (SEM). Vickers hardness tests were carried out using a load of 10 kg to measure the hardness of the composites. Each hardness value represents the average value of five such measurements. The tensile testing on all composites was also tested by using a 30 KN Instron 5567 at room temperature with a cross-head speed of 0.5 mm/min.

RESULTS AND DISCUSSION

The three point bend tests were performed to reveal fracture behaviour of the Al-Si/SiC_p composites with increasing of SiC_p. Fig. 1(a) and (b) shows the effect of SiC_p on flexural strength of composites with containing of 0-20% SiC_p as reinforcement materials.

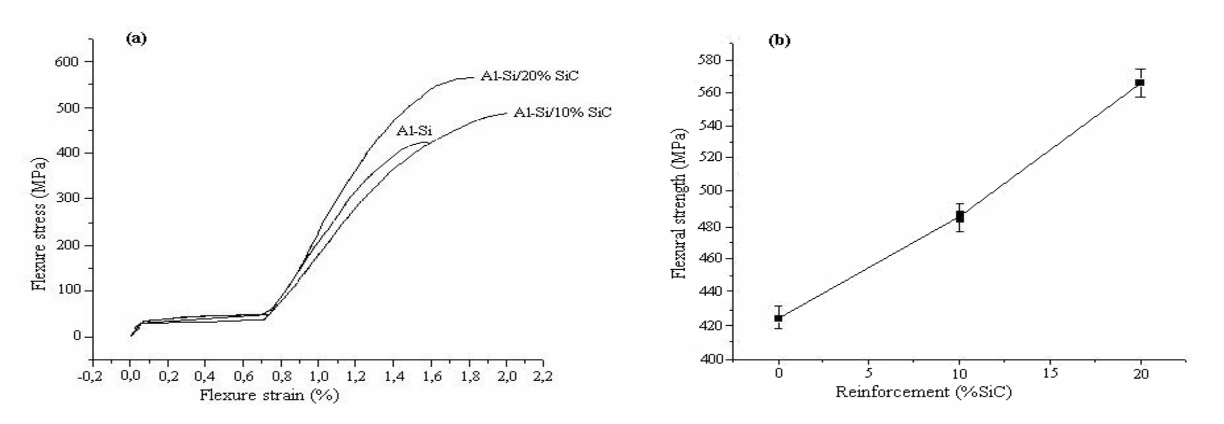

Fig. 1: (a) The flexural behaviour of Al-Si/SiC_p composites and (b) Flexural strength versus SiC_p content of the composites.

The results were clearly shown that the flexural strength increased with increasing SiC_p composition. Thus, fine SiC_p was contributed an increase of flexural strength of Al-Si/SiC_p composites. The flexural strength of Al-7Si and Al-7Si/10% SiC was 425 MPa and 485 MPa respectively. The composites showed enough ductility to attain more strength by addition of 20 wt% SiC_p in the composites caused an increase in flexural strength to 566 MPa.

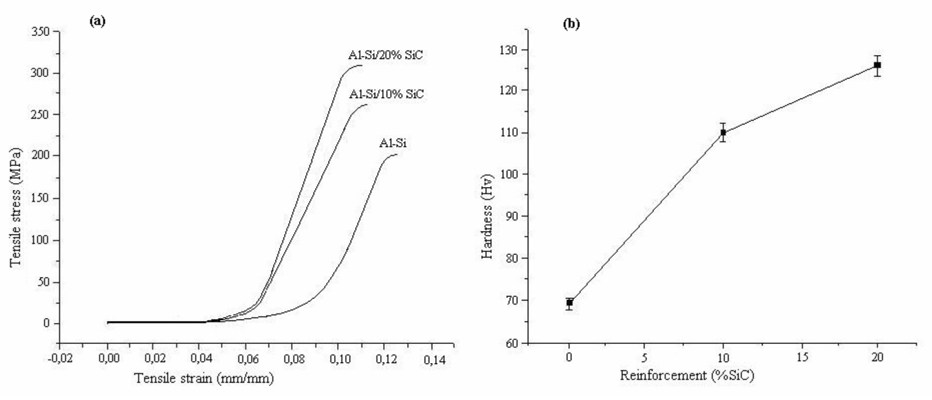

Fig 2: Mechanical properties of composites; (a) Tensile properties of Al-Si/SiC_p composites (b) Effect of SiC_p on hardness of Al-Si/SiC_p composites.

Fig. 2 shows the tensile strength and hardness of the composites. The average hardness of composites with different SiC_p percentage ranges from 110 to 127 MPa. Again, the hardness of composites increases with increasing SiC_p content. In this case, Zhang *et al.* (2004) believed that the fracture stress of silicon carbide particles, with fine size can prevent the quick expansion of cracks through the composite and limit the deformation of the composite. These will improve the hardness and flexural strength of the composites. Meanwhile, the tensile strength of composites was higher than that of Al-7Si alloy. The ultimate tensile strength of Al-Si, Al-Si/10% SiC and Al-Si/20% SiC was 203 MPa, 259 MPa and 308 MPa respectively.

The microstructures of composites are illustrated in Fig. 3. Some agglomeration was observed and the porosity was seen at SiC_p especially in an agglomeration of particles. In general, the microstructure was characterized by agglomeration of SiC_p grains. However, the distribution of SiC_p was inhomogeneous. Moreover, the fractography indicated that SiC_p in the composites result in prone to cluster together. Although the in homogeneously distribution of SiC_p in the composites, these conditions leads to improvement in the flexural strength, tensile strength and hardness of Al-Si/SiC composites by increasing the SiC_p content. These results could be effect of good interface bonding between SiC_p and matrix as shown in Fig. 4.

Fracture of SiC_p occurred in large particle or in regions with clusters. Fracture of the reinforcing particles depends on the local stress acting on the particle. The large mismatch in the elastic modulus between the reinforcing particles and the metal matrix generates a constrained deformation in the matrix and a consequent concentration of stresses near the reinforcing particles. These stresses can determine cracking of the particles, fracture of the matrix and interfacial decohesion.

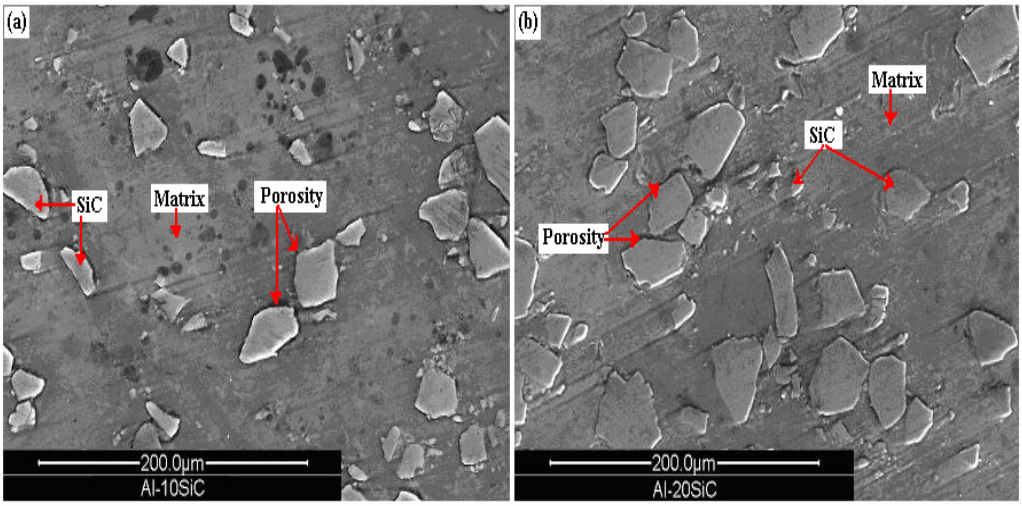

Fig. 3: Fractographs of crack initiation at the fractured SiC particles for Al-Si/10% SiC_p and Al-Si/20% SiC_p respectively.

Fracture of the particle is greater in region with large particles and clusters where there is a concentration of stresses and where the short interparticle distance facilitates linkage between voids and cracks in the particle (Srivatsan and Al-Hajri, 2002; Hong *et al.*, 2003)]. The larger particles generate high load transfer from the plastically deforming aluminium matrix and the elastically deforming particle, which can cause crack. On the other hand, the smaller particle usually does not crack, but because of the strain differences between matrix and particle the matrix can fail by decohesion.

Razaghian *et al.* (1998) investigated the fracture behaviour of SiC particulate reinforced 7075 aluminium alloy under uniaxial tensile loading. They indicated particle fracture was the main damage mechanism prior to final fracture at room temperature. Large particles and regions of clustered particles were found to be the locations prone to damage in the composites at room temperature (Razaghian *et al.*, 1998). Particle fracture was observed at clusters of particles as well as in large particles and can be attributed to the high local stress in these regions as shown in Figs. 3 and 5. It was believed that there are two mechanisms of crack initiation in Al-Si/SiC composites. First, cracks may initiate at the interface between the Al-Si matrix and SiC_p. The second crack initiation mechanism was due to cracking of the large SiC_p. As the SiCp size increases the tendency for particle fracture increases. Thus, larger SiC_p will have a higher probability of faults and can fracture under stress.

The using of 3.0 µm SiC_p as reinforcement materials has contributed to the increase in composites strength. The spacing between particles is reduced when the particle size is fine. The fine particles will exert more constraint on grain growth during cooling and more restriction on plastic flow during deformation which contributes to the increase in strength.

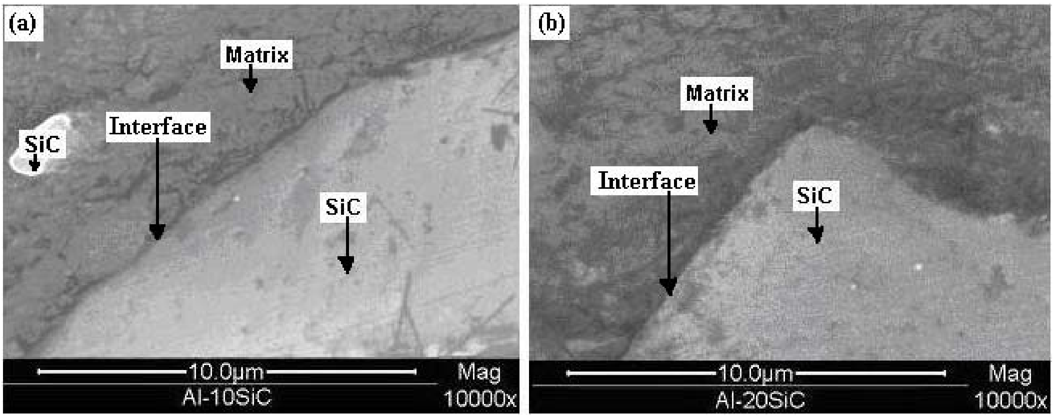

Fig. 4: Micrograph showing good cohesion at the interface between the Al-Si matrix and SiC particles for Al-Si/10% SiC_p and Al-Si/20% SiC_p respectively.

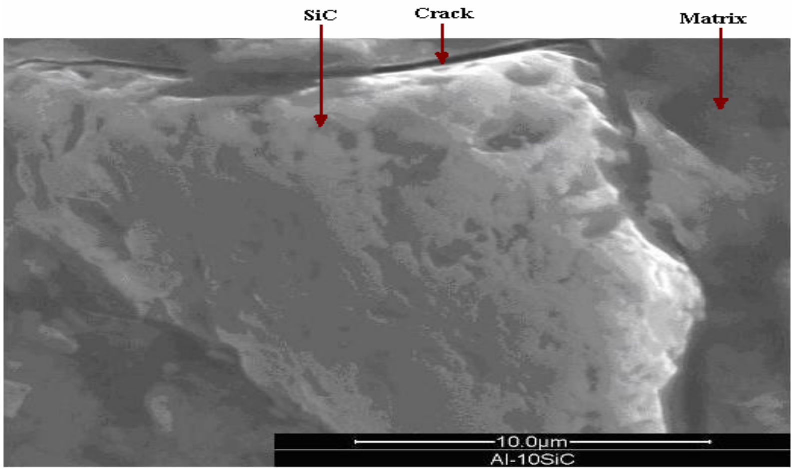

Fig. 5 Crack initiation at SiCp due to debonding at the interface

Voids nucleation was concentrated at the interface with the particles, where there is a high matrix strain and at the clusters where high local triaxial stresses are present. It was believed that under higher stress levels the interfacial bond strength was not sufficient and cracks initiated at the debonded interface. The initiation of cracks was also increase due to porosity and the fracture of coarse SiC_p.

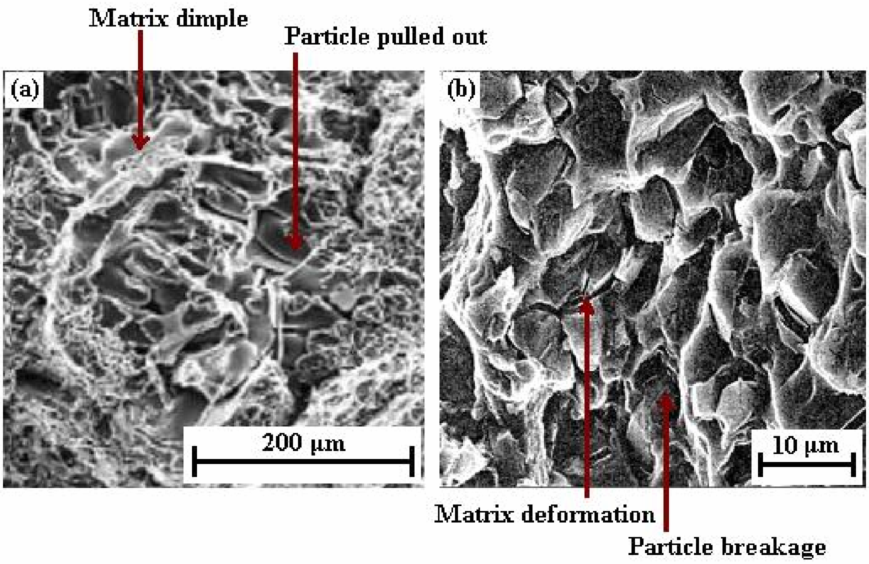

Fig. 6: Micrographs of the fracture surfaces show (a) matrix dimples and pullout of the SiCp (b) broken particle and matrix failure or deformation.

A typical fracture morphology in Fig. 6 shows that the bending fracture mainly consists of broken SiC_p and matrix dimples, a small amount of debonded particle/matrix surface or decohesion, indicating a fracture occur dominated by SiC_p breakage followed by matrix failure, possibly due to the constraint of the SiC_p to the matrix plastic deformation. As shown in Fig. 6(a), the fracture surface exhibited many dimples in the matrix regions around the particles and in the voids formed due to the particle decohesion. Large voids and dimples are caused by fracture and decohesion of particles while the ductile dimples can be attributed to the constraints in plastic flow of the aluminium matrix or to the reduction of strains induced by the particle cracking which lead to the formation of tear ridges. Fracture in metal matrix composite is control by three main mechanisms; (i) interfacial decohesion, (ii) fracture of reinforcing particles and (iii) void nucleation and growth (Ceschini *et al.,* 2006). The interfacial decohesion is often due to the presence of undesired interfacial reaction products, such as Al_4C_3 (Taya and Aresenault, 1989). These intermetallic materials promoting voids nucleation at the particles interface, interfacial decohesion and failure of the particles.

During flexure test SiC_p acts as crack stoppers or points deflecting the growth plane of the main and secondary cracks. This means that SiC_p changed the growth plane of the cracks. Beside that, there are different types of stress field in Al-Si/SiC composites. Macrostresses are considered to be continuous across the phases or grains in the materials. The mechanical properties of composites were also influenced by the existence of interfacial reaction. For interfacial reaction, the most reaction occurred since Al-7Si alloy contains Mg and due to the presence of SiO_2 phase on the SiC_p is the formation of MgO or $MgAl_2O_4$. The possible reactions are (Peng *et al.,* 2004);

$$SiO_2 + 2Mg \rightarrow 2MgO + Si \qquad (1)$$

$$Mg + 2Al + 2O_2 \rightarrow MgAl_2O_4 \qquad (2)$$

The presence of SiO_2 owes of heat treatment of SiC_p prior casting process. The interfacial reaction results in a higher viscosity of molten Al-Si alloy, to which slow moving during casting process, thus inhomogeneous distribution of SiC_p is attributed. Generally, as the result of without particles homogeneity properties, the flexural strength could be reduced. However, flexural strength depends on the debonding of SiC-Al interfaces and porosity presence in the composites. Therefore, if debonding of SiC-Al interface and porosity increase, the flexural strength was also decreased. Furthermore, if pores presence in the composites, it could makes the composites weaker. However, it was suggested that the dislocations generated owing to the thermal mismatch between the reinforcement and the matrix during quenching after solution treatment may promote the formation of precipitates (Hwu et al., 1996). Due to the presence of fine reinforcement particles in the composites thus more particles surface area and less particle spacing so that more dislocations are generated after quenching. The fine particle sizes will provide more interface area which serves as the nucleation sites of grain formation. Fine particles will exert more restriction on plastic flow during deformation that will contribute to the increase in strength.

CONCLUSION

Good quality of Al-7Si/SiC_p composites containing up to 20% SiC_p can be produced using stir casting process. The flexural strength, hardness and tensile strength of Al-Si/20% SiC_p were higher than that of Al-Si/10% SiC_p. It can be concluded that the flexure strength, hardness and tensile of the composites increased as the SiC_p reinforcement was increased due to using fine SiC_p and good bonding at particle-matrix interface. SiC_p improved the flexural strength mainly by acting as barriers to cracks in which cracks normally initiated at the debonded particle/matrix interface. Good interface particle/matrix cohesion properties were achieved due to the addition of Mg alloy during processing and heat treatment of SiC_p prior mixing in such a way improve the wettability behaviour of the particle/matrix interface.

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