1. Introduction

Metal matrix composites (MMCs) are attractive for numerous applications in several fields as aerospace, automotive and sports industries since they exhibit excellent mechanical properties combined with relatively low cost [1]. Al-based MMCs are well known for their high specific strength, hardness and wear resistance [2]. Among various discontinuous dispersoids, SiCp have been found to have excellent compatibility with the aluminium matrix, making their composites reference materials [3]. The enhancement in mechanical properties, regarding the unreinforced alloy, is a result of the ability of Al-Si/SiCp composites to withstand high tensile and compressive stresses by the transfer and distribution of the applied load from the ductile matrix to the reinforcement phase [4]. The improvement in properties also is dependent on mutually interactive influences of the intrinsic properties of the composite constituents, and the size, shape, orientation, volume fraction and distribution of the reinforcing phase in the metal-matrix. However, there are some disadvantages in properties upgrading such as inadequate fracture toughness, limited damage tolerance and poor tensile ductility compared to the unreinforced counterpart. Between available reinforcement morphologies, particulate reinforced composites as Al/SiCp exhibit improved physical and mechanical properties since they are generally isotropic and they can be processed through conventional methods used for metals. The presence of the reinforcement has been found to result in smaller grain size, accelerated aging and in precipitate distribution, morphology and size in the metal-matrix. A high density of dislocations arises near the interfaces between the reinforcement and the metal-matrix as a result of the mismatch in the coefficient of thermal expansion between the SiC particle and the aluminium alloy. The contraction of the matrix during cooling induces plastic deformation, causing an increase in the density of dislocations [5, 6].

Many studies have dealt with fracture of metal-matrix composites with reinforced ceramic particles, between them Razaghian et al. [7] investigated the fracture behaviour of a SiC particulate reinforced 7075 aluminium alloy under uniaxial tensile loading in the temperature range 25 – 400 °C. The ductility of the composite was found to be much lower than that of the monolithic alloy at all temperatures, but both materials exhibited similar strength levels above 300 °C. Particle fracture was the main damage mechanism prior to final fracture at room temperature, while interface debonding together with interparticle voids were dominant features in fracture at high temperature. Large particles and regions of clustered particles were found to be the locations prone to damage in the composite at both room and high temperatures. At room temperature, particle fracture was observed at clusters of particles as well as in large particles, whereas at high temperature voids nucleated in the matrix closely adjacent to particles and at particle ends. This can be attributed to the high local stress in these regions and the high probability of flaws in large particles.

Several manufacture methods for Al-Si/SiCp fabrication are currently in use. In liquid-phase processing routes, which are the most widely used, the solidification synthesis of metal matrix composites involves producing a melt of the selected matrix followed by the introduction of a reinforcement material into the melt, obtaining a suitable dispersion. Casting composite materials, also called compocasting, is a promising route which
has advantages as simplicity, low cost, flexibility and applicability to large quantity production. In preparing Al-Si/SiCp composites by the casting route, there are numerous factors that need considerable attention, including the difficulty of getting a uniform distribution of the reinforcement material, wettability between the two main substances, porosity in the cast metal matrix composites, and chemical reactions between the reinforcement material and the matrix alloy. In order to achieve the optimum properties of the metal matrix composite, the distribution of the reinforcement material in the matrix must be uniform, and the wettability or bonding between them should be optimized [8].

Processing Al-Si/SiCp composites has some difficulties related to the interaction between the matrix and the reinforcement. The low wetting of SiC by molten aluminium and the high reactivity between both constituents, especially at high temperatures (>752 °C), can be considered as limiting factors which make the manufacture of Al/SiCp composites difficult [9]. At low interaction temperatures, wetting of SiC by molten aluminium is small and SiC particles are segregated to the interdendritic spaces during solidification. Raising interaction temperature encourages the reactivity strongly leading to an increase in wettability between aluminium and SiCp, and therefore, a better reinforcement distribution [10]. At high interaction temperatures, direct reaction between SiC and Al occurs to form hexagonal platelet-shaped Al₄C₃ crystals and free Si according to:

\[
4 \text{Al} + 3 \text{SiC} \rightarrow \text{Al}_4\text{C}_3 + 3 \text{Si}
\]  

The interfacial reaction is known to have undesirable effects on the overall composites properties: (i) poor ambient resistance due to the unstable nature of Al₄C₃ in presence of humidity, methanol, HCl, etc., (ii) degradation of the reinforcement owing to the formation of Al₄C₃, may cause a decrease in strength and modulus, and (iii) the free Si, formed as a result of the interfacial reaction, produces the Al-Si eutectic during fabrication or heat treatment stage, resulting in unintended mechanical properties of the matrix alloy. For that reason, fabrication of Al-Si/SiCp composites devoid of Al₄C₃ has been one of the major concerns [11].

Three methods are widely used to control reaction (1): (i) addition of a large amount of Si into the Al, normally 10-12%, to displace the reaction to the left; nevertheless this alternative limits the number of alloys that can be used as matrices. (ii) Coating of the reinforcement with metal or ceramics has been successful to some extent both for preventing the detrimental interfacial reaction and enhancing the wetting characteristics. Silica is one of the most successful coatings used. It can be obtained either through direct oxidation or via sol-gel. SiCp can be oxidized in air at temperatures higher than 1100 °C to obtain a partially crystalline nanometric thick silica coating that goes under stress during cooling because of the volume change associated to the transformation. Although this layer protects the reinforcement, its rapid consumption gives place to a preferential etching of SiCp and to the formation of aluminium carbide in the oxidation layer cracks due to their stress state. The sol-gel procedure gives place to a continuous micrometric amorphous silica coating that works out more properly, because of its wider thickness and continuous and free stress structure. Moreover, it is also possible to reach a higher degree of design in its inner structure, by means of heat treatments, which allows controlling its reactivity. (iii) Reducing the interaction temperature during processing [3, 12].

The mechanical properties of composites are generally depending on the interface bonding between matrix and reinforcement. To achieve superior mechanical properties in Al-Si/SiCp composites, it is essential to form interfaces that do not degrade the reinforcement during fabrication and that retain the structural stability both for corrosive environments and at elevated temperatures. The characteristic of the interface bond are affected by factors such as processing temperature, time and method, composition of the matrix and characteristic of the reinforcement surface [12, 13].

In this study, the effect of time and temperature during liquid casting been studied on the degradation of different Al/SiCp composites. Sol-gel silica coatings have been deposited on the SiCp to evaluate its effect in the characteristics of the composite. All the samples have been evaluated by means of a three-point bending test in notched
samples carried out in Scanning Electron Microscope (SEM). Values such as yield and maximum stress, fracture toughness have been determined. The use of SEM images has also allowed determining the main fracture mechanisms that take place during crack propagation in each system [14, 15].

2. Experimental procedure

2.1 Processing

Al/SiCp composites were fabricated by stir casting with two Al-Si alloys, A356 and A380, reinforced with SiCp particles in a 10 % volume fraction with an average particle size of 15.3 µm.

The following methodology was used for Al/SiCp fabrication. A resistance furnace was used for melting and mixing. First, the aluminium alloy was melted at 700 °C and a 1 % of Mg was added for improving SiCp wetting by the aluminium alloy. For homogenisation purposes, 15 minutes of stirring was performed. After that, fluxes were added and manual skimming of the melting surface was carried out. Then SiCp were poured slowly and continuously into the molten metal via a vortex made by mechanical agitation. The impeller was frequently moved vertically within the mixture to ensure a uniform distribution of the SiCp.

At this stage, two different procedures were followed and two kinds of specimens were produced, apart from the own alloy without reinforcement that was used for comparative purposes. A380/SiCp composites were fabricated with an hour of stirring time at 700 °C. In this case, SiCp reinforcement was used in the as-received state and after being coated by SiO₂ using a sol-gel route. The second type of composites evaluated were A356/SiCp composites fabricated with 21 h of stirring time and melting temperatures of 700, 750 and 850 °C. In this case, uncoated particles were used.

2.2 SiCp coating procedure

The sol-gel method was used for coating the SiCp reinforcement with a SiO₂ layer by using the following procedure. First the sol was prepared by mixing TEOS (tetraethyloorthosilicate) with ethanol and then acidulated with drops of diluted HCl. The mixture was then maintained in constant stirring for 2 hours. After that, SiCp were immersed in the mixture and stirred for 2 hours while the formation of the gel takes place. Once coatings were formed on the surface of SiC particles, they were filtered, cleaned with ethanol and dried for 1 hour at 120 °C. Finally, the coating was consolidated by heating the coated particles at 500 °C for one hour.

2.3 Hardness testing

The Vickers hardness of the processed composites was measured using a Testor 2100, Instron hardness tester with a load of 50 N applied for 15 s. The values shown are the average of ten measurements for each condition.

2.4 Microbending tests

Samples of 33 × 5 × 2.5 mm³ were polished, etched and notched with a U-notch. The bending tests were conducted using a three-point bending stage, with a maximum load of 200 N from Deben, on a Hitachi S-3400N to observe the in situ crack growth during the test carried out at room temperature. Loading was applied under displacement control with a displacement speed of ~0.5 mm/min. During test flexure load, flexure extension and SEM images were recorded. In situ crack growth was recorded with the SEM during the test.

The stress, σ, and the strain, ε, were converted from the recorded raw data according to the following relations:

\[ \sigma = \frac{3p}{2ae^2}, \quad \varepsilon = \frac{6ed}{l^2}, \]

where \( a \) is the initial width of the sample, \( e \) is the initial thickness of the sample minus the U-notch, \( l \) is the span length between the supports, \( p \) is the force applied on the sample and \( d \) is the midspan displacement of the sample.

In order to calculate fracture toughness, linear elastic fracture mechanics (LEFM) was used to estimate the
fracture toughness of the composites according to the ASTM Test Method E 399. The property $K_{lc}$ determined by this test method characterizes the resistance of a material to fracture in a neutral environment in the presence of a sharp crack under severe tensile constraint. A $K_{lc}$ value is believed to represent a lower limiting value of fracture toughness, and it is described by the following expression:

$$K_{lc} = \frac{P_Q S}{B W^2} f\left(\frac{a}{W}\right)$$

$$f\left(\frac{a}{W}\right) = 3 \sqrt{\frac{a}{W} \left(\frac{1.99 - \frac{a}{W}}{1 - \frac{a}{W}}\right)^{2.15 - 3.53 \left(\frac{a}{W}\right)^2 + 2.7\left(\frac{a}{W}\right)^2}}$$

In which, $K_{lc}$ is the plane-strain fracture toughness, $P_Q$ is the force at a 5% secant offset from the initial slope (corresponding to about 2.0% apparent crack extension) and is established by a specified deviation from the linear portion of the record, $S$ is the span, $B$ is the specimen thickness, $W$ is the specimen width and $a$ is the crack length.

The achievement of the plane strain condition and, by extension, the reliability of the testing method was evaluated according to:

$$B, a > 2.5 \left(\frac{K_{lc}}{\sigma_{YS}}\right)^2$$

$$1 \leq \frac{W}{B} \leq 4, \quad \frac{W}{B} = 2$$

Where $\sigma_{YS}$ is the 0.2% offset yield strength of the material.

### 2.5 Microstructure and fracture characterization

Microstructures of the different specimens were studied by both Optical (OM, Leica DMR) and Scanning Electron Microscopy (SEM, Hitachi S-3400N), also with X-Ray Energy Dispersion Spectroscopy (EDS, XFlash 2010 Bruker). After the tests, the fracture path formed during bending was characterized by SEM.

### 3. Results and discussion

#### 3.1 Starting material microstructures

Figure 1 shows the original silicon carbide starting powder. The as-received particles present many sharp fractured surfaces (Fig. 1a) whereas the roughness of the coated SiCp seems to be reduced due to the formation of a homogeneous SiO$_2$ layer surrounding the SiCp (Fig. 1b).

The composition of the matrix alloys A356 and A380 used in this work are shown in Table 1. The typical dendritic microstructures of A380/SiCp composites fabricated at 700 °C and one hour of stirring time are presented in Figure 2. The A380 alloy (Fig. 2a) consists of $\alpha$-Al matrix, eutectic silicon and intermetallic phases. Eutectic Si exhibits platelet morphologies which appear as needles in section. Several intermetallic compounds are present in the A380 alloy, $\alpha$-Al$_2$Cu, $\beta$-Al$_5$FeSi needles along with polyhedral Al(Fe,Mn,Cr)Si and typical dendritic Chinese script $\alpha$ phase. Figure 2b shows the as-cast microstructure of the A380/10% uncoated SiCp, where the main phases are large primary $\alpha$-Al dendrites (light gray), eutectic silicon (dark gray) between dendrite arms, geometrically-shaped silicon carbide particles (darker gray) and intermetallic compounds (lighter gray) as $\beta$-Al$_3$FeSi needles and polyhedral Al(Fe,Mn,Cr)Si. The reinforcement shows a uniform distribution although some segregation of SiCp particles into the eutectic liquid during solidification is observed. Small SiC particles were also present, which seem to be formed by the fracture of bigger ones during processing. In general, there is good wetting between the SiCp and the metal matrix. Usual casting defects appear in the as-cast composites such as micro-segregations and porosity. Composites porosity appeared as a result of shrinkage during solidification, and it was also observed in the inner part of SiCp clusters where SiCp wettability was poor. In the composites fabricated with sol-gel coated SiCp (Fig. 2c), wettability was improved and a more homogeneous particle distribution was observed. This resulting reinforcement distribution is expected to provide better mechanical properties. The formation of Al$_2$C$_3$ from SiC particles was not observed in A380/SiCp composites fabricated at 700 °C and one hour.
Similar microstructures were obtained in the as-cast A356/SiCp composites using the same process but using longer interaction times and higher processing temperatures (Fig. 3). Figure 3a shows the microstructure of the A356/SiCp with 10% of uncoated SiC fabricated at 700 °C and 21 hours of stirring time. The same constituents were observed although coarse polyhedral Al(Fe,Mn,Cr)Si intermetallic compounds appeared as a result of the reaction between the steel stirrer and the molten aluminium during processing. An increased interaction time did not seem to degrade the reinforcement and no Al₄C₃ formation was observed.

In A356/SiCp composites fabricated at 750 °C (Fig. 3b) little Al₄C₃ (the darkest phase) formation took place, whereas in the A356/SiCp composites fabricated at 850 °C (Fig. 3c) porosity was observed in the samples. This porosity is caused by the massive formation of Al₄C₃ that detaches during sample preparation. The amount of Al₄C₃ formed was higher as temperature increased. Finally, also large polyhedral Fe-rich intermetallics were formed.

### 3.2 Hardness tests

Figure 4 shows the measured hardness of the tested composites. The hardness of A380 was 90 HV and increased by 25% with the addition of as-received SiCp and by 15% with the addition of sol-gel coated particles when processed at 700 °C and 1 hour. Longer processing times applied to the composites fabricated with the A356 alloy indicate that there was not a clear tendency between hardness and processing temperature.

Two mechanisms may have participated in the hardness increase apart from the hardness increase caused by the incorporation of the SiCp, which hardens the composite by the presence of a second phase much harder than the matrix alloy and an increase in dislocation density because the mismatch between the matrix and the reinforcement [16].

On the one hand, several authors have indicated that the interfacial reaction between reinforcement and aluminium matrix may form Si and Al₄C₃ in the alloy, which are hard. On the other hand, composites fabricated at the highest temperature showed the formation of Fe-rich intermetallic precipitates. The former mechanism may increase hardness, but it causes the degradation of the composites by the modification of the reinforcement, the formation of brittle phases, and by the formation of carbide that degrades in the presence of humid environments.

### 3.3 Microbending tests

The microbending tests provide more reliable mechanical properties of the material. Flexural stress – strain curves for the tested composites are collected in Figure 5. From the bending curves, bending yield and maximum stresses have been calculated and the results are represented in Figure 6.

The shape of the curves is similar for all the tested specimens, but for the composite fabricated with the highest processing temperatures and the longest times, which showed very poor bending resistance.

The bending yield stress of the as-cast A380 alloy was 175 MPa and increased by 6% with the addition of SiCp and by 50% with sol-gel coated SiCp, processed at 700 °C and 1 hour. The samples treated for 21 hours showed a very different behaviour as the resistance clearly reduced when increasing the processing temperature. Reduction in resistance of 60% and 80% were observed in the samples processed at 750 °C and 850 °C, respectively.

The measured values for the fracture toughness for the Al/SiC composites are presented in Figure 7. The fracture toughness of the as-cast A380 was 4.3 MPa m¹/² and increased by 7% with the addition of SiCp and by 50% with sol-gel coated SiCp, processed at 700 °C and 1 hour. The fracture toughness of the A356/SiCp composites fabricated at longer times and temperatures exhibit a decreasing trend with the processing temperature. Reductions of 60% and 80% were observed at composites fabricated at 750 °C and 850 °C, respectively.

In general, the reasons for the improvement in the flexural behaviour of the composites regarding the unreinforced alloy are mainly the same than for hardening, particularly in the short time processed composites.
On the other hand, composites fabricated at temperatures of 750 ºC and above, and for the longest holding times showed very small bending fracture resistance. This behaviour can be explained by the formation of Al₄C₃ which has well known detrimental effects on mechanical properties because of its brittle and hygroscopic nature.

The fracture toughness (which is a measure of the composites resistance to crack propagation) of the unreinforced alloy is expected by some authors to drop by the SiCp addition due to the increased sites for crack nucleation (particles, particle/matrix interfaces, and particle clusters) [17]. However, other authors propose that it is possible to achieve good fracture properties in MMCs with a uniform distribution of SiCp and fine particulate sizes [18]. In the present, study A380 toughness slightly increased by the SiCp addition and a strong improvement was reached by coating the SiCp when processing is carried out at low temperatures and times, which may suggest that a good wetting, distribution of the reinforcement in the composites and strong bond especially in the coated ones was achieved. Whereas in the long time and temperature composites, the reason for the decrease in fracture toughness with the temperature seem to be the same proposed for the reduction in hardness.

Thus, such results indicate that strongest interfacial bonds are formed between the reinforcement and the matrix with the selected processing method when the processing temperature is kept at 700 ºC and the stirring time is one hour. Therefore, casting at 700 ºC and mixing for one hour seems to be the optimal parameters to fabricate Al/SiC composites with good mechanical properties. Also, it is possible to determine that coating the particles with a SiO₂ layer gives rise to the formation of improved interfaces which allows a great enhancement in mechanical properties of the composites, regarding the uncoated reinforced composites.

### 3.4 Fracture surfaces

Figure 8 shows a typical notched sample before and after the bending test. The crack started at the tip of the notch. To locate with more precision the initiation of the cracks, images with higher magnification were also taken.

Figure 9 resumes the initial stages of the crack formation in the A380 alloy (Fig. 9a) and composites (Fig. 9b and 9c). In these systems it can be seen that cracks initiated in small defects present within the notch tip surroundings, such as casting defects or reinforcement clusters. These zones should have caused as stress concentrators and initiated the formation of the crack in the material.

The fracture of the as-cast A380 alloy (Figure 10) it was found that cracks initiated at the precipitates present in the alloy, which cracked before the Al-matrix. Fracture occurs by a combination of brittle fracture of the large Fe-rich intermetallics with a ductile mechanism of matrix rupture. The inner part of the fracture (Fig. 10 a) demonstrates the matrix voiding resulting in pronounced dimpling of the fracture surface as a result of the ductile matrix breakage. Fig. 10 b shows the cracks in a β-Al₅FeSi needle and the microcracks which emerge from them to the aluminium matrix; whereas in the Fig. 10 c, a fractured polyhedral Al(Fe,Mn,Cr)Si is remarked. Finally, Fig. 10 d shows the edge of the main crack and beneath small cracks appear from the main one.

SEM images of the crack propagation during the bending test of the composite A380/SiCp with 10 % of uncoated SiCp (Fig. 11) after initiation at the defects present in the notch (Fig. 11 a), revealed the coalescence of fine cracks to form a mayor crack that emerged and propagated through the aluminium matrix in a ductile way (Fig. 11 b and c). Then, the crack progressed by the interface between particles and the matrix, causing its decohesion (Fig. 11 d). This was the dominating mechanism, although some large particles fractured, probably because of pre-existing defects in them. The crack path and fracture characteristics were driven by properties of the matrix in a ductile way, although crack propagation through the eutectic Si and Fe-rich intermetallics at interdendritic spaces was also observed. In this case, crack propagates faster than when progressing though the α-Al phase.

SEM images of the crack propagation during bending test of the A380/SiCp with a 10 % of sol-
gel coated SiCp (Fig. 12) showed similar mechanisms for the initiation (Fig. 12 a) and propagation of the crack (Fig. 12 b) in the Al matrix, but the behaviour in the surroundings of the SiCp was different. In this case, cracks propagated around the SiCp in a more cohesive way. Figures 11c and d show that the detached particles were surrounded by material from the matrix (arrowed zones). This indicates that the presence of the sol-gel coating favoured the bonding between the particles and the aluminium matrix.

The beginning of cracks in the A356/SiCp composites processed using high temperatures and holding times (Fig. 13) was similar to that of the previously observed composites, although the presence of a big amount of Fe-rich intermetallics seems to be the main reason for crack initiation. Also defects in notch machining (Fig. 13 a), casting defects as SiCp clusters (Fig. 13 c) and porosity which appear at the notch tip favour the crack initiation and propagation through them.

The images of the progression of the cracks in the composite (Fig. 14) revealed that a fast growing of the main crack after initiation took place by the fracture of the brittle intermetallic compounds (Fig. 14 b), the decohesion of SiCp (Fig. 14 a) and the ductile fracture of the alloy between intermetallics. The presence of brittle phases increased as processing temperature also did. The presence of Al₄C₃ which precipitates at the particle/matrix interfaces has a weakening interface effect which favours the particle decohesion during bending.

To corroborate the evidence of Al₄C₃ formation, the figure 15 shows a EDS mapping of the microstructure of the A356/SiCp composite fabricated at 850 °C for 21 hours and where it is possible to see the oxygen distribution which indicates the presence of the aluminium hydroxide formed during the hydration of the Al₄C₃, according to the following reaction:

$$\text{Al}_4\text{C}_3 + 18\text{H}_2\text{O} \rightarrow 4\text{Al(OH)}_3 + 3\text{CO}_2 + 12\text{H}_2$$  (8)

4. Conclusions

Al/SiCp composites were successfully fabricated by the stir casting method. Good incorporation and distribution of the reinforcement is achieved. Mixing times of one hour along with temperatures of 700 °C are the optimal parameters for Al/SiCp composites fabrication since the best mechanical properties are achieved. The highest hardening effect is managed when the reinforcement is uncoated, whereas coating the SiCp with a SiO₂ layer allows the best bending response of the composite as a result of the strengthening bonding effect.

The main fracture mechanisms observed during fracture of Al/SiCp composites are the ductile fracture of the aluminium matrix, fracture of the brittle phases, decohesion of the SiCp and particle fracture.

References


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**Figure 1.** a) As-received SiCp and b) sol-gel coated SiCp.

**Table 1.** Composition of used alloys.

<table>
<thead>
<tr>
<th>Material</th>
<th>Si</th>
<th>Fe</th>
<th>Cu</th>
<th>Mn</th>
<th>Mg</th>
<th>Zn</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>A356</td>
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<td>0.117</td>
<td>0.003</td>
<td>0.001</td>
<td>0.310</td>
<td>0.003</td>
<td>Rest</td>
</tr>
<tr>
<td>A380</td>
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<td>2.248</td>
<td>0.126</td>
<td>0.908</td>
<td>0.908</td>
<td>Rest</td>
</tr>
</tbody>
</table>
Figure 2. Optical micrographs of the: a) A380 alloy, b) A380/SiCp composites with as-received SiCp and c) A380/SiCp with sol-gel coated particles. All them fabricated at 700 °C and 1 hour of stirring time.

Figure 3. Optical micrographs of the A356/SiCp composites fabricated with uncoated SiCp and 21 hours of stirring time at: a) 700 °C, b) 750 °C and c) 850 °C.
Figure 4. Vickers hardness (HV₅) of Al/SiCp composites.

Figure 5. Bending curves of the tested composites.

Figure 6. Bending yield stress of tested composites.

Figure 7. Fracture toughness of tested composites.

Figure 8. Notched samples of A380/10% coated SiCp before and after bending tests.
Figure 9. SEM micrographs before and after bending tests of the A380/SiCp composites: a) without reinforcement, b) with uncoated SiCp, and c) with sol-gel coated SiCp. All them fabricated at 700 °C and 1 hour of stirring time.

Figure 10. SEM images of the crack propagation during bending test of the A380/unreinforced sample.

Figure 11. SEM images of the crack propagation during bending test of the A380/10% uncoated SiCp composite.

Figure 12. SEM images of the crack propagation during bending test of the A380/10% coated SiCp composite.
Figure 13. SEM micrographs before and after bending tests of the A356/SiCp composites fabricated with uncoated SiCp and 21 hours of stirring time at a) 700 °C, b) 750 °C and c) 850 °C.

Figure 14. SEM images of tested A356/SiCp composites fabricated at 21 hours and a) 700 °C, b) 750 °C and c) 850 °C.

Figure 15. EDS mapping of the A356/SiCp composite fabricated at 21 hours 850 °C.