Fabrication of aluminum matrix composites reinforced with nano- to micrometer-sized SiC particles

R. Taherzadeh Mousavian a, R. Azari Khosroshahi a,b, S. Yazdani a, D. Brabazon b, A.F. Boostani c

a Faculty of Materials Engineering, Shahed University of Technology, Tabriz, Iran
b Advanced Processing Technology Research Centre, School of Mechanical & Manufacturing Engineering, Dublin City University, Dublin 9, Ireland
c School of Mechanical, Materials and Mechatronic Engineering, University of Wollongong, NSW 2522, Australia

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abstract

In this study, the hot extrusion process was applied to stir cast aluminum matrix–SiC composites in order to improve their microstructure and reduce cast part defects. SiC particles were ball milled with Cr, Cu, and Ti as three forms of carrier agents to improve SiC incorporation. Large brittle ceramic particles (average particle size: 80 μm) were fragmented during ball-milling to form nanoparticles in order to reduce the cost of composite manufacturing. The experimental results indicate that full conversion of coarse micron sized to nanoparticles, even after 36 h of ball milling, was not possible. Multi-modal SiC particle size distributions which included SiC nanoparticles were produced after the milling process, leading to the incorporation of a size range of SiC particle sizes from about 50 nm to larger than 10 μm, into the molten A356 aluminum alloy. The particle size of the molten powders and the amount of released heat from the reaction between the carrier agent and molten aluminum are inferred as two crucial factors that affect the resultant part tensile properties and microhardness.

1. Introduction

The use of Al-Si (in particular A356) alloys in the manufacture of automotive engine components has increased considerably in recent years [1–6]. Although high-strength aluminum alloys have been developed, the addition of alloying elements and microstructural modification, can be costly, contain toxic elements, and often results in properties which only result in a slight increase in stiffness. The demands for lightweight, high-modulus, and high-strength materials have therefore led to the development of aluminum matrix nanocomposites (AMNCs) [7,8–18].

Various methods have been developed for the production of metal- matrix composites. An economical way of producing metal-matrix composites is the incorporation of the particles into the liquid or semi-solid metal and casting [19–22]. Semi-solid stir casting process transforms from dendritic to globular the phase structure such that fine ceramic particles can be mechanically trapped in within a highly viscous aluminum melt [23,24].

A good distribution of micron-particles was obtained in our previous studies by using various wettability improvement methods such as interfacial-active element addition by ball milling of ceramic powders with Al and Mg powders [19] and electroless deposition of Ni-P [25], Cu [26], and Co [27,28] on the ceramics. In these cases, improved strength and hardness were obtained for the micron-SiC p/Al composites compared with the unreinforced Al alloys, while a significantly reduced ductility limits their widespread application. It is of interest to use SiC nanoparticles to improve the strength and hardness of the matrix, while maintaining good ductility at room temperature [8,10]. However, due to their large surface-to-volume ratio, attractive van der Waals interactions, and poor wettability of fine ceramic particles by molten aluminum, achievement of their uniform dispersion is not an easy task [7,10,22].

Stuet al. [29] suggested an idea, in which nanosized particles could be uniformly dispersed on the surface of a carrier using ball milling process. They used aluminum powders as the carrier agent. Nanoparticle engulfing as increasing the thermal conductivity of ceramics [30–32], using graphene encapsulated nanoparticles [7], and rapid solidification [33]. A significant problem that has not yet received sufficient attention is that the as-received nanoparticles are in the agglomerated form before casting. Therefore, the idea of carrier agent has two important advantages: first, it could prevent the nanoparticles from agglomeration due to their possible capturing by a layer of carrier agent and secondly, the presence of carrier agent could be used to increase the wettability of the ceramic by molten aluminum. The carrier agent efficiency could therefore be enhanced by:

• Utilizing a high density of carrier agent.
• Utilizing a carrier agent which has a reasonable life time within the melt such that it would not be immediately lost by entering at the vortex (for this purpose, it must have a higher melting temperature than aluminum). In fact, immediate and simultaneous releasing of fine particles would lead to formation of severe clusters during stirring.
• Providing as low release of the nanoparticles into the molten metal to allow them make a suitable bond with the matrix.

The idea of using large micron reinforcement particles and their conversion to nanoparticles by fracturing during the ball milling process was achieved with size and uniformly achieved the micron particles are milled with a carrier agent.
In this study, large micron SiC particles were ball milled with three kinds of carrier agents (Ti, Cr, and Cu) to produce nanocomposite superstructures [34] before incorporation into the molten A356 aluminum alloy. Semi-solid stir casting followed by the hot extrusion process were then applied for AMNCS manufacturing to ensure a lower porosity content, a better interfacial bonding between particle and matrix, possible separation of agglomerated particles, and refinement of the matrix structure.

2. Experimental procedure

A356 ingot was used as a matrix aluminum alloy. Table 1 presents the chemical composition of the ingot used in this work as obtained using a M5000 optical emission spectrometer, Focused Photonics Inc., China.

Table 1

<table>
<thead>
<tr>
<th>Element</th>
<th>Al</th>
<th>Si</th>
<th>Mg</th>
<th>Fe</th>
<th>Cu</th>
<th>Mn</th>
<th>Zn</th>
<th>Ni</th>
<th>Ti</th>
</tr>
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<tbody>
<tr>
<td>wt.%</td>
<td>7.2</td>
<td>0.33</td>
<td>0.11</td>
<td>0.01</td>
<td>0.02</td>
<td>0.02</td>
<td>0.03</td>
<td>0.01</td>
<td></td>
</tr>
</tbody>
</table>

Fig. 1 shows the morphology of as-received SiC particles with an average particle size of 80 μm and 99.9% purity as well as those of pure Cr (99.1% purity, Fig. 1b), Cu (98.9% purity, Fig. 1c), and Ti (99.0% purity, Fig. 1d) carrier agents that were supplied from Shanghai Dinghan Chemical Co., Ltd. China. It could be seen that in contrast with Cr powders that were coarse in particle size, both the Cu and Ti powders have an average particle size below 20 μm.

Based on the idea suggested by Su et al. [29], three samples were fabricated in this study, containing ball-milled Cr–SiC, Cu–SiC, and Ti–SiC powders. These samples were produced with 15 g of SiC and 10 g of carrier metals, which were Cr (S1), Cu (S2), and Ti (S3) for this study. The ball-milling process was performed in this study by using a planetary ball mill (Sepahan 84D) with steel vial and a blend of hardened steel balls (10 mm and 20 mm diameters) in argon atmosphere. Xu et al. [34] suggested that four types of nanocomposite superstructures (Al-core/nanoparticle-shell superstructure, nanocomposite-core/nanoparticle-shell superstructure, metal-linked core/shell nanostructures, etc.)
and metal-linked nanoparticles) could be formed after mixing nanoparticles with aluminum powders. The milling process was therefore applied for 1, 12, 24, and 36 h using 1 wt.% stearic acid as process control agent (PCA) to evaluate the conversion of coarse SiC powders to fine particles as well as study on the prepared nanocomposite superstructures by increasing the milling time. These powders were incorporated into the molten A356 aluminum to fabricate AMNCs reinforced with 3 wt.% SiC and 2 wt.% carrier agents. For this purpose, one gram of the reinforcement powders was encapsulated carefully in an aluminum foil packet before the stir casting process. These packets were then pre-heated at 300 °C for 1 h in order to remove the moisture and impurities from the powders. The A356 alloy (475 ± 5 g) was heated to 640 °C using a resistance furnace in order to have uniform melt condition. The melt was then cooled to 600 °C corresponding to a 0.3 solid fraction, based on analysis with Thermo-Calc software which was shown in our previous study [2].

A ZrO coated-stainless steel stirrer was placed below the surface of melt, rotated with a speed of 500 rpm, while high purity argon gas was simultaneously applied as a protective shroud on the melt surface. During the stir-casting process, the aluminum foil packets were added to the vortex center at 600 °C during the first minute of the semi-solid stirring ring. The temperature of the melt was then increased to 630 °C, which is above the melting point of A356 alloy, and the stirring was continued for another 5 min. An addition of 1 wt.% Mg was also added during the reinforcement feeding process, to improve the reinforcement wettability by molten metal. The composite slurry in the liquid state was poured into a low-carbon steel mold (at 25 °C). A schematic of the ball-milling process as well as the stir-casting set-up and its bottom-pouring system was shown in detail in Fig. 2. This figure also shows the central location from which the samples were taken from the composite casting for machining before the extrusion process and subsequent characterization.

Disk-shape samples with 27 mm diameter were machined from the were then extruded to 9 mm diameter bars, through shear-faced dies, the hot extrusion process. The tensile tests were performed at room temperature using a universal testing machine of 100kN operating at a constant rate of cross-head displacement, with an initial strain rate of 2 × 10−3 s−1. The YS elongation to break) were measured and averaged over three test samples of SiC reinforced aluminum composite extrusion in diameter and 24 mm in length, according to ASTM B557-10.

Microhardness testing was conducted according to ASTM E384 using an applied load of 100 g for a 15 s duration. At least ten measurement of the samples were taken from the as-cast and extruded samples. Microstructural electron microscopes (SEM, KYKY-EM3200 and SEM, Philips XL 30, FESEM Mira Tescan), and an optical microscope (OM, Olympus equipped with digital camera model DP73). For this purpose, the samples were ground, polished, and etched with Keller's reagent (190 ml water, 5 ml HNO3, 3 ml HCl, and 2 ml HF). The average porosity contents of the produced samples in as-cast and extruded conditions were determined using the Archimedes method. The phase compositions of the 36 h ball milled powders were characterized by using X-ray diffraction meter (Bruker's D8 advance system, Germany) using Cu Kα (λ = 0.15405 nm) radiation to indicate if any reaction between SiC powders and the carrier agents had taken place during the 36 h milling process.

<table>
<thead>
<tr>
<th>Samples</th>
<th>Ball milling time (h)</th>
<th>Ball milling speed (rpm)</th>
<th>Ball to powder weight ratio</th>
<th>Stir casting time (min)</th>
<th>Reduction ratio for extrusion</th>
<th>Extrusion temperature (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>S1</td>
<td>36</td>
<td>220</td>
<td>8:1</td>
<td>6</td>
<td>9:1</td>
<td>500</td>
</tr>
<tr>
<td>S2</td>
<td>36</td>
<td>220</td>
<td>8:1</td>
<td>6</td>
<td>9:1</td>
<td>500</td>
</tr>
<tr>
<td>S3</td>
<td>36</td>
<td>220</td>
<td>8:1</td>
<td>6</td>
<td>9:1</td>
<td>500</td>
</tr>
</tbody>
</table>

3. Results and discussion

As noted, the ball-milling process was used in this study for two purposes: firstly to prepare fine SiC powders from large micron sized SiC particles and secondly to aid incorporation of fine ceramic particles, three metal carrier agents for 1, 12, 24, and 36 h to study the progress and structures. Fig. 3a shows the morphology of Cr-SiC mixture after 1 h be seen in the low-magnification image (left-hand image) of Fig. 3a. A red-colored rectangles, showing the effect of milling process on fragment magnification shows the surface of a large SiC particle that our EDAX analysis detected the presence of Cr element on this surface. As it can be seen, some small SiC particles with a particle size from 300 nm to about 3 μm were embedded on these large Cr-coated SiC particles. A schematic of a large Cr-coated SiC particles with the presence of small fragment SiC particles on its surface is presented in the right-hand image to indicate the prepared structure after 1 h of ball-milling.
Fig. 3. The morphology of Cr–SiC powder mixture after 1 h (a), 12 h (b), 24 h (c), and 36 h (d) of ball milling process.
Fig. 4. The morphology of Cu–SiC powder mixture after 1 h (a), 12 h (b), 24 h (c), and 36 h (d) of ball milling process.
Fig. 5. The morphology of Ti–SiC powder mixture after 1 h (a), 12 h (b), 24 h (c), and 36 h (d) of ball milling process.
More fine particles were prepared (see Fig. 3b) after 12 h ball milling. Fragmentation had occurred, resulting in more formation of submicron SiC. The same structure was formed again after a further 12 h as shown schematically in Fig. 3a. The number and the size of the large SiC particles were reduced after 12 h, while still some of the particles larger than about 10 μm could be found (see red-colored rectangles), which shows that prolonged milling might be required for obtaining fine composite particles. The particles became finer in size after 12 h and almost the same size after 24 h ball milling, while as can be seen the full conversion of large SiC particles to nano- and submicron particles seems not to be possible using the ball-milling parameters of this study. Fig. 3c
shows the FESEM images of 36 h milled powders. It could be observed reason of which is not fully clear. It could be seen that some large com-
shows high-magnification morphology of 36 h ball milled powders
SiC particles could be seen in this powder mix. The yellow-colored circle
formed by adhesion of some nanoparticles. This shows that the pres-
cles from re-agglomeration and associated complex structures form
matic phase.
By considering Figs. 4a–d and 5a–d, it was concluded that almost the
same trend like Cr–SiC mixture could be seen for Cu–SiC and Ti–SiC mix-
tures after 1, 12, 24, and 36 h ball milling. In all of these mixtures, more fine SiC particles seem to be produced by increasing the milling time. The same structure as schematically shown in Fig. 3a was also formed for all of these mixtures, showing that the type of the carrier agent has no effect on the formed structures and large SiC particle fracturing. In addition, it was found that the full conversion of large micron sized SiC particles to nano- and submicron particles was not possible with any of the different carrier agents used. However, Fig. 4d indicates that for the Cu carrier agent, a spheroidal morphology was obtained after 36 h milling. The question arises as to why such coarse composite powders were obtained between 24 and 36 h milling for this mixture. Brun et al. [35] reported that copper powders easily undergo dynamic recovery during ball milling due to their low recovery and recrystallization temperatures, and coarse copper powders were ob-
tained in their study. Therefore, ductility of the copper powders might be increased after 24 h due to a decrement in the dislocation density as a result of dynamic recovery, leading to the formation of such morphology.

There was a concern whether 36 h ball milling could lead to a reac-
indicated that no other phases were revealed except the corresponded for use with a carrier agent due to its thermodynamic stability with Fig. 6 is the copper and SiC peaks with a low background thickness leads to consideration that the copper powders undergo dynamic recov-
incorporation of fine fragmented SiC particles into the molten A356 alu-
solid stir casting for microstructural improvement. The performance of the carrier agents could be compared using optical OM and FESEM anal-
the Cr–SiC 36 h milled powder. The eutectic structure contains sil-
nanoparticles with no agglomeration had been present in the matrix,
red-colored circles in Fig. 7a show the presence of micron sized ceramic
is shown in the red-colored rectangles of Fig. 3d. Fig. 7b shows the microstructure of extruded sample S1. Significant fragmentation of sil-
con eutectic phase parallel to the extrusion direction is the most impor-
tant point that could be understood from this figure. Our previous study [2] indicated that this fragmentation of silicon platelets considerably in-
creases the ductility and strength of the A356 aluminum alloy.

Another important point of Fig. 7b (shown with white-colored arrows) is the de-
tection of agglomerated composite particles. In fact, the Cr–SiC compos-
ites powders with large particle did not release the nano- and submicron SiC particle. Based on the required factors of a suitable carrier agent that were mentioned in the Introduction section, it seems that Cr could not release the fine particles into the molten aluminum even dur-
ing stirring at 630 °C. This duty of the carrier agent (releasing the ceram-
ic particles) is very complex, depended on three factors. First, the
rier agent for a better ceramic releasing; secondly, the particle size of the not be as well unbraided during stirring to release the particles; and
minum is very significant. During an exothermic reaction, the greater releasing process, while endothermic reactions negatively affect this process. Therefore, by utilization of smaller composite reinforcement powders and using a carrier agent resulting in an exothermic reaction with the molten aluminum, the method could be more successful.

![Image](https://via.placeholder.com/150)
Fig. 8. The optical microscopy images of as-cast sample S2 (a), extruded sample S3 (b).

Fig. 9. The optical microscopy images of as-cast sample S3 (a), extruded sample S1 (b).

Fig. 10. The FESEM images of extruded sample S1, (a) low-magnification, (b) magnified image a, (c) magnified image b, and (d) magnified image a from another area.
Fig. 8 shows the OM images of as cast and extruded sample S_2 contains Cu-SiC powders. The presence of dark ceramic phases (marked in white-colored squares) and the detection of an orange colored phase that could be related to the copper (marked in white-colored circle) are the most important points of Fig. 8a. Fig. 8b shows the microstructure of sample S_2 after extrusion process. Considerable presence of the copper metallic phase (orange-colored) covered by a dark layer of SiC fine particles and its deformation toward the extrusion direction backs the idea of core-shell structure formation as suggested by Xu et al. [34]. A dark layer could be seen around the orange-colored phases in Fig. 8b, suggesting that a core (carrier agent)-shell (nano- and submicro-SiC particles) structure could be formed after ball milling of these two phases.

Fig. 4d showed that 36 h milled Cu-SiC powders contain large particles with an endothermic nature reaction between copper coating and these two subjects caused the considerable presence of large-sized agglomerated Cu-SiC powders after casting and extrusion.

Fig. 9 is related to the microstructure of sample S_3 after casting (Fig. 9a) and extrusion (Fig. 9b). By comparing Figs. 7a, 8a, and 9a, it was obtained that more dark-colored phases with a lower particle size and a semi-ideal distribution was revealed for the sample S_3 (Fig. 9a). As very little particle rejection was equally observed for all the samples during stirring, it seems that the more presence of the ceramics in the
matrix of sample S₁ might be related to their lower agglomeration be-
posite particles with lower agglomeration compared with Figs. 3d and
melt. In addition, based on the literature [37–40], irrespective of the
reaction between Cu and Al with an endothermic nature, the Gibbs free
ative) those of different kinds of Al–Cr binary intermetallic compounds
might be taken place between the Ti powders and aluminum melt, lead-
S₁ than the other samples. Fig. 9b also shows the lower detection of dark phases compared with the other samples after extrusion.

The engulfment or pushing of the fine particles could be understood
-sample S₁ after extrusion process. The presence of some large com-
pared to the extrusion direction, is clearly visible. Fig. 10b shows one of
these Cr–SiC particles in a higher magnification, in which an interesting
phenomenon was taken place. It could be observed that some phases
energy for formation of Al/Ti at 630 °C is much higher than (more neg-
such Al/ Cr). Therefore, exothermic reactions with a higher intensity
ning to the entrance of higher amount of heat during stirring for the sam-
ple S₁ than the other samples. Fig. 9b also shows the lower detection of dark phases compared with the other samples after extrusion.

from these results. Fig. 10a shows the low-magnification microstructure of
posite powders (shown by white-colored arrows), which were aligned
which Cr–SiC powders shown in Fig. 3d were able to re-

icle in the range of 100–200 nm. The second, indicated in red-colored cir-
cycle and pointed to the presence of single nanoparticles below 100 nm. The
third, indicated in a white-colored square that shows a cluster of submi-
crystal particles and, finally a white-colored arrow points to agglomera-
tion of three submicron particles. In fact, it could be concluded that single nano-
submicron particles could be distributed in the matrix using the carrier agent method, and agglomerated submicron particles (after ball-milling)
with strong bonding were also incorporated into the melt. The last important point, shown in the white-colored square, is the formation of
submicron SiC clusters with no or weak bonding. In fact, these types of clusters were formed during solidification. Ferguson et al. [32] reported that
the ceramics with a particle size between 100 nm to 10 μm is highly prone to be pushed by the solidification front in contrast with the
nanoparticles and micron-sized particles above 10 μm size. In fact, submicron particles have Stokesian motion behavior. Fig. 10d
shows that no such clusters were revealed for nanoparticles.

which Cu–SiC composite powders were incorporated into the molten
ike particles that were not released, and did not distribute the fine com-
process was not effective when unreleased composite particles were
are the composite particles depended on the bonding between the SiC
boundaries in this case could act as weak points in terms of the
composite mechanical properties.

A finer FESEM microstructure was obtained for the sample S₁. Fig. 12a shows the incorporation of fine SiC particles into the
with the same previous samples and the presence of nano-
- and submicron particles with a better distribution of fine particles are the two main differences between the sample S₁ and the previous sam-
EDAX analysis confirmed that the phases shown by white-colored rectangles in Fig. 12a mainly contain Al and Ti elements, showing that they are Al–Ti intermetallic compounds with a chemical composition close to Al/Ti intermetallic phase, which extrusion process might cause their severe fragmentation after casting because our previous study indicated that such intermetallics have considerable
large size with edge-like morphology. Both Fig. 12a and b confirmed that irrespec-
tive of suitable distribution and entrance of fine particles for the sample S₁, the presence and formation of agglomerated particles as well as solidification-induced clusters is inevitable even for this sample. Fig. 12b shows that a relatively better distribution was obtained for the sample S₁ while due to the incomplete releasing of fine particles during stirring, incomplete conversion of micron to nanosized ceramic particles, and formation of large composite powders after ball milling, the achievement to the fabrication of an ideal AMNC seems not to be easy using this method.

In order to show the concept of carrier agent, micron to nanosized particle conversion, and the structures that were formed in this study after casting, a schematic was designed and prepared. Fig. 13 indicates ceramic (SiC), five types of composite structures could be prepared. The first type (type 1) is the micron sized SiC particles with a lower par-
ticle size than that of the as-received SiC particles. These smaller parti-
colored rectangles in Fig. 3b showed such composite type. As could be

melt and its entrance into the molten metal. It could be suggested that endothermic reactions or very weak exothermic reactions caused the
formation of type 1–2 (2–1, 3–2, and 4–1), while exothermic reac-
tions may lead to formation of type 1–2 (2–2, 3–2, and 4–2). Therefore,
based on the high Gibbs free energy for the formation reaction of Al/Ti
of Ti could not be found around the micron-sized SiC particles using
after ball milling process. The difference between types 2 and 3 is the
particles (type 2) have a higher chance for releasing the fine ceramic
class can be named as nanocomposite-core/nanoparticle-shell struc-
tural particles on the composite surface would be entered into the
vealed in Figs. 10–12. It seems that both types 3-1 and 3-2 are highly

the relative densities of the samples before and after extrusion are
given in Table 3. It could be seen that sample S₁ has the minimum rela-
tive density, while sample S₄ has a maximum relative density. Based on the characteristics of the samples, it seems that the formation of types 2 and 3 superstructures (see Fig. 13) is highly responsible for the entrance of air gaps and porosities into the composites. Fig. 5d showed a spheroid
morphology that contains air gaps and such structures are po-
tential for reduction of the relative density of the final composites. It could be
observed that hot extrusion process could not highly remove the pores from the as-cast samples. Fig. 10b shows composite particles that seems to have sintered during casting and for which the extrusion process could not successfully remove the resultant pores.

The tensile properties and hardness of composites were obtained and reported in Table 4. It can be seen that the methods performed in this study could not be valuable if copper and chromium were used as the carrier agents. Microstructural characterizations indicated that why these two metals are not potential for this purpose. In fact, 8 and 13% increases in the UTS values of the samples S3 and S4, respectively, are not valuable, while very low ductilities were obtained for them. However, the hardness of both the mentioned samples is much higher than that of A356 aluminum alloy, showing that the unloosed composite particles have poor tensile properties, while they were much effective for increasing the composites hardness. On the other hand, more than about 45% increases in the YS and UTS values and 100% increase for the hardness were obtained for the sample S5 and its ductility (8.5%) is also suitable, showing that Ti could be considered as a suitable carrier agent of the fine particles. The extrusion process overall slightly enhanced the hardness of the samples mainly due to the removal of porosity.

Titanium was found to be a suitable carrier agent and further studies on Ti might therefore lead to the development of improved AMNCs using this method. Lower sized composite powders after ball milling could be obtained by optimization of ball milling parameters and by changing the ratio of Ti to SiC mass percent. In addition, by using a cryomilling process or lower Ti particle size, more valuable mechanical properties might also be obtained.

### Table 3
The relative density of the samples before and after extrusion.

<table>
<thead>
<tr>
<th>Sample</th>
<th>Relative density (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>As cast sample S1</td>
<td>95.3</td>
</tr>
<tr>
<td>As cast sample S2</td>
<td>95.1</td>
</tr>
<tr>
<td>As cast sample S3</td>
<td>96.4</td>
</tr>
<tr>
<td>Extruded sample S1</td>
<td>98.3</td>
</tr>
<tr>
<td>Extruded sample S2</td>
<td>98.0</td>
</tr>
<tr>
<td>Extruded sample S3</td>
<td>99.1</td>
</tr>
</tbody>
</table>

### Table 4
The mechanical properties of the samples after the casting and extrusion process.

<table>
<thead>
<tr>
<th>Sample</th>
<th>UTS (MPa) after extrusion</th>
<th>YS (MPa) after extrusion</th>
<th>Hardness (HV) after extrusion</th>
<th>Hardness (HRC) after extrusion</th>
</tr>
</thead>
<tbody>
<tr>
<td>A356</td>
<td>155 ± 6</td>
<td>88 ± 3</td>
<td>74 ± 2</td>
<td>79 ± 4</td>
</tr>
<tr>
<td>S1 (Cu)</td>
<td>175 ± 10</td>
<td>101 ± 5</td>
<td>110 ± 11</td>
<td>122 ± 9</td>
</tr>
<tr>
<td>S2 (Cu)</td>
<td>108 ± 8</td>
<td>99 ± 7</td>
<td>103 ± 11</td>
<td>110 ± 8</td>
</tr>
<tr>
<td>S3 (Ti)</td>
<td>228 ± 17</td>
<td>85 ± 0.5</td>
<td>147 ± 16</td>
<td>161 ± 9</td>
</tr>
</tbody>
</table>

### 4. Conclusions

The idea of using a carrier agent was applied for incorporation of fine ceramic particles into the molten aluminum. SiC encapsulated with carrier agents of Cr, Cu, and Ti metallic powders were utilized for this. In the idea that they could be fragmented by increasing the milling time into the semi-solid melt of A356 alloy and the final composites were extruded. The large particles converted to submicron and nanoparticles at prolonged milling time.

1. The particle size of the coarse SiC powders was decreased by increasing the milling time. However, fine micron SiC particles around 10–20 μm were detected even after 36 h milling, indicating that not all the large particles converted to submicron and nanoparticles at prolonged milling time.

2. The presence of nanoparticles, submicron particles, and fine micron particles in the matrix of A356 alloy showed that multimodal sized SiC particles in the range of 50 nm to 20 μm were produced during milling with metallic carrier agents.

3. XRD results indicated that the SiC particles are stable with the Cr, Cu, and Ti carrier agents from the reaction point of view after 36 h ball milling process, showing that the idea of using a carrier agent could be applied with the SiC reinforcing phase.

4. It was found that extrusion process could not alter the distribution of nanoparticles or submicron particles. However, a significant fragmentation of silicon platelets was observed after extrusion, leading to a complete change in the composite microstructures.

5. It was found that the method of using a carrier agent could be successful if the carrier could release the ceramic particles during stirring. The melting temperature of the carrier, the composite powder particle size, the stirring time and temperature, the viscosity of the melt, and the reaction tendency of the carrier with the melt affect the success of this method. It was also found that the Ti carrier agent was more successful than the Cu and Cr agents regarding the formation of a composite with higher amount of fine ceramic particles with a more ideal ceramic distribution.

6. Even the complete release of ceramic particles by the carrier agent could not guarantee the formation of AMNCs with ideal ceramic distribution due to segregation and agglomeration during stirring and solidification of nano- and in particular for sub-micron sized SiC particles.

7. More than 45% and 100% increases were obtained for the strength and hardness of a composite contains Ti carrier agent, while it was found that Cr and Cu are not suitable for increasing the mechanical properties of the final composites. A maximum UTS of 228 MPa and YS of 131 MPa were obtained for sample S3 and a maximum break strain of 8.5% was recorded.
References

[1] M. Javidian, D. Larouche, Application of cast Al-Si alloys in internal combustion en-

erties of rolled A356 based composites reinforced by Cu-coated bimodal ceramic

426.


hanced tensile properties of aluminium matrix composites reinforced with graphene

[8] L.-J. Zhang, F. Qiu, J.G. Wang, Q.-C. Jiang, High strength and good ductility at elevat-
ted temperature of nano-SiC/pAl2O3 composites fabricated by semi-solid stir cast-

sile properties of a novel network-structure AI2A2O6 matrix composite by hybridis-

[10] Y. Yang, J. Lan, X. Li, Study on bulk aluminium matrix nano-composite fabricated by
ultrasonic dispersion of nano-sized SiC particles in molten aluminium alloy, Mater. Sci.

435.

[12] K. Kalaiselvan, N. Murugan, S. Parameswaran, Production and characterization of

based metal matrix composite fabricated by direct metal laser sintering process, Mater.
Des. 32 (2011) 139–145.

[14] O. El-Kady, A. Fathy, Effect of SiC particle size on the physical and mechanical prop-

prediction of stir cast Al-TiB2 composites using response surface methodology, Mater.

[16] R. Zheng, H. Yang, T. Liu, K. Ameyama, C. Ma, Microstructure and mechanical prop-
erties of aluminium alloy matrix composites reinforced with Fe-based metallic glass

[17] B.A. Kumar, N. Murugan, Metallurgical and mechanical characterization of stir cast

[18] D. Kumar, R. Narayanansamy, C. Loganathan, Effect of glass and SiC in aluminium ma-
tal on workability and strain hardening behavior of powder metallurgy hybrid

composites reinforced with Al2O3–TiB2 composite ceramic particulates using
299–305.


[21] S. Kumar, R.S. Parwar, O. Pandey, Effect of dual reinforced ceramic particles on high
temperature tribological properties of aluminium composites, Ceram. Int. 39 (2013)
6333–6342.


[23] K. Kalyani, M. Fehlbier, A. Ludwig, A. Bührig-Polczek, P. Sahm, Non-equilibrium
microstructure stable for semisolid casting of light metal alloys by rapid

[24] S. Nafisi, R. Ghoshehcha, Effect of stirring on solidification pattern and alloy distribu-


nanoparticle reinforced aluminium matrix composite by solid–liquid mixed casting


[33] M. Dehnabi, B. Nitoumand, F. Ashrafizadeh, P. Roharg, Effects of continuous and dis-
continuous ultrasonic treatments on mechanical properties and microstructural
73–83.

[34] J. Liu, L. Chen, H. Chiu, H. Konish, X. Li, Assembly of metals and nanoparticles into


[37] N. Yao, L. Liu M. Lai, Application of thermodynamic calculation in the in-situ process of

2104–2110.
