ISSN 1318-0010
KZLTET 32(5)397(1998)

A COST EFFECTIVE FOUNDRY METHOD FOR THE PREPARATION OF STRUCTURAL GRADE DISCONTINUOUSLY REINFORCED AlMCs

NOV, TRŽNO ZANIMIV LIVARSKI POSTOPEK PRIPRAVE DISKONTINUIRANO OJAČANIH KOMPOZITOV NA OSNOVI ALUMINIJEVIH ZLITIN

VARUŽAN M. KEVORKIJAN[1]**, R. HEXEMER**[2]

[1]Dr. Varužan M. Kevorkijan, independent research, Lackova 139, 2341 Pekre, Slovenia
[2]R&D Group, ART, Buffalo, NY, USA

Prejem rokopisa - received: 1997-10-01; sprejem za objavo - accepted for publication: 1997-12-19

In this work, it was demonstrated that the wetting between ceramic particles and molten metal can be activated by exothermic interfacial reactions (e.g. Si+C $\rightarrow$ SiC or $3TiO_2+3C+4Al \rightarrow 3TiC+2Al_2O_3$). In this way, two methods of processing for introducing larger amounts of fine SiC particles into a Al alloy melt without rejection of the reinforcement, and without unwanted chemical reactions between the matrix and the reinforcement were developed. The introduction of fine (less than 6 μm) rounded-off SiC particles combined with a significant reduction in casting and extrusion defects caused by cost effective HIP treatment without encapsulation results in superior tensile properties. Moreover, in a defect-free composite, high aspect ratio reinforcements like SiC whiskers and platelets can be successfully used in order to improve the mechanical properties of final parts.

Key words: discontinuously reinforced metal matrix composites, liquid metal processing modes, wetting activated by exothermic interfacial reactions, hipping

V delu je predstavljena nova livarska tehnologija za pridobivanje diskontinuirano ojačanih kompozitov z osnovo iz najrazličnejših Al zlitin. Postopek je zasnovan na eksotermnih reakcijah (npr. Si+C $\rightarrow$ SiC ali $3TiO_2+3C+4Al \rightarrow 3TiC+2Al_2O_3$), ki potekajo na meji med keramičnimi delci in talino, oz. na kemijsko aktiviranem procesu omakanja keramičnih delcev s talino. Ugotovljeno je, da je izboljšanje mehanskih lastnosti končnih izdelkov mogoče doseči le z zmanjšanjem napak v materialu, ki nastajajo pri obdelavi taline in pri ekstrudiranju. Napake je najboljše odpraviti s HIP obdelavo končnih izdelkov. Raziskave so pokazale, da zagotavlja kombinacija livarskega načina priprave kompozita in HIP obdelave končnih izdelkov zelo ugodno razmerje med ceno in kvaliteto. Zaradi tega je omenjena tehnologija primerna za nadaljnje uvajanje diskontinuirano ojačanih kompozitov na osnovi aluminija v avtomobilsko industrijo.

Ključne besede: diskontinuirano ojačani kompoziti, livarski postopki, kemijsko aktivirano omakanje, eksotermna reakcija na meji med keramiko in talino

1 INTRODUCTION

Recently, it was suggested[1] that the wetting between ceramic particles and molten metal can be activated by exothermic interfacial reactions (e.g. Si+C $\rightarrow$ SiC). In this way, a method of processing for introducing larger amounts of fine SiC particles into an Al alloy melt without rejection of reinforcement, and without unwanted chemical reactions between the matrix and the reinforcement was presented.

Chemically treated fine SiC particles were first dispersed in a Si-Al melt (at 1473 K in vacuo). The wetting of SiC by the Si-Al melt was enhanced by exothermic chemical reaction between a carbon layer doped with MgO, previously deposited on the surface of the SiC particles, and the matrix. Improvements in fluidity was achieved by increasing the temperature of the melt. By using a very high Si content in the starting melt composition (60 wt. %), the formation of $Al4C_3$ was inhibited even at high operating temperatures. When a sufficient concentration of SiC particles had been incorporated into the Si-Al melt, an Al-Mg alloy was carefully added at a controlled rate in a nitrogen atmosphere, as to obtain the final eutectic composition. It was found that once dispersed, ceramic particles will not be rejected during the introduction of the second melt, if the process is performed with careful proceesing control. In this particular case, the processing control was performed by a specialy designed cooling regime and by using a nitrogen atmosphere. As a result, Al-SiC composites with 20 vol.% of SiC particles in the size range less than 6 μm were routinely prepared.

However, the effectiveness of the solution offered was still close to the existing commercial foundry procedures which operate with particles in the size range 10 to 20 μm. Moreover, a recent investigation performed by the same author has confirmed that the real improvement of mechanical properties of cast discontinuously reinforced metal matrix composites (DR MMC) achieved by introduction of fine or even high aspect ratio ceramic dispersoids is very modest. It was confirmed that the damage accumulation processes, caused by casting and extrusion defects, limit further improvement of the mechanical properties of cast MMCs, especially when fine ceramic particles with a high aspect ratio have been used for the reinforcement.

The aim of this work is twofold: (i) to confirm further that an exothermic interfacial reaction is a key factor in sucessfull introduction of ceramic particles into a melt; and (ii) to demonstrate that the introduction of a sufficient concentration of fine ceramic particles and platelets with higher aspect ratio can result in a DR MMC with superior tensile properties, if the casting and extrusion defects are previously eliminated by HIP treatment.

2 EXPERIMENTAL PROCEDURES

Two types of β-SiC powders produced by low temperature carbothermal reduction (AP-1 consisting of fine particles with a diameter range between 2-12 µm, and AP-2 consisting of coarser particles with about 20 µm mean diameter), commercially available M-Grade ART SiC whiskers and laboratory scale prepared SiC platelets (size range: 30-50 µm, thickness: 3-5 µm, aspect ratio 8-10, purity: max. 1000 ppm of metallic impurities, particulate content 5-10 wt%, oxygen 1.1 wt% and free carbon 0,53 wt%) were used in this study. Standard 356-T6 alloy was used in all experiments. Note that 356-T6 alloy has a nominal composition of 7 wt% of Si and 0.3 wt% of Mg.

SiC reinforcements used in this study were surface treated by the following procedures: Coating SiC reinforcements with a carbon layer: Phenolic resin and SiC reinforcements were first wet mixed in acetone. Then the acetone was evaporated with constant stirring in order to maintain the homogeneous distribution of species. The dry mixture was then pyrolyzed in flowing argon for 4h at 450°C to permit the formation of an amorphous carbon layer on the surface of the SiC reinforcements; Coating the SiC reinforcements with a carbon layer doped with TiO_2: Phenolic resin, SiC reinforcements and TiO_2 powder (99.9% pure, rutile, average particle size 1 µm) were wet-milled in ethanol for 24 h using alumina milling media. Using the same procedure as just described, SiC reinforcement were covered with an amorphous carbon layer doped with TiO_2. Note that in a separate set of experiments, the wetting tendency of the as received SiC reinforcements was also evaluated.

Both treated and untreated SiC reinforcements (except SiC reinforcement doped with TiO_2 + C) were first dispersed into the Si-Al melt at 1150°C in vacuo. The composition of the melt was usually 60 wt% Si - 40 wt% Al. The SiC reinforcement was introduced into the melt through a ceramic tube using a flow of carrier gas (argon with some additives). When a sufficient amount of SiC reinforcement was incorporated into the Si - Al melt, an Al alloy (Al-4 wt% Mg) and pure Al were added at a controlled rate and under vigorous mixing conditions in vacuo, or under a protective atmosphere, in order to achieve the 356-T6 alloy composition. SiC particles doped with TiO_2 + C were dispersed directly into the Al alloy melt in an argon atmosphere. The volume fraction of different SiC reinforcements successfully incorporated in the Al 6061-T6 matrix is listed in **Table 1**.

Table 1: The concentration of SiC reinforcement which leads to the spontaneous rejection of ceramic dispersoids from the melt

Tabela 1: Koncentracija SiC delcev v talini ob kateri prihaja do njihovega spontanega izločanja

Introduction of SiC phase directly to a Al-Si melt				Immersion of SiC phase to a Si-Al melt - the new processing route			
Type of reinforcement				Type of reinforcement			
AP-1	AP-2	whiskers	platelets	AP-1	AP-2	whiskers	platelets
Coating SiC with a carbon layer							
/	/	/	/	40-50	50-60	50-60	50
				10-20	10-25	10-25	10-20
Coating SiC with a carbon layer doped with TiO_2							
18	25	23	23	/	/	/	/
As received SiC phase							
1-2	2-3	1	1	20-30	25-35	25-35	20-30
				5-8	5-7	5-6	3-5

*concentration in Si-Al melt,
**concentration in final eutectic composition

The apparatus consisted of a graphite crucible surrounded by an induction heating coil. The mixing assembly consisted of a DC variable speed motor, the spindle and the dispersion impeller having the blades angled from about 15° to about 45° from a line perpendicular to the shaft. The crucible was provided with a protective cover and an inert gas or vacuum protection chamber.

Hot extrusion was carried out using a 4 MN (400 t) direct pilot extrusion press. Before extrusion, the samples were preheated in air to 420°C. The extrusion container was also preheated to 400°C and, before each pressing cycle, lubricated with an oil suspension of flaked graphite. The hot extrusion of samples was performed at a deformation ratio $\delta \approx 1 : 20$. Therefore, the resulting final diameter of the extruded bars was approximately 16 mm. A typical dependence between extrusion pressure and ram travel was obtained; starting with a higher peak pressure and continuing with a stable working (running) pressure. The peak pressure was 330 to 470 MPa (3.3 - 4.0 t/cm^2). The extrusion speed was maintained as low as possible (4-10 mm/s) and regulated manually, depending on the extrudability of each individual sample.

HIP treatment (without encapsulation) of extruded specimens was performed in a HIP 2000 unit (National Forge, Belgium) using a graphite furnace. Typical experimental conditions were: temperature 30-50 K below the melting point of Al-Si-Mg alloy, argon pressure: 150-180 MPa, holding time: 1-2 hours. Samples for HIP treatment were cut parallel to the direction of extrusion from an extruded billet, and then machined on a lathe to a rod with diameter 50 mm and length 100 mm.

The mechanical properties investigated included Brinell and Vickers hardness tests and tensile tests performed at room temeprature. The composite Brinell hardness (62.5 kgf, preload 10 kgf, sphere 2.5 mm) was measured at the centre of the samples and near the edges. Vickers microhardness measurements (20 gf load) were taken from the aluminium matrix. The average of three measurements is reported in **Table 3**. The tensile tests were carried out in accordance with ASTM B557. Three parallel tensile tests for each specimen were performed. NonHIPped samples were first rough cut parallel to the direction of extrusion, directly from an extruded billet, using a carbide band saw, and then machined on a lathe using a tungsten carbide cutting tool. The HIPped and nonHIPped samples were sand cast first, and than machined on a lathe. The load versus strain, was measured using an extensometer and plotted. The measured yield strength (σ_y), ultimate tensile strength (σ_{uts}), and % elongation are reported in **Table 3**, along with values for unreinforced 356-T6 alloy.

Extruded and HIPped material was also investigated in optical and scanning electron microscopes with regard to parameters such as SiC - distribution, surface tearing, grain size and primary constituents. However, these results will be published later.

Table 2: Room temperature tensile properties of extruded and hipped MMCs based on an 356 alloy (T6 condition). Grade A-1: fine (6 μm) SiC particles, Grade A-2: Coarser (15 μm) SiC particles, Grade B-1: SiC whiskers, Grade B-2: platelets

Tabela 2: Mehanske lastnosti ekstrudiranih in vroče stisnjenih vzorcev kompozita na osnovi 356 zlitine (T6 stanje). Vzorec A-1: fini (6 μm SiC delci, Vzorec A-2: Grobi (15 μm) SiC delci, Vzorec B-1: SiC kratka vlakna, Vzorec B-2: SiC ploščice

Composite	V_f	Aspect ratio	σ_y (MPa)	σ_{uts} (MPa)	E (GPa)	Elongation %
Extruded						
A-1	10	<2	293	309	83	0.6
A-2	10	<2	278	301	81	0.6
B-1	10	12	307	328	88	0.6
B-2	10	8-10	302	322	83	0.6
A-1	20	<2	359	377	103	0.4
A-2	20	<2	337	355	98	0.3
B-1	20	12	372	391	108	0.4
B-2	20	8-12	368	386	106	0.3
A356(T6) unreinforced	/	/	205	280	76	6
Hipped						
A-1	10	<2	320	351	109	0.8
A-2	10	<2	318	335	93	0.8
B-1	10	12	465	475	168	0.7
B-2	10	8-10	467	481	173	0.7
A-1	20	<2	438	465	136	0.5
A-2	20	<2	431	461	124	0.4
B-1	20	12	593	565	212	0.4
B-2	20	8-12	595	562	226	0.4
A356(T6) unreinforced	/	/	262	364	97	9

Table 3: Brinell and Vickers microhardness tests results

Tabela 3: Rezultati merjenja mikrotrdote vzorcev kompozita po Brinell-u in Vickers-u

Composite	V_f (%)	Aspect ratio	Brinell hardness		Vickers microhardness	
			as extruded	annealed	as extruded	annealed
Extruded						
A-1	10	<2	103±10	91±6	89±3	89±3
A-2	10	<2	96±6	83±6	81±3	81±3
B-1	10	12	107±10	95±6	92±3	93±3
B-2	10	8-10	106±10	95±6	91±3	92±3
A-1	20	<2	122±10	103±10	91±3	91±3
A-2	20	<2	110±10	94±10	85±3	86±3
B-1	20	12	119±10	105±10	92±3	92±3
B-2	20	8-10	117±10	104±10	93±3	92±3
Hipped						
A-1	10	<2	153±10	152±6	142±3	154±3
A-2	10	<2	148±6	147±6	139±3	149±3
B-1	10	12	162±10	154±6	148±3	154±3
B-2	10	8-10	171±10	156±6	149±3	153±3
A-1	20	<2	203±10	167±10	213±3	211±3
A-2	20	<2	189±10	159±10	208±3	206±3
B-1	20	12	212±10	182±10	207±3	202±3
B-2	20	8-10	1214±10	189±10	202±3	202±3
A356 (T6)	/	/	112-144	98-127	109-141	106-147

3 RESULTS AND DISCUSSION

Table 1 summarizes the lowest experimentally monitored concentrations of SiC reinforcement which result in the spontaneous rejection of the reinforcing phase from the melt. In this way, the efficiency of the direct immersion of fine SiC reinforcements into an Al-Si melt was quantitatively compared with the potential of the new processing route. As described, the new processing route is based on the immersion of SiC reinforcements in a Si-Al melt and its further alloying with Al-Mg to the final eutectic composition.

The preliminary results in **Table 1** show that, as expected, only a very limited level of uncoated fine SiC particulates can be successfully introduced into a melt. Rejection occurs spontaneously at a particulate concentration between 2-3 wt%. More promising results were obtained with fine SiC particles previously oxidised in air (at 1200°C for 1 h). In that case, 6-9 wt% of fine SiC particles could be successfully incorporated into the matrix prior to their spontaneous rejection. However, note that oxidation of fine SiC particles at 1200°C for 1 h changes their chemical composition dramatically. Thus, the wetting kinetics of strongly oxidised fine SiC particles in an Al melt are rather governed by the wetting tendency of the equivalent amorphous SiO_2 particulates. Using an exothermic reaction between carbon, previously deposited on the surface of SiC particles, and silicon in the melt, a high portion of ceramic filler can be sucessfully immersed, **Table 1**. For example, it was found that only 10-15 wt% uncovered fine SiC particles could be successfully incorporated into a Si melt without significant rejection. In contrast, under the same experimental conditions more than 30 wt% fine SiC particles covered by a carbon layer were successfully introduced into a Si melt, in accordance with the exothermic reaction between the carbon layer and silicon in the melt. The details of the reaction paths are not beyond the scope of this work[2]. However, it is important to note that the overall reaction is highly exothermic. Therefore, using a recently postulated model of the wetting process[3], one can conclude than an exothermic interfacial reaction which proceeds rapidly at the surface of the ceramic phase should enhance its wetting tendency.

Very recent experiments performed on SiC reinforcements covered with a TiO_2 + C layer strongly support this conclusion. Note that the reaction:

$$3TiO_2+4Al+3C \rightarrow 3TiC+2Al_2O_3 \qquad \Delta H^{\circ}_{f,} = -1076{,}6 \text{ kJ/mol} \qquad (1)$$

is very exothermic. As reported[4], this could result in a local rise in temperature of up to 2600 K. The reaction occurred at a temperature approximately 250 K higher than the melting point of Al (933.4 K). For details of the reaction mechanism see[4]. As evident from **Table 1**, approximately 25 vol% of fine SiC particles or SiC reinforcement with high aspect ratio were successfully introduced directly into the melt by the very exothermic chemical reaction between the melt and a mixture of TiO_2 and carbon deposited on the surface of SiC particles. This clearly confirms that an exothermic interfacial chemical reaction can significantly improve the immersion of ceramic particles into a melt.

Mechanical properties

An effort has been made to compare the mechanical properties of HIPped and nonHIPped Al-SiC composites prepared by liquid-metal processing. Types A-1 and A-2 are composites reinforced with low aspect ratio SiC particles (AP-1 SiC powder and AP-2 SiC powder with mean particle diameters of 10 μm and 20 μm, respectively). Type B are composites reinforced with high aspect ratio ceramic fillers like SiC whiskers (B-1) or SiC platelets (B-2). Although there is significant scatter in some of the experimental data (especially in data generated using nonHIPped specimens), probably caused by the presence of casting defects, defects introduced by hot extrusion and differences in residual porosity at the interface, some general conclusions regarding the effect of the reinforcement and HIP treatment can be made.

Tensile data obtained from composites A-1 and A-2 containing a volume fraction of reinforcement from 0.10 to 0.20 and SiC particles of mean diameters from about 10 μm to 20 μm indicate that the mechanical properties depend on both the size and volume fraction of the particles. However, it seems that the strength of nonHIPped particulate MMCs prepared by liqud metal is dependent most strongly on the volume fraction of reinforcement, with a weaker dependence on the particle size and aspect ratio.

The modulus of elasticity of the composites investigated increases with increasing reinforcement content. Again, in nonHIPped specimens the reinforcement content is the dominant factor in increasing the modulus of elasticity.

Based on collected experimental data, one can also conclude that the modulus appears to be insensitive to the type of reinforcement used. Note that in defect-free material prepared by P/M, several authors[5] reported that the modulus of particulate MMCs is much less than that of whisker MMC's due to the particulates much lower aspect ratio (≈1-2), and therefore, their inability to pick up load from the matrix as effectively as the whiskers. Moreover, the variation in the elastic modulus of MMC's with the aspect ratio of the reinforcement has been also theoretically predicted[5]. According to this model, significant improvements in the elastic modulus should be obtained by increasing the aspect ratio of the reinforcement.

However, as reported[5], the model is very sensitive to the orientation of high aspect ratio particles with respect to the axis of measurement. The same author[5], using SiC platelets with different aspect ratios, reported thet modulus values fall very rapidly with degree of platelet misorientation.

The results of the present investigation mostly confirm this conclusion. The values of UTS, YS and elongation measured in nonHIPped specimens are almost the same for both high aspect ratio composites (B-1 and B-2) and are only 10% higher than for the corresponding particle reinforced composite (A-2). Moreover, there is no significant diference in mechanical properties between B-1 and B-2 grades and A-1 grade prepared by fine and, at least one order of magnitude cheaper, rounded-off SiC particles. Note that this effect could strongly affect the cost competiveness of MMC's prepared by the liquid metal route and, consequently, the future market prospects for MMC commercialisation.

It seems that in MMCs prepared by casting the key factor which needs to be controlled is the successful immersion of fine SiC particles. The introduction of high cost, high aspect ratio particles not can improve the mechanical properties significantly.

Obviously, the presence of casting defects and defects caused by hot extrusion in the nonHIPed samples tested played a major role in the actual results obtained. Both high aspect ratio composites (B-1 and B-2) showed a significant amount of whisker and platelet fracture after extrusion, probably due to high loading and to the high selected aspect ratio of the reinforcements.

From these considerations, the mechanical properties in nonHIPped MMC's prepared via the liquid metal route are most strongly dependent on the volume fraction of reinforcement, with a somewhat weaker dependence on particle size, and are almost insensitive to the aspect ratio of the ceramic filler.

The results of Brinell hardness and Vickers microhardness tests are similar for both as extruded and annealed nonHIPped composites, **Table 3**. This fact suggest that dynamic recrystallization might have occurred during extrusion.

The Brinell hardness increased with the SiC volume fraction, as would be expected. For purposes of comparison, both hardness values of unreinforced A 356-T6 are also given. The hardness values presented in **Table 3** indicate that even the hardness of cast Al-SiC composite is not strongly affected by the morphology and the aspect ratio of the reinforcing phase. Again, in nonHIPped specimens defects in the material seems to play the dominant role in the actual results obtained.

However, the immersion of fine SiC particles in an A356 matrix (using AP-1 SiC powder with a mean particle diameter less than 10 μm) resulted in some improvement of hardness. This suggests that by limiting the size of the particles and by improving their dispersion and alignment, it should be possible to improve the composite properties, even at a low aspect ratio of the reinforcing phase. Moreover, the introduction of SiC platelets with an average particle size less than 10 μm, employed in order to minimize platelet fracture during extrusion, could result in significant improvement of the composite properties. Again, in addition to cost, some improvement of mechanical properties in nonHIPped specimens, caused by using a ceramic phase with higher aspect ratio, should be compared with the eventual increase of defects simultaneously introduced in this way.

In contrast, inspection of the room temperature tensile properties of HIPped specimens **(Table 2)** showed a significant improvement in yield strength (σ_y), ultimate tensile strength (σ_{uts}) and modulus of elasticity (E). In HIPped specimens Brinell hardness and Vickers microhardness **(Table 3)** were also improved (in some cases it was doubled).

The improvement in tensile properties and hardness of HIPed specimens was more marked in specimens reinforced with a high volume fraction (20 vol%) of SiC whiskers and platelets (B-1 and B-2), i.e. in specimens with a high concentration of casting defects. Note that in HIPped specimens a significant difference between the modulus of particulate MMCs with low aspect ratio (A-1, A-2) and MMCs with high aspect ratio (B-1, B-2) was found. A possible explanation is that in nonHIPped specimens the casting and extrusion defects masked the real correlation between the aspect ratio and modulus.

Hence, it seems that ceramic reinforcements with high aspect ratio, such as low cost SiC platelets prepared by $\beta \rightarrow \alpha$ phase transformation, can be effectively used in the liquid metal foundry route for the preparation of DR MMCs.

In this way, it was recognised that the supplementing of conventional foundry method by HIP treatment, which reduces casting defects, offers a new commercial route for further cost effective production of cast DR MMCs with superior mechanical properties.

4 CONCLUSION

Two processing routes for introducing larger amounts of fine SiC particles into an Al alloy melt without rejection of the reinforcement, and without unwanted chemical reactions between the matrix and the reinforcement were described. Both processes are based on an exothermic interfacial reaction, i.e. on the chemically activated wetting of ceramic dispersoids into a melt.

In the first processing route chemically treated fine SiC particles were first dispersed in a Si-Al melt (at 1200°C in vacuo). The wetting of SiC by liquid silicon was enhanced by chemical reaction between a carbon layer previously deposited on the surface of the SiC particles and the matrix. When a sufficient amount of SiC patricles had been incorporated into the Si matrix, an Al alloy was carefully added at a controlled rate under vigorous stirring conditions and in a protective atmosphere, so as to obtain the final matrix composition of A 356 without SiC particle rejection. In this way, a processing route for the production of Al-SiC composites with 10-20 wt% of SiC particles in the size range less than 10 μm was developed.

In the second processing route SiC reinforcements covered with a TiO_2+C layer were directly immersed into a Al alloy melt, introducing a very exothermic reaction between the melt and the surface reactants.In this way more than 25 vol% of fine SiC particles or SiC reinforcement with a high aspect ratio were successfully introduced into the melt. The process is also suitable for introducing other ceramic reinforcements such as AlN, Al_2O_3, etc. in molten light metals.

The extrusion tests performed showed that the cast MMCs, based on A356 alloy reinforced with silicon carbide particles, platelets or whiskers, can be extruded with good results following the same procedure as that for unreinforced aluminium.

A study of the influence of silicon carbide particle size, aspect ratio and volume fraction on room temperature tensile properties showed that, according to the strengthening mechanism, a higher volume fraction and a smaller particle size are beneficial to improving tensile strength, because of the reduction of space between the particles. On the other hand, because silicon carbide particles were introduced and dispersed into a molten aluminium alloy in liquid form by a mechanical stirring method, adding particles resulted in casting defects which deteriorate the properties of composites. In this way, the damage accumulation processes, caused by casting and extrusion defects, limit further improvement of the mechanical properties of nonHIPped cast MMCs based on high aspect ratio reinforcement. Because of this, the use of costly reinforcement like SiC platelets or whiskers in nonHIPped cast MMCs seems unreasonable and not cost effective.

On the contrary, the elimination of casting and extrusion defects achieved by HIP treatment of specimens resulted in superior tensile properties. Moreover, in defect-free composites, it was found that high aspect ratio reinforcements like low cost SiC platelets can be sucessfully used in order to improve the mechanical properties of the final material. From the acumulated data it is also evident that the mechanical properties of the resulting composites matched the same characteristics of similar composites prepared via the PM route.

One can conclude that supplementing the conventional foundry method for DR MMCs by the low cost HIP technique results in DR MMCs with a very advanced quality/cost ratio and, therefore, it could represent an important market factor for further application of DR MMCs in the automotive sector.

5 REFERENCES

[1] Kevorkijan, M. K. and Sustarsic, B., The Introduction of Fine SiC Particles into a Molten Al Alloy Matrix: Application to MMCs Preparation via a Foundry Route, *Proceedings of the American Society for Composites - Tenth Technical Conference*, Santa Monica, California, USA, October 18-20, 1995, Composite Materials, Mechanics and Processing, Technomic, p. 569-578

[2] Ness, J. N. and Page, T. F., Microstructural Evolution in Reaction-Bonded Silicon Carbide, *Journal of Materials Science*, 21 (1986) 1377-1397

[3] Kevorkijan, M. V. and Sustarsic, B., The Introduction of Fine SiC Particles into a Molten Al alloy Matrix: Application to Composite Material Casting, *Metals, Alloys, Technologies*, 29 (1995) 5-6, 447-454

[4] Choi, Y. and Rhee, S. W., Reaction of TiO_2-Al-C in the Combustion Synthesis of TiC-Al_2O_3 Composite, *Journal of the American Ceramic Society*, 78 (1995) 4, 986-992

[5] Jain, M. K., Lloyd, D. J. and Tremblay, S. P., Evaluation of Silicon Carbide Platelets in Metal Matrix Composites, *Proceedings of International Conference on Metal and Ceramic Matrix Composites*, Anaheim, California, February 19-22, 1990, Metal & Ceramic Matrix Composites: Processing, Modeling & Mechanical Behavior, Bhagat, R. B., Clauer, A. H., Kumar, P. and Ritter, A. M., Eds, p. 549-560