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Interfaces in Discontinuously Reinforced Metal-matrix Composites

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ABSTRACT

The paper overviews the fundamental and engineering aspects of the interfaces in discontinuously reinforced metal-matrix composites. The major findings of studies undertaken at the Defence Metallurgical Research Laboratory, India, and Northwestern University, USA, are highlighted in addition to a detailed survey of the literature. The various issues such as the nature of interfacial bond, chemical reaction at the interfaces, effect of alloying and processing on structure of the interfaces and the properties of the composites are examined in this paper. Strategies are suggested to exploit the full potential of reinforcing the metallic matrix.

1. INTRODUCTION

In recent years, the development of metal-matrix composites (MMCs) has been receiving worldwide attention on account of their superior strength and stiffness, wear resistance, corrosion fatigue behaviour and creep resistance in comparison to the corresponding wrought alloys. The idea involved in the design of a composite material is to combine the good attributes of metals with those of ceramics. Metals have low stiffness values, and are ductile. Ceramics are stiff and strong, but are brittle and fail catastrophically. In MMCs, we exploit the great strength of the ceramic while avoiding the brittle failure. The performance of the composite depends, besides the matrix microstructure and the nature of the ceramic reinforcement, very critically on the matrix-reinforcement interface. The interface plays a crucial role in the load transfer between the matrix and the reinforcement, very critically as well as dislocation-particle interactions, which are significant in strengthening and stiffening the composite. Moreover, the physical properties such as thermal conductivity, CTE, dimensional stability, etc are also closely related with the nature of the interface. Thus, it is desirable to have a clear understanding of the interfacial characteristics of specific MMCs to tailor them to achieve optimum performance in applications.

At present, MMCs can be classified into either continuous fibre composites or discontinuously reinforced composites. This paper addresses the issues pertaining to the interfaces in discontinuously reinforced MMCs.

2. NATURE OF MATRIX-REINFORCEMENT BONDING AND FACTORS AFFECTING IT

Strengthening by the reinforcing phase in MMCs is critically dependent on the strength of the bond between matrix and reinforcement. Interfacial bonding can be categorised as mechanical and chemical[1]. Mechanical bonding is significant only in the case of fibre reinforced composites, when fibres have rough or faceted surfaces. Chemical bonding is important for all kinds of reinforcements, viz. fibres, whiskers and particulates. A chemical bond is possible only if the atoms of the matrix and reinforcement are in direct contact and is accomplished by an exchange of electrons, and the type of exchange determines the character of the bond. It

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can be metallic, which is non-directional, and ionic or covalent which are directional. An interface with a metallic bond is thus more ductile than that with ionic or covalent bonds. For example, $Fe\text{-}Fe_3C$ interfaces in steel do not show fatigue crack initiation[2] as found in $Al\text{-}Al_2O_3$ interfaces[3]. Interface segregants, and chemical reaction can affect the strength of the individual atomic bonds at the interface, which ultimately reflects on the macroscopic strength.

3. CHEMICAL REACTION AT INTERFACES

A chemical reaction at the interface may lead to a strong bond between the matrix and the reinforcement, but a brittle reaction product can be highly detrimental for the performance of the composite. Table 1 gives matrices and reinforcements[4-28] with reaction products seen at the interfaces. A ceramic-ceramic bond (reaction product-ceramic reinforcement) is weaker and more

Table 1. Reaction products at some of the matrix/reinforcement interfaces

Matrix	Reinforcement	Reaction products and precipitates	Reference
Aluminium			
Al	SiC	Al_4C_3, Al(Si)	4,5
$Al\text{-}Mg$	SiC	Mg_2Si, Al_4C_3, MgO	6
$Al\text{-}Cu\text{-}Mg$	SiC	$CuMgAl_2$, MgO	7-10
Al	B_4C	AlB_2, Al_4C_3	10
$Al\text{-}Mg$	B_4C	$Al_xMg_{(1-x)}B_2$, Al_4C_3, $Al_x(B\text{-}C\text{-}O)_{1-x}$, MgB_6	10
Al	TiC	Al_4C_3, Al_3Ti	11-14
Al	TiB_2	No reaction	13-15
Al	Al_2O_3	No reaction	
$Al\text{-}Mg$	Al_2O_3	$MgAl_2O_4$	16,17
Al-Cu	Al_2O_3	$CuAl_2O_4$	17
Al-Li	Al_2O_3	$a\text{-}LiAlO_2$, $LiAl_5O_8$, Li_2O	1,17
Titanium			
Ti	SiC	TiC, Ti_5Si_3, Ti_xSi_y	18,19
Ti	TiC	Ti_2C	20
$Ti\text{-}Al\text{-}V\text{-}Nb$	TiC	$Ti_2(Al,Nb)C$	21
Ti	TiB_2	TiB	19
Magnesium			
Mg	Al_2O_3	$MgAl_2O_4$	1
Mg	SiC	Mg_2Si, C	22
$Mg\text{-}Li$	SiC	Mg_2Si, Li_2C_2	23
$Mg\text{-}Li$	SiC with SiO_2 coating	Mg_2Si, Li_2O	23
Intermetallics			
$NiAl$	SiC	Ni_xSi_y, Al_4C_3	1
$NiAl$	TiB_2	No reaction	24
$NiAl$	TiC	No reaction	25
$Ni\text{-}Al\text{-}Cr\text{-}Zr\text{-}B$	Al_2O_3	No reaction	25
$Ni\text{-}Al\text{-}Cr\text{-}Zr\text{-}B$	TiB_2	$(Ti,Zr)B_2$	25
$Ni\text{-}Al\text{-}Cr\text{-}Zr\text{-}B$	TiC	Segregation of Zr and B	25
Al_3Ti	TiC	No reaction	26
$TiAl/Ti_3Al$	TiB_2	No reaction	27
$MoSi_2$	SiC	No reaction	28
$MoSi_2$	TiC	No reaction	28
$MoSi_2$	TiB_2	No reaction	28
$MoSi_2$	Al_2O_3	$SiO(g)$, $Al(g)$	28
Ti_5Si_3	SiC	$TiSi$, TiC	28
Ti_5Si_3	TiC	No reaction	28

brittle than metal-ceramic or metal-intermetallic bond. Hence, cracks are initiated by cracking of the reaction product or the reaction product-matrix interface. For example, Al_4C_3 which forms as a reaction product at Al matrix-SiC fibre interface during processing by melting and casting acts as site for crack initiation[29]. Thus, the matrix and reinforcement can coexist in thermodynamic equilibrium under processing or service conditions and this is the first criterion in choice of ceramic reinforcements for any metal matrix. Alloying can be used to prevent progress of chemical reactions. Addition of 3 per cent Si raises significantly its activity in Al and completely stops the reaction[17] from progressing in forward direction according to the law of mass action.

$$Al(1) + SiC(s) \rightarrow Al_4C_3(s) + Si(Al) \qquad (1)$$

4. TYPE OF REINFORCEMENT AND ELECTRONIC BONDING

Even if the reinforcement is chemically compatible with the matrix, the strength and nature of the electronic interaction between matrix and reinforcement, and structure and chemistry of interface are important for efficient load transfer. Thus, it has been observed that for similar process variables and in the absence of any chemical reaction product at the interface, the increase in Young's modulus by incorporating the reinforcement in a metallic matrix provides a measure of the interfacial bond integrity between the matrix and the reinforcement. Two ways of averaging have been proposed; Voight[30], assumes uniform strain in both the constituent phases, and Reuss[31] assumes uniform stress in both the phases. The Voight averaging, also known as 'rule of mixtures' (ROM) leads to

$$E_c = V_m E_m + V_p E_p \qquad (2)$$

where, E_c, E_m and E_p are Young's modulus values of the composite, matrix and reinforcement particles respectively and V_m and V_p are volume fractions of matrix and reinforcement phases. The Reuss averaging gives

$$E_r = 1/\{(V_m/E_m) + (V_p/E_p)\} \qquad (3)$$

Figure 1 consists of bar diagrams for pure Al matrix powder metallurgy (PM) composites with 20 vol per cent SiC, TiC, TiB_2 and B_4C particulate reinforcements showing the ratios of experimental Young's modulus[32,33] values to those predicted by the ROM. A modified Tsai-Halpin model[34] was also used to calculate the composite elastic moduli for the above mentioned MMCs. This model incorporates the effect of aspect ratio of the reinforcement into the elastic modulus of the composite. The composite Young's modulus, E_c according to this is given by

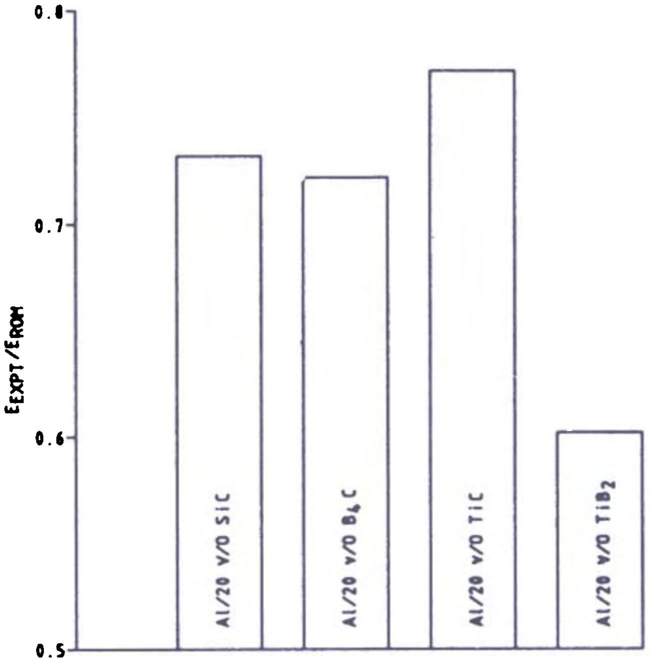


Figure 1. E_{EXPT}/E_{ROM} for Al matrix composites with different reinforcements (ref. 32). Al/TiC shows the highest ratio.

$$E_c = E_m (1 + 2SqV_p)/(1 - qV_p) \qquad (4)$$

where S is the aspect ratio, and

$$q = (E_p/E_m) - 1)/\{(E_p/E_m) + 2S)\} \qquad (5)$$

Figure 2 shows the comparison of experimental Young's modulus with the values calculated from the modified Tsai-Halpin model. Transmission electron microscopy (TEM) study did not reveal the presence of any reaction product at the matrix/reinforcement interface in these composites in the hot rolled condition[15]. The process details have been published elsewhere[15,35]. It is clear from Figs 1 and 2 that Al/TiC shows the best correlation between theoretical and experimental values and Al/TiB_2 composite shows the worst, suggesting that the former has the strongest interfacial bond and the latter has the weakest among the four systems studied. In another investigation[36] with P/M Al-Ti alloy containing Al_3Ti, Al_4C_3 and Al_2O_3 as dispersoids, only changes in the volume fraction of the intermetallic compound, Al_3Ti could be correlated with change in Young's modulus, whereas those of Al_4C_3 and Al_2O_3 had a negligible effect. It is also interesting

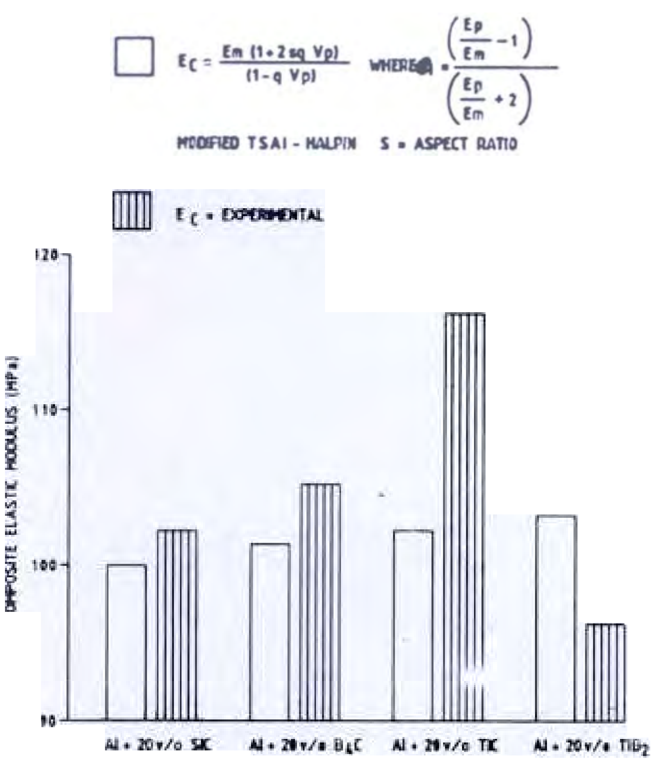


Figure 2. **Comparison between experimental and calculated (modified Tsai-Halpin model) elastic moduli for *Al* matrix composites (ref. 32).**

to note that ROM (Voight law of averaging) was closely followed. Observations of Zedalias *et al*[37] with Al_3Zr dispersoids and Skinner[38] with $Al_{13}(FeV)_3Si$ dispersoids in pure *Al* agree that there is an excellent correlation between experimental values of Young's modulus and those predicted by ROM in the case of intermetallic dispersoids. Fine[39] has shown that changes in Young's modulus values with volume fraction of $MnAl_6$ and Al_2O_3 dispersoids in pure *Al* follow the Voight and Reuss laws of averaging, respectively. A well-bonded interface would allow the maximum transfer of load from the matrix to the reinforcement and is expected to exhibit a high value of E_{EXPT}/E_{ROM} ratio, whereas, a weakly-bonded interface will result in lower E_{EXPT}/E_{ROM} ratio. In the latter case, values predicted by Reuss law of averaging will be closer to experimental values. Figure 3 shows[37,40-42] a bar diagram of E_{EXPT}/E_{ROM} for 4 systems, namely, Al/Al_3Zr[37], *Cu/Fe*[40], *Al/SiC*[41] and Al/Al_2O_3[42]. It is seen that metal or intermetallic and metal-metal systems show E_{EXPT}/E_{ROM} values close to unity. Also, E_{EXPT}/E_{ROM} for *Al/TiC* was higher than those of *Al/SiC*, Al/TiB_2 and Al/B_4C (Figs 1 and 2). In case of *TiC* and intermetallic compounds as dispersoids, it is obvious that the matrix-reinforcement bond is stronger. Of course, the bonds between *Al* and the intermetallic compounds are metallic. In metal-ceramic bonding, transition metal carbides like *TiC*, which have partial metallic nature[43], thus show stronger bond with *Al*, a reactive metal than more covalent bonded ceramics[44] like TiB_2 (Table 2)[33].

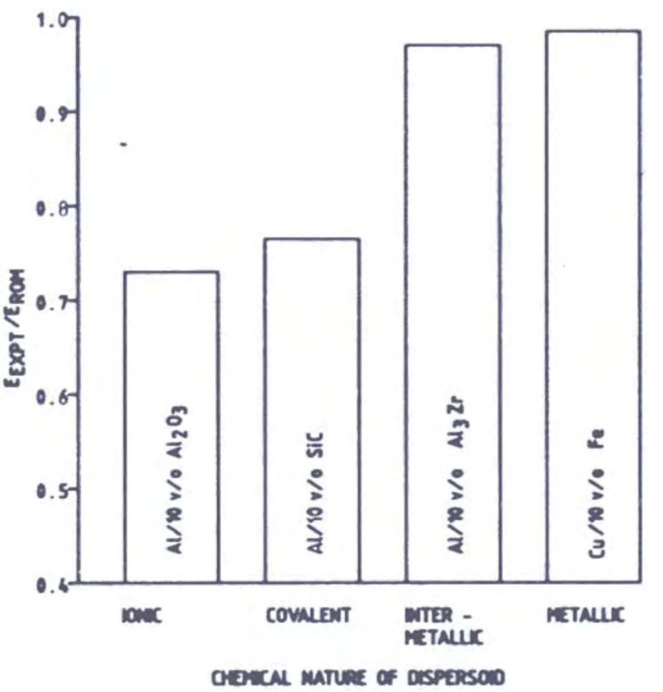


Figure 3. **Bar diagram showing E_{EXPT}/E_{ROM} for *Cu/Fe*, Al/Al_3Zr, *Al/SiC* and Al/Al_2O_3 systems.**

Table 2. **Bonding in refractory[33] compounds, *TiC* and TiB_2**

Compound	Heat of formation (kcal/mol)	% covalent bond
TiC	44	80
TiB_2	82	92

It is also clear from Table 2 that TiB_2 has a higher heat of formation and is more stable. It has also been reported by Ramqvist[45] that carbides with lower heat of formation are wetted better by metals than those with higher heat of formation. Also, *TiC* reacts with *Al* under certain conditions (Table 1), whereas TiB_2 does not. Thus, higher stability of TiB_2 in *Al* as compared to that of *TiC* is another reason, why the latter is wetted better.

Ni–16Al–8Cr–1Zr–0.1B matrix with *TiC*, Al_2O_3 and TiB_2 reinforcements prepared by P/M route has been also investigated[25]. While TiB_2 was not chemically compatible with the matrix, both TiC and Al_2O_3 were stable in the matrix (Table 1). Interestingly, reinforcing with TiC has shown excellent bond integrity leading to much higher modulus and yield strength compared to that with Al_2O_3. This again can be explained by metallic nature of TiC and strong interaction of Ti with both Ni and Al.

The above observations can be better appreciated with an understanding of wetting mechanisms. Wetting defines the extent of interaction between a liquid and a solid during fabrication of a composite and determines the bond strength. The mechanisms include, apart from chemical reaction, Van der Waals and image forces, adsorption and electronic bonding[46]. The adhesive forces between a reactive metal and a reinforcement is dependent on the electronic properties of the latter, as the former will always have free electrons in the conduction band. Thus, a ceramic reinforcement with more metallic nature is wetted better than those which are more covalent or ionic. Thus TiC and Al_3Ti act as effective nucleants during solidification of Al as has been investigated by Cisse *et al*[47]. The fact that Al solidifies with an epitaxial orientation relationship with respect to Al, that is, (001) Al // (001)TiC and [001]Al // [001]TiC, of course, suggests that under equilibrium conditions, interfaces of low energy form between Al and TiC. Similar phenomenon was observed for Al_3Ti too. And it has been seen from the data reported earlier, that bond integrity between Al and these reinforcements are excellent.

Chemistry and crystallography of surfaces, which play a significant role in determining electronic density of states of surface atoms, also affect the wetting behaviour. For example, photo-emission spectrum from TiC{111} surfaces has shown high density of states near Fermi level similar to Ti {0001} surfaces[48] and hence their behaviour with respect to wetting by Al is similar. Thus, good wetting of TiC by Al has been reported in the literature[43,49,50]. Rhee[43] observed that wetting increased in the order $AlN < TiB_2 < TiN < TiC$. An interesting observation in exothermic dispersoid (XD) Al/TiC composite was that the interface plane of TiC was mostly {111} (Fig. 4)[51]. Besides its metallic nature which improves wetting, these planes are densely packed and result in maximisation of a number of bonds with matrix and lowering the interfacial energy. Similar observations were also made by Gao and Merkle[52] for equilibrium and non-equilibrium metal-metal oxide interfaces.

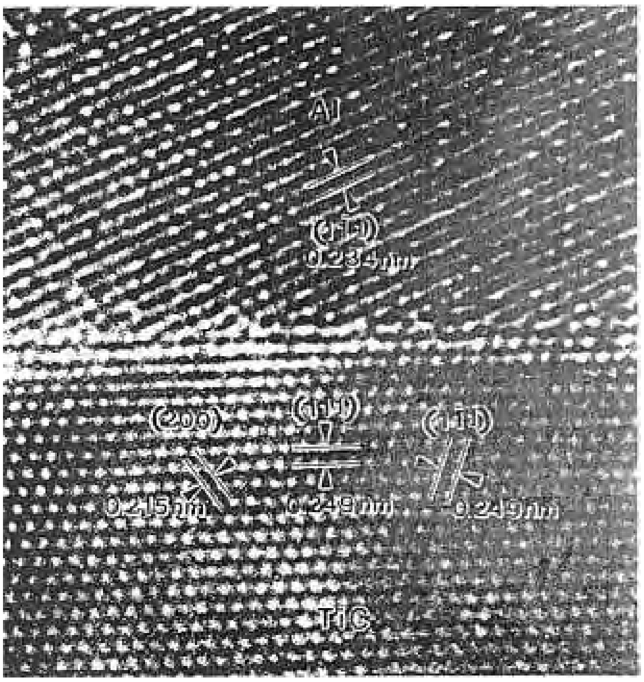


Figure 4. High resolution TEM micrograph showing the particle/matrix interface in Al/TiC composite prepared by XD process. The interface is abrupt on an atomic scale and is parallel to TiC (111) plane, which is densely packed (Ref 51).

In another experiment, Ohuchi [53]observed that Cu adhered more strongly to AlN {0001} surfaces as compared to AlN {1012} or AlN {1011} surfaces. The reason for this was the fact that Cu bonds to Al atoms during interaction and hence {0001} surfaces containing only Al atoms were more favourable.

The differences in dielectric constants of the metal and the ceramic reinforcements (mainly; oxide and carbide) also play a leading role in chemical bonding as this gives rise to image forces at the interface[54,55]. The effect is strengthened if there are charged defects on or near the surfaces of ceramic as these will have lower energy of formation at the interface than at a free surface. Thus, an oxide or carbide, which tends to lose stoichiometry more easily or have lower binding energy will be wetted better as compared to that which is more stable. In a study of wetting characteristics of MgO single crystal by Pb, Sn and Bi liquid metals by Nogi *et al*[56], it was observed that degree of polarization, type of charge distribution and atomic arrangements on the surface controlled the degree of wetting by the liquid metal.

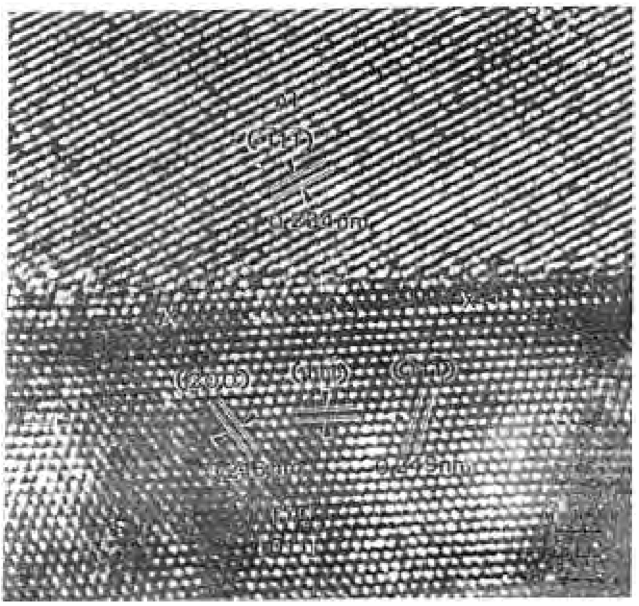


Figure 5. **High resolution TEM micrograph showing the particle/matrix interface in *Al/TiC* composite prepared by XD process. First 3 atomic layers inside *TiC* near the interface appear distorted due to a stacking fault bound by *X* on the (111) *TiC* planes (shown with an arrow). Position of an extra half plane of an edge dislocation is shown with an arrow (ref. 51).**

TiC also is an excellent example. Atom fraction of carbon in *TiC* varies from 33 to 48 per cent without any change in its crystal structure[57]. This also explains why *TiC* is more strongly bonded to *Al* as compared to TiB_2 and *AlN*. Stacking faults and dislocations have been seen close to *Al-TiC* interface using high resolution transmission electron microscopy (HRTEM) and Fig. 5 shows an example. The fact that charged defects at the surface of ceramic reinforcement improves wetting suggests that irradiation of ceramic substrates or fibres will lead to better bonding with metal as has been observed by Tombrello[58].

5. EFFECT OF ALLOYING ELEMENTS ON WORK OF ADHESION AND BONDING

Work of adhesion, W_{ad} of a particle-matrix interface is strongly affected by alloying element additions. Work of adhesion is defined as the energy of creation of free surfaces of particle and matrix by rupturing of the interfacial bond. Work of adhesion for solid/solid interface is given by

$$W_{ad} = \gamma_{mv} + \gamma_{pv} - \gamma_{mp} \qquad (6)$$

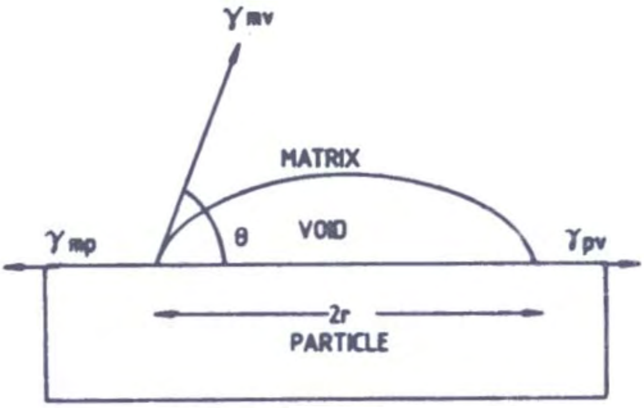


Figure 6. **A spherical cap shaped void formed at the interface a *TiC* particle facet and matrix (ref. 14).**

where γ_{mv} and γ_{pv} are the surface energies of matrix and particle and γ_{mp} is the particle-matrix interfacial energy. Figure 6 is a schematic diagram of a void formed at a particle matrix interface. If ϕ is the contact angle between a particle facet and a spherical cap shaped void, the force equilibrium in the horizontal direction gives

$$\gamma_{mp} = \gamma_{mv} \cos \phi + \gamma_{pv} \qquad (7)$$

This is Young's equation[59]. Substituting Eqn (6) in Eqn (5), we get

$$W_{ad} = \gamma_{mv}(1 - \cos \phi) \qquad (8)$$

Easterling *et al*[60] used Eqn (8) for determining the W_{ad} between the particle and the matrix in the $Fe\text{-}Ni/Al_2O_3$ composite. Contact angle was seen to change very sharply with the changes in atomic fraction of alloying elements in the matrix and so the work of adhesion as shown in Table 3. It was also observed that the flow stress was directly related to the work of adhesion.

Table 3. **Variation in critical strain to interface cavity formation, contact angle and work of adhesion with alloying elements in an $Fe\text{-}alloy/Al_2O_3$ composite[60]**

Matrix	E_c	θ (degrees)	W (mJm^{-2})
Fe-40Ni	0.05	<10	44
Fe-40Ni-5Cr	0.05	16	112
Fe-40Ni-5Mn	0.10	15	99
Fe-40Ni-5Mo	0.15	65	1674
Fe-1Ti		35	524
Fe-10Co	0.08	60	1450
Fe-5Cr	0.10	80	2397

Segregation of alloying elements to the interface reduces interfacial energy due to the Gibbs adsorption phenomenon. However, this may be detrimental to interface toughness if the alloying elements lead to an increase in the ionic or covalent character of the interface bonding[61]. Alloying elements play significant role in coarsening behaviour of precipitates in metal-intermetallic composites[62]. For example, in *Fe-Ni-Al* alloys containing a few per cent of *Mo*, the *NiAl* particles remain spherical upon ageing even though the particle-matrix lattice misfit is considerable. This is in contrast to that in absence of *Mo*, when the precipitates turn cuboidal. It has been suggested that *Mo* segregation at the interface relaxes the elastic misfit strains or changes the interfacial energy such that even large precipitates remain spherical.

The nature of the chemical bond also depends on alloying elements. For example, addition of *Mg* to $Al\text{-}Al_2O_3$ composites leads to the formation of *MgO* and $MgAl_2O_4$ at the interface[1,16,17] and this enhances the wettability, which is otherwise poor. The reactions are

$$3Mg + Al_2O_3 \rightarrow 3MgO + 2Al \qquad (9)$$

$$Mg + Al_2O_3 \rightarrow MgAl_2O_4 \qquad (10)$$

Mg has also improved the wetting between *Al* and *SiC* particles, by reducing the SiO_2 layer on the surface of *SiC* and aiding in the removal of Al_2O_3 film on *Al* melt as per the reaction[63]

$$4Mg + SiO_2 \rightarrow Mg_2Si + 2MgO \qquad (11)$$

It has been reported by Sritharan *et al*[17] that ideal matrices for *SiC* reinforcement, which show the maximum strengthening effect, are high *Mg* containing 5XXX and 6XXX alloys as more of Mg_2Si can be precipitated. On the other hand, reduction in strength has been reported in case of *Al-Cu-Mg* alloys (with respect to monolithic alloy) as *Mg* segregation at interfaces result in the depletion of *Mg* content in the matrix. Similar lowering of strength is reported for 7XXX alloys as *Mg* content is depleted by formation of $MgZn_2$ and Mg_2Si at the interfaces. In certain other studies, addition of *Mg* to *Al-SiC* composites has proved to be detrimental due to the formation of *Mg*-rich amorphous region[7-8] or intermetallic precipitates like $CuMgAl_2$[7-10,64] at the interface. In cast alloy-matrix MMCs, coarse intermetallic particles are quite often seen at the interfaces because the residual melt enriched in solute solidifies near poorly heat conducting ceramic reinforcement.

Other than *Mg*, *Li* also has been successfully used as an alloying element to enhance the wetting between *Al* and Al_2O_3[65] and *SiC* [66] reinforcements. The mechanism is similar to that of *Mg*, that is, reduction of oxide layers at the interface, which otherwise prevent direct contact between *Al* and the ceramic reinforcement. Webster[66] has systematically investigated the effect of *Li* additions (3-5 per cent) to *Al* on mechanical properties of Al/SiC_w composites. Strengthening effect was significantly enhanced by additions of *Li* as shown in Table 4 and this has been attributed to increase in the matrix-reinforcement bond strength. *Li* increases the reactivity of *Al*, and as a consequence, promotes its interaction with *SiC* at the interface. Thus, Young's modulus values were close to those predicted by the ROM. No chemical reaction product could be observed at the interface using TEM. Alloying with indium, lead and thallium have improved wetting between *Al* and carbon fibres by reducing the surface tension of liquid metal[67]. However, to our knowledge, these have not been used for *Al/SiC* composites.

Table 4. Variation in Young's modulus with addition of *Li* to Al/SiC_w composites[66]

Matrix	Modulus (GPa)	% increase with respect to matrix
Al	70	
6061-*Al* + 20% SiC_W	100	43.0
Al + 3.5% *Li*	86	
Al + 3.5% *Li* + 20.9 SiC_W	139	62.0

Al/TiC composites show Young's modulus values closest to those theoretically predicted because of excellent bond integrity. However in *Al-Fe-V-Si* alloy/*TiC* composite containing 6 vol per cent *TiC* and 27 vol per cent $Al_{13}(Fe,V)_3Si$, the Young's modulus of the composite is slightly lower than that predicted by Reuss law of averaging, showing that the interfacial bond is weak[68] (Table 5). The effect of individual alloying elements is not fully known. But it definitely suggests that the interfacial bond strength is a strong function of matrix composition.

Effect of matrix, type of reinforcement and alloying on bond integrity can also be appreciated by analysing

Table 5. Young's modulus values of *Al/TiC*[32,33] and *Al-Fe-V-Si/TiC*[68]

Material	Young's modulus (GPa)	E_{EXPT}/E_{ROM}
Al/20 v/o *TiC*	115	0.79
Al-Fe-V-Si/15 v/o *TiC*	102	0.72

the modulus values of *Mg*-alloy MMCs. Figure 7 is a comparison between E_{EXPT}/E_{ROM} of PM *Mg/SiC*[69] and PM *Al/SiC*[70], where the matrices are commercially pure metals. It is evident that in *Mg/SiC* composites have smaller value of the ratio as compared to the *Al/SiC* composites, indicating that the interfacial bond is stronger in the latter. The possible reason is matrix-reinforcement chemical interaction is stronger in the latter as *Al* forms a stable carbide as well as alloys with *Si*.

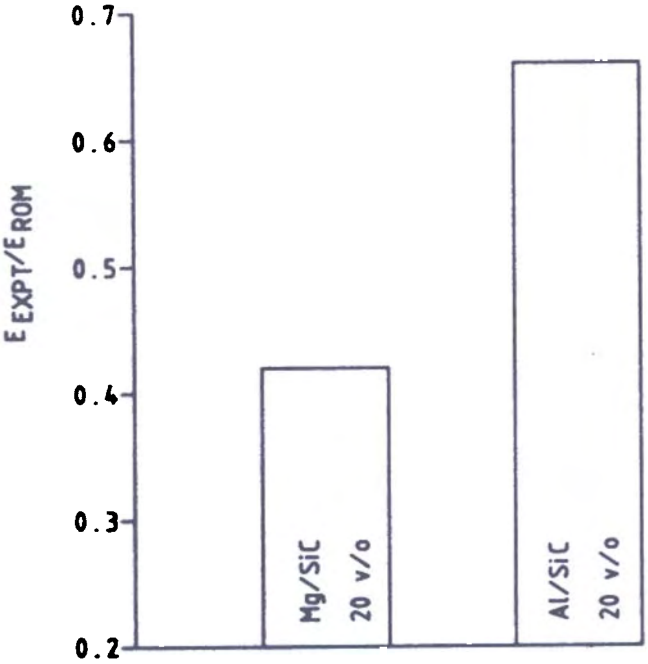


Figure 7. Comparison between E_{EXPT}/E_{ROM} ratios of *Mg*/20 v/o *SiC* and *Al*/20 v/o *SiC* (ref. 69) metal matrix composites.

The effect of matrix alloying elements like *Al* in *Mg* MMCs is also evident from the data shown in Table 6[71]. *AZ91-Mg* as the matrix shows remarkably larger increase in Young's modulus than commercially pure *Mg* as the matrix. *AZ91-Mg* contains 9 per cent *Al*, which strongly enhances the extent of interaction of *Mg* with *SiC*. Of course, alloying elements in the matrix alter the dielectric constant of the matrix. Comparison of effect of different reinforcements on modulus of *AZ91-Mg* matrix composites has also been demonstrated by Schroder *et al.* Figure 8 shows the variation in E_{EXPT}/E_{ROM} with different reinforcements.

Table 6. Change in Young's modulus of the compo-ite with change in the volume fraction of *SiC* in commercial purity *Mg* and *AZ91 Mg* alloy[71].

Volume fraction of *SiC* particles (%)	Modulus values (GPa) Commercially pure *Mg*	*AZ91-Mg* alloy
0	43	46
5	45	53
10	47	60
15	50	65

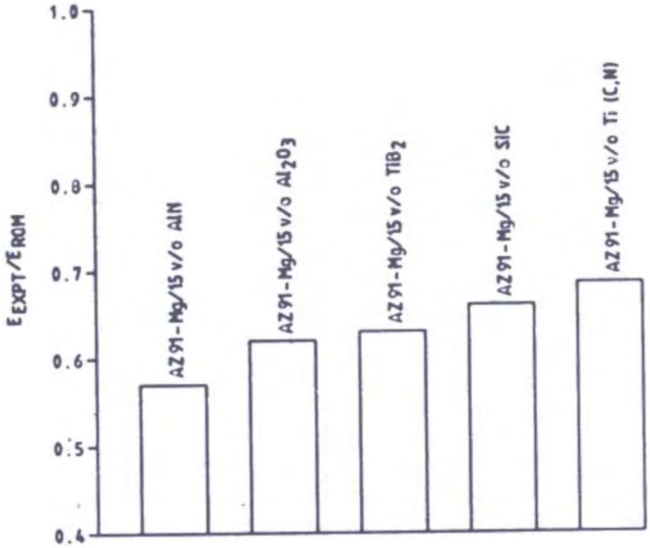
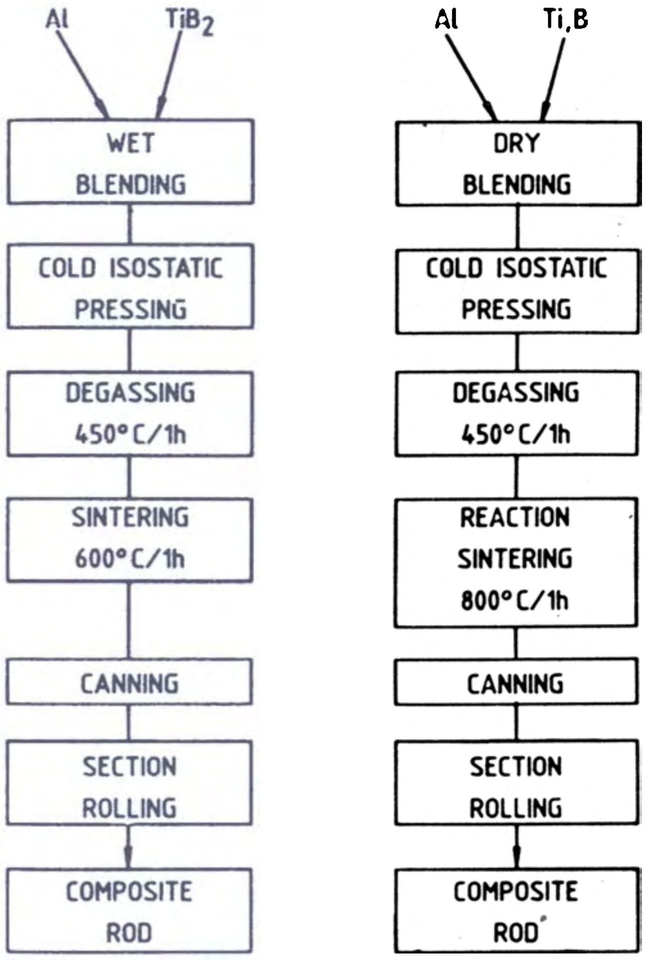


Figure 8. Comparison of E_{EXPT}/E_{ROM} ratios of *AZ91-Mg* matrix composites with different reinforcements; reinforcement of *Ti(C,N)* appears to give the best properties (data from ref. 71).

It is clear that both *Mg* matrix as well as *Al* play a role in determining the strength of the interfacial interaction with the reinforcement. It was observed that *SiC*, *Ti(C,N)* and Al_2O_3 show relatively higher modulus, that is, closer to the value expected from the ROM as compared to TiB_2 and *AlN*. *Al* strongly interacts with *SiC* and *Ti(C,N)*. TiB_2, Al_2O_3 and *AlN* are very stable due to their higher heat of formation and binding energy and interact weakly with both *Mg* and *Al*. Hence, the matrix reinforcement bond is weaker for TiB_2 and *AlN* reinforcements in *AZ91-Mg* composites.

6. EFFECT OF PROCESSING: CONVENTIONAL AND IN-SITU PROCESSING

Processing variables play a very significant role in the control of interfacial characteristics. Based on processing, composites can be classified as 'synthetic' or 'Natural'. Synthetic composites are prepared by artificial mixing of matrix and reinforcement powders followed by cold compaction and hot pressing. Flow chart in Fig. 9 shows the process developed and followed at Defence Metallurgical Research Laboratory (DMRL), Hyderabad. Hot pressing is carried out in two steps, the first being above solidus and for a short duration and the next for longer time below the solidus. Without a liquid phase formed during hot pressing or sintering, high modulus values are not achievable[72]. The mechanisms of bonding in absence of a liquid phase are mechanical interlocking of particles and solid-state diffusion leading to a chemical bond. Table 7 shows results of mechanical characterisation of 2124-*Al/SiC* after just vacuum hot pressing (VHP) and after both VHP and extrusion[35]. The microstructures representative of both the secondary processing conditions are shown in Fig. 10.



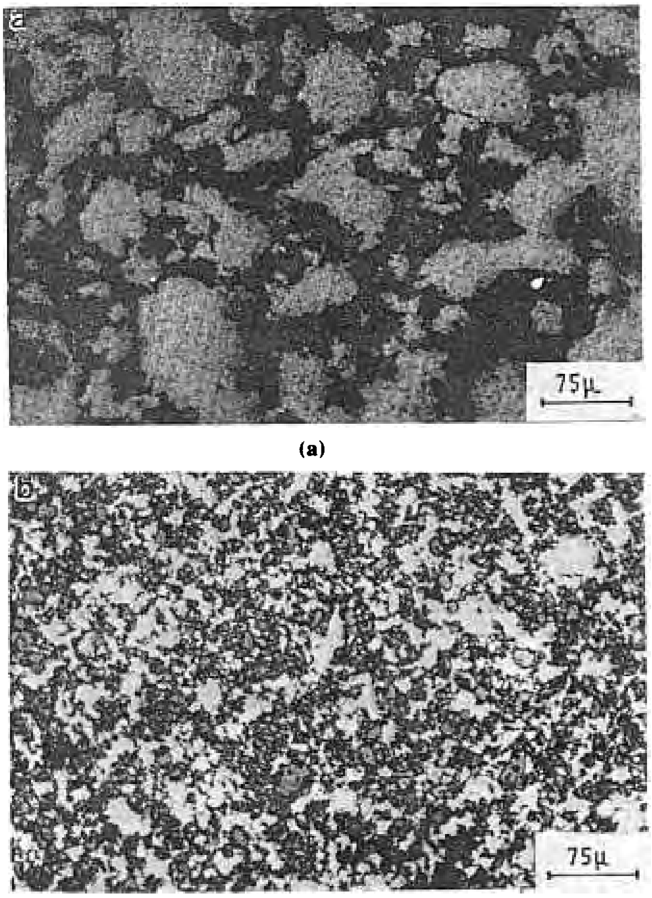
(a) (b)

Figure 9. Flow chart showing conventional (P/M) and XD process developed at DMRL for preparing metal matrix composites.

Table 7. Mechanical properties of 2124 *Al*/30 v/o *SiC* composites after VHP and VHP followed by extrusion[35]

Processing condition	Young's Modulus (Gpa)	0.2% YS (MPa)	UTS (Mpa)	% elongation
VHP	108	–	435	0.1
VHP + Ext.	120	382	497	1.7

YS: yield strength, UTS: ultimate tensile strength



(a)

(b)

Figure 10. Optical micrographs showing microstructures of *Al*/20 v/o *SiC* in (a) vacuum hot pressed condition and (b) vacuum hot pressed and extruded condition (ref. 35).

There is a significant increase in the Young's modulus after extruding the VHP billet. This can be understood from a comparison of the microstructures. In the VHP condition, SiC particles are seen at the oxide-coated prior particle boundaries (PPBs), whereas after extrusion, the PPBs are broken and intimate contact between matrix and reinforcement is established. This helps in improving the bond integrity and increases the modulus and strength of the system.

The other routes for preparing 'synthetic' composites are adding reinforcements to molten metal and distributing using a stirrer, rheocasting or melt infiltration of ceramic preform or green compact by liquid metal. Higher processing temperature may help wetting in many cases, but often lead to degradation of reinforcements by chemical reactions as is observed in Al-SiC system. Chiou and Chung[73] prepared Al/SiC composite by infiltration of liquid Al into a porous preform of SiC whiskers in vacuum under inert gas pressure at 665, 690 and 720 °C and observed that tensile strength, Young's modulus as well as the extent of elongation decreased with increase in the processing temperature. Also, significant whisker pullout could be observed during tensile failure of the composites prepared at higher temperature. All this is attributed to formation of brittle reaction product, Al_4C_3 at the interface.

Sometimes, processing at certain temperature ranges can only lead to chemical reactions between matrix and reinforcement. For example, Al and TiC react as[11-14] shown in Eqn (12) at temperatures below 752°C, above which the free energy of formation is positive and the chemical reaction is not feasible. Extent of chemical reaction also varies with time and a systematic study has been performed for Al-TiC[12-14] and Al-SiC[74].

$$7Al + 3TiC \rightarrow Al_3Ti + Al_4C_3 \tag{12}$$

It has been discussed in the earlier section that coarse intermetallic particles precipitate at the matrix-reinforcement interface in MMCs prepared by casting as the solute rich residual melt solidifies at the interface towards the end. However, in MMCs prepared by spray deposition[75], TEM investigations have shown clean interfaces with negligible precipitation. This can obviously be explained on the basis of the fact that contact time between liquid matrix phase and reinforcement is of the order of a few seconds in spray deposition processing and a few minutes in case of squeeze casting. Thus in Al-alloy composites prepared by spray deposition, attractive combination of strength and modulus could be achieved because of excellent bond integrity.

In the above mentioned processing techniques, reinforcement particles or whiskers are prepared in a separate process earlier to processing of the composites. During this or subsequent handling, the surfaces of the particles pick up impurities or get oxidised, which continue as a third continuous or discontinuous phase at the interface. XPS studies on isolated Si_3N_4 whiskers have shown Si_2N_2O and oxygen impurities[76]. These oxide layers participate in interfacial reactions during processing of MMCs and interface chemistry becomes complex. The interface in these cases fail to reach the state of thermodynamic equilibrium.

Amorphous oxide layers often found at interfaces[7,8,77] are responsible for interfacial void formation or cracking during external loading at room and elevated temperatures[78-80]. A comparative study was carried out by Ribes *et al*[79], where Al/SiC composites with SiC particles with and without a SiO_2 layer at the surfaces were compared. Particulate cracking resulted in failure in the former, whereas interfacial decohesion was the mechanism for the latter. SiO_2/SiC interface cracked and this resulted in lowering of the strength of the composite as compared to that with unoxidised SiC. Interestingly, ductility of the composite containing oxidized SiC was better than that with non oxidised SiC and this has been explained on the basis of shear band formation at a higher overall strain in the former as compared to that in the latter. Cracking at interfaces relaxed the stress concentration.

Back in 1954, Mcdonald and Ransley[72], through a series of investigations found that incorporation of high modulus dispersoids in the Al matrices resulted in moduli less than that of the matrix by itself. It is obvious that increase of dispersoid content fails to affect the modulus. It was realised by Mcdonald and Ransley that wetting of the reinforcement by the matrix is the key and that certain other additives were necessary to promote it. For example, addition of Ni and Co to Al improved wettability in the Al/TiC system. A chemical bond of electronic nature needs to be established between matrix and reinforcement phases for efficient load transfer.

It was also concluded by Mcdonald and Ransley that wetting took place automatically if the high modulus phase was an 'intermetallic compound precipitated *in-situ* by reaction between elements' and the same was difficult to induce by mechanical mixing of powders and purely solid-state processing. This is because the composites prepared by *in-situ* processing have clean interfaces; chemical reaction between elements to produce the dispersoid takes place inside the solid or liquid metal. These are examples of 'natural' composites as thermodynamically stable phase precipitates in the matrix and the process is driven by the nature of the system to reach equilibrium. Some of the *in-situ* processing techniques involving molten metal and developed in recent years are XD[81] developed at Martin Marietta Laboratory, Baltimore, USA and gas injection process developed by Koczak[82] at Drexel University, Philadelphia, USA. Similarly, solid-state processing techniques like internal oxidation or reduction[83] and displacement reactions[84] have also been developed. Solid-state processing normally leads to second phase precipitates with an equilibrium low energy orientation relationship with respect to the matrix and interfaces can show varying degree of coherency based on lattice misfit and strength of interaction between the matrix and the reinforcement[83].

Atomic resolution TEM studies of the interface between *Al* matrix and *TiC* particle in XD *Al/TiC* composite have proved interface is abrupt on an atomic scale[14] (Figs 4 & 5). No impurities or sign of oxidation of particle surfaces could be seen by EDS and EELS at the interface. This is possible only because the particles were precipitated *in-situ*, which prevents them from contamination.

EDS investigation using a probe size of 3 nm (including effect of beam broadening) was used to investigate interface in XD 2024 *Al/TiB*$_2$ composite in a field emission analytical TEM[51]. No segregation of *Mg* or *Cu* alloying elements could be found near interface. This suggests that the interfaces formed are of sufficiently lower energy, and there is little driving force for interfacial segregation.

Apart from the fact that the interfaces are clean, some other interesting features are associated with the XD composites. Many of the *Al/TiC* interfaces in 0.7 μm particle XD composite (cast and extruded at a ratio of 27:1 at 375 °C) were semicoherent and showed localised strain contrast in the TEM[51,85]. Figure 11 is a weak beam dark field TEM image of an *Al/TiC* interface showing misfit strain localisation. Normally, particles which were finer and located inside *Al* grains and surrounded by some *Al* subgrains had semicoherent interfaces. On cold rolling to 75 per cent of its original thickness followed by annealing, the material recrystallised leading to a microstructure, in which faces *Al* of the *TiC* particles were surrounded by one or more subgrains[85,86]. The interfaces between these and *TiC* particles were in most cases semicoherent with misfit strain localization (Fig. 12). As shown in Fig. 4, interfaces were always parallel to {111} TiC planes. In another investigation, Wang and Arsenault[24] observed that the *NiAl/Al*$_2$*O*$_3$ interfaces in XD *NiAl/Al*$_2$*O*$_3$*/TiB*$_2$ composites were semicoherent and showed misfit localisation. Presence of semicoherency, of course,

Figure 11 Weak beam dark field TEM micrograph of a semicoherent interface in cast and extruded XD *Al/TiC* composite (ref. 51). Misfit strain contrast can be seen at the interface.

Figure 12 Dark field TEM micrograph of a semicoherent interface between an *Al* subgrain and *TiC* particle in the XD *Al/TiC* composite. The composite was cold rolled and annealed for recrystallization (ref. 86). Interface dislocations can be seen.

supports the idea that particle-matrix bonding is strong in XD composites, where there is direct contact between the atoms of the two phases.

In-situ composites have consistently shown higher modulus as compared to those prepared by conventional techniques like casting and powder metallurgy[87,88]. Table 8 shows comparison of Young's modulus, yield strength and elongation for 20 vol per cent XD *Al/TiB*$_2$ and PM *Al/TiB*$_2$ composites measured by Kuruvilla *et al*[88] at DMRL. XD *Al/TiB*$_2$ also had some *Al*$_3$*Ti* particles which resulted in poorer elongation properties in the present situation. However, Young's modulus of *Al/TiB*$_2$ prepared by XD process is also much higher as is clear form the bar chart for $E_{EX.T}/E_{ROM}$ shown in Fig. 13. Also because of finer particle size, yield strength was greater in the XD composite. Young's modulus value of 15 vol per cent XD *Al/TiC* was found to be 94 GPa[14,15], which is well within lower and upper limits provided Hashin-Shtrikman model[89], Similar observations were made by Aikin for 15 vol per cent XD *Al/TiB*$_2$[90]. A significant improvement in Young's modulus over the monolithic metal or alloy is achievable mainly because of excellent chemical bond between atoms of metal and the reinforcement, which are again in direct contact as has been seen in XD *Al/TiC*.

Table 8. Comparison of mechanical properties between conventionally processed and XD *Al*/20 v/o *TiB*$_2$ composites[88]

Material	E (Gpa)	YS (Mpa)	UTS (Mpa)	% elongation
Al	70	64	90	21
Al/TiB$_2$ (PM)	96	121	166	85
Al/TiB$_2$(XD)	131	235	334	7

E : Young's modulus, YS: yield strength; UTS: ultimate tensile strength.

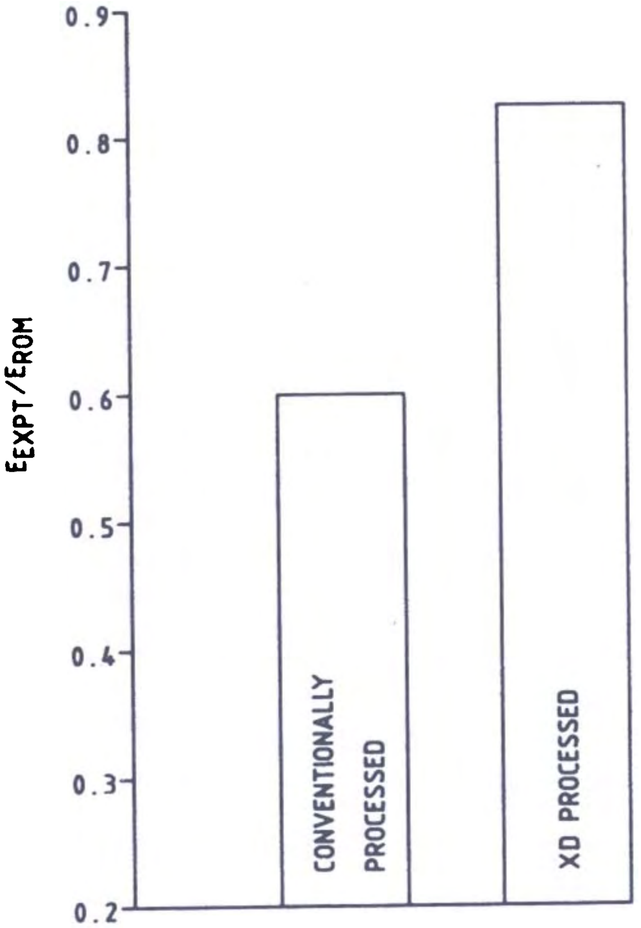


Figure 13. **Comparison of E_{EXPT}/E_{ROM} ratios of *Al/TiB*$_2$ composites prepared by conventional (PM) and XD process. It is clear that the latter has higher modulus.**

7. EFFECT OF AGEING AND HEAT TREATMENT

Ageing heat treatments in 7XXX and 2XXX *Al-SiC* composites, particularly the former, significantly alter the structure and chemistry of the interface[91]. Interfacial segregation of *Cu, Mg* or *Zn* could be seen in both underaged as well as overaged conditions. In underaged 2XXX matrix composite, the region of segregation was in the form of diffuse layer 20 nm into the matrix from the interface. The profiles of *Mg* and *Cu* were fairly broad and peaked away from interface because of non-equilibrium segregation of vacancies. However, after overageing, vacancies were segregated at the interface to meet the equilibrium conditions and peak shifts to the interface. Similarly, a diffuse interface precipitate layer 10-20 nm wide was seen near interface. On overageing, fairly regularly spaced particles, ($Mg_{32}(Al,Zn)_{49}$ appeared with a distinct PFZ next to them extending 35-40 nm in the matrix, which is depleted in solute. In response to equilibrium vacancy segregation at the interface, there is a decrease in their concentration in the region where PFZ forms as solute atoms migrate by diffusion and segregate at the interface.

Recently, isothermal heat treatment of *Al* MMCs containing 20 vol per cent *SiC*, *TiB*$_2$, *B*$_4$*C* or *TiC* was performed at 600 °C for different periods of time and the results are shown in Table 9[92]. *TiB*$_2$ is thermodynamically stable in *Al* at 600°C and obviously,

Table 9. Effect of isothermal annealing treatment on tensile ductility of *Al* matrix composites

Material	Time exposed (hr)	% elongation Initial	% elongation After exposure
Al/20 v/o $B_4C^{(a)}$	48	9.0	2.5
Al/20 v/o $TiC^{(a)}$	96	6.8	0.8
Al/20 v/o $SiC^{(b)}$	96	10.0	4.0
Al/20 v/o $TiB^{(b)}$	100	12.9	12.2

vigorous reaction, and b: no reaction

the elongation properties are not affected. *TiC* and B_4C roacts extensively with *Al* resulting in reaction products at the interface (Al_3Ti and Al_4C_3 in case of *TiC* and Al_4 C_3 and *AlB* in case of B_4C), which adversely affect tensile ductility. *Al/SiC* composite also shows lower elongation after heat treating and this is surprising as no reaction product could seen at the interface using conventional TEM. However, some discontinuous precipitates of *MgO* and $MgAl_2O_4$ could be seen, which possibly originated from *Mg* impurities in the matrix. *Al/SiC* also showed some changes in Young's modulus and a significant change in yield strength after heat treating as shown in Fig. 14. This suggests strengthening of the interfacial bond during heat treatment. While, it appears that at a temperature of 600 °C (0.94 *Tm*) where diffusivity of *Al* is large, rearrangement on an atomic scale is possible near interface leading to a structure closer to equilibrium. Slight, but quite noticeable increase was reported[92] in the yield and ultimate tensile strengths of the Al/TiB_2 composite also. A detailed investigation of the interface is necessary to explain this. Recently, Warner and Stobbs[93] noticed an increase in elastic modulus in *Al*-1.9 per cent Mg/14 v/o SiC_p composite on isothermal heat treatment, with the 0.2 per cent yield stress remaining unchanged. Lavernia *et al*[94] on the other hand, saw noticeable increase in yield and ultimate tensile strength of the MMCs prepared by spray deposition process on annealing at 560 °C for 22 hr prior to extrusion. It was observed that the failure mechanism involved fracture of *SiC* particles, rather than pullout. This was explained based on the fact that annealing helped in improving the bond strength between matrix and reinforcement. Interdiffusion between *Al* and *SiC* was suggested as the reason for increase in bond strength. This was reported earlier by Arsenault[95].

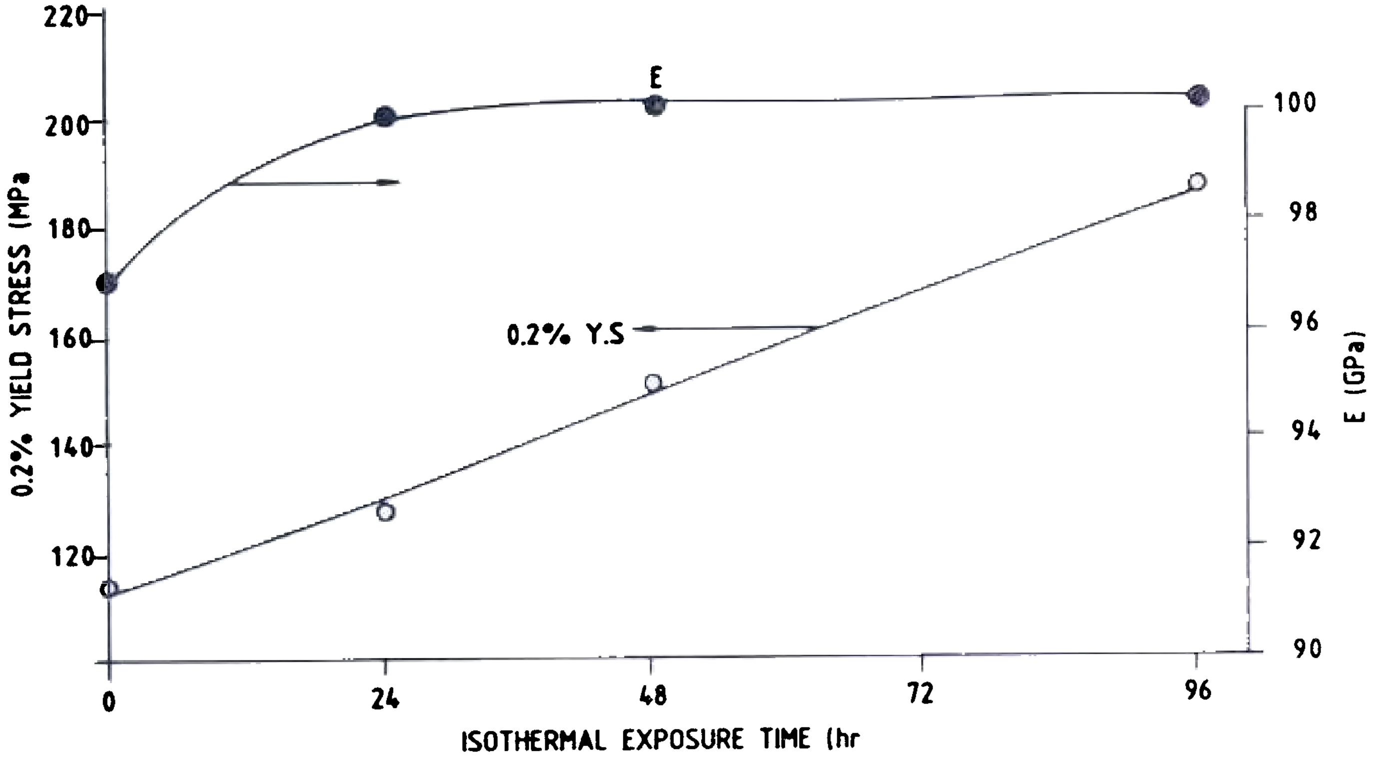


Figure 14. Variation of 0.2 per cent yield stress and Young's modulus of *Al*/20 v/o *SiC* metal matrix composites with time of exposure at 600 °C (ref. 33).

8. INFLUENCE OF THE INTERFACE ON PHYSICAL PROPERTIES

Interfaces play a significant role in determination of the physical and mechanical properties of the composite. The physical properties include coefficient of thermal expansion, thermal conductivity and damping.

Recently, coefficients of thermal expansion were measured for 15 vol per cent XD *Al/TiC* composite with 0.7 or 4.0 μm particle sizes and it was observed that the former had a smaller coefficient of thermal expansion[96] The matrix being pure *Al* in either situation, it is surely due to more interface area in smaller particle size composite. Coefficient of thermal expansion is an important criterion for the design of dimensionally stable composites.

Hasselman *et al*[97] have reported thermal conductivity in *Al/SiC* MMCs for particle sizes varying from 0.7 to 28 μm. It was observed that thermal conductivity decreased with smaller particle size. This phenomenon can again be explained based on the fact that finer particle size composite has larger interface area. The interface acts as a thermal barrier. Thus for maximising thermal conductivity, particle size needed is the largest possible as this will reduce the total interface area. Geiger *et al*[98] observed that 6090 *Al/SiC* composites with 10.2 and 28 μm particles had higher thermal conductivity than that of the unreinforced matrix. This was possible probably due to the excellent bonding. However, with decrease in the *SiC* particle size, thermal conductivity of the composite decreased. Geiger and Jackson[99] have also reported that 6061 *Al/SiC* shows higher thermal conductivity than 2124 *Al/SiC*. Besides the interface, the temper of the matrix alloy also affects the thermal conductivity and thus the above behaviour is difficult to explain.

Interfaces with reaction products or precipitates are likely to act as stronger barriers for heat conductivity than one with cleaner ones. This was investigated in detail by Reeves *et al*[100] for the case of *Ti/SiC* composites, where the reaction products were a mixture of *TiC* and Ti_5Si_3. For reaction layer of 0.5 μm or thinner, the thermal conductivity of the composite was similar to that of the unreinforced matrix. This could be due to the fact that *SiC* particles possessed lower thermal conductivity than typical dense bulk *SiC* or because of the barrier created by reaction layer. If the reaction layer was relatively thick (greater than 1 μm), thermal conductivity values of the composite were markedly reduced. The values were below those expected for composites reinforced with similar volume fractions of insulating particles.

To estimate the role of interface, Hasselman and Johnson[101] proposed a model for composite conductivity, K_c. This includes the interfacial heat transfer coefficient or thermal conductance, h ($Wm^{-2}K^{-1}$).

$$K_C = K_M \frac{2f(K_I/K_M - K_I/rh - 1) + K_I/K_M + 2(K_I/rh) + 2}{f(1 - K_I/K_M + K_I/rh) + K_I/K_M + 2(K_I/rh) + 2} \tag{13}$$

where f is the volume fraction of inclusions and the subscripts C, M and I refer to composite, matrix and inclusion, respectively, and r is the radius of the inclusions assumed to be spherical. This equation is valid only for 'dilute' composites, in which disturbance to the thermal field around an inclusion does not overlap with the disturbance from surrounding inclusions.

The effect of reinforcement/matrix interface on the damping behaviour of the composites has been investigated by Zhang *et al*[102]. Schoek theory, proposed to explain the phenomenon of internal friction in alloys, has been used to explain the same for MMCs. According to this theory, internal friction is increased by the relaxation at semicoherent or incoherent precipitate/matrix interfaces and the unelastic strain contributed by dislocations close to the interface. Resultant composite internal friction was found to be proportional to the volume fraction of the reinforcement. In the MMCs, not only the interface area is large, but also CTEs between particulates and matrix result in stress concentration at the incoherent interfaces, which increases the internal friction. At higher temperatures, when the matrix softens relative to the hard reinforcement, a reversible movement of vibrations is expected to occur at the interface and this can be used to explain increase in interface damping with temperature. In 2519 *Al* MMCs, the largest role of interface could be seen at temperatures between 250 and 350 °C.

9. EFFECT ON YIELD STRENGTH AND ELONGATION

The effect of brittle reaction products at the interface on the load transfer efficiency of the interface, which is adversely affected, has been discussed earlier in the paper. Similarly, coarse intermetallic precipitates at the *Al-Cu-Mg* alloy/*SiC* interfaces are detrimental to all the mechanical properties. The feedback from the above reports is that a clean interface is always desirable.

Even if there are no reaction products at the interface and the matrix, particle sizes and shapes are the same, the yield strength, modulus and tensile ductility vary based on the nature of chemical bond. This is evident from the data shown in Table 10 for mechanical properties of pure *Al* and *Al/SiC*, Al/B_4C, *Al/TiC* and Al/TiB_2 composites prepared by P/M process at DMRL[70]. The stress-strain curves are shown in Fig. 15. It is obvious that the highest increase in yield and ultimate tensile strength is observed for *Al/TiC*. Improvement in strength is through load transfer at the interface as well as dislocation-particle and dislocation-dislocation interactions. Of course, the excellent bond integrity of *Al/TiC* interface explains why *Al/TiC* composite has the highest increase in yield and ultimate tensile strengths.

Table 10. Mechanical properties of *Al* MMCs with 20 volume per cent of reinforcement prepared at DMRL[70]

Material	Young's modulus (GPa)	σ_y (MPa)	σ_{UTS} (MPa)	ε_f (%)
Al	70	71	134	21.0
Al/SiC	97	123	212	11.4
Al/B_4C	101	129	215	9.0
Al/TiC	106	153	262	15.6
Al/TiB_2	94	121	191	14.8

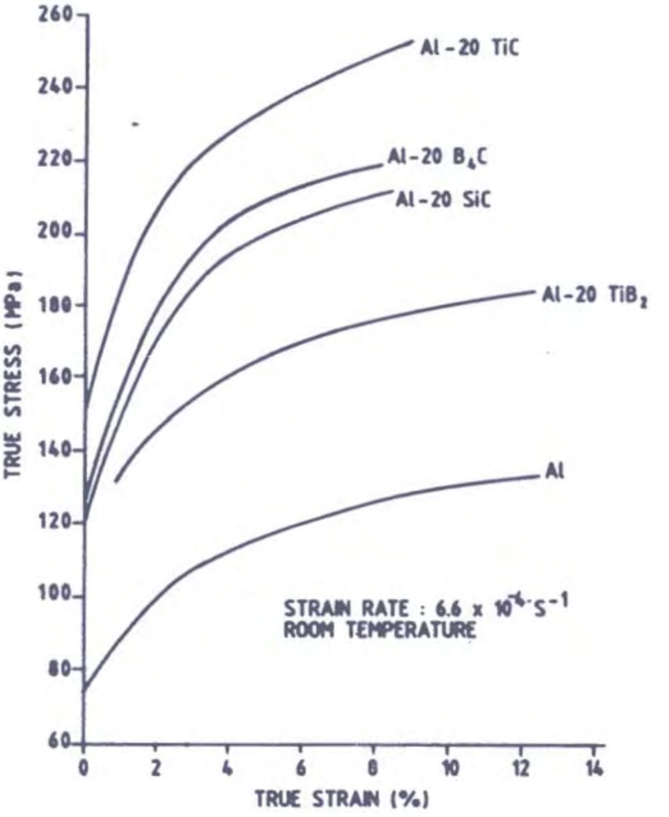
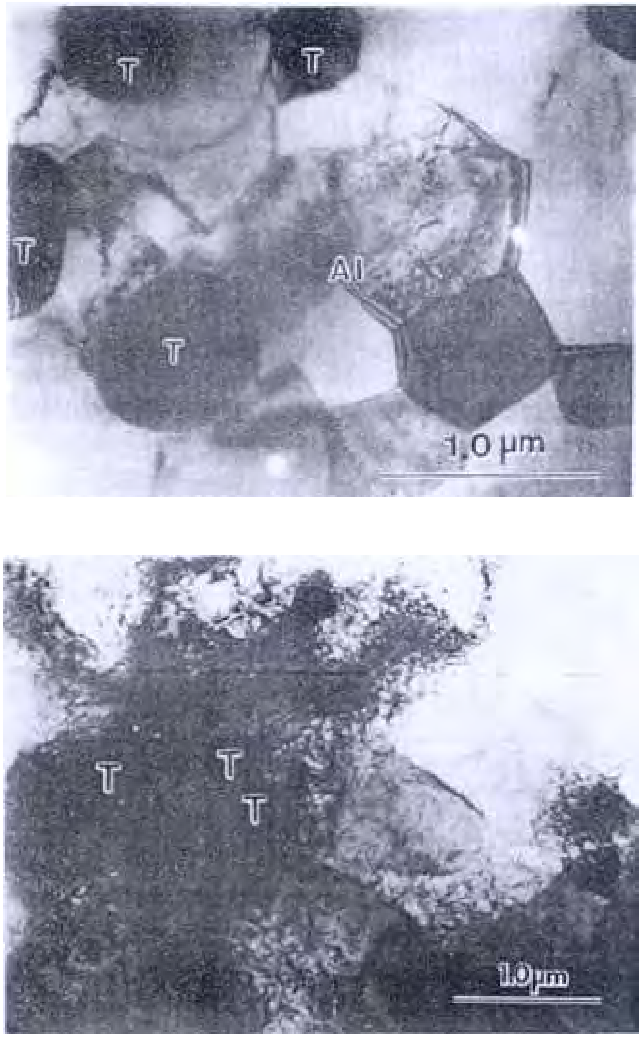


Figure 15. True stress versus true strain curves for *Al*, Al/TiB_2, *Al/SiC*, Al/B_4C and *Al/TiC* metal matrix composites. *Al/TiC* has the highest strength and elongation to failure (ref. 70).

The resistance of the interface to cracking even at high strains in the surrounding matrix is also essential for significant role in load transfer. A brittle interface will crack at lower strains leading to little load transfer. In Table 9, *Al/TiC* composite also shows maximum strain to fracture. Remarkable elongation of 20 per cent or more has been seen in 15 vol per cent XD *Al/TiC* at room temperature, partly as the particles were equiaxed and also because interfacial bonding was excellent. Even after cold rolling to 75 per cent reduction, cracks could be seen at only few interfaces, where particles formed clusters. This is unlike the situation seen in many other interfaces like Al/Al_2O_3[103] and Cu/SiO_2 and Cu/Al_2O_3[104]. The regions around the interface are highly misoriented with respect to surrounding regions, resulting from local lattice rotation as shown[85] in Fig. 16. The mechanism of local lattice rotation in dispersion hardened alloys has been discussed by Ashby[105] and Humphreys[106]. The high ductility of the *Al/TiC* interface can only be explained based on the metallic nature of bonding. In another investigation[107], *Al* containing 15 vol per cent tetragonal DO_{23} structured $Al_3Zr_{25}Ti_{75}$, prepared by casting, was cold rolled to 0.017 per cent of the original thickness with a few intermediate anneals. While the intermetallic was fractured, the *Al* flow kept the interface intact and no cracks could be seen at *Al* intermetallic interfaces. The Al/Al_2O_3, Cu/SiO_2 and Cu/Al_2O_3 interfaces, which are less ductile will crack at smaller strains and not much local lattice rotation can take place as has been explained by Humphreys[106].

10. INTERFACE ENGINEERING

It is well understood how important the interfaces are in determining performance of MMCs. How processing methods, ageing and heat treatments can



(b)

Figure 16. TEM micrographs of microstructures of XD *Al/TiC* composite after cold deformation: (a) rolling; (b) tension test (area close to fracture surface). Regions around the particles are misoriented with respect to each other and hence the contrasts are different (ref. 14).

affect the interfacial bond and properties has been discussed in previous sections. Hence, there is a drive for developing the technology to tailor the interfaces to achieve desired properties. Of course, processing methods and heat treatments can be devised and alloying elements can be added in order to modify interfaces. The other option is to treat the surfaces of the reinforcements, which play a crucial role in wetting. We have previously discussed how wetting can be improved by irradiating and thus introducing charged point and line defects on the surfaces of substrates like fibres and particles. A few more examples of treating reinforcement surfaces are discussed in the following paragraphs.

Mcdonald and Ransley[72] pretreated *TiC* particles by heating in moist hydrogen at 1000 °C leading to the removal of the oxide layer on the surface and partial decarburisation of surfaces. This leaves a metallic *Ti* layer on the surface. It has already been discussed in an earlier section that *TiC* {111} surfaces ending with *Ti* have same surface electronic structure as *Ti* {0001} and *Ti* sites on *TiC* {100} surfaces are more active, when these are carbon deficient. Removal of oxide layer also probably resulted in establishment of direct contact and an equilibrium between the atoms of matrix and reinforcing phase and hence a strong chemical bond. Thus pretreatment with hydrogen resulted in significant increase in Young's modulus. This also proves once again our earlier proposition that metallic bonds result in a more efficient load transfer.

Another and quite popular surface modification method is coating with a third phase to improve wetting or prevent adverse chemical reaction. Coating material is normally wetted well by the *Al* matrix, such as *Ni* and *TiC*. TiB_2 coating applied by chemical vapour deposition (CVD), which is also an excellent wetting agent works on a different principle[108]. Bulk TiB_2 is not wetted as well by *Al* as CVD TiB_2 coating on *SiC* and *C* fibres. It has been reported by Wu[109] recently that CVD TiB_2 contains significant amounts of chlorine and wetting in *C/Al* system is related to the concentration of chlorine. Thus, it was hypothesised that enhanced wetting action was due to the fluxing action of chlorine, which destabilised the surface aluminium oxide and allow direct contact between liquid *Al* and the reinforcement.

Recently, Johnson and Sonuparlak[110] investigated the effect of coating diamond particles (modulus = 1050 GPa) with *SiC* using chemical vapour infiltration. *SiC* not only helps in improving wettability, but retards formation of Al_4C_3, which is detrimental because of its tendency to corrode in moist environments. Also, Al_4C_3 has poor thermal conductivity and it defeats the purpose of reinforcing with diamond for preparing composites having high thermal conductivity. The effect of increasing the *SiC* coating thickness of diamond reinforcements on the physical properties of the composite is shown in Table 10. Al_4C_3 content remained low between 0.047 and 0.12 wt per cent. It is visible that Young's modulus increases with increase in coating thickness as well as there is an overall decrease in the coefficient of thermal expansion. These are due to the

Table 11. Effect of increase in thickness of SiC coatings on diamond particles on the physical properties of 50% diamond/Al MMCs[110]

Coating thickness (μm)	Thermal conductivity (W/m^2-K)	CTE (25-100 °C (ppm/K)	Young's modulus (GPa)
0.41	239	6.8	368
0.53	242	6.5	385
0.97	259	5.2	407
1.23	240	4.6	398
1.42	225	5.0	413
1.60	234	4.5	427

changes in the interface bond because of the coating as well as increase in the volume fraction of SiC from ≈ 3 to ≈ 11 vol per cent.

SUMMARY

In this paper, we have analysed some of the existing information about interfaces in the composites of interest to understand their nature and relationship with mechanical properties. The present paper is based on the ideas evolved from the work carried out at the DMRL, Hyderabad, India and Northwestern University, USA. The variables controlling the characteristics of interfaces are quite large in number, such as processing (technique, temperature, etc) as well as composition and chemical nature of matrix and reinforcement. Processing variables can be manipulated to enhance the wetting between matrix and reinforcement, like addition of alloying elements, pretreatment or coating of reinforcements, increase the temperature of molten metal, hot pressing the blended metal and ceramic powders above solidus temperature, etc. Ageing of the matrix with high solid solubility also affects the interface structure and chemistry.

The chemical bond at the interface can be through a chemical reaction or can be electronic. The former is detrimental as it leads to a brittle reaction product at the interface, whereas, the latter is desirable. For electronic bonding, intimate contact between the atoms of the matrix and of the ceramic phase is necessary and this is found more easily in *in-situ* composites. That is the reason, some matrix-particle interfaces in XD composites are semicoherent and improvement in Young's modulus values are higher than those seen in composites prepared by conventional processes.

Of a variety of reinforcements, the ones having metallic character form metallic bonds with the metal matrix. In such cases, the matrix-reinforcement interfacial bonding is stronger than in other metal-ceramic composites and this leads to a greater improvement in Young's modulus, yield and ultimate tensile strength values over those of the matrix material. Tensile ductility is also impressive in case of metallic bonding at interfaces as they can withstand higher strains. Although, the incorporation of the reinforcements which give rise to metallic bonding at the interfaces results in superior properties as compared to others, most of the potential ceramic reinforcements possess ionic or a covalent bonding. Hence, modification of interfaces is necessary to enhance the metallic character of the bond. Because of the critical role played by interfaces in the composites, it is necessary to plan the choice of matrix and reinforcement and processing conditions while keeping an eye on the nature of the interfaces expected.

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REFERENCES

Chawla, K.K. Composite materials: science end engineering. Springer-Verlag, New York, 1987. p. 83.

2 Sunwoo, H.; Fine, M.E.; Meshi, M. & Stone, D.H. Cyclic deformation of pearlitic eutectoid rail steel. *Metall. Trans. A*, 1982, **13**, 2035-47.

3. Eid, N.M.A. & Thomson, R.F. The nucleation of fatigue cracks in a low-alloy steel under high-cycle fatigue conditions and universal loading. *Acta Metall.*, 1979, **27**, 1239-49.

4 Iseki, T.; Kameda, T. & Maruyama, T. Interfacial reactions between SiC and aluminium during joining. *J. Mater. Sci.*, 1984, **19**, 1692-98.

5 Lee, D.J.; Vaudin, M.D.; Handwerker, C.A. & Kattner, U.A. Phase stability and interface

reactions in the *Al-SiC* system, *Mat. Res. Soc. Symp. Proc.*, 1988, **120**, 357-65.

6. Rack, H.J. P/M aluminium matrix composites. *In* Dispersion strengthened aluminium alloys, edited by Y.W. Kim & W.M. Griffith, TMS, Warrendale, PA, 1988, 649-80.

7. Nutt, S.R. Interfaces and failure mechanisms in *Al-SiC* composites. *In* Interfaces in metal matrix composites, edited by A.K. Dhingra & S.G. Fishman, TMS, Warrendale, PA, 1986, 157-67.

8. Nutt, S.R. & Carpenter, R.W. Non-equilibrium phase distribution in an *Al-SiC* composite. *Mater. Sci. Eng.*, 1985, **75**, 169-77.

9. Henriksen, B.R. & Gionnes, J. Microstructure characterisation and mechanical properties of two *SiC* reinforced composites. *In* Advanced structural inorganic composites, edited by P. Vincenzini. Elsevier Science Publishers, 1991, pp. 251-58.

10. Lucas, J.P.; Yang, N.Y.C. & Stephens, J.J. Interface and near-interface microstructure of discontinuous reinforced metal matrix composites. In structure and properties of interfaces in materials: *In* Mater. Res. Soc. Proc. Symp. 238, edited by W.A.T. Clark, U. Dahmen & C.L. Briant, MRS, Pittsburgh, PA, 1992, pp. 877-83.

11 Yokokawa, H.; Sakai, N.; Kawada, T. & Dokiya, M. Chemical potential diagram of *Al-Ti-C* system: Al_4C_3 formation on *TiC* formed in *Al-Ti* liquids containing carbon. *Metall. Trans. A*, 1991, **22**, 3075-77.

12. Satyaprasad, K.; Mahajan, Y.R. & Bhanuprasad, V.V. Strengthening of *Al*/20 v/o *TiC* composites by isothermal heat treatment. *Scripta Metall.*, 1992, **26**, 711-16.

13. Mitra, R.; Fine, M.E. & Weertman, J.R. Chemical reaction strengthening of *Al/TiC* metal matrix composites by isothermal heat treatment at 913 K. *J. Mater. Res.*, 1993, **8**(9), 2370-790.

14. Mitra, R. Microstructure and interfaces in XD *Al/TiC* metal matrix composites, Northwestern University, Evanston, Illinois, USA, 1993. PhD Dissertation, 170 p.

15 Kuruvilla, A.K.; Bhanuprasad, V.V.; Prasad, K.S. & Mahajan, Y.R. Effect of different reinforcements on composite-strengthening in aluminium. *Bull. Mater. Sci.*, 1989, **125**, 495-505.

16. Levi, C.G.; Abbaschian, G.J. & Mehrabian, R. Interface interactions during aluminium alloy-alumina fibre composites, *Metall. Trans. A*, 1978, **9**, 697-711.

17. Sritharan, T.; Xia, K.; Heathcock, J. & Mihelich, J. Matrix/reinforcement development for aluminium-based composites. *In* Metal and ceramic matrix composites: processing, modelling and mechanical behaviour, edited By R.B. Bhagat, A.H. Clauer, P. Kumar & A.M. Ritter, TMS, Warrendale, PA, 1990, 13-22.

18 Jones, C.; Kiely, C.J. & Wang, S.S. The characterisation of an *SCS6/Ti-6Al-4V* MMC interphase, *J. Mater. Res.*, 1989, **4**, 327-35.

19. Reeves, A.J.; Stobbs, W.M. & Clyne, T.W. The effect of interfacial behaviour of *Ti* reinforced with *SiC* and TiB_2 particles. *In* Metal matrix composites-processing, microstructure and properties, edited by N. Hansen; D. Juul Jensen; T. Leffers; H. Liholt; T. Lorentzen; A.S. Pederson; O.B. Pederson & B. Ralph, Riso National Laboratory, Rosklide, Denmark, 1991, pp. 631-36.

20. Konitzer, D.G. & Loretto, M.H. Microstructural assessment of *Ti6Al4V-TiC* metal matrix composite, *Acta Metall.*, 1989, **37**, 397-406.

21 Konitzer, D.G. & Loretto, M.H. Microstructural assessment of interaction zone in titanium aluminide/*TiC* metal matrix composite. *Mater. Sci. Tech.*, 1989, **5**(7), 627-31.

22 Brown, N.R. Reaction zones at the interfaces in Magnesium metal matrix composites, Northwestern University, Evanston, Illinois, USA, June, 1993, PhD Dissertation.

23. Warwick, M. & Clyne, R.T.W. Microstructural stability of fibrous composites based on magnesium-lithium alloys. *In* Developments in the science and technology of composite Materials: Third European Conference on Composite Materials, edited by A.R. Bunsell, P. Lamicq & A. Massiah, Bordeaux, France, Elsevier Applied Science, 1989, pp. 205-12.

24 Wang, L. & Arsenault, R.J. Interfaces in XD processed $TiB_2/NiAl$ composites. *Metall. Trans. A*, 1991, **22**, 3013-18.

25. Fuchs, G.E. Development of *Ni* Aluminide based composites for elevated temperature applications. *In* Metal & ceramic matrix composites: processing, modelling and mechanical behaviour, edited by R.B. Bhagat, A.H. Clauer, P. Kumar and A.M. Ritter, TMS, Warrendale, PA, 1990, pp. 391-400.

26. Norman, J.H.; Reynolds, G.H. & Brewer, L. Chemical stability of fiber-metal matrix composites. *In* Mater. Res. Soc. Symp. Proc., MRS, Pittsburgh, PA, 1990, **194**, 369-377.

27. Feng, C.R.; Michael, D.J. & Crowe, C.R. Microstructural characteristics of two-phase titanium aluminides, *Mater. Sci. Eng. A*, 1991, **145**, 257-64.

28. Meschter, P.J. & Schwartz, D.S. JOM, 1989, **41**(11), 52-55.

29. Clough, R.B.; Biancaniello, F.S.; Wadley, H.N.G. & U.R. Kattner, U.R. Fiber and interface fracture in single crystal aluminium/*SiC* fiber composites. *Metall. Trans. A*, 1990, **21**, 2747-57.

30. Voight, W. *Lehrbuch der Kristallphysik Leipzig, Teubner*, 1928, p. 716.

31. Reuss, A. Z. Angew. *Math. Mech.*, 1929, **9**, 55.

32. Mahajan, Y.R.; Jain, M.K. & Bhanuprasad, V.V. DMRL, Hyderabad, 1990. (Unpublished.)

33. Mahajan, Y.R. & Rama Rao, P. Discontinuously reinforced metal matrix composites. *In* New Materials, edited by S.K. Joshi; C.N.R. Rao; T. Tsuruta & S. Nagakura, Narosa Publishing House, New Delhi, 1992, p. 322.

34. Lagace, H. & Lloyds, D.J. Microstructural analysis of *Al/SiC* composites. *Canadian Met. Quart.*, 1989, **28**, 145-52.

35. Jain, M.K.; Bhanuprasad, V.V.; Kamat, S.V.; Pandey, A.B.; Varma, V.K.; Bhat, B.V.R. & Mahajan, Y.R. Processing, microstructure and properties of 2124 *Al-SiC*$_p$ composites. *Inter. J. Powder. Metall.*, 1993, **29/3**, 267-75.

36. Hawk, J.A.; Mirchandani, P.K.; Benn, R.C. & Wilsdorf, H.G.F. The effect of *Ti* additions on the elevated temperature strength and microstructural stability of MA aluminium alloys. *In* Dispersion strengthened aluminium alloys, edited by Y.W. Kim & W.M. Griffith. TMS, Warrendale, PA, 1988, p. 551.

37. Zedalias, M.S.; Ghate, M.V. & Fine, M.E. Elastic moduli of Al_3Zr. *Scripta Metall.*, 1985, **19**, 647-50.

38. Skinner, D.J. The physical metallurgy of dispersion strengthened *Al-Fe-V-Si* alloys. *In* Dispersion strengthened aluminium alloys, edited by Y.W. Kim & W.M. Griffith. TMS, Warrendale, PA, 1988, p. 181.

39. Fine, M.E. Elastic moduli of two phase aluminium alloys. *Scripta Metall.*, 1981, **15**, 523-24.

40. Krock, R.H. The elastic modulus of some dispersed-phase composite materials. *In* Modern developments in powder metallurgy, edited by H.H. Hausner, Plenum Press, New York, 1966, pp. 105-13.

41. Bonollo, F.; Guerriero, R.; Sentimenti, E.; Tangerini, I. & Jang, L.M. A study on P/M *Al/SiC*$_p$ metal matrix composites. *In* Advanced structural inorganic composites, edited by P. Vincenzini, Elsevier Science Publishers, 1991, pp. 259-68.

42. Lyle, J.P., Jr. In aluminium, edited by K.R. Van Horn, ASM, Metals Park, OH, 1967. V. 1, Chap. 10, p. 337.

43. Rhee, S.K. Wetting of ceramics by liquid aluminium. *Amer. Ceram. Soc.*, 1970, **53**, 386-89.

44. Delannay, F.; Froyen, L. & Deruyttrere, A. The wetting of solids by molten metals and its relation to molten metals. *J. Mater. Sci.*, 1987, 22, 1-16.

45. Ramqvist, J.V. Wetting of metallic carbides by liquid copper, nickel, cobalt and iron. *Inter. J. Powder Metall.*, 1965, **1**(4), 2-21.

46. Russel, K.C.; Oh, S.Y. & Figureds, A. Theoretical and experimental studies of ceramics : metal wetting. *MRS Bull.*, 1991, **XVI**(4), 46-52.

47. Cisse, J.; Bolling, G.F. & Keer. H.W. Crystallographic orientations between aluminium grown from melts and different titanium compounds. *J. Cryst. Growth*, 1972, **13/14**, 771-81.

48. Bradshaw, A.M.; Van der Veen, J.F.; Himpsel, F.J. & Eastman, D.E. Electronic properties of the clean and hydrogen covered *TiC* (111) Ti-terminated polar surface. *Sol. State Comm.*, 1980, **37**, 37-40.

49. Samsonov, G.V.; Panasyuk, A.D. & Kozina, G.K. Wetting of refractory carbides with liquid metals. *Sov. Powder Metall. Met. Ceram.*, 1968, **71**, 874-78.

Manning, C.R. & Gurganus, T.B. Wetting of binary aluminium alloys in contact with B, B_4C and graphite. *J. Amer. Ceram. Soc.*, 1969, **52**, 115-18.

51 Mitra, R.; Chiou, W.A.; Fine, M.E. & Weertman, J.R. Interfaces in as-extruded XD *Al/TiC* and Al/TiB_2 metal matrix composites. *J. Mater. Res.*, 1993, **8**(9), 2380-920.

52 Gao, Y. & Merkle, K.L. Equilibrium and non-equilibrium metal-ceramic interfaces. *In* Structure and properties of interfaces in materials, edited by W.A.T. Clark, U. Dahmen & C.L. Briant, MRS Fall Meeting, 2-6 December, 1991, Boston. MRS, Pittsburgh, PA, 1992, 238, pp. 775-80.

53 Ohuchi, F.S. *Cu-AlN* interface: electronic structure and adhesion. *In* Interface science and engineering: J. de. Physique, edited by R. Raj & S.L. Sass, New York, 1987, V. C5, pp. 783-89.

54 Stoneham, A.M. & Tasker, P.W. Metal-non-metal and other interfaces: the role of image interactions. *J. Phys.*, 1985, **C18**, L543-47.

55 Finnis, M.W.; Stoneham, A.M. & Tasker, P.W. Approaches to modelling metal/ceramic interfaces. *In* Metal-ceramic interfaces: acta-scripta metallurgica proceedings series, V. 4, edited by M. Ruhle, A.G. Evans, M.F. Ashby & J.P. Hirth, Pergamon Press, 1990, pp. 35-44.

56. Nogi, K.; Tsujimoto, M.K.; Ogino & Iwamoto, N. Wettability of *MgO* single crystal by liquid pure *Sb, Sn* and *Bi*. *Acta Metall.* , 1992, **40**, 1045-50.

57. Goretzki, H. Neutron diffraction studies on titanium-carbon and zirconium carbon alloys. *Phys. Stat. Sol.*, 1967, **20**, K141-43.

58. Tombrello, T.A. *In* Mater. Res. Soc. Symp. Proc., 1984, **25**.

59. Young, T. *Trans. Royal Soc.*, 1805, **94**, 65.

60. Easterling, K.E.; Fischmeister, H.F. & Navara, E. The particle-to-matrix bond in dispersion-hardened austenitic and ferritic iron alloys. *Powder Metall.*, 1973, **16**, 128-45.

61 Fine, M.E.; Bourell, D.L.; Eliezer, Z.; Persad, C. & Marcus, H.L. Basic principles for selecting phases for high temperature metal matrix composites: interfacial considerations. *Scripta Metall.*, 1988, **22**, 907-10.

62. Calderon, H.A.; Fine, M.E. & Weertman, J.R. Coarsening and morphology of particles in *FE-Ni-Al-Mo* ferritic alloys. *Metall. Trans. A*, 1988, **19**, 1135-46.

63 Quigley, B.F.; Abbaschian, G.J.; Wunderlin, R. & Mehrabian, R. A method for fabrication of aluminium-alumina composites. *Metall. Trans. A*, 1982, **13**, 93-100.

64 Bhanuprasad, V.V.; Prasad, K.S.; Kuruvilla, A.K.; Pandey, A.B.; Bhat, B.V.R. & Mahajan, Y.R. Composite strengthening in 6061 and *Al-4Mg* alloys. *J. Mater. Sci.*, 1991, **26**, 460-66.

65. Dhingra, A.K. *Phil. Trans. R. Soc.*, 1980, **A294**, 151.

66 Webster, D. Effect of lithium on the mechanical properties and microstructure of *SiC* whisker reinforced aluminium alloys. *Metall. Trans. A*, 1982, **13**, 1511-19.

67 Kimura, Y.; Mishima, Y.; Umekawa, S. & Suzuki, T. Compatibility between carbon fibre and binary aluminium alloys. *J. Mater. Sci.*, 1984, **19**, 3107-14.

68. Skinner, D.J.; Zedalias, M.S.; Frazier, W.E.; Kockzak, M.J. & Sahoo, P. The microstructural stability of *TiC* reinforced rapidly solidified *Al-Fe-V-Si* composite structures. *In* Metal and ceramic matrix composites: processing, modelling & mechanical behaviour, edited by R.B. Bhagat, A.H. Clauer, P. Kumar & A.M. Ritter, TMS, Warrendale, PA, 1990, 535-41.

69 Schroder, J.; Kainer, K.U. & Mordike, B.L. Production technologies, microstructures and mechanical properties of *SiC* particle reinforced magnesium alloys. *In* Proc. P/M aero 91: Int. Conf. on P/M aerospace materials 1991, 4-6 Nov, 1991, Liausanne, Switzerland, paper 38, pp. 38-1-38-15.

70. Bhanuprasad, V.V.; Roy, M.; Kamat, S.V. & Mahajan, Y.R. DMRL, Hyderabad. (Unpublished).

71 Schroder, J.; Kainer, K.U. & Mordike, B.L. Particle reinforced magnesium alloys. *In* Developments in the science and technology of composite materials. Proceedings of the third European Conference on Composite Materials, Bordeaux, France, March 20-23, 1989, edited by A.R. Bunsell, P. Lamicq & A. Massiah, Elsevier Applied Science, 1989, pp. 221-26.

72 Mcdonald, N.F. & Ransley, C.E. Preparation of high-modulus aluminium alloys by powder metallurgy. *In* Metal powder review: symposium on powder metallurgy, 1954, pp. 242-48.

73 Chiou, J.M. & Chung, D.D.L. Characterization of metal-matrix composites fabricated by vacuum infiltration of a liquid metal under an inert gas pressure. *In* Metal and ceramic matrix composites: processing, modelling and mechanical behaviour, edited by R.B. Bhagat, A.H. Clauer, P. Kumar & A.M. Ritter, TMS, Warrendale, PA, 1990, pp. 107-115.

74 Lloyd, D.J. & Jin, I., A method of assessing the reactivity between *SiC* and molten *Al, Metall. Trans. A*, 1988, **19**, 3107-09.

75 White, J.; Willis, T.C.; Hughes, I.R. & Jordan, R.M. Metal matrix composites produced by spray deposition. *In* Dispersion strengthened aluminium alloys, edited by Y.W. Kim & W.M. Griffith, TMS, Warrendale, PA, 1983, 693-708.

76. Homeny, J.; Neergard, L.J.; Harasek, K.R.; Donner, J.T. & Bradley, S.A. Characterization of silicon nitride whiskers. *J. Amer. Ceram. Soc.*, 1990, **73**(1), 102-05.

77 Ning, X.G.; Pan, J.; Hu, K.Y. & Ye, H.Q. High resolution electron microscopy studies on the microstructure of $Si_3N_{4w}/6061Al$ composite. *In* Structure and properties of interfaces in solids, proceedings of MRS fall meeting, 2-6 December, 1991, Boston, USA, edited by W.A.T. Clark, U. Dahmen & C.L. Briant, MRS, Pittsburgh, PA, 1992, V. 238, pp. 865-70.

78 Marcus, H.L. & Rabenberg, L.K. Interface characteristics and mechanical properties of metal-matrix composites, ONR Contract N00014-84-K-0687, Technical report UTCMSE-87-3, p. 160-65.

79 Ribes, H.; Da Silva, R.; Suery, M. & Bretheau, T. Effect of interfacial oxide layer in *Al-SiC* particle composite on bond strength and mechanical behaviour. *Mater. Sci. Tech.*, 1990, **6**, 621-28.

80 Xia, K.; Nieh, T.G.; Wadsworth, J. & Langdon, T.G. *In* Fundamental relationships between microstructure and mechanical properties of metal-matrix composites, edited by P.K. Liaw & M.N. Gungor, TMS, Warrendale, PA, (1990), p. 543.

81 Westwood, A.R.C. Materials for advanced studies and devices. *Metall. Trans. A*, 1988, **19**, 749-58.

82. Koczak, M.J. & Kumar, K.S. U.S. Patent 4, 808, 1989. 372.

83 Mader, W. Structural relaxations at metal/metal oxide interfaces. Structure and properties of interfaces in solids,. *In* Proceedings of MRS fall meeting, 2-6 December, 1991, Boston, USA, edited by W.A.T. Clark, U. Dahmen and C.L. Briant, MRS, Pittsburgh, PA, 1992, 238, pp. 763-74.

84. Henager, C.H. Jr; Brimhall, J.L. & Hirth, J.P. Synthesis of composites in situ using displacement reactions. *Ceram. Sci. Eng. Proc.*, 1992, **13**, 596-604.

85 Fine, M.E.; Mitra, R. & Weertman, J.R. Interfaces in *Al-TiC* composites prepared by *in-situ* processing. *Z. Metallkd.*, 1993, **4**, 282-85.

86. Mitra, R.; Chiou, W.A.; Weertman, J.R.; Fine, M.E. & Aikin, Jr., R.M. Study of interfaces in XD *Al/TiC* metal matrix composites. Structure and properties of interfaces in materials. *In* Proceedings of MRS fall meeting, 2-6 December, 1991, Boston, USA, edited by W.A.T. Clark, U. Dahmen and C.L. Briant, MRS, Pittsburgh, PA, 1992, V. 238, pp. 871-76.

87 Westwood, A.R.C. & Winzer, S.R. Advanced ceramics. *In* Advancing materials research, edited by P.A. Paras and H.D. Langford, Nat. Acad. Press, Washington, D.C. 1987, pp. 225-43.

88. Kuruvilla, A.K.; Prasad, K.S.; Bhanuprasad, V.V. & Mahajan, Y.R. Microstructure property correlation in Al/TiB_2 (XD) composites *Scripta Metall.*, 1990, **24**, 873-78.

89 Hashin, Z. & Shtrikman, S. A variational approach to the theory of the elastic behaviour of multiphase materials. *J. Mech. Phys. Solids*, 1963, **11**, 127-40.

90. Aikin, Jr., R.M. The mechanisms of dispersion strengthening and fracture in *Al*-based XD^{TM} alloys. Martin Marietta Laboratory, Baltimore, MD,NASA Contractor Report 4365, Contract NAS 1-18531.

91 Manoharan, M. & Lewandowski, J.J. Effects of ageing condition on the fracture toughness of 2XXX and 7XXX series aluminium alloy composites. *Scripta Metall.*, 1989, **23**, 301-04.

92. Satyaprasad, K.; Varma, V.K.; Bhanuprasad, V.V. & Mahajan, Y.R. DMRL, Hyderabad. (Unpublished.)

93. Warner, T.J. & Stobbs, W.M. Clean or dirty interfaces? The influence of interfacial morphology on the mechanical properties of MMCs. *In* ICCM: Proc. 7th Inter. Conf. on Comp. Mater. Nov. 22-24, 1989 at Gaungzhou, China, edited by Wu Yunshu, Gu Zhenlong, & Wu Renjie, International Academic Press and Pergamon Press, 1989, 1, pp. 503-08.

94. Wu, Y. & Lavernia, E.J. Spray-atomised and codeposited 6061 *Al/SiC$_p$* composites. *JOM*, 1991, **43**(8), 16-23.

95. Arsenault, R.J. Interfaces in metal matrix composites. *Scripta Metall.*, 1984, **18**, 1131-34.

96. Chawla, K.K.; Xu, Z.R.; Mitra, R. & M.E. Fine. Dept. of Materials and Metallurgical Engineering, New Mexico Institute of Mining and Technology, Socorro, NM, USA. (Unpublished).

97 Hasselman, D.P.H. & Donaldson, K.Y. Effect of reinforcement particle size on the thermal conductivity of a particle-silicon carbide-reinforced aluminium matrix composite. *J. Amer. Ceram. Soc.*, 1992, **75**(11), 3137-40.

98 Geiger, A.L.; Hasselman D.P.H. & Donaldson, K.Y. Effect of reinforcement particle size on the thermal conductivity of a particulate silicon carbide-reinforced aluminium-matrix composite. *J. Mater. Sci. Lett.*, 1993, **12**, 420-23.

99 Geiger, A.L. & Jackson, M. Low expansion MMCs boost avionics. *Adv. Mater. Proc.*, 1989, 7, 23-30.

100. Reeves, A.J.; Taylor, R. & Clyne, T.W. The effect of interfacial reaction on thermal properties of titanium reinforced with particulate *SiC. Mater. Sci. and Eng.*, 1991, **A141**, 129-38.

101 Hasselman, D.P.H. & Johnson, L.F. Effective thermal conductivity of composites with interfacial thermal barrier resistance. *J. Compos. Mater.*, 1987, **21**, 508-15.

102. Zhang, J.; Perez, R.J.; Gupta, M. & Lavernia, E.J. Damping behaviour of particulate reinforced 2519 *Al* matrix composites. *Scripta Metall.*, 1993, **28**, 91-96.

103 Ruedl, E. Void formation at the interface between particles & matrix in deformed Al-Al2O3 foils. *J. Mater. Sci.*, 1969, **4**, 814-16.

104 Humphreys, F.J. & Stewart, A.T. Dislocation generation at *SiO$_2$* particles in an brass matrix on plastic deformation. *Surf. Sci.*, 1972, **31**, 389-421.

105. Ashby, M.F. Work hardening in dispersion hardened crystals, *Phil. Mag.*, 1966, **14**, 1157-78.

106. Humphreys, F.J. Local lattice rotation at second phase particles in deformed metals. *Acta Metall.*, 1979, **27**, 1801-14.

107. Earvolino, P.A.; Fine, M.E.; Weertman, J.R. & Parameswaran, V.R. Processing an $Al/Al_3Zr_{0.25}Ti_{0.75}$ metal matrix composite by conventional melting, casting & rolling. *Scripta Metall.* , 1992, **26**, 945-48.

108. Maruyama, B.; Barrera, E.V. & Rabenberg, L. Characterization and modification of composite interfaces. *In* Metal matrix composites: processing and interfaces, edited by R.K. Everett & R.J. Arsenault, Academic Press, San Diego, CA, 1991, pp. 181-216.

109. Wu, R. Interfacial behaviors and its influence on the mechanical properties of carbon fiber reinforced aluminium composites. *In* Interfaces in polymer, ceramic and metal matrix composites, edited H. Ishida, Elsevier, New York, 1988, pp. 43-56.

110. Johnson, W.B. & Sonuparlak, B. Diamond/*Al* metal matrix composites formed by the pressureless metal infiltration process. *J. Mater. Res.*, 1993, **8**, 1169-73.