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SiC as Base of Composite Materials for Thermal Management

J.M. Molina

Instituto Universitario de Materiales de Alicante, Universidad de Alicante

Departamento de Química Inorgánica, Universidad de Alicante

Spain

1. Introduction

Some of the high-end applications in energy-related topics such as electronics, aeronautics and research in elementary particles have reached their technological limits because of the impossibility of finding materials capable of removing the excessive heat generated in running their equipments and, at the same time, maintaining their dimensional stability in environments often extremely aggressive, namely, wide temperature range of use (218-423K), corrosive environments (>98% humidity), fast heating-cooling cycles or interaction with accelerated particles. The growing needs for thermal control are a consequence of the unlimited increasing power consumption in the operation of their equipments. These applications, that exclude the use of monolithic materials given their required unique combination of properties, force the use of composite materials that exhibit high heat transport and, at the same time, do maintain their dimensional stability under operational conditions. Despite the significant progress in the development of composite materials for these applications in recent years, nowadays there is a need to find new materials capable of withstanding the extreme conditions that are impingingly demanded for the new heat sinks. Power electronics and optoelectronics demand thermal conductivities (TC) above 350 W/mK and 450 W/mK respectively; aeronautics is less demanding (>250 W/mK). The coefficients of thermal expansion (CTE) should be matched to those of the architectures on which they are mounted to prevent failures by thermo-mechanical fatigue (it is required 3-6 ppm/K for optoelectronics and less than 12 ppm/K for power electronics, both in the 293-500K range, and 10-14 ppm/K in the range 233-344 K for aeronautics). In applications such as collimators in particle accelerators, apart from a great resistance to radiation damage, materials with more than 300 W/mK of CT and approximately 10-12 ppm/K of CTE are needed. All applications require isotropic materials with a flexural strength of >120 MPa and the lowest possible density, especially those for aeronautics.

SiC-based metal matrix composite materials have shown, up to now, a perfect combination of properties such that to cover the increasing demanding of the energy-related industries. Among them, Al/SiC has become a leader material and nowadays is considered the state-of-the-art in thermal management. This material, in which SiC is present as particulate, cannot meet the future requirements for heat dissipation and dimensional control given its limited thermal conductivity (about 180 W/mK) and relatively high coefficient of thermal

expansion (around 10 ppm/K). For this reason, during the last years research has been directed to develop new and alternative materials that can replace the traditional Al/SiC and allow the emergence and growth of new and more efficient equipments. Several solutions have been proposed, most of them based on the use of different finely divided reinforcement materials embedded in a metallic matrix. The use of metals as matrix seems to be a question with no discussion, since metal consolidates the preform, is thermally stable (something important for some applications), is lighter than many ceramics and at the same time allows in most cases an easier machining process. Moreover, the global properties of the material can be widely varied by playing with the metallurgical state of the metallic matrix. Al, Ag and Cu and their corresponding alloys with interfacial active elements have proven to be appropriate matrices for composites conceived and designed for the above mentioned applications. Nowadays, among the different options considered as reinforcements in composites for electronics we find that SiC is still leading the choice. Different combinations of SiC with other reinforcements (such as alumina or diamond) or the use of mixtures of SiC particles of different sizes (bimodal or multimodal distributions of particles) have proven to be essential to match the extreme requirements of electronics. A very recent composite material, developed and patented at the University of Alicante, is based on the use of mixtures of graphite flakes and SiC particles (or alternatively other reinforcements) in order to make a preform in which flakes tend to form layered structures. The SiC particles act as a separator between layers of flakes and on the other hand allow reduce the thermal expansion coefficient in the transversal direction which otherwise would be inadmissibly high. One clear competitor for the SiC-based composites is the family of those fabricated with diamond particles. Even though their thermal properties are very attractive they pose important problems related to obtain pieces with complicated geometries, as diamond is very difficult or even impossible to be machined. Within this scenario the new research on composites based on machinable reinforcements seems to be the only industrially attractive option for many applications.

Most of these composites are fabricated by pressure infiltration of the metal into the preform, assisted either by gas or by mechanical means (squeeze casting). The selection of proper materials quality as well as of optimal fabrication conditions is completely essential to meet the target properties.

The present chapter presents different SiC-based composite materials which have been evolving over time aiming to be useful for thermal management. It also analyzes the different aspects of the fabrication that affect the thermal properties of these composites.

2. Fabrication procedures of composites for thermal management

The limitations of metallic materials that had traditionally been used as heat sinks in the electronic industry very soon attracted the attention of other alternative systems. The composite materials made out of metals as matrix and ceramics as reinforcement became immediately potential candidates for a wide variety of applications in electronic packaging (Clyne, 2000a). The following Ashby's map allow us to think about the different possibilities of metal-ceramic combinations for the aforementioned applications.

Among ceramics, SiC is one with relatively high thermal conductivity (500 W/mK for monocrystals, 250-350 for polycrystals) and a coefficient of thermal expansion (around 4.7 ppm/K) very close to that of silicon (2.5-3.6 ppm/K). Its low price, especially for the non-most pure forms of SiC, makes this ceramic to have a very high performance/price ratio. For

the sake of comparison we shall mention that very pure SiC particles can be as costly as three times the price of those of normal grades (less than 99.8% purity). Their thermal conductivity is somehow higher but nevertheless not sufficient to justify their use in comparison to other less expensive ceramics (like diamond). Diamond, in its turn, has a very high thermal conductivity but the performance/price ratio is still its limiting factor. However, seemingly this ceramic represents a good choice for the coming future in the electronics industry given that their price has followed a continuously decreasing tendency during the last ten years.

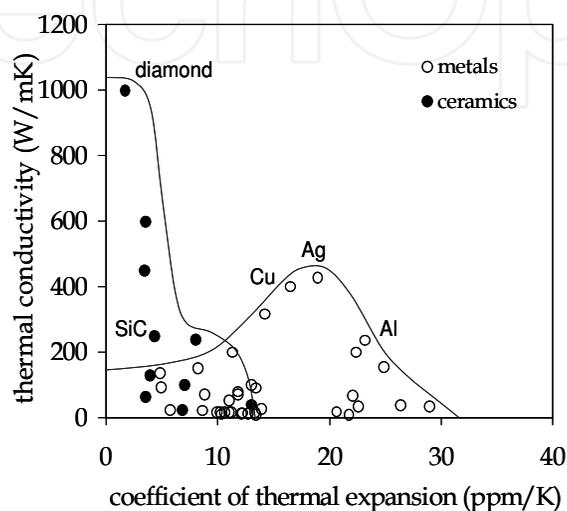


Fig. 1. Ashby's map of thermal conductivity and coefficient of thermal expansion for different metallic and ceramic materials

On the side of metals, aluminium turns out to be one of the most attractive. Although its thermal properties are not excellent, is a light metal and has a low melting point. Its combination with SiC in proper amounts may generate composite materials with the desired properties for thermal management. Given that the thermal conductivities of both SiC and Al are very similar, the expected value of this property for their composites is rather in the same range. The most accounting effect on composites is the thermal expansion coefficient, which can be varied over a wide range by playing with the volume fraction of Al and SiC phases. In fact, in view of the Ashby's map, it becomes apparent that high volume fractions of SiC are necessary if the coefficient of thermal expansion needs to be considerably reduced. It is this last condition what limits in practice the number of fabrication procedures that can be chosen in order to manufacture composites for heat sinking.

As a general rule, only those processing techniques that allow obtain a high volume fraction of reinforcement are useful. In this sense, infiltration of the molten metal into packed preforms has been recognized as the most appropriate procedure. Since most ceramics are not wetted by molten metals, infiltration typically requires be pressure-assisted. The way in which the molten metal is forced into the open space of the ceramic preform determines the two main infiltration techniques, namely gas-pressure infiltration and squeeze casting (or mechanically-assisted infiltration). Alternatively, the powder-metallurgy technique has also been used for those systems where the configuration of the preform is such that intrusion techniques become difficult or where the preform might suffer of dimensional damage when high pressures are required.

2.1 Gas-pressure assisted infiltration

In general, pressure does not exceed 15MPa and this allows fabricate pieces with complex shapes without taking the risk of a considerable deformation of the preform. The method is highly versatile because, with a simple modification of the main mould, pieces of different geometries can be fabricated. Final machining is sometimes needed although, if moulds are properly designed in such a way that the demoulding process turns out not to be difficult, neat-shape fabrication is feasible. The main drawback of this technique stems from the limited rate of metal penetration into the preforms, resulting difficult the manufacture of large pieces. This restriction is a consequence of the technology characteristics; in particular of the low pressures applied and the wettability-reactivity characteristics of the system at hand.

2.2 Mechanically-assisted infiltration

The metal is forced to penetrate into the ceramic preform at very high pressures (in the range 50-100 MPa) by means of a piston mechanically driven. This method is called “squeeze casting” and its application into the fabrication of MMC’s is very extended, although it is not free of drawbacks related with the high pressures used. This may cause deformations in the preforms that alter the global shape and the relative presence of metal and ceramic phases. The solution to this problem not always seems to be found by diminishing the working pressure because infiltration rate or metal-ceramic reaction are for some systems important issues to be considered. The necessity of huge installations, occasionally very expensive, is another drawback of squeeze casting. Its main advantage is that the high working pressures effectively ensure infiltration and the final composite materials can be free of remaining porosity.

2.3 Powder-metallurgy

Although powder metallurgy has become an excellent technique for the manufacturing of relatively complicate shaped metallic pieces, in the field of composite materials is not an extensively used fabrication method. This technique allows obtaining high volume fractions of reinforcement (75%) and moreover offers a perfect control of reactivity between metal and ceramic phases. However, a clear drawback is the difficulty encountered for the control of porosity in the material, which seems a phase that inherently appears when using this technique. The control of the oxygen content (as metallic oxides existing concomitantly in the metallic powder) seems also to limit the possible massive use of this technique in the industrial fabrication of composite materials.

Recently, another technique derived from the already mentioned powder-metallurgy has become a matter of interest. This technique is called “spark-plasma” and it consists of heating the metallic or graphitic mould, as well as the powder compact in case of conductive samples, by means of an electrical current that flows through it. This technique allows a fast processing of the materials but, nevertheless, it suffers from the same disadvantages of the classical powder-metallurgy route.

3. Measurement and estimation of thermal properties in composites for thermal management

3.1 Property needs in materials for electronics

As already explained, there are several requirements for those materials considered for electronics. An ideal heat sink must extract the heat generated in excess in a given running

equipment but at the same time has to be thermally stable in order to maintain the dimensionality of the architecture on which they are mounted. This already points out that the main properties that are essential for those materials are two: thermal conductivity and thermal expansion coefficient. While the former needs to be as high as possible the later is restricted to a certain range, typically between the limits of 3 and 14 ppm/K. Although these two properties are limiting, other characteristics of the materials (i.e. density, cost) may also take a significant importance for certain applications. In the present context only the two main properties are reviewed.

3.2 Measurement of thermal properties

Thermal conductivity is measured by mainly two means. The simplest one consists of comparing the thermal conduction of the sample with a reference material by connecting the sections of both and establishing a thermal gradient. It is called the comparative stationary method and provides accurate and relatively fast measurements. Another more sophisticated method is based on the use of a laser flash, which measures the heat propagation in non-stationary conditions caused by the flash of a laser impacting the surface of the material.

The coefficient of thermal expansion is typically accessed through the use of a thermo-mechanical analyser, which subjects the sample to a thermal cycling over a certain range of temperature. The dimensions of the sample are followed in one, two or alternatively three axis. In fact, what is measured is the dimension change with the increase in temperature (units of ppm/K).

3.3 Estimation (modelling) of thermal properties

3.3.1 Coefficient of thermal expansion

A very simple and tentative model for the estimation of this property in composite materials consists of averaging over the volume fraction of the corresponding phases:

$$\alpha = \alpha_r V_r + \alpha_m (1 - V_r) \quad (1)$$

where the subscripts m and r denote matrix and reinforcement, respectively. This simple approach is called the linear rule of mixtures and offers a back-of-the-envelope calculation of what the coefficient of thermal expansion must approximately be.

More sophisticated treatments are all based on thermoelasticity theory (Clyne, 2000b). Schapery's model gives upper (+) and lower (-) bounds on the CTE. The specific expression for the former is:

$$\alpha^{(+)} = \alpha_r + (\alpha_m - \alpha_r) \frac{K'_m (K'_r - K_c^{(-)})}{K_c^{(-)} (K'_r - K'_m)} \quad (2)$$

where K' refers to bulk modulus; $K_c^{(-)}$ is Hashin and Shtrickman's lower bound to the bulk modulus of the composite, namely,

$$K_c^{(-)} = K'_m + \frac{V_r}{\frac{1}{K'_r - K'_m} + \frac{V_m}{K'_m + \frac{4}{3}\mu_m}} \quad (3)$$

where μ_m is the shear modulus of the matrix. The upper bound to the bulk modulus is obtained by interchanging the subscripts m and r everywhere in Eq. (3) which, when inserted in Eq. (2), gives the lower bound on the CTE.

3.3.2 Thermal conductivity

The thermal conductivity of composite materials containing thermally conductive inclusions has been extensively studied. One of the simplest analytical models that assume a non-idealized interface between matrix and reinforcement is that derived by Hasselman and Johnson for spherical particles in a pore-free infinite matrix (Clyne, 2000b):

$$K_c = \frac{K_m \left[2K_m + K_r^{eff} + 2V_r (K_r^{eff} - K_m) \right]}{2K_m + K_r^{eff} - V_r (K_r^{eff} - K_m)} \quad (4)$$

where K is the thermal conductivity, K^{eff} is the effective thermal conductivity, V is the volume fraction, and the subscripts c , m and r refer to composite, matrix and reinforcement, respectively. This model has been proved to give accurate predictions when particles and matrix exhibit a low ratio of conductivities, which is the case for SiC particles and pure Al or many Al-based alloys. The effective thermal conductivity of particles K_r^{eff} is defined as:

$$K_r^{eff} = \frac{K_r^{in}}{1 + \frac{2K_r^{in}}{D \cdot h}} \quad (5)$$

where K_r^{in} stands for the intrinsic thermal conductivity of the particles, and D and h are the average particle diameter and the interfacial thermal conductance, respectively.

The differential effective medium (DEM) scheme has been applied with success to model and interpret transport properties in different composite materials (Clyne, 2000b; Molina et al., 2008a; Molina et al., 2008b; Molina et al., 2009; Tavangar et al., 2007; Weber et al., 2010). Its predictive capacity is in many cases considerably superior to other available models in literature (see for example (Tavangar et al., 2007)). The leading integral equation of the DEM approach for the thermal conductivity of a multi-phase composite material is:

$$\int_{K_m}^{K_c} \frac{dK}{K \sum_i f_i \frac{-(K - K_{r_i}^{eff})}{(K - K_{r_i}^{eff}) \cdot P - K}} = -\ln(1 - V_r) \quad (6)$$

where K_c is the thermal conductivity of the composite material, f_i is the fraction of the i inclusion in the total amount of inclusions of the composite (hence, $\sum_i f_i = 1$) and P is the polarization factor of the inclusion. $K_{r_i}^{eff}$ is the effective thermal conductivity of the i inclusion, which is related to its intrinsic thermal conductivity, $K_{r_i}^{in}$, the matrix/inclusion interface thermal conductance h , and the diameter D of the inclusion by Eq. (5).

In general, the integral on the left hand side of Eq. (6) has no analytical solution and has to be solved numerically by means of appropriate mathematical software.

4. SiC-based composite materials for thermal management applications

4.1 SiC reinforcements

Nowadays it is easy to find many suppliers of silicon carbide in different forms (i.e. particles, fibers, sheets, rods, tubes) with a relatively huge variety of dimensions. The market of these products is related to abrasive or refractory applications. When speaking about fabrication of composite materials, different sources of silicon carbide, all finely divided (like particles or fibers), are mainly searched (Clyne, 2000a). Among all these forms of silicon carbide being used as reinforcements for composite fabrication, particles of different purity grades and shapes have been mostly used. The commonest particles are angular in shape (Fig. 2), although some have been processed with perfect spherical geometry. Though spherical shapes are always interesting for modelling purposes its use has been restricted due to its prohibitive price. SiC particles are mainly fabricated in two grades, differentiated by its colour: green powder for purities higher than 99% and black powder for purities in the range 98%-99%.

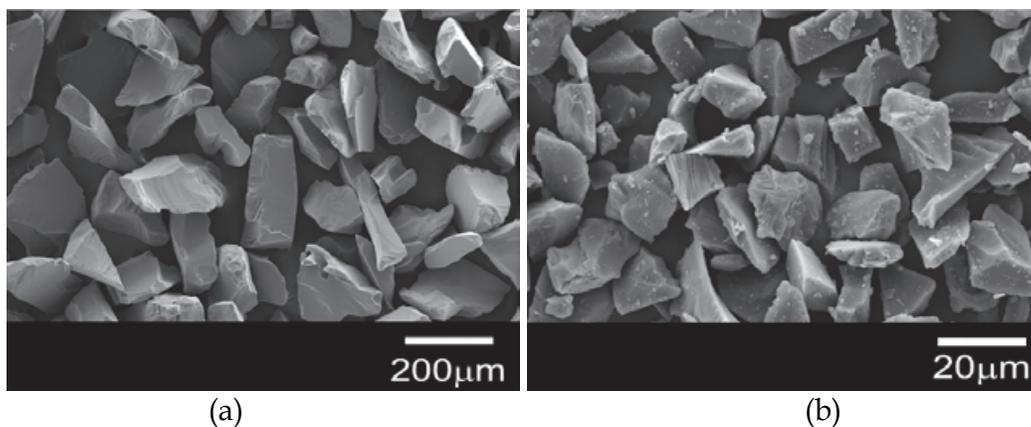


Fig. 2. Scanning electron microscopy images of commercial particles of SiC of two different average size diameters: (a) 170 μm (corresponding to 100 mesh; SiC100) and (b) 17 μm (corresponding to 500 mesh; SiC500)

An important parameter of particulate systems is the relation between their average diameter and specific surface area, measured by means of a laser diffraction technique and gas adsorption, respectively. The importance of this correlation will be shown later on in next sections.

Mathematically, these two intrinsic properties are related as follows:

$$D = \frac{6\lambda}{\rho \cdot S_r} \quad (7)$$

where D is the average particle diameter, ρ is the density of SiC (with a nominal value of 3210 kg/m^3), S_r is the specific surface area (m^2/kg) and λ is a geometrical factor introduced to account deviations from sphericity, surface roughness and particle size distribution. λ typically takes values greater than unity for angular particles prepared by milling of SiC blocks and turns out to be especially important for the modelling of different properties.

4.2 Metal matrix composites with reinforcement of mono or multimodal distribution of SiC particles

For years, the only materials able to mitigate the increasing demanding of heat removal in energy-related industrial applications have been composite materials made out of a proper combination of a metal and SiC particles.

Al/SiC composites have been extensively studied. Less studied materials are those with other metals such as copper or silver. We find only few references to those materials that point out their possible use in thermal management. Their potential as heat-dissipating materials is very high given the intrinsic properties of their constitutive elements. However, several problems arise when manufacturing these materials by liquid state processing routes. Estimates based on standard composite models indicate that the Cu/SiC system containing 50-60 vol.% SiC should yield materials with very good thermal properties: around 275 W/mK of thermal conductivity and 7-9 ppm/K of thermal expansion coefficient. Nevertheless, studies carried out on this system have demonstrated that those thermal properties are not achieved, mainly due to the detrimental reactivity at the metal-ceramic interface, which conducts to a solid solution of Si into the Cu matrix. An attempt to overcome the problem by protecting the SiC by an oxide layer has been demonstrated to be helpful but still not sufficient to attain the required properties (Narciso et al., 2006).

On its hand, the Ag/SiC system presents a problem of reactivity as well (Narciso et al., 1997, as cited in Molina et al., 2010). Ag dissolves a large amount of oxygen that can react with the SiC particles. The gas evolved in the reaction considerably reduces the metal/ceramic contact area and, hence, affects dramatically the thermal properties of the composite. The solution of oxidizing the SiC particles seems to improve, as in the case of Cu/SiC, the final properties of the material. However, for both systems, Cu/SiC and Ag/SiC, seemingly the only way to achieve the expected thermal properties by liquid state routes is to reduce the contact time between SiC particles and molten metal (details of the importance of the processing route will be given in following sections).

Given that in the Al/SiC system the problems related with reactivity are less accused, admissibly high thermal properties can be easily obtained. We will briefly summarize here two main kinds of materials obtainable when only SiC is used as reinforcement: composites based on a monomodal distribution of SiC particles and those based on bimodal distributions.

4.2.1 Al/SiC with monomodal distribution of particles

The traditional heat sinks in electronics, based on metallic combinations, promptly found a serious competitor in the Al/SiC family. These materials experimented a great impact on the industries and still nowadays represent the state-of-the-art in thermal management.

One of the most complete data collections on thermal properties of Al/SiC composites with monomodal distribution of particles can be accessed through (Arpon et al., 2003a; Arpon et al., 2003b; Molina et al., 2003a). The materials were fabricated by gas-pressure assisted infiltration of compacted SiC particles with molten Al at 750°C. Metallographic studies give evidence of the proper distribution of particles in the metal and the absence of particle breaking, given the special care taken during the packing procedure (Fig. 3).

The thermal properties achieved by those composites have been reviewed in the following table (Table 1). It is clearly seen that the dependence of the thermal properties on the material characteristics is different for thermal conductivity and for the coefficient of thermal expansion. While the former depends strongly on the volume fraction and average size of the SiC particles, the latter is only dependent on the volume fraction of SiC.

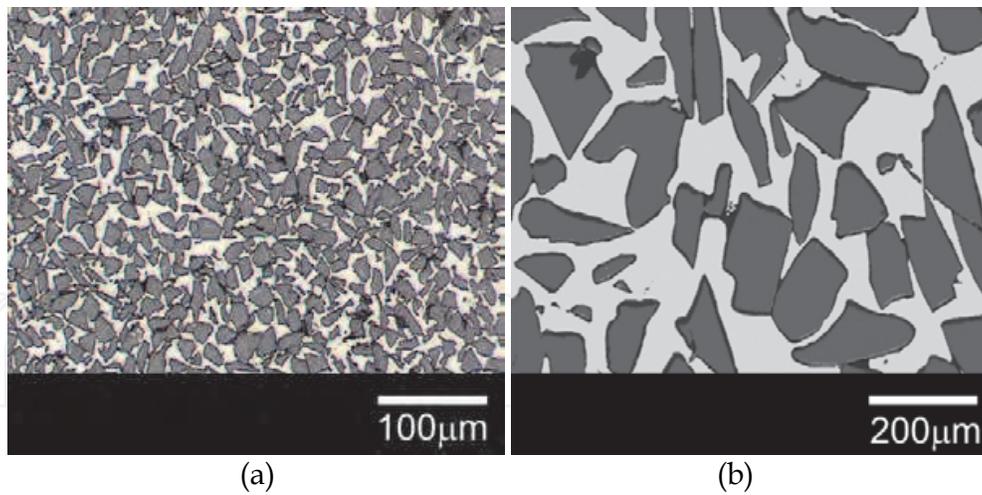


Fig. 3. Optical micrographs of cross sections of different samples of Al/SiC composites containing SiC particles of different average size diameters: (a) 17 μm (corresponding to 500 mesh; SiC500) and (b) 170 μm (corresponding to 100 mesh; SiC100)

SiC particles	V_r	D (μm)	TC (W/mK)	CTE (ppm/K)
SiC100	0.58	167	221	12.7
SiC180	0.58	86	209	13.2
SiC240	0.6	57	203	12.5
SiC320	0.59	37	204	12.9
SiC400	0.58	23	194	13.1
SiC500	0.55	17	193	13.4
SiC800	0.53	9	154	14.9

Table 1. Main characteristics of Al/SiC composites with monomodal distribution of particles; V_r is the particle volume fraction, D is the average size determined by laser diffraction, TC is the thermal conductivity (W/mK) and CTE refers to the thermal expansion coefficient

Modelling thermal conductivity in these composites is relatively simple. The two model schemes here reviewed are both applicable with similar results given that, as already explained, the phase contrast for aluminium and SiC is close to 1. In this sense, the plot in Fig. 4a might be obtained from Eq. (4) or Eq. (6) with almost indistinguishable results.

If now we speculate about how thermal conductivity can be improved in these composites, we may think that the thermal conductivity of the SiC filler cannot be increased (for a given quality purity) and that the interface thermal conductance between Al and SiC is indeed of the order $10^8 \text{ W/m}^2\text{K}$ (in agreement with other experimental works and calculated data with the acoustic mismatch model (Molina et al., 2009)). The largest potential comes from the metal, in improving the matrix conductivity to the value of pure aluminium (273 W/mK). In the experiments presented here, gas pressure infiltration was used to fabricate the composites. Seemingly, there was some chemical interaction between the metal and the quartz crucible during the time metal was molten prior to infiltration. The result of this

reaction was that metal was alloyed with Si and the matrix conductivity diminished down to a value as low as 185 W/mK (Molina et al., 2008b).

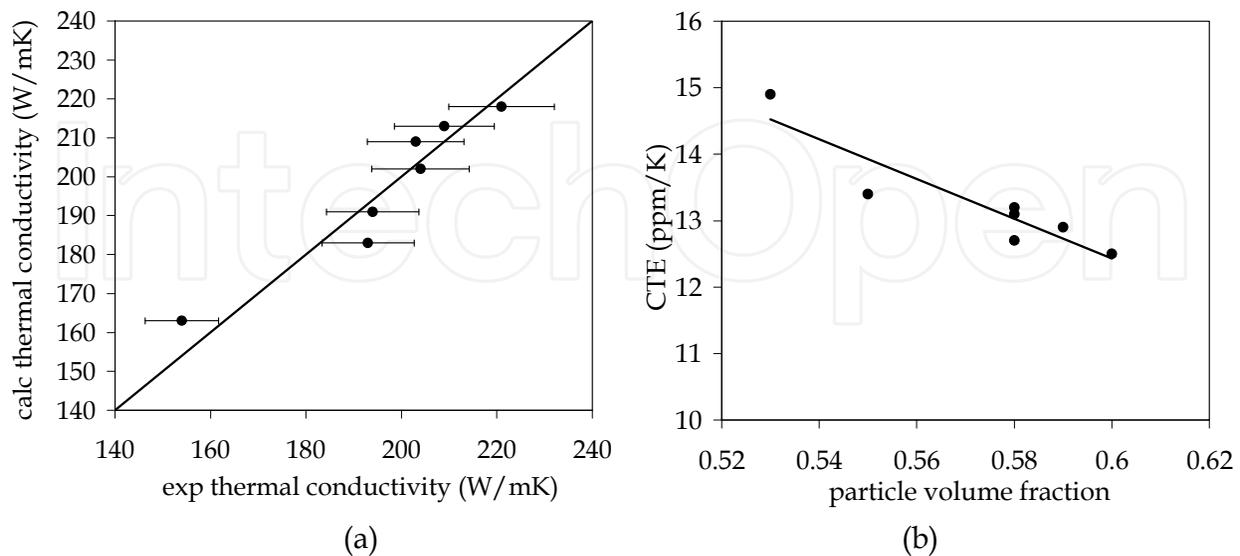


Fig. 4. Experimental results for Al/SiC composites containing SiC particles of different average size diameters: (a) thermal conductivity compared with predicted values obtained with the DEM scheme (the values taken for modelling are: $K_m = 185$ W/mK, $K_r^{in} = 253$ W/mK and $h = 7.5 \cdot 10^7$ W/m² K); and (b) coefficient of thermal expansion (α) versus volume fraction of reinforcement (the line corresponds to a fitting with equation $\alpha = -29.88V_r + 30.36$)

The coefficient of thermal expansion of the different composites versus the SiC volume content is shown in Fig. 4b. All composite materials showed a low value of thermal expansion, increasingly lower when the SiC content is increased. The linear relation encountered for this property with the SiC volume fraction reinforces the conclusion that CTE is only dependent on the volume fraction of the different phases present, in qualitative agreement with most analysis of this property (Arpon et al., 2003a; Molina et al., 2003a), no matter the size, shape or size distribution of the particles.

4.2.2 Al/SiC with bimodal distribution of particles

For certain high-demanding electronic applications Al/SiC composites based on a monomodal distribution of SiC particles were unable to properly operate as admissible heat sinks. For those applications where the heat generated was in excess and/or the material had to accomplish with very restricted thermal expansion, new methods for increasing the volume fraction of SiC particles in the metal were developed.

It has been established that the minimum volume fraction needed for these composites is about 0.6. The logical reasoning to achieve high volume fractions of particles consists of mixing and packing particles of largely different sizes (size ratio >10). During the past few years diverse groups have been working in the ceramic particle packing of multimodal mixtures (Molina et al., 2002).

A very simple and intuitive model for the prediction of particle packing was presented and validated in (Molina et al., 2002) and the main features can be summarized as follows. When

a bimodal mixture is considered (two different particle sizes), the following model with two limits mathematically separated can be applied:

a. Fine particle end

$$V_r = \frac{1}{X_{cp} + \frac{1 - X_{cp}}{V_{sp}}} \quad (8)$$

b. Coarse particle end

$$V_r = \frac{V_{cp}}{X_{cp}} \quad (9)$$

where V_r is the total volume fraction; V_{cp} and V_{sp} are the volume fractions in compacts of only coarse and small particles, respectively; and X_{cp} is the fraction of coarse particles in the powder mixture.

Other mathematical schemes, like that proposed by Yu and Standish (Yu & Standish, 1987, as cited in Molina et al., 2002), make use of experimental information through an empirical parameter with adjustable value. Fig. 5 shows experimental results of packing obtained with mixtures of two largely different sizes of SiC (17 μm and 170 μm). Fig. 5 puts in clear evidence that the particle volume fraction of the compacts made from bimodal mixtures can be tailored by simply controlling the relative amount of large and small particles. Moreover, the experimental results for the particle volume fraction are rationalized using the simple model (Eq. (8) and Eq. (9)) that assumes that small particles can be easily accommodated in the free space left by large particles. Therefore, the local compactness of fine particles is nearly constant for coarse particle percentages up to that for which the maximum global compactness is reached.

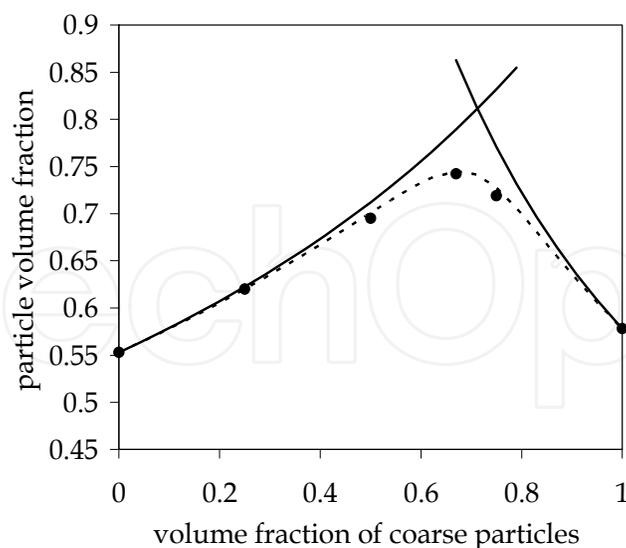


Fig. 5. Volume fraction of particles attained in packed preforms of a bimodal mixture of SiC formed by angular particles of 17 μm and 170 μm . The continuous lines correspond to the two branches of the simple geometrical model described by Eq. (8) and Eq. (9); the broken line refers to a semiempirical scheme proposed by (Yu & Standish, 1987, as cited in Molina et al., 2002).

For gas-pressure Al-infiltrated bimodal preforms, optical microscopy revealed that, up to a 60% of coarse particles, the SiC particles appear homogeneously distributed in the Al matrix, being the coarse particles embedded in a bed of small particles (Fig. 6a). For higher percentages of coarse particles, the spaces between the touching large particles were somewhat irregularly filled with the small particles (Fig. 6b).

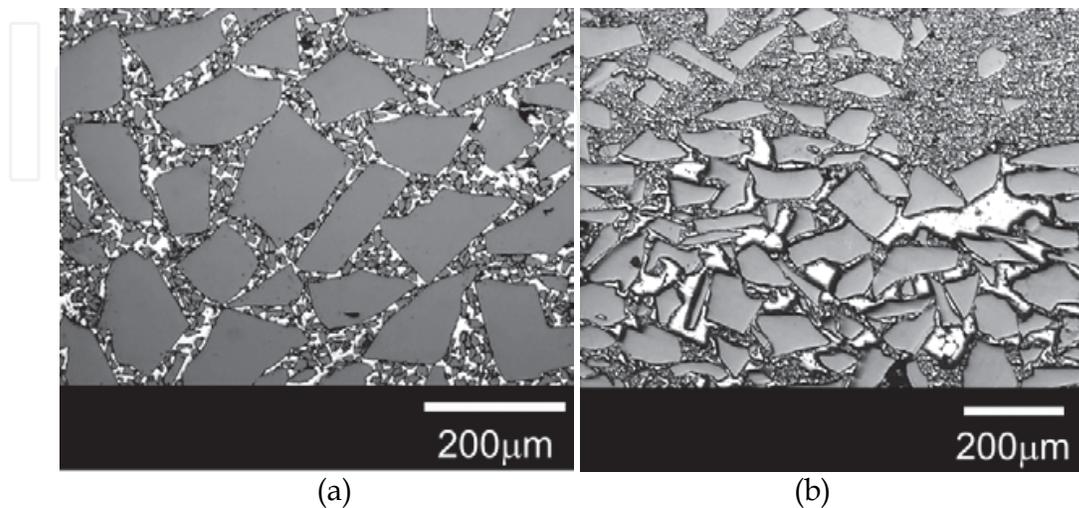


Fig. 6. Optical micrographs of cross sections of different samples of Al/SiC composites containing a mixture of SiC particles of two largely different sizes (17 μm and 170 μm) in different proportions: (a) 67% coarse particles + 33% small particles; and (b) 75% coarse particles + 25% small particles, showing inhomogeneous filling

The main thermal properties of composite materials obtained by gas pressure infiltration of ceramic preforms containing a bimodal mixture of SiC particles (170 μm + 17 μm) with pure Al are gathered in Table 2.

% of coarse particles	V_r	TC (W/mK)	CTE (ppm/K)
25	0.62	215	10.8
50	0.69	220	9.0
67	0.74	228	7.8
75	0.72	225	8.4

Table 2. Main characteristics of Al/SiC composites with bimodal distribution of particles (SiC100/SiC500); V_r is the particle volume fraction, TC is the thermal conductivity (W/mK) and CTE refers to the coefficient of thermal expansion (ppm/K)

The experimental results of thermal conductivity for these composites are plotted in Fig. 7 versus the percentage of coarse particles. The results follow a trend similar to that of particle volume fraction (see Fig. 5) illustrating the close relationship between the two magnitudes. The figure shows as well the predictions offered by the DEM model, Eq. (6), for three different series of inputs (Molina et al., 2008b).

The best results are obtained when the model takes into account that, instead of inserting in Eq. (5) the average particle diameter D measured by laser diffraction, the characteristic length scale of the finely divided powder might be better represented by a calculated value of D obtained from the specific surface area of the particles (Eq. (7)) (Molina et al., 2005). When considering bimodal mixtures, a proper length scale characteristic of the particle reinforcement might be obtained by averaging the ratio λ/S with the volume fraction for each present particle, as follows:

$$D = \frac{6}{\rho} \left[\frac{\lambda_{cp}}{S_{cp}} V'_{cp} + \frac{\lambda_{sp}}{S_{sp}} V'_{sp} \right] \quad (10)$$

V' is the volume fraction of a given particle size in the mixture; the subscripts cp and sp refer to coarse and small particles, respectively. Based on experimental measurements the values of λ deduced for the two particle types here considered are: $\lambda_{cp}=6.8$ and $\lambda_{sp}=2.7$; the values used for the specific surface area are: $S_{cp}=91 \text{ m}^2/\text{kg}$ and $S_{sp}=338 \text{ m}^2/\text{kg}$.

Fig. 8a reports results for the coefficient of thermal expansion versus particle volume fraction for all Al/SiC composites presented in Table 1 and Table 2 containing either a single particle size or bimodal particle distributions. As shown in the figure the whole set of data can be satisfactorily accounted for by means of a linear fitting. As already pointed out in the former section, this indicates that what matters, as long as the CTE is concerned, is the global particle volume fraction, neither the particle size and shape nor the particle size distribution (Arpon et al., 2003a).

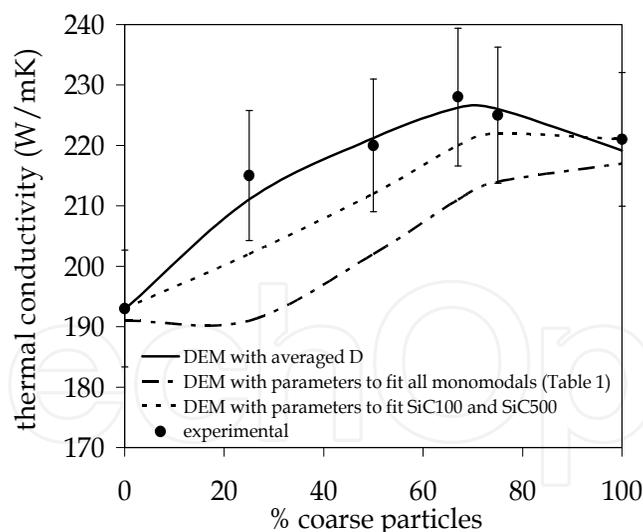


Fig. 7. Thermal conductivity of Al/SiC composites obtained by gas pressure infiltration of aluminium into preforms of bimodal SiC mixtures (SiC100/SiC500) versus the percentage of coarse particles. Dots correspond to experimental results; the dashed lines are different calculations within the DEM scheme. The values taken for modelling are: $K_m = 185 \text{ W/mK}$, $K_r^{in} = 253 \text{ W/mK}$ and $h = 7.5 \cdot 10^7 \text{ W/m}^2 \text{ K}$

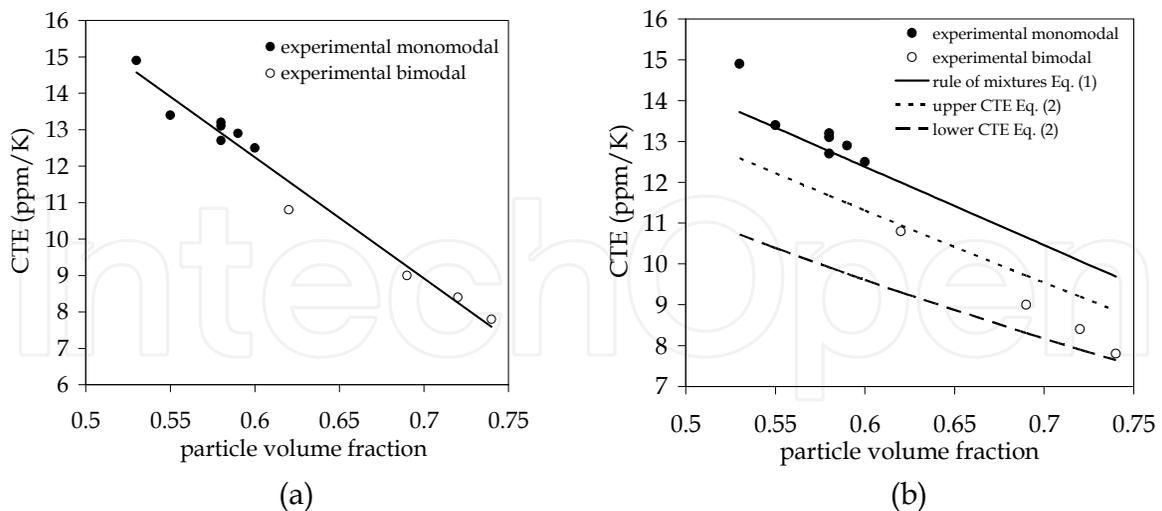


Fig. 8. Thermal expansion coefficient in the range 50-300°C versus particle volume fraction for Al/SiC composites having either a single particle size (full circles) or bimodal particle distributions (empty circles): (a) fitting by a straight line of equation $\alpha = -33.20V_r + 32.16$; (b) theoretical results predicted by different models

Different theoretical results for the coefficient of thermal expansion have been depicted in Fig. 8b for composites containing mono- and bimodal distributions of SiC particles. The present discrepancies between experimental results and predictions can be qualitatively understood by noting that those two models have been developed using thermoelasticity theory. In this sense, experimental results of CTE obtained at lower temperatures might be closer to the predictive values, given that plastic deformation becomes greater at higher temperatures. These results, in consequence, illustrate the failure of the thermoelasticity framework upon which most theories are based.

4.3 Metal matrix composites with a hybrid mixture of reinforcement (SiC+ceramic)

4.3.1 Metal/SiC-diamond composites

Being diamond a very interesting ceramic due to its superior thermal properties its use as particulate in metal matrix composites for thermal management applications has been largely evaluated. Many have been the composites fabricated with excellent thermal properties. However, these composites have no lack of some disadvantages: the price of diamond particles makes the material cost too high and machinability is very difficult or even impossible. In an attempt to overcome some of these problems and at the same time combine the unique properties of diamond and SiC together, a new material containing both types of ceramics was presented in (Molina et al., 2008a). The proposal consists of using as reinforcement a mixture of diamond and SiC particles of similar size. The preform is afterwards consolidated with Al by gas pressure infiltration. The aim of using this particular mixture of ceramics is to achieve a decrease in the material costs maintaining the thermal conductivity at a certain high value. In Fig. 9a it is shown the predicted property of thermal conductivity for three different combinations of SiC and diamond particles. As clearly shown, there is no big advantage in using a mixture of SiC100 (170 μm) and diamond-500 (16 μm) in respect to the use of monomodal SiC100 (170 μm) particles. Even when only the most conducting diamond phase is present, the thermal conductivity of the composite is still

very low, given that interface resistance becomes very important for the small particle size here considered (16 μm). A different situation is found when mixtures of diamond-100 (170 μm) and SiC500 (16 μm) are considered, for which the maximum thermal conductivity is about 550 W/mK (that corresponds to a monomodal diamond particle distribution of 170 μm of average diameter). Fig. 9a also includes the calculation for a system in which SiC and diamond particles have the same average diameter (170 μm).

In (Molina et al., 2008a) powder mixtures containing 20, 40, 50, 60, 70, 80 and 90 vol.% of diamond (of about 200 μm of average diameter) were prepared by dry mixing with SiC (of same average size). Given the similarity in shape and size of the diamond and the SiC particles, the relative density of the preforms was in the range of 58 ± 1 vol.% for all samples.

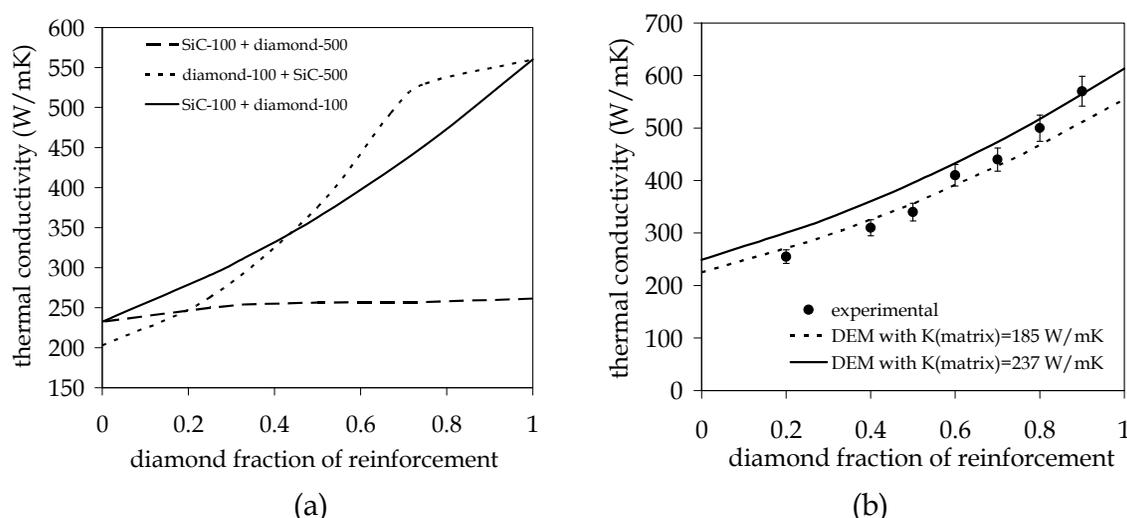


Fig. 9. Thermal conductivity of aluminium-matrix composites containing different mixtures of diamond and SiC particles: (a) calculated values with the DEM scheme for three sets of mixtures; (b) experimental and calculated values with the DEM scheme for mixtures of powders of similar size (200 μm) for two intrinsic values of matrix thermal conductivity

	SiC	Diamond
Intrinsic thermal conductivity (W/mK)	259	1450
Interface thermal conductance with pure aluminium (W/m ² K)	1.05×10^8	5×10^7

Table 3. Parameters used in the calculations of Fig. 9 for the intrinsic thermal conductivity and interface thermal conductance with pure aluminium for SiC and diamond

The thermal conductivity of these composites was characterised and the results are gathered in Fig. 9b, along with the prediction obtained with the generalized differential effective medium scheme. The composites fabricated with this combination of ceramics exhibit high thermal conductivities for the whole range of ceramic proportions. Modelling in Fig. 9 has been developed within the parameters collected in Table 3.

Fig. 9b shows two predictive lines corresponding to different inputs for the intrinsic thermal conductivity of the metal. While for the SiC-rich mixture content end a thermal conductivity of the matrix of 185 is needed to fit experimental data, this value increases to the thermal conductivity of pure aluminium (237 W/mK) for those mixtures with predominant presence

of diamond particles. The reason behind this is that the thermal conductivity of the nominally pure aluminium matrix is influenced by take-up of some silicon from the silicon carbide. It was shown in (Molina et al., 2008b) that the intrinsic value of the thermal conductivity of pure aluminium in composites fabricated via gas-pressure infiltration might be as low as 185 W/mK, because of the presence of silicon in solid solution in combination with precipitated silicon phase in the matrix. Reactivity of liquid aluminium and SiC in the “as-received” condition seems unavoidable in gas-pressure infiltration, since the time elapsed during pressurization of the chamber and posterior solidification is of the order of some minutes, depending on the special characteristics of the equipment at hand. Decreasing as much as possible the infiltration temperature seems then to be a successful way to avoid metal-ceramic reactivity in SiC-based systems.

4.3.2 Metal/SiC-graphite flakes composites

A very recent family of composite materials has been developed and patented at the University of Alicante (Narciso et al., 2007; Prieto et al., 2008). The invention is concerned with a composite material with high thermal performance and low cost which has a layered structure achieved by proper combination of different components. The components of the material are three: 1) a phase mainly formed by graphite flakes (phase A); 2) a second phase (phase B) involving particles or fibers of a material which can act as a phase separator of phase A (phase B is a ceramic material preferably selected from the group of SiC, BN, AlN, TiB₂, diamond and carbon fibers); and finally, 3) a third phase (phase C) formed by a metallic alloy. The three present phases must have good thermal properties, although their main function is different for each one: phase A (graphite flakes) is the principal responsible of the properties of the final material, phase B acts as a separator of the layers of phase A and phase C has to consolidate the preform.

The resultant layered structure of these composites is mainly due to the fact that graphite flakes naturally tend to lie on top of each other, especially when a given pressure is applied. It is in fact due to this tendency to get densely packed that, when only flakes are present, they almost leave no space between them and infiltration becomes an almost unfeasible task. The presence of another ceramic (phase B), like SiC particles, allows molten metal infiltration by keeping the graphite flakes separated. The feasibility of the fabrication procedure of these composites was demonstrated in (Prieto et al., 2008).

A representative illustration of the microstructures of these composites is given in Fig. 10.

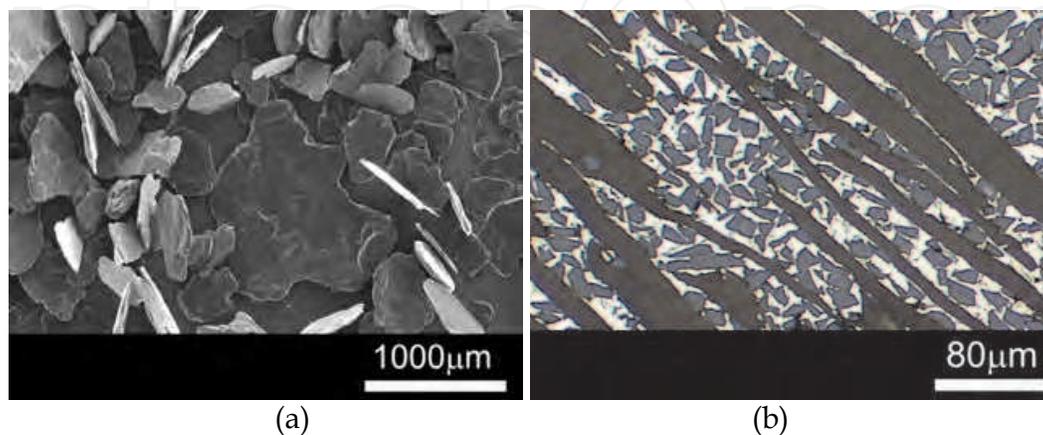


Fig. 10. Micrographs of: (a) graphite flakes and (b) composite material obtained by infiltration with Al-12%Si of preforms obtained by mixing graphite flakes (60%) with SiC particles (40%)

Table 4 shows the most significant results of the thermal properties for some of these materials. These composites recently developed present exceptionally high values of thermal conductivity. The thermal properties are clearly anisotropic, given the fact mentioned before that graphite flakes are oriented in a plane, for which they exhibit the maximum thermal conductivity and the lowest CTE (xy-plane in Table 4).

Reinforcement	Metal	V_r	CT (W/mK)	CTE (ppm/K)
60% graphite flakes + 40% SiC	Al-12%Si	0.88	xy: 368 z: 65	z: 11 xy: 7.0
63% graphite flakes + 37% SiC	Ag-3%Si	0.88	xy: 360 z: 64	z: 11 xy: 8.0

Table 4. Thermal properties of metal/SiC-graphite flakes composites. xy refers to the graphene planes, while the direction perpendicular to it is denoted by z

Although the properties presented in Table 4 are very good, the authors of the patent (Narciso et al., 2007) have already encountered even more promising values when special conditions of infiltration are used. The thermal properties of these composites are currently being evaluated by means of different modelling schemes, conveniently adapted to account for both the anisotropic microstructure of the materials at the mesoscale and the anisotropy in the intrinsic thermal properties of the graphite flakes. Modelling on this system arouses special interest since it is a very cheap and machinable material which has attracted the interest for many applications and represents a clear alternative candidate for heat sinking.

5. Selection of processing conditions for fabrication of SiC-based composite materials for thermal management

The thermal properties of composite materials are mainly determined by the intrinsic properties of their constituents and the characteristics of the matrix-reinforcement interface. Aside an appropriate selection of the constituents it is essential to control the processing conditions during fabrication of the material in order to generate a proper interface able to effectively transfer the heat across the different constituent phases. During fabrication of a composite material by infiltration or squeeze casting some of these processing conditions concern:

- i. Ceramic particulate (average diameter, size distribution, shape, packed volume fraction)
- ii. Liquid metal (surface tension, viscosity)
- iii. Liquid-solid interface (wettability, reactivity)
- iv. Experimental variables (maximum applied pressure, pressurization rate, temperature, infiltration atmosphere)

Next sections will focus on different aspects in regard to the most important parameters that mainly determine the thermal properties of SiC-based composite materials: wettability-reactivity at the liquid-solid interface, maximum applied pressure and pressurization rate.

5.1 Threshold pressure for infiltration

An outstanding technologically relevant parameter in composite materials processing is the threshold (P_0), or minimum, pressure to achieve the entrance of the molten metal into the porous preform. Being essential for materials validation, its measurement is, however, not simple. One of the methods to get the threshold pressure of a given system is to infiltrate a

perform at various applied pressures and measure, for a fixed time, the infiltrated height for each pressure (Garcia-Cordovilla et al., 1999; Molina et al., 2004; Molina et al., 2008; Piñero et al., 2008). Data are then analysed by means of Darcy's law:

$$h^2 = \frac{2k \cdot t}{\mu \cdot (1 - V_r)} (P - P_0) \quad (11)$$

where k is the permeability of the porous solid, t is the infiltration time and μ the viscosity of the liquid metal. P_0 can be easily derived from plots of h^2 vs P .

Threshold pressure and contact angle are intimately correlated by means of the so-called capillary law:

$$P_0 = 6\lambda \cdot \gamma_{lv} \cdot \cos \theta \frac{V_r}{(1 - V_r) \cdot D} = \gamma_{lv} \cdot \Sigma \cdot \cos \theta \quad (12)$$

being θ the contact angle and γ_{lv} the surface tension of the molten metal at the infiltration temperature. The value of P_0 is clearly dependent on the wetting characteristics of the system and, hence, may be strongly affected by the reactive phenomena occurring at the interface between metal and substrate while the infiltration front moves over the substrate surface. The study of this parameter becomes especially interesting for those systems where infiltration front movement and reaction cannot be decoupled in time. A remarkable fact that has to be taken in consideration is that if infiltration occurs too rapidly, reaction could be prevented and the system may behave as non-reactive. A conclusive study regarding these points was presented in (Molina et al., 2007b; Tian et al., 2005), which discusses results for infiltration of pure Al and Al-12wt%Si into compacts of as-received and thermally oxidized SiC particles. The main results of this study are summarized in Fig. 11a.

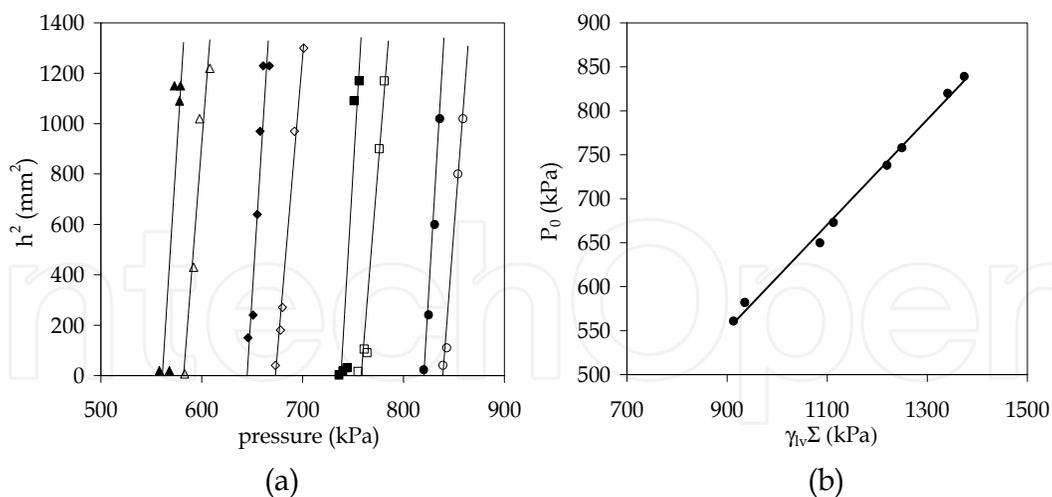


Fig. 11. (a) Plots of the square of the infiltrated height h^2 as a function of applied pressure P for gas pressure infiltration at 700°C of Al and Al-12%Si in preforms of SiC particles in the as-received and oxidized conditions: Al/SiC500 (○), Al-12%Si/SiC500 (●), Al/SiC500ox (□), Al-12%Si/SiC500ox (■), Al/SiC400 (◇), Al-12%Si/SiC400 (◆), Al/SiC400ox (△) and Al-12%Si/SiCox (▲). The straight lines are linear fittings of experimental data; (b) Threshold pressure P_0 versus $\gamma_{lv} \Sigma$ for the different systems in (a). The line corresponds to a fitting with equation $P_0 = 0.603 \gamma_{lv} \Sigma + 32.9$ kPa

The most important conclusion is that the contact angle derived from a fitting of the experimental data (Fig. 11b) by means of Eq. (12) is the same for all cases studied. The infiltration behaviour of the different systems, governed by a unique contact angle, indicates that the metal/particle interface is in both cases the same. Instead of being a metal/SiC contact, there exists an interlayer of silica between both. The very thin silica layer that covers naturally the SiC particles seems to be thick enough to partly remain after reaction with the metal during infiltration at these low temperatures and relatively rapid infiltration kinetics. Another system with remarkable interest is Ag/SiC (Garcia-Cordovilla et al., 1999; Molina et al., 2003b). Silver is a metal with high capacity for dissolution of oxygen in the molten state. This oxygen can rapidly oxidize the SiC particles. This was observed to affect directly the threshold pressure of the system by increasing its value. The apparent contact angle derived from the data was 168°. The authors suggested that the gas evolved during the oxidation of SiC reduced the contact area and, in consequence, wetting.

5.2 Drainage curves for gas-pressure infiltration

Determination of threshold pressures is often not sufficient to fully characterize wetting in infiltration processing. Intrinsic capillary parameters, characteristic of dynamic wetting of a discrete reinforcement, are not, per se, equal to those derived in near-static conditions (i.e. sessile drop measurements). Furthermore, preforms are invaded over a range of pressures that is governed by the complex internal geometry of open pores within the preform (Rodriguez-Guerrero et al., 2008). A more thorough characterization of wetting is obtained by the so-called drainage curves. These are plots of the metallic saturation (fraction of non-wetting fluid in the porous medium) versus the pressure difference between the fluid and the atmosphere in the pores. These drainage curves contain all information related to wetting of the porous preform by the non-wetting liquid. With the assumption that irreversibility effects (e.g. Haines jumps) and other inertial losses can be neglected, the work of immersion (W_i) can be calculated as the work necessary to fully infiltrate the preform (W) divided by the total preform/infiltrant interface created per unit volume of reinforcement:

$$W_i = \frac{W}{A_v \cdot V_r} = \frac{(1 - V_r) \cdot \int P dS}{A_v \cdot V_r} \quad (13)$$

where P , S and V_r are saturation, applied pressure and volume fraction of reinforcement, respectively; A_v is the particle specific surface area per unit volume of preform. The contact angle can be easily derived by making use of the following relationship:

$$W_i = \gamma_{lv} \cdot \cos \theta \quad (14)$$

Recently, a new technique was proposed for the direct measurement of capillary forces during the infiltration process of high-temperature melting non-wetting liquids into ceramic preforms. In essence, the equipment is a high-temperature analogue of mercury porosimetry. The device can track dynamically the volume of metal that is displaced during pressurization and hence allows obtaining in a single experiment the entire drainage curve characterizing capillarity in high-temperature infiltration of particles by molten metal (Bahraini et al., 2005; Bahraini et al., 2008; Molina et al., 2007a; Molina et al., 2008d). The technique was validated in an study of wetting of silicon carbide by pure aluminium and by aluminium-silicon eutectic alloy using drainage curves obtained during gas pressure infiltration at 750°C.

With relatively fast pressurization rates the drainage curves for a metal/SiC system that can be obtained are shown in Fig. 12 for SiC320 particles of about 37 μm of average diameter.

The shape of the curves is determined by the shape of particles in the preform. Any change of (i) the work of immersion, (ii) the particle volume fraction and/or (iii) the particle size (which is accounted for by A_v parameter) will cause a predictable shift over the pressure axis. The values of contact angle derived from the drainage curves for different sizes of SiC particles with Al and Al-12%Si are in the range 110-113°. These values are fully consistent with measurements with the sessile drop method for the wetting of oxide-covered SiC by molten aluminium free of a surface layer of oxide. In these infiltration experiments the triple line is forced to move at a motion rate which is well above the “natural” rate dictated by reaction kinetics in the sessile drop method. Hence, infiltration and reaction processes are decoupled in time and the SiC surface is covered before reaction can take place at the interface.

Nevertheless, when pressurization rate is decreased, the interfacial reaction can take place concomitantly with the motion of the triple line and both phenomena may interact to provoke different behaviours in drainage curves. Fig. 13 shows drainage curves for the infiltration of SiC particles with molten Al and Al-Si eutectic at a reduced pressurization rate of 0.05MPa/s together with the curves obtained for the same systems at 0.13 MPa/s.

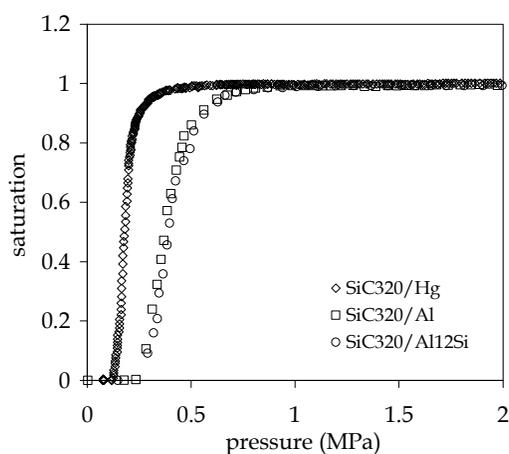


Fig. 12. Drainage curves of SiC320 infiltrated with Hg, Al and Al-12%Si at 750°C

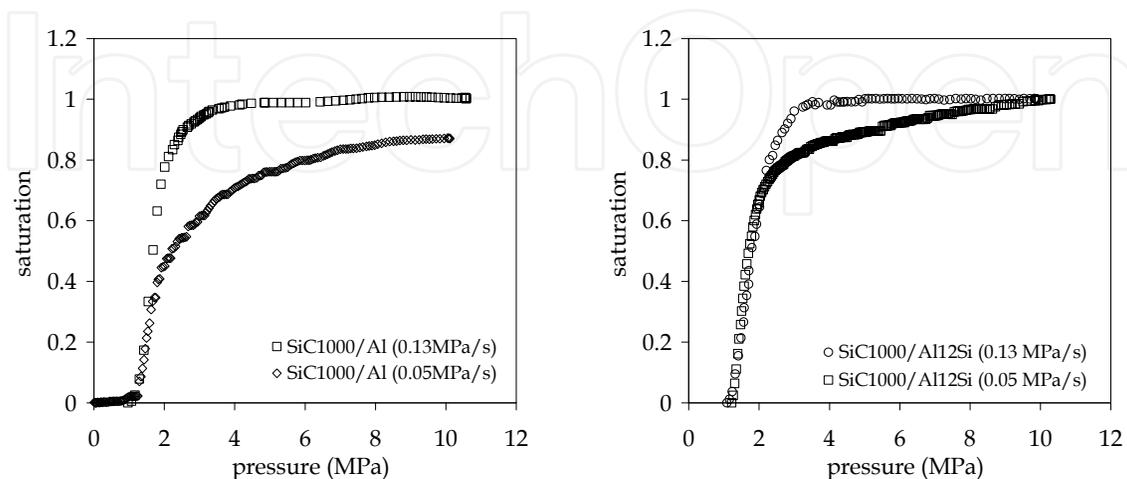


Fig. 13. Drainage curves at 750°C of (a) SiC1000/Al and (b) SiC1000/Al-12%Si, measured at the two pressurization rates of 0.13 and 0.05 MPa/s

The curves of Fig. 13 show that interfacial reactions, which have proven in sessile-drop experiments to aid wetting, under forced pressure-driven infiltration can hinder infiltration of SiC preforms by aluminium-based melts. These effects can be due to the fact that chemical interactions can cause morphological changes at the solid/liquid interface. As a corollary, rapid pressure infiltration is preferable in processing metal matrix composites featuring interfacial reactivity.

5.3 Gas pressure infiltration vs squeeze casting

It is interesting to compare the resulting materials processed by two different liquid-state routes, namely gas-pressure infiltration and squeeze casting, which make use of different pressures and pressurization rates (this having a direct implication on the contact time between molten metal and particles before metal is solidified).

In (Weber et al., 2010) it is presented a complete study of comparison of the different properties encountered for Al/SiC composites processed by these two fabrication techniques. In this work, bimodal powder mixtures of green quality SiC powders with average sizes of 170 μm and 17 μm , respectively, were used. A set of samples was processed by squeeze casting while other two sets were prepared by gas pressure-assisted infiltration at two largely different infiltration kinetics. Fig. 14a resumes the thermal conductivities for both series of composites together with modelling predictions using the DEM scheme.

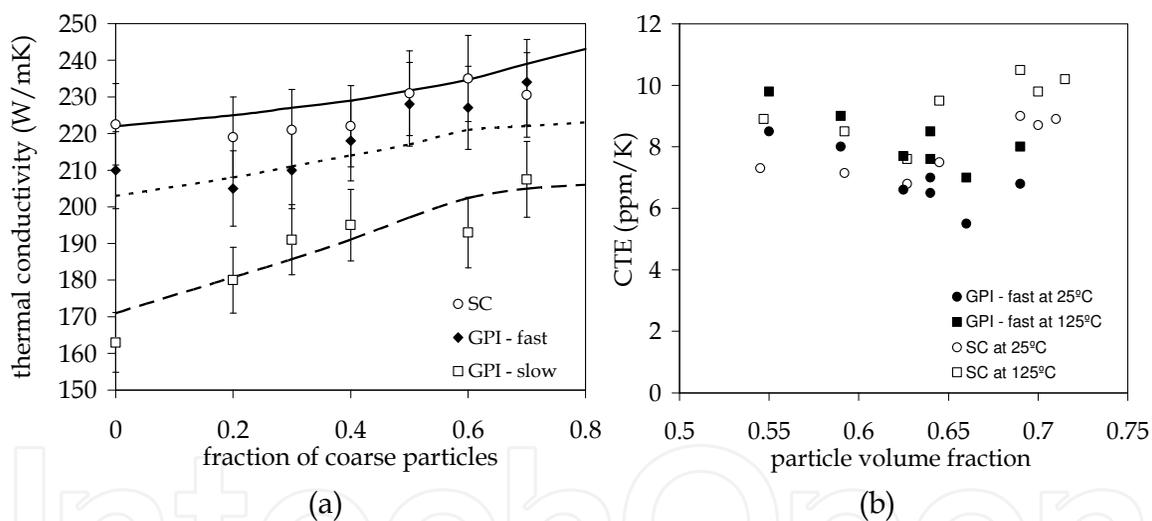


Fig. 14. (a) Thermal conductivity of the Al/SiC composites produced by infiltration of aluminium into preforms of mono- and bimodal SiC mixtures (SiC100/SiC500) versus the fraction of coarse particles. SC refers to squeeze casting and GPI refers to gas pressure infiltration. The lines correspond to calculations with the DEM scheme; (b) Coefficient of thermal expansion for the Al/SiC composites in (a) for temperatures of 25°C and 125°C

For the squeeze cast samples, thermal conductivity was in between 225 and 235 W/mK with a slight tendency to increase with the amount of large particles. For the samples prepared by fast GPI, the thermal conductivity increased from around 200 W/mK for the composite containing only small particles with increasing fraction of large particles up to 230 W/mK. For the slow GPI samples, values increased from 160 to 205 W/mK with increasing fraction of large particles. For the modelling of thermal conductivity, different matrix conductivities have been taken into account. While for SC samples the matrix conductivity is that of pure

aluminium (237 W/mK) due to the lack of time to react with the reinforcement, for the GPI samples values of 190 W/mK and 170 W/mK for GPI fast and GPI slow, respectively, have been used. Interestingly enough, the interface thermal conductance varies as well its value with the contact time corresponding to each processing technique. For SC and fast GPI the interface thermal conductance is found to be 1.4×10^8 W/m²K. For the slow GPI this parameter has a value which is about the half, most probably due to the abundant reaction product (Al₄C₃) at the interface (Weber et al., 2010).

The results of the CTE measurements are collected in Fig. 14b. The physical CTEs (measured in a range of $\pm 5^\circ\text{C}$ around the indicated temperature) are given for the SC and the fast GPI samples only, yet for two temperatures of technical interest, i.e., 25°C and 125°C. The CTE decreases in general with increasing SiC volume fraction and is typically 1–1.5 ppm/K higher at 125°C than at ambient temperature.

5.4 Effect of porosity

In a non-wetting system like Al/SiC infiltration of the metal into the open channels of the preform does not take place at a single, well-defined pressure but, as already seen, it rather takes place progressively with the applied pressure when this pressure exceeds a certain threshold (threshold pressure). In order to obtain a hundred percent filling of the porous space of the preform by the metal an infinitely large pressure, impossible to obtain in laboratory, would be needed. For a given infiltration pressure, therefore, defects at the contact area of particles will exist and porosity will hence be unavoidable.

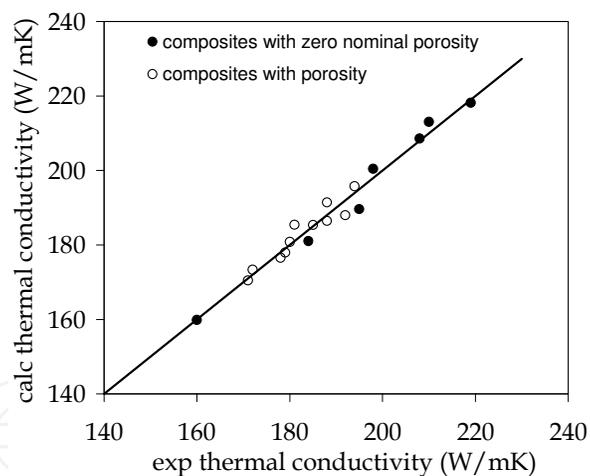


Fig. 15. Plot of the thermal conductivity calculated with the two-step Hasselman-Johnson model versus that determined experimentally. The line represents the identity function. DEM scheme offers identical results

Porosity does affect the two main properties which are important in materials for thermal management and, hence, may limit its use for this application. Depending on the nature of both, metal and reinforcement, voids in the material may increase or decrease the coefficient of thermal expansion of the composite material, being this effect very dependent on the geometry of the pores. On the other hand, the presence of porosity does decrease strongly the thermal conductivity of any material, being monolithic or composite. The voids, present in the metallic phase, can be treated as inclusions of zero conductivity in the metal. In a

recent paper (Molina et al., 2009) it has been demonstrated that a simple application of the Hasselman-Johnson model in a two-step procedure (which accounts for the presence of two types of inclusions, reinforcement particles and voids, and the metallic matrix) offers a good approximation of the experimental results of thermal conductivity obtained for Al-12%Si/SiC composite materials. Alternatively, the DEM model may be used as in (Molina et al., 2008a; Molina et al., 2008b) for accounting for the two types of inclusions (SiC particles and pores) at the time. Results of both models are equivalent since the phase contrast in the Al/SiC (or Al-Si/SiC) system is too low. It has been recently demonstrated (Tavangar et al., 2007) that the Hasselman-Johnson scheme increasingly offers inconsistent predictions for the thermal conductivity of composites as the effective phase contrast - ratio between effective thermal conductivity of reinforcement and matrix thermal conductivity - exceeds roughly four.

6. Conclusion

Several composite materials containing SiC as reinforcement, either single or combined with other ceramics, have been presented as serious candidates to cover the specific demand of heat dissipation for thermal management applications. Aside from the metal/SiC composites with monomodal distribution of SiC particles, which nowadays define the state of the art in materials for electronics, those derived from combinations of SiC with either SiC of another largely different size (bimodal mixtures) or other ceramics (hybrid mixtures with diamond or graphite flakes) present high values of thermal conductivity and coefficients of thermal expansion extremely low such as to represent the future generation of heat sinks for electronics. The use of these composites is mainly determined by the specific requirements for every application, taking into account not only the thermal properties but also density, isotropy or ease of machinability (when complex shapes are needed). The spectrum covered by the SiC-based composites aims to offer specific solutions for the different problems of heat dissipation encountered in the energy-related industries such as electronics or aeronautics.

This contribution emphasizes the fact that the choice of a proper fabrication processing is as important as a good selection of the constituents of the composite material. Being aluminium a very used metal for the fabrication of SiC-based composites, processing by liquid state routes must take into account the high reactivity between Al and SiC at the temperature of molten aluminium. In these sense, squeeze casting, which operates allowing very short contact times between metal and reinforcement, offers composites with the highest values of thermal conductivity. Several specific conditions should be taken into account in gas pressure infiltration to give appropriate materials with acceptable thermal properties. In any case, porosity has to be avoided because dramatically decreases the thermal conductivity of the materials. For this purpose, a certain minimum pressure that ensures complete saturation is needed along with a certain pressurization rate in order to force that infiltration and reactivity can be decoupled in time, since interfacial reaction can hinder infiltration.

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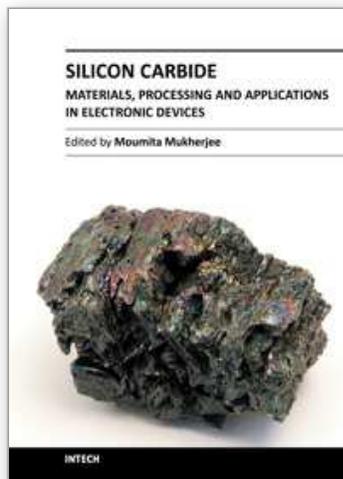
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Silicon Carbide (SiC) and its polytypes, used primarily for grinding and high temperature ceramics, have been a part of human civilization for a long time. The inherent ability of SiC devices to operate with higher efficiency and lower environmental footprint than silicon-based devices at high temperatures and under high voltages pushes SiC on the verge of becoming the material of choice for high power electronics and optoelectronics. What is more important, SiC is emerging to become a template for graphene fabrication, and a material for the next generation of sub-32nm semiconductor devices. It is thus increasingly clear that SiC electronic systems will dominate the new energy and transport technologies of the 21st century. In 21 chapters of the book, special emphasis has been placed on the “materials” aspects and developments thereof. To that end, about 70% of the book addresses the theory, crystal growth, defects, surface and interface properties, characterization, and processing issues pertaining to SiC. The remaining 30% of the book covers the electronic device aspects of this material. Overall, this book will be valuable as a reference for SiC researchers for a few years to come. This book prestigiously covers our current understanding of SiC as a semiconductor material in electronics. The primary target for the book includes students, researchers, material and chemical engineers, semiconductor manufacturers and professionals who are interested in silicon carbide and its continuing progression.

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InTech Europe

University Campus STeP Ri
Slavka Krautzeka 83/A
51000 Rijeka, Croatia
Phone: +385 (51) 770 447
Fax: +385 (51) 686 166

InTech China

Unit 405, Office Block, Hotel Equatorial Shanghai
No.65, Yan An Road (West), Shanghai, 200040, China
中国上海市延安西路65号上海国际贵都大饭店办公楼405单元
Phone: +86-21-62489820
Fax: +86-21-62489821

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