JOURNAL OF MATERIALS SCIENCE **36** (2001) 4519–4530

Review on TiC reinforced steel composites

K. I. PARASHIVAMURTHY
Siddaganga Institute of Technology, Tumkur 572103, India
R. K. KUMAR, S. SEETHARAMU
Central Power Research Institute, Bangalore 560079, India
M. N. CHANDRASEKHARAIAH
Siddaganga Institute of Technology, Tumkur 572103, India
E-mail: mnchandra@usa.net

Particulate reinforced ferrous-based metal matrix composites exhibit both excellent wear and cutting properties, and can give significant cost reductions over existing materials used in heavy wear applications. This paper reviews the current status of literature on titanium carbide–reinforced steel matrix composites and looks into the different types and methods of reinforcements being used, together with other alternative processing routes. Mechanical properties such as elastic modulus, low and high temperature strengths and wear properties are discussed as a function of the volume fraction of reinforcement. The review concludes by underlining the importance of further research in some critical areas to fully realize the industrial potential of these composites.

1. Introduction

The attractive physical and mechanical properties that can be obtained with metal matrix composites (MMC's), such as high specific modulus, strength, and thermal stability have been documented extensively [1–7]. MMC's are the focus of intense research and development world wide for aerospace, automotive and other applications. Much of this interest is centered on the lighter structural metals such as titanium, magnesium, and aluminum, primarily to attain improved strength and stiffness [8]. The incorporation of high modulus and high strength reinforcements into steel significantly improves the specific modulus, specific strength, and wear resistance of composites [9]. The addition of vanadium and niobium which produces carbide dispersions in steel alloy matrices, also offers encouragement [10–12].

Titanium carbide bounded by steel is a unique metal matrix composite. Its special properties such as strength retained at elevated temperatures and increase in machinability in the annealed condition, were studied as early as the 1960s [13]. Use of this composite was restricted to highly specific applications due to complex fabrication routes, limited formability coupled with the prohibitive cost of carbides.

The advantages of the MMC's over the conventional cemented tungsten carbide, as well as their application to high performance toolings and other engineering structures have been extensively discussed [14]. Thus, titanium carbide reinforced steel matrix composites were excluded form consideration in many particle component configurations due to their complex fabrication route and cost constraints. Typical applications for those particulate composites include wear parts such as gears, bearings, and shafts, cutting tools and forming tools [15].

There are many complex problems in the development of particulate-reinforced steel composites. The fabrication process of this kind of MMC's by conventional powder metallurgy has several limitations for the homogeneity of the material. In recent studies, solidification processing has emerged as one of the most economical and versatile technique to produce steel based MMC's.

A number of investigators [9, 16–19] have studied the mechanics and precipitation kinetics to understand the solid/liquid type of reaction. Recently alternative-reinforcing phases were tried and the processing parameters on the dispersion of titanium carbide in liquid iron alloy matrix was also examined [18]. A novel rapid assessment technique has been developed to evaluate the degree of wetting and reaction of solid particles with liquid metals [20]. However, reports [16] suggest that for pure iron or low carbon alloys, dispersion of titanium carbide was minimal. Titanium carbide tends to dissolve and super saturation causes some TiC to be precipitated on cooling. The dispersion of TiC improved with increasing carbon content in the iron.

The physical and mechanical properties are sensitive to the type of reinforcement, mode of manufacturing, secondary processing and heat treatment. While there are still many areas which lack understanding, recent work has highlighted new directions to achieve optimum properties in these composites.

This paper reviews the current understanding of particulate reinforced steel composites, examines the

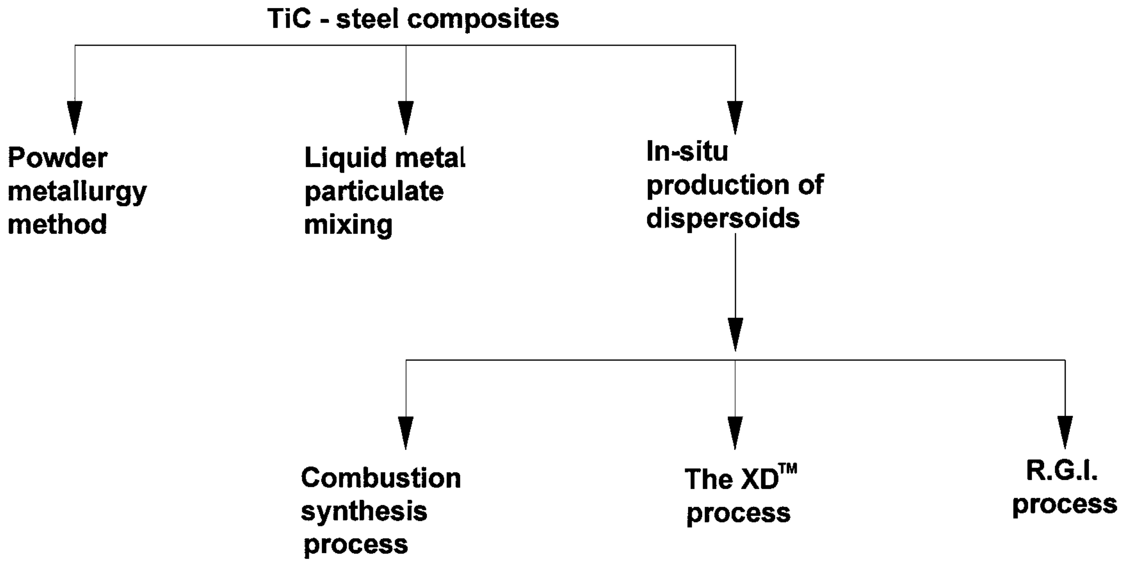

Figure 1 Methods of producing TiC reinforced steel matrix composites.

various factors affecting (a) processing and (b) mechanical behaviour of steel composites. In particular, this work will address the issue of the compatibility between reinforcement and matrix, and that of the interfacial activity on the resulting strength, and wear life of MMC's at both ambient and elevated temperatures.

2. Composite processing

A variety of processing techniques have evolved over the last two decades in an effort to optimize the structure and properties of particulate reinforced MMC's [21–35]. Different processes for fabrication of steel matrix based composites are given in Fig. 1.

2.1. Powder metallurgy method

Composites are produced in their final form by pressing metal powder into the desired shape, usually in a metal mould, and then heating the compacted powder, either concurrently or subsequently for a period of time at a temperature below the melting point of the major constituent [36–38]. Solid phase processes invariably involve the blending of rapidly solidified powder with particulates, using a number of steps as shown in Fig. 2.

In the elemental powder blending technique, reinforcement and base elemental powder are blended to obtained a uniform composition, pressing to approximately 75% density, and final consolidation by extrusion, forging, rolling or some other hot working method. Cold and hot isostatic pressing (CHIP) involves, cold isostatic pressing (CIP) of blended powder in a reusable elastomeric mould followed by vacuum sintering and then hot isostatic pressing (HIP) without the need for additional expensive tooling [40, 41].

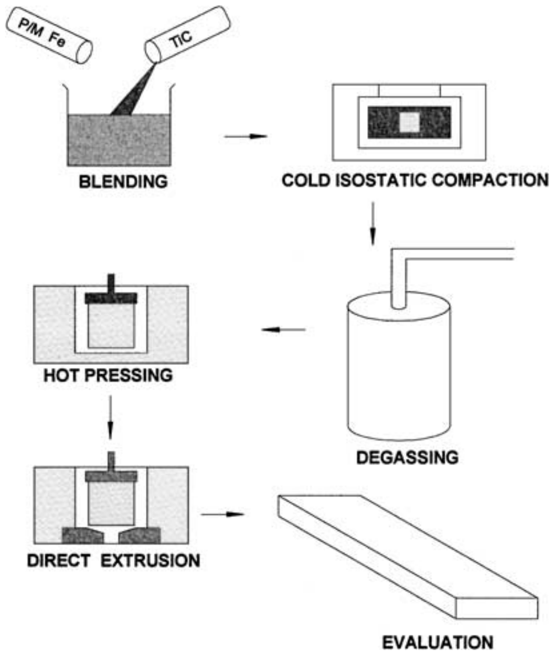

Figure 2 Schematic diagram of PM-MMC fabrication steps [39].

Powder metallurgy roots are currently employed on an industrial scale to produce iron based TiC metal matrix composites [42–46] marketed under a number of trade names. Seilstofer (Germany) have recently produced TiC ALLOY (trade name) composites consisting typically 40 vol. % TiC in a heat treatable steel alloy matrix by hot isostatic pressing. In addition, materials containing 45–50 vol. % TiC embedded in high alloy steel matrix are available from Thyssen (Germany) under the trade name FERRO-TITANIT.

The powder metallurgy method has some potential benefits in obtaining near-net shaped products but the capital investment is still high compared to conventional and other emerging techniques. It also has several limitations for the homogeneity of the material including density and intricate shape casting.

2.2. Liquid metal – carbide particulate mixing

Solidification processing route for particulate reinforced steel composites have the advantages of simplicity, flexibility, cheapness and ease of production of composites with complex shapes. Liquid metallurgy method of making a particulate composite consists of introduction, retention and suspension of solid particles in liquid alloy followed by casting into suitable moulds. However, there is very little discussion in the literature about attempts to produce particulate-reinforced steel composites through the solidification route.

In one such attempt [8], composite was produced by induction heating of iron and TiC powder mixed with molten iron to get TiC-steel composites. The interface formed between the matrix and the reinforcement is of interest since the characteristics of this region determines the load transfer and crack resistance of the MMC during deformation. Systematic studies of the metal-ceramic interfaces were initiated in the late sixties [47]. It is now widely accepted that, in order to maximize interfacial bond strength in MMC's, it is necessary to promote wetting, control chemical interactions and minimize oxide formation.

The fundamentals of wetting of solids by molten metals in relation to the preparation of metal matrix composites have been reviewed by Delannay *et al.* [48]. Wetting is effected between a metal and a liquid when the strength of the interfacial bond exceeds the surface tension of the liquid.

The wettability can be obtained by measuring the contact angle θ, formed between a solid and liquid as defined by Young's equation [49]. (Fig. 3)

$$\gamma_{sg} = \gamma_{lg} \operatorname{Cos} \theta + \gamma_{sl} \qquad (1)$$

Where γ_{sl}, γ_{sg}, and γ_{lg} are the interfacial energies between solid and liquid, solid and gas, and liquid and gas phases, respectively.

The driving force D_f for wetting and dispersion is affected by the surface tension of the liquid metal and the strength of the solid-liquid interface as measured by the work of adhesion W_{ad}. When the surface tension of the liquid is known, the work of adhesion can be determined experimentally from the measurement of the contact angle θ, using the equation [50].

$$W_{ad} = \gamma_{lg}(1 + \cos\theta) \qquad (2)$$

Hence, wetting is achieved when $\theta < 90°$ (i.e. when $\gamma_{lg} > \gamma_{sl}$) or when the driving force (D_f) for wetting exceeds the liquid interface energy ($D_f > \gamma_{lg}$) [50]. The value of D_f depends on the surface tension of the liquid and the strength of the liquid-solid interface. Wetting is difficult to achieve in molten metal ceramic system as a result of the high surface tension commonly associated with the metal. For example, wetting of carbon and SiC by aluminum and its alloys has been measured and found to be poor below 950 °C [51]. Measured contact angles for iron and TiC, according to Naidich [52], are given in Table I.

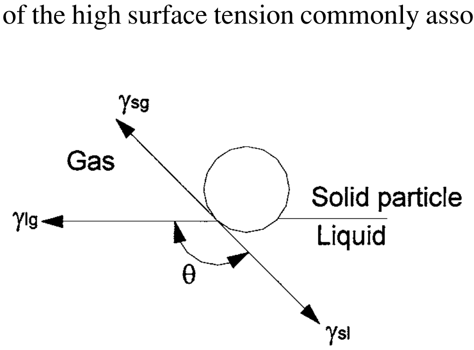

Figure 3 Contact angle formed between solid, liquid and gas phases [49].

TABLE I Wettability of TiC by iron [52]

T °C	Atmosphere	Contact angle θ in deg.
1550	Hydrogen	49
1550	Helium	36
1490	Vacuum	28

In sessile drop measurements of the interfacial tension of the tool steel on titanium carbide, Kiparisou [53, 54] found that less wetting occurs in hydrogen than in argon atmosphere.

The dispersion test developed and conducted by Terry and Chinyamakobvu [8, 55] to produce TiC-reinforced steel composites showed that the use of low carbon iron alloy favours dissolution of the TiC whereas their dispersion is favoured by high contents of dissolved carbon or titanium in liquid iron. Under such conditions, good wetting of the TiC by the liquid alloy permits the production of a well dispersed TiC composites [55].

Pseudobinary section of $TiC_{0.9} - Ti_{0.015}\ Fe_{0.985}$ as shown in Fig. 4 (according to Booker [56]) indicated that the solubility of TiC in liquid iron decreases with decreasing temperature from about 22 mol-% at 1600 °C to about 13 mol-% at the eutectic temperature of 1475 °C. The iron rich corner of the Fe-Ti-C phase diagram according to Chambers *et al.* [57] based on the data obtained by Murakami [58] is shown in the Fig. 5.

The figure gives information about the effect of dissolved carbon and titanium on the solubility of TiC in liquid Fe-Ti-C. Fig. 6 shows the vertical section parallel to the Fe-C side, calculated at various mole fractions of titanium showing the dissolution temperature of TiC_y as a fraction of carbon content by Jonsson [59]. From Figs 4, 5 and 6 it is seen that the use of low carbon iron alloys favours dissolution of TiC. Terry and Chinyamakobvu [55] also showed that the TiC would

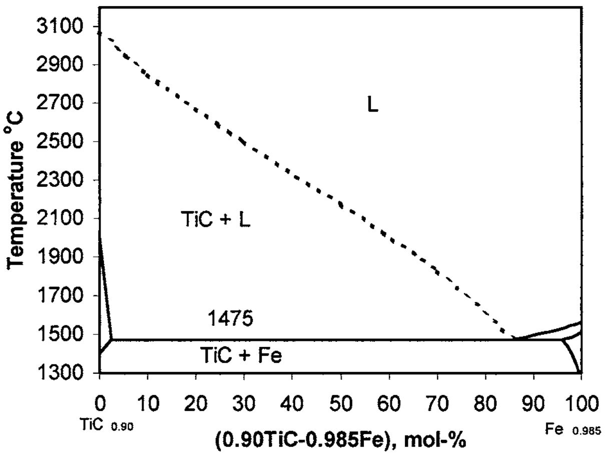

Figure 4 Pseudobinary section $TiC_{0.9}$–$Ti_{0.015}\ Fe_{0.985}$ according to Ref. [56].

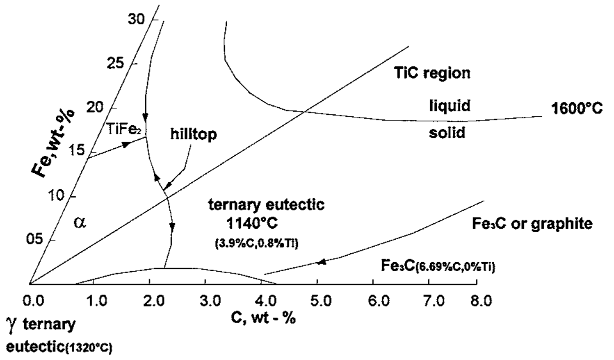

Figure 5 Iron rich corner of Fe-Ti-C phase diagram according to Ref. [57] based on data from Ref. [58].

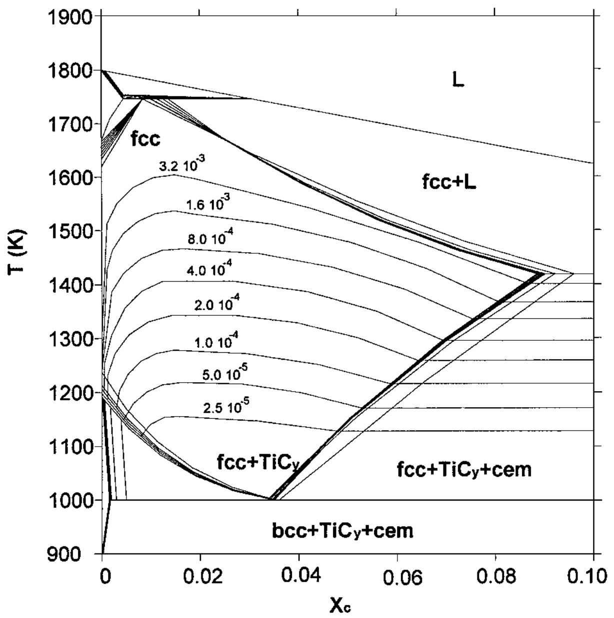

Figure 6 Vertical section parallel to the Fe-C side, calculated at various mole fraction of Ti showing the dissolution temperature of TiC_y as a function of carbon content [59].

be expected to dissolve if the proportions added are below the solubility limits of TiC in pure molten iron.

It appears, therefore, that the kinetics of TiC dissolution presents a barrier to the reaction of TiC, especially in the higher carbon ranges in metals. Precipitation of TiC from the Fe-Ti-C matrix would be expected to occur on cooling, both in the liquid and solid states, as the solubility decreases with temperature.

2.3. *In situ* production of dispersoids

2.3.1. Combustion synthesis process

During 1972, Merzhanov and Borovinskaya [60] and others in the Soviet Union pioneered to produce intermetallic compounds and ceramics, through exothermic reactions where adequate heat is liberated to sustain the reaction by the rapid propagation of combustion without further addition of energy [61, 62]. This process is called Self-propagating High-temperature Synthesis (SHS) or combustion synthesis process.

Systems for *in situ* production of reinforcements could be designed on the basis of thermodynamic and kinetic studies. A reacting constituent A, may be in solution in a molten alloy M, and the other constituent B, may be dispersed as solid particle which may also dissolve in M and react as [63]

$$A(M) + B(M) \rightarrow AB(S) \tag{3}$$

At equilibrium, the free energy change is given by

$$\Delta G^\circ = -RT \ln(1/a_A a_B) \tag{4}$$

Where T is temperature of the molten alloy, a_A and a_B are activities of A and B in M, and R is the universal gas constant. Whenever activities of A and B are such that

$$a_A a_B > e^{\Delta G^\circ/RT} \tag{5}$$

there will be formation of compound AB. At higher temperatures, the exponential term in Equation 5 will be less, and thus the extent of reaction will be more.

The SHS process involves the mixing and compacting powders of the constituent elements and igniting at a reasonable temperature thus creating a combustion wave which propagates throughout the entire portion, involving the uniform heating of the whole sample until the ignition point is reached [64]. The source of heat can be electrical sparks, induction heating, laser, heating wire, conventional furnace etc. [65]. The basic advantage of this process is the self-generation of energy which avoids costly equipments and also, as very high temperature is reached in many combustion synthesis reactions, all volatile impurities are evaporated at these temperatures producing high purity products [66–70].

An important parameter, the highest temperature to which the product phase may be raised is the adiabatic temperature T_{ad} as a consequence of the evolution of heat due to the chemical reaction. T_{ad} can be calculated from the enthalpy of the reaction ΔH_{T0} using the equation

$$\Delta H_{T0} = \int_{T_0}^{T_{ad}} C_p \mathrm{d}T_{ad} \tag{6}$$

where T_0 is the initial temperature (temperature at which the reaction is initiated) and C_p is the combined heat capacity of the products. The calculation of T_{ad} in the preparation of composites is described in detail in the literature [71, 72].

Combustion synthesis reaction has been classified as, solid-solid, solid-liquid, liquid-liquid and solid-gas reactions, depending upon the physical state. Among these reactions, of particular relevance to this review, is the solid-liquid reaction which is common in several carbides and borides. The main problem encountered in combustion synthesis process, as applied to carbides, is the process control. Some times the temperature in the process can be too high resulting in explosion or can be too low to initiate the reaction. High reaction heat resulting in very high T_{ad} can be reduced by the addition of an inert dilute metal normally a constituent of the product itself. If the T_{ad} is too low, supplementary heating is required [73]. When the calculated T_{ad} is higher than the melting point of any constituent of the product, Equation 6 should be redefined to take this factor into consideration. Thus

$$\Delta H_{T0} = \int_{T_0}^{T_m} C_p dT + f\Delta H_m \tag{7}$$

Where ΔH_m is the enthalpy of the fusion of the product having a melting point T_m and f is the molten fraction.

Many investigators [74–77] have conducted experiments involving thermal explosion of titanium and carbon mixture. Carbon and titanium is heated in an inert atmosphere. At higher temperatures (> 1450 °C) the reaction produced by solid state diffusion of carbon to the Ti/TiC interface is given by

$$Ti_{(s)} + C \longrightarrow TiC_{(s)} \quad \text{and} \quad \Delta H_{298} = -184\,\text{kJ mole}^{-1} \tag{8}$$

Where ΔH_{298} is the standard heat of reaction at 298 K. The standard state enthalpy change can be calculated, using the following equation.

$$\Delta H_T^\circ = \Delta H_{298} + \int_{298}^{T} \Delta C_p dT \tag{9}$$

Where $\Delta C_p = C_p(\text{TiC}) \longrightarrow [C_p(\text{Ti}) + C_p(\text{C})]$.

Sato and Munir [78] have investigated the construction of a SHS diagram (Fig. 7) which indicates that the stability of SHS depends on the dilution and starting temperature of the reaction.

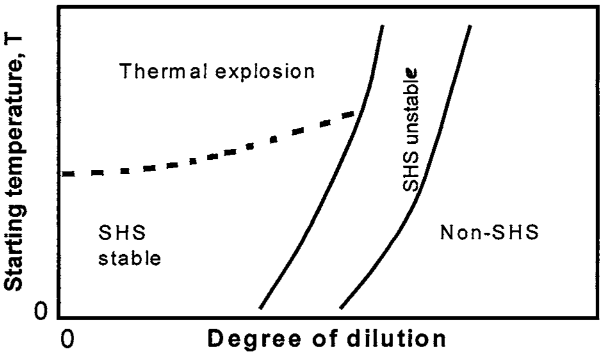

Figure 7 Schematic representation of an SHS diagram [78].

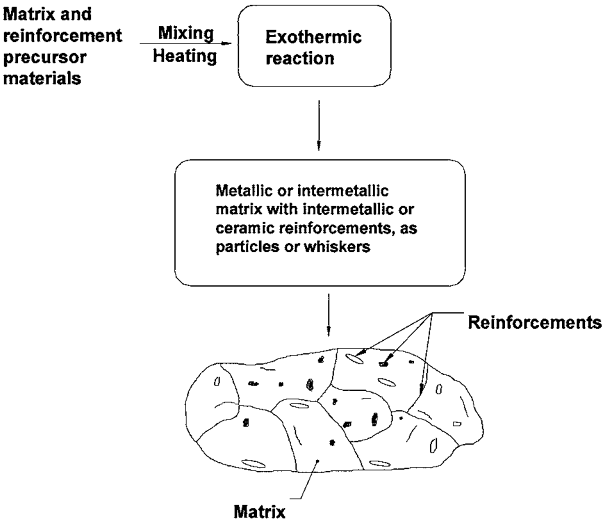

Figure 8 Schematic diagram of XD^{TM} process for production of metal and intermetallic–matrix composites [79].

Saidi *et al.* [64] have investigated the effect of many process variables on the Fe-Ti-C reaction. The authors note that, if pure titanium plus carbon is reacted, there is no ignition below 1570 °C, but that as little as 2.7 wt% iron addition can reduce the ignition temperature to 1250 °C. The heat released by the exothermic reaction between titanium and carbon rises the temperature of the sample to such a level that the reaction becomes self sustaining and complete with in a few seconds.

2.3.2. The XD^{TM} process

Martin-Marietta Laboratories [79–83] developed *in-situ* composite materials belonging to the XD^{TM} family particulate and whisker-reinforced metals and intermetallics. Original XD^{TM} concept was very good in the sense of providing thermally stable dispersion of reinforcements in materials, but of late very little has come of this. In the XD^{TM} process, matrix alloy and reacting constituents are mixed in the solid state and ignited to generate a self-propagating reaction throughout the mixture, as shown in Fig. 8.

It was found, however, that most of the reinforcements degraded the oxidation resistance of the matrix significantly, with the exception of SiC particles, which actually improved the oxidation resistance of the matrix alone.

The XD^{TM} process is the first step in making materials for conventional metal working processing such as casting, forging, extrusion, rolling etc. During the *in situ* development of reinforcements, the process eliminates oxide formation that could weaken the interface between the reinforcement and the matrix. Composite containing TiC particle introduced via XD^{TM} processing offers substantial grain refinement and improved microstructural uniformity.

The production of a variety of metal matrix composites by the XD^{TM} process has been reported including the matrices of Ti, Ti_3Al, and Fe [79]. The reinforcing phases include borides, carbides, and nitrides.

Even though the XD^{TM} process is inexpensive, there is some possible limitation with regard to chemical and thermal stability of a dispersoid phase. Large particles will grow at the expense of smaller particles leading to an increase in the average interparticle spacing which will then reduce the yield strength of the composites. Another problem may arise when the reaction between composites forming elements is less exothermic, in which case efforts to produce composites might be unsuccessful.

2.3.3. Reactive gas injection (RGI) process

The RGI process [84–86] involves the injection of carbon or nitrogen, which reacts with the alloying elements and forms carbide or nitride (Fig. 9). The synthesis route has the advantage of thermodynamic stability, fine reinforcement in the matrix and clean interface leading to improved ambient and elevated temperature strength and fracture toughness.

Base metal, with titanium, is melted in a crucible at an appropriate processing temperature and the carbonaceous gas is introduced into the melt via a gas injector

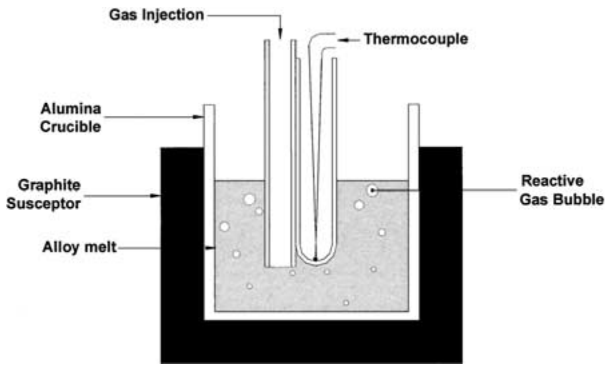

Figure 9 RGI process [84].

system. The reaction is conducted at a constant temperature for an appropriate length of time to ensure complete conversion of Ti to TiC and the melt is allowed to solidify.

The processing time and temperature depends upon the gas particle pressure and alloy composition. Processing times and temperatures ranged from 6 min to 8 hours and 1450 °C to 1600 °C respectively, depending upon the matrix composition as well as melt quantity.

A series of Ni-based carbide reinforced composites were produced by the RGI synthesis process. The primary alloy considered was Ni-Ti while feasibility studies were also conducted on Fe-Ti alloys. Summary of processing method, structure details and properties of TiC reinforced steel matrix composites is given in Table II.

3. TiC–Matrix interface

As stated earlier, small TiC particles distributed in a steel matrix is the source of strengthening of the composites. Adequate and coherent bonding between matrix and particles is essential to permit loading the carbides to the maximum strength.

Metcalfe [88] has classified interface bond based on the type of chemical reaction between particle and matrix, as

a. *Mechanical bond:* Particle and matrix mutually non-reactive, insoluble and bond develops in the absence of chemical reaction.

b. *Dissolution and wetting bond:* Particle and matrix mutually non-reactive but soluble and the bond develops on the wetting and characteristic of dissolution of a particle in the matrix. For good wetting of particle, the contact angle should be less than 90 degrees.

c. *Reaction bond:* Particle and matrix react to form new compound or compounds at the interface.

For better understanding of the bonding behaviour, strength data are needed as a function of fabrication parameter, metal properties, chemical and mechanical compatibility and characteristic of the interface region such as flaws and unbonded areas.

Fig. 10 shows the scanning electron micrograph of the commercial FERROTIC Fe-TiC composite developed by powder metallurgy method. It shows uniform distribution of TiC in iron matrix with regularly shaped carbides. Interface may develop with absence of dissolution and chemical reaction of particle and matrix [87].

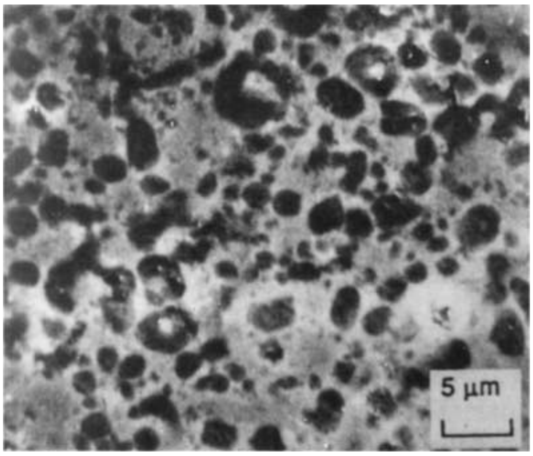

Figure 10 Scanning electron micrograph of the commercial FERROTIC Fe-TiC composite developed by powder metallurgy method [87].

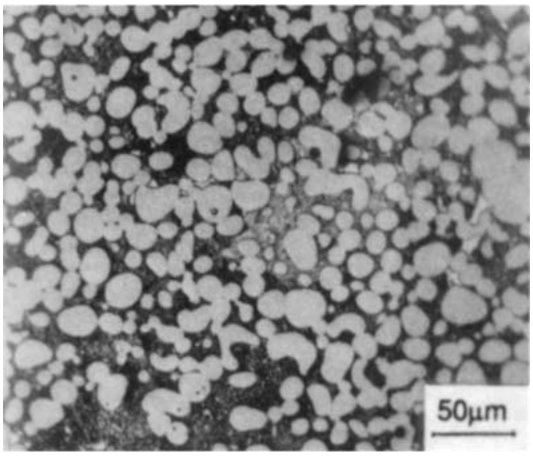

Figure 11 Optical micrograph of composite developed by dispersing carbides in the steel matrix by casting route [9].

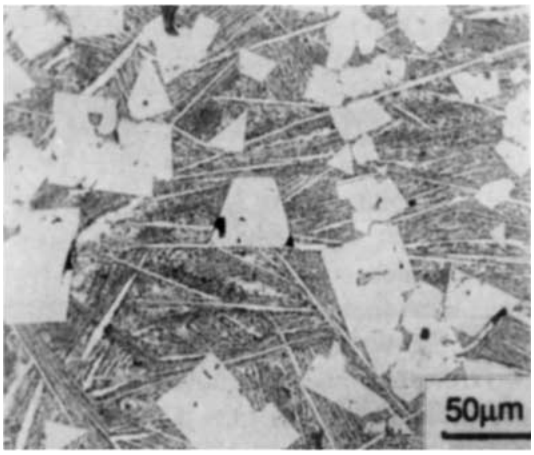

Figure 12 Optical micrograph of Fe-TiC composites developed by the reaction between carbon and titanium in molten steels [9].

The second case is illustrated in Fig. 11 for a Fe-TiC composite developed by dispersing carbides in the steel matrix by casting route where the particles partially dissolved in the matrix but coalesced extensively [9].

Fig. 12 shows Fe-TiC composites developed by the reaction between carbon and titanium in molten steels [9]. The interface may develop with reaction of matrix constituents and carbide in steel matrix. Galgali *et al.* [89] have conducted a TiC-Fe interface analysis

TABLE II Summary of processing methods, structure details and properties of TiC reinforced steel matrix composites

SI. No	Processing method	Composite particulars	Structural details	Properties determined	Objectives	Ref.
01	Precipitation of TiC from an Fe-Ti-C melt	Small specimens (10-15g) were processed by melting cast iron containing 3.8 wt % C in a high frequency induction furnace under argon and adding the appropriate amounts of Ti and Fe yielding a vol. fraction of 0.15-0.30	Structure consists of dispersion of TiC particles in a steel matrix, which is martensitic for water quenched specimen and pearlite for argon cooled specimen.	Wear	To develop wear resistant materials	9
02	Mixing of TiC powder in an Fe-C melt	Small specimen of composite (2-5 g) were processed by melting iron, low carbon (0.4 wt%) steel in a high frequency furnace under argon and mixing TiC powder at nominal volume fraction of 0.2-0.7 through electromagnetically induced sturring.	Structure consists of dispersed TiC particles in a steel matrix which is pearlitic.	Wear	To develop wear resistant materials	9
03	Dissolution of TiC rich Fe-TiC master alloy in low carbon steel melt	TiC reinforced steel composite were processed by dissolving iron –TiC master alloy melt	Clearly reveals uniform dispersion of carbide particles in a steel matrix	UTS, Elongation, Impact, strength	Preparation and characterization of high strength steel based composite.	12
04	TiC precipitation in liquid iron	Small samples of 25 g of an Fe-Ti alloy was melted in a high frequency induction furnace under argon atmosphere	Structure consists of TiC in a steel matrix	—	To study mechanism and precipitation kinetics during reaction Fe- Ti-C.	21
05	Obtain TiC in iron matrix by direct reaction in liquid iron alloys	4 g of graphite added to 33 g of Fe-24 wt% Ti alloy and mixture was heated to 1600 °C in a graphite crucible by induction furnace for 15 min.	Microstructure consists of well-rounded TiC particles in a ferritic matrix.	—	To get initial assessment of the feasibility of obtaining good dispersion of TiC in iron matrix	16
06	Dispersion and reaction of TiC in liquid iron alloys	20 g of mixture (metal + TiC pellets) heated by induction melting with the iron alloys in an alumina crucible under argon atmosphere with varying carbon.	Microstructure consists of TiC. Dispersion improves with increasing carbon content in the matrix.	Hardness	To examine the effect of process parameters on the dispersion.	55
07	Dispersion and reaction of TiC in liquid iron	Carbon black and titanium powder were mixed with iron powder composition ranging from 2.7% to 85% and pre compacted samples are heated in graphite crucible using induction furnace.	TiC produced by SHS reaction	—	Characteristics of the combustion synthesis of TiC and Fe-TiC composites.	64
08	Powder metallurgy method (FERRO-TiC)	About 0.45 vol. fraction TiC distributes in tool-steel matrix composite produced by powder metallurgy method	Micrographs show distribution of TiC particles in the tool steel matrix.	UTS, Strength, Fracture behavior	—	87

with energy dispersive analysis of a few composite samples.

The above micrographs illustrate the three bonding characterization mentioned earlier and superior mechanical and wear properties were achieved in all these cases.

4. Mechanical properties

Particulate reinforced MMC's exhibit attractive mechanical properties, such as high tensile strength, creep, and fatigue resistance [39]. Variety of titanium carbide reinforced steel composites are currently being investigated because of the possibility of utilizing cost-effective processing techniques to produce materials which have good combination of strength, ductility and fracture toughness. Titanium carbide is a covalently bonded material which can disperse uniformly in a steel matrix on account of its easy availability, high hardness, high melting point and good thermodynamic stability [90, 91].

4.1. Elastic modulus

To find the elastic properties of composites from the individual components a number of mathematical models

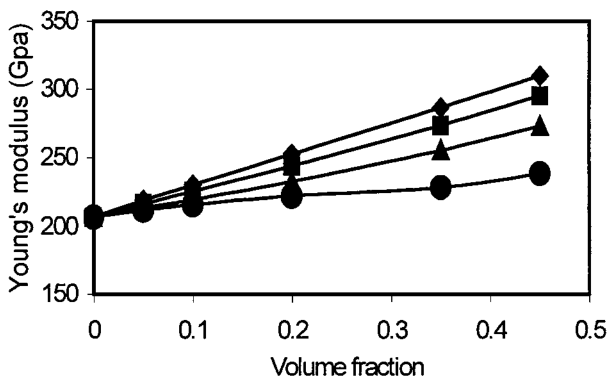

Figure 13 Shows comparison of predicted and experimentally determined Young's module of Fe-TiC composites. ◆ Iso strain rule, ■ Halpin - Tsai model [95], ▲ Iso stress model, ● Experimental data [87].

have been developed. Comparison with theoretical expectations is difficult due to uncertainty in the exact value for the modulus of the reinforcements [65]. Seetaramu *et al.* [92] have studied the effect of particle size and spacing upon the elastic constants of ferrous materials through a finite element analysis of two-dimensional idealized configuration of inclusion and matrix.

The titanium carbide reinforced steel leads to a substantial improvement in the elastic modulus [93, 94]. The elastic modulus can be measured from the elastic portion of the stress-strain curve or else from the pulse echo overlap method [71] which gives a more accurate measure of modulus.

Variation in young's modulus with volume fraction of titanium carbide reinforcements in steel produced by powder metallurgy method is shown in Fig. 13 [87]. For comparison, the iso-stress model, the iso-strain and the Halpin-Tsai models [95] were used to evaluate the elastic modulus and the results were compared with the experimental data of Srivatsan *et al* in Fig. 13 [87]. The evaluation was done using the following relations.

Iso Stress model $E_c = E_p E_m/(V_p E_m + E_p V_m)$ (10)

Iso strain model $E_c = V_p E_p + E_p V_m$ (11)

where E_c, E_m, E_p are elastic modulus of the composite, matrix and particles respectively and V_m, V_p are the volume fraction of matrix and particle respectively and Halpin-Tsai equation

$$E_c = E\mathrm{m}(1 + 2sqV_p)/(1 - qV_p) \quad (12)$$

where $q = (E_p/E_m - 1)/(E_p/E_m + 2s)$ and $s =$ aspect ratio of particles. When the aspect ratio $s = 1$, the calculated Young's modulus values are higher than the experimental data. Generally, Halpin-Tasi equation can be used to predict the composite young's modulus. However, when the reinforcing phase possesess a large aspect ratio, a negligible increase in strength is predicted.

Srivatsan *et al.* [87] have reported an increase in stiffness of a titanium carbide reinforced steel matrix with 45% TiC from room temperature to 100 °C. Higher modulus at higher temperature is beneficial in pushing up the useful temperature limit of these composites.

4.2. Hardness

Ferro-Titanit is the trade name of Edelstahl Witten Krefeld GmbH. Germany and produced through powder metallurgy method. Ferro-Titanit has a carbide content of about 0.45 volume fraction and, in heat treated condition, it has up to 69 HRC [96]. Terry and Chinyamakobvu [16] have reported that hardness increases with increasing volume percentage of TiC. The hardness increased from 380 HV10 to 483 HV10 with the volume fraction of TiC varying from 3 to 11.8 percent.

Dogan *et al.* [97] have conducted hardness test on 0.4 volume fraction of TiC reinforced steel with varying carbon percentage. They found that hardness varies from 54 HRC to 71 HRC with carbon content varying from 0.34 to 2.66 percentage while keeping TiC content constant. It is clear from their study that the hardness is not only a function of volume fraction of TiC particles but also that of matrix type and hardenability.

4.3. Strength

There are numerous factors influencing the yield and tensile strengths of particulate reinforced MMC's. These factors are complex and interrelated. For example, one of the important factors influencing the mechanical behaviour of MMC's is the alloy matrix. Increasing the volume fraction of reinforcements generally increases the strength of the composites. Mechanical properties of composites are also affected by the residual stress which forms as a result of the difference in the thermal expansion coefficient between the matrix and reinforcement.

Srivatsan *et al.* [87] have investigated the mechanical properties of different tool-steel/TiC composites with 0, 0.35 and 0.45 volume fractions. They found that the room temperature yield strength of 45% volume fraction composite was higher than that of 35% volume fraction steel. Also, the elevated temperature (100 °C) yield strength of 45% composite decreased by 20% where as that of 35% TiC composites decreased only by 10%. Fracture stress follows the same trend as the ultimate tensile strength. They further observed the cross-section of the fracture surface of a tensile tested sample and noticed that fracture of these composites occurs due to the crack nucleation, growth and influences of local plastic strain and the hydrostatic stress generated during tensile loading.

Galgali *et al.* [89] have produced composite with varying contents of TiC by reaction method. They reported a significant increase in the ultimate tensile strength of composites. It increased from 363 MPa to 775 MPa with increasing TiC in the matrix. Wood *et al.* [98] have reported tensile strength increases from 881 Mpa in 5% carbide steel to 1488 MPa in 10% carbide steel.

There are two models used to predict strengthening behavior of particulate MMC's [65] (i) load transfer model; and (ii) matrix strengthening model.

Load transfer models were developed on the basis that the hard and nondeformable reinforcements in particulate reinforced composites may carry more loads than the relatively soft matrix. This model can again be

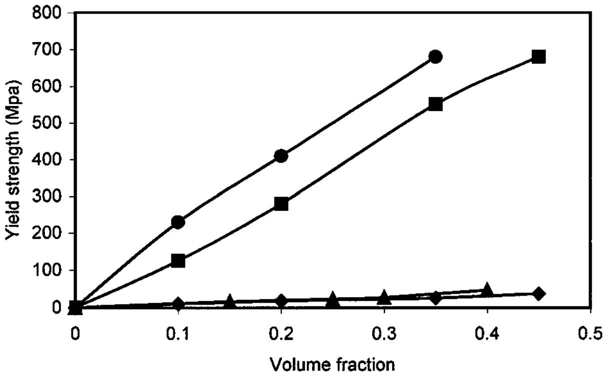

Figure 14 Shows strength comparison with model calculation and experimental values. (■) Experimental value [87], (●) Nardone and Prewo [101], (♦) Ashby [102], and (▲) Arsenault [103].

grouped as, Eshelby model [99] and the shear lag model [100]. Nardone and Prewo [101] have developed a modified shear lag model. According to this model yield strength of the composite is given by

$$\sigma_{\text{cy}} = \sigma_{\text{my}}[V_{\text{p}}(s+4)/4 + (1 - V_{\text{p}})] \qquad (13)$$

Where $\sigma_{\text{cy}} =$ Yield strength of the composite, $\sigma_{\text{my}} =$ Matrix yield strength, $V_{\text{p}} =$ Volume fraction of particles, $s =$ Aspect ratio of particles

Ashby [102] and Arsenault [103] also developed a model for calculating yield strength of the composites. Validity of the above strengthening models have been verified for Fe-TiC composites using the experimental data of Srivatsan *et al.* [87] in Fig 14. Comparison shows that the data which is calculated by using modified shear lag model grossly overestimates the experimental values while that calculated by Ashby and Arsenault underestimates the yield strength value σ_{cy}. It is also observed that if the tensile transfer of load is ignored, then the modified shear lag model [101] is in complete agreement with the experimental values.

In the matrix strengthening model, the strength changes with varying percentage of reinforcements in the matrix. The strengthening effects are changes with microstructural changes expected in the matrix due to the presence of the reinforcements.

Several models have also been proposed for explaining the increase in yield strength of discontinuous reinforced composites.

The continuum mechanism model of composite behaviour [100] assumes that load is transferred from the matrix to a nonplastically deforming reinforcement. Load transfer occurs between a reinforcement and the matrix by means of shear stress at the particle-matrix interface.

In many composites, dislocations are generated in the alloy matrix upon cooling or quenching from the casting process, due to mismatch of the coefficient of thermal expansion between the matrix and reinforcement. The higher the dislocation density higher the strength of the alloy. The amount of dislocation generation is affected by coefficient of thermal expansion, particle size, particle volume fraction and matrix strength [104]. The Orowan bypassing of particles by dislocations may also lead to strengthening [100, 101, 104–106]. The yield stress of single -phase polycrystalline materials is observed to be proportional to $d^{-1/2}$, where d is the particle diameter. This leads to the Hall-petch type relation:

$$\sigma_{\text{y}} = \sigma_0 + k\,d^{-1/2} \qquad (14)$$

where $\sigma_0 =$ frictional stress resisting the motion of gliding dislocation or back stress, $k =$ is the Hall Petch slope which is considered to be a measure of the resistance of the grain boundary to slip transfer [103, 104].

In forest hardening model [107], dislocations could be generated from the interface of a misfitting particle and a ductile matrix. Kamat *et al.* [107], have examined the tensile properties of MMC's and suggested that dislocations link up to form dislocation cells resulting in increased strengthening due to a high dislocation density within matrix with a diameter proportional to the interparticle spacing and strength.

The overall response of the particulate and whisker reinforced composites, and the evolution of stress-strain in the matrix have also been computed using finite element models [108–110].

5. Wear properties

Kattamis and Suganuma [9] have developed a TiC based MMC which exhibits both excellent wear performance while achieving significant cost reduction over existing materials. They have adapted both dispersive and reaction methods for producing the MMC's and have investigated their wear properties. Galgali *et al.* [89, 111] have reported that adhesive wear of titanium carbide reinforced cast iron composites depends both on the amount of titanium carbide and graphite present. The particle size distribution also plays an important role. They have also observed that as the volume of reinforced particles increases there is a decrease in the wear rate.

The finer titanium carbide particle distribution in composites may be responsible for uniform wear rate at all applied loads and sliding distance. It is clear that the contribution from carbide size will be most significant in improving the wear properties. Ashby [112] has noted that composite steels offer very attractive properties if the reinforcement is of a lower density than steel. The reason is that many components are designed using $E^{1/2}/\rho$ or $E^{1/3}/\rho$ criteria (where E is Young's modulus and ρ is the density). While increasing the modulus and wear properties is beneficial, the low density of TiC (4.93 g cm^{-3}) added at relatively low cost to a steel would provide very positive design benefits. Summary of wear tests conducted on Fe-TiC composites produced by various methods is given in Table III.

6. Applications

Many applications have been identified for the family of tool–steel based composites [36–38]. These includes palletized knives, mould inserts and mould gates in plastic industry; guide, roller, worm heading punches and extrusion dies in the hot-working industry; rotors and pressure plates to make fuel pumps; landing gear components in the aerospace industry and dies for stamping, forging and swaging in metal forming. Also particulate –reinforced TiC-ferrous matrix composites have potential application in waste – derived fuel plants

TABLE III Summary of wear tests conducted on Fe-TiC composites produced by various methods

Sl No.	Material	Processing method	Wear test method	TiC volume fraction	Results	Remarks	Ref.
01	Fe-TiC composite	Precipitation of TiC from Fe-Ti-C melt	Pin on disk	0.22-0.30	Sp. Wear rate decreases with increasing carbide particle size spacing and increasing steel matrix microhardness.	Tested both in as cast and heat-treated conditions.	9
02	Fe-TiC composite	Mixing of TiC powder in an Fe-C melt	Pin on disk	0.22-0.30	Sp. Wear rate decreases with increasing carbide particle size spacing and increasing steel matrix microhardness	Tested both in as cast and heat treated condition	9
03	Fe-TiC composite (FERRO-TITANIT)	Powder metallurgy method	—-	0.35-0.45	High wear resistant	Tested in heat treated condition	98
04	Fe-TiC composite	Powder metallurgy method	Abrasive wear test	0.35-0.45	Wear loss of the TiC particle reinforced composite decreases with increasing hardness	Tested in heat treated condition	96
05	Fe-(W,Ti)C	Mixing of Fe-(W,Ti)C powder in molten iron	Abrasive wear with different abrasive media	—-	Addition of (W,Ti)C increases the abrasive wear properties	Tested in as cast and in hot rolled condition	98
06	Fe-(W,Ti)C	Mixing of Fe-(W,Ti)C powder in molten iron	Pin on disk	—	Addition of (W,Ti)C considerably improves the wear properties of iron based alloys	Tested in as cast condition	113
07	Fe-TiC composite	Mixing of TiC rich, Fe-TiC master alloy in to molten iron	Pin on disk	—-	Wear resistance of TiC reinforced steel composite in the normalized condition is superior to same steel composite oil quenched and tempered condition.	Tested in as cast and in tempered condition.	114
08	Fe-TiC composite (Ferro-TiC alloy)	Powder metallurgy method	—	Up to 0.45	High wear resistant	Tested in heat treated condition	115

which are faced by such problems as excessive wear of shredding and milling components as well as pneumatic pipes and duck work. Benefits could be expanded to other industries such as mining, paving, coal handling and power station etc [9]. Use of these TiC–reinforced composites in high technology application is attractive primarily because of the many exceptionally good and wear properties [87].

7. Concluding remarks

Large body of literature reviewed in this paper presents a cross-section of views and experimental results obtained over the years by numerous investigators in the field of TiC–steel composites. This review has been primarily concerned with processing and mechanical property evaluation of these composites.

Regarding the processing of TiC–steel composites, a variety of techniques have been developed over the last decade. Available literature shows that these processing techniques can be classified as powder metallurgy, powder mixing with liquid metal and exothermic reaction synthesis. Advantages and disadvantages of these roots have been examined and commented upon. The review has shown that the state of the interface between particles and matrix contributes to the enhancement of the elastic modulus, yield strength and wear resistance. Unfortunately, these parameters cannot be satisfactorily modelled.

Fundamentally, the metal-matrix interface coherence and the particle distribution decides the strengthening mechanism. In this context, high-resolution transmission electron microscopy and lattice-imaging techniques seem to be very useful. It is essential to perform such experimental investigations.

The successful application of TiC – steel composites depends on overcoming various technological difficulties. There is hardly any investigation on the correlation between processing parameters and product quality. It is necessary to carry out such investigations to realize the full potential of TiC based ferrous composites.

Acknowledgements

The authors wish to thank Dr. M. N. Channabasappa, Principal and Prof. H. N. Rangaraj, Head of the Department of Mechanical Engineering, Siddaganga Institute of Technology, Tumkur for their constant support and encouragement. They also thank Mr. S. Sampathkumar, Materials Technology Division, Central Power Research Institute, Bangalore for his useful and valuable suggestions.

References

1. S. G. FISHMAN, *J. Met.* **38** (1986)26.
2. Y. FLOM and R. J. ARSENAULT, *ibid.* **38** (1986)31.
3. *Idem., Mater. Sci. Eng.* **77** (1986) 191.
4. A. H. M. HOWES, *J. Met.* **38** (1986)28.
5. V. C. NARDONE and K. W. PREWO, *Scripta. Metall.* **20** (1986) 43.
6. A. MORTENSEN, J. A. CORNIE and M. C. FLEMINGS, *ibid.* **40** (1988) 12.
7. A. MORTENSEN, M. N. GUNGOR, J. A. CORNIE and M. C. FLEMINGS, *ibid.* **38** (1986) 30.
8. B. S. TERRY and O. S. CHINYAMAKOBVU, *Mater. Sci. Technol.* **07** (1991) 842.
9. T. Z. KATTAMIS and T. SUGANUMA, *Mat. Sci and Engg.* **A128** (1990) 241.
10. P. A. BLACKMORE, A. SEGAL, A. J. BAKER and P. R. BEELY, in "Solidification and Casting of Metals" (The Metals Society, London, 1977) p. 533.
11. P. R. BEELEY, A. J. BAKER, P. A. BLACKMORE and A. SEGAL, in "Solidification Technology in Foundry and Casthouse" (The Metals Society, London, 1983) p. 398.
12. S. R. TITTAGALA, P. R. BEELEY and A. N. BRANLEY, *Met. Technol.* **10** (1983) 257.
13. M. EPNER and E. GREGORY, *Trans, TMS-AIME* **218** (1960) 117.
14. J. M. PANCHAL, T. VELA and T. ROBISCH, in "Fabrication of Particulate Reinforced Metal Composites," edited by J. Masounave and F. G. HAMEL (ASM International, Materials Park, OH, 1991) p. 245.
15. M. K. MAL and S. E. TARKAN, in "Progress in Powder Metallurgy" (Metal Powder Industries Federation, New York, 1972) Vol. 28.
16. B. S. TERRY and O. S. CHINYAMAKOBVU, *Mater. Sci. Letter.* **10** (1991) 628.
17. B. V. CHAMBERS, T. Z. KATTANIS, J. A. CORNIE and M. C. FLEMINGS, in "Solidification Processing" (The Institute of Metals, London, 1987) p. 453.
18. A. LAWLEY and D. APELIAN, in "Spray Forming," edited by J. V. Wood (Abingdon, Woodhead Publishing, 1994) p. 267.
19. J. LATUSZKIEWICZ, P. G. ZIELINSKI and A. ZALUSKA, *Mater. Sci. Eng.* **97** (1988) 181.
20. F. DELANNAY, L. FROYER and A. DERUYTTERE, *J. Mater. Sci.* **22** (1987)1.
21. ZHENSHENG LIU and H. FREDRIKSSON, *Metall. and Mater. Tran. A* **28A** (1997) 471.
22. A. MORTENSEN, M. J. C. FLEMINGS and J. A. CORNIE, *Met. Trans* **19A** (1988) 709.
23. R. L. MAHAR, R. JAKASAND and C. A BRUCH, Tech. Rep. AFML-TR-68, May 1968.
24. H. JONES, "Rapid Solidification of Metals and Alloys," Monograph Series No. 8 (The Institute of Metallurgists, 1982).
25. T. DONOMOTO, N. MIURAS, K. FUNATANI and N. MIYAKE, SAE Technical Paper no. 83052 (Detroit, MI, 1983).
26. T. W. CLYNE, M. G. BADER, G. R. CAPPLEMAN and P. A. HUBERT, *J. Mater. Sci* **20** (1985) 85.
27. Duralcan, "Metal Matrix Composites May-June Data Report Package" (Dural Aluminum Composites Corporation, San Diego, CA 92121).
28. G. ELKABIR, L. K. RABENBERG, C. PERSAD and H. L. MARCUS, *Scripta. Metall.* **20** (1986) 1411.
29. "Rapidly Solidified (RS) Aluminum Alloys—Status and Prospects NMAB-368" (National Academy Press, Washington, DC, 1981).
30. B. FERGUSON, A KUHN, O. D. SMITH and F. HOFSTATIER, *Int. J. Powd. Metall. Powd. Tech* **20** (1984) 131.
31. T. W. CLYNE and J. F. MASON, *Met. Trans.* **18A** (1987) 1519.
32. J. A. CORNIE, A. MORTENSEN and M. C. FLEMINGS, in Proceedings of the 6th International Conference on Composite Materials, ICCM and ECCM, edited by F. L. Matthews, N. C. R. Buskell, J. M. Hodgkinson and J. Morton (Elsevier Applied Science, London, 1987) p. 2.297.
33. S. KRISHANAMURTHY, Y. W. KIM, G. DAS and F. H. FROES, in "Metal Ceramic Matrix Composites; Processing, Modeling and Mechanical Behavior," edited by R. B. Bhagat, A. H. Claver, P. Kumar and A. M. Ritter (The Minerals, Metals/Materials Society, Warrendale, PA, 1990) p. 145.
34. M. S. NEWKIRK, A. W. URGUHART, H. R. ZWICKER and E. BREVAL, *J. Mater. Res* **1** (1986) 81.
35. J. WEINSTEIN, in Proceeding of the International Symposium on Advances in Processing and Characterization of Ceramic Metal Matrix Composites, CIM/ICM, Vol. 17, edited by H. Mostaghaci (Pergamon, Oxford, 1989) p. 132.
36. J. M. PANCHAL, T. VELA and T. ROBISCH, in "Fabrication of Particulate—Reinforced Metal Composites," edited by J. Masounave and F. G. Hamel (ASM International, Metals Park, OH, 1991) p. 245.
37. M. K. MAL and S. E. TARKAN, in "Progress in Powder Metallurgy," Vol. 28 (Metal Powder Industries Federation, New York, 1972).
38. P. MANDALIS, S. E. TARKAN and M. K. MAL in "Progress in Powder Metallurgy," Vol. 31 (Metal Powder Industries Federation, New York, 1975).
39. I. A. IBRAHIM, F. A. MOHAMED and E. J. LAVERNIA, *J. Mater. Sci.* **26** (1991) 1137.
40. S. ABKOWITZ and P. WEIHRAUCH, *Adv. Mat. Proc.* **136** (1989) 31.
41. S. ABKOWITZ, *Met. Pow. Rep.* **48** (1993) 34.
42. O. RIDGER, J. STICKFORTH and G. LOHRKE, *Tech, Mitt, Krupp* **19** (1961) 154.
43. K. H. MISKA, *Mater. Engg.* **11** (1975) 67.
44. J. L. ELLIS, *Tool Engg.* April (1957) 103.
45. T. J. ROBISCH, M. K. MAL and S. E. TARKAN, *Ind. Heat.* **5** (1982) 18.
46. S. E. TARKAN and M. K. MAL, *Met. Progr.* **28** (1974) 99.
47. M. RUHLE and A. G. EVANS, *Mater. Res. Symp. Proc.* **120** (1988) 293.
48. F. DELANNAY, L. FROYEN and A. DERUYTTERE, *J. Mater. Sci.* **22** (1987) 1.
49. R. E. JOHNSON JR., *J. Phys. Chem.* **63** (1959) 1655.
50. S. Y. OH, J. A. CORNIE and K. C. RUSSEL, *Ceramic Engg. Sci. Proc.* **8** (1987) 912.
51. A. MORTENSEN, J. A. CORNIE and M. C. FLEMINGS, *ibid.* **40** (1988) 12.
52. J. V. NAIDICH, *Prog. Surf. Membr. Sci.* **14** (1981) 354.
53. S. S. KIPARISOV, *Sov. Powder Metall., Met. Ceram.* **15** (1976) 618.
54. *Idem., ibid.* **15** (1976) 460.
55. B. S. TERRY and O. S. CHINYAMAKOBVU, *Mater. Sci. And Tech.* **8** (1992) 399.
56. P. H. BOOKER, Ph.D. dissertation, Graduate Center, Oregon, 1979.
57. B. V. CHAMBERS, T. Z. KATTAMIS, J. A. CORNIE and M. C. FLEMINGS, in "Solidification Processing" (The Institute of Metals, London, 1987) p. 435.

58. Y. MURAKAMI, *Trans. Natl. Res. Inst. Met. (Jpn)* **1** (1959) 7.
59. S. JONSSON, *Metall. Mater. Trans. B* **29B** (1998) 361.
60. A. G. MERZHANOV and I. P. BOROVINSKAYA, *Doklady Nauk* **204** (1972) 366.
61. W. J. FRANKHOUSER, K. W. BRENDLEY, M. C. KIESZEK and S. T. SULLIVAN "Gasless Combustion Synthesis of Refractory Compounds" (Noyes, NJ, 1985).
62. Z. A. MUNIR and U. ANSELMI-TAMBURINI, *Mater. Sci. Rep.* **3** (1989) 277.
63. S. RAY, *J. Mater. Sci.* **28** (1993) 5397.
64. A. SAIDI, A. CHRYSANTHOU, J. V. WOOD and J. L. F. KELLIE., *ibid.* **29** (1994) 4993.
65. S. RANGANATH, *ibid.* **32** (1997) 1.
66. Z. A. MUNIR, "Reviews in Particulate Materials," edited by A. Bose, R. M. German and Lawley, Vol 1 (1993) p. 41.
67. A. G. MERZHANOV, "Combustion and Plasma Synthesis of High Temperature Materials," edited by Z. A. Munir and J. B. Holt (VCH Publishers, NY, 1990) p. 1.
68. L. L. WANG, Z. A. MUNIR and Y. M. MAXIMOV, *J. Mater. Sci.* **28** (1993) 3693.
69. H. C. YI and J. J. MOORE, *J. Mater. Sci.* **25** (1990) 1159.
70. J. SUBRAHAMANYAM and M. VIJAYAKUMAR, *ibid.* **27** (1992) 6249.
71. S. RANGANATH, Ph.D. dissertation, Banaras Hindu University, India, 1993.
72. J. SUBRAHMANYAM, M. VIJAYAKUMAR and S. RANGANATH, *Met. Mat. Proc.* **1** (1989) 105.
73. S. RANGANATH, M. VIJAYAKUMAR and J. SUBRHMANYAM, *Mater. Sci. Engng.* **A149** (1992) 253.
74. V. M. SHKIRO, V. N. DOROSHIN and I. P. BOROVINSKAYA, *Combust. Explos. Shock Waves* **16** (1980) 370.
75. D. C. HALVERSON, K. H. EWALD and Z. A MUNIR, *J. Mater. Sci.* **28** (1993) 4583.
76. M. E. MULLINS and E. RILEY, *J. Mater. Res.* **4** (1986) 408.
77. A. H. ADVANI, N. THADHANI, H. A. GREBE, R. HEAP, C. COFFIN and T. KOTTKE, *Scripta. Mettall. Mater.* **25** (1991) p. 1447.
78. N. SATO and Z. A. MUNIR, in "Proceeding of the Symposium on High Temperature Matrials V" Vol. 9-18, edited by W. B. Johnson and R. A. Rapp (Electro Chemical Society, Pennington, NJ, 1990) p. 99.
79. L. CHRISTODOULOU, P. A. PARRISH and C. R. CROME, *Mat. Res. Soc. Symp. Proc.* **120** (1988) 29.
80. R. K. VISWANADHAM, S. K. MANNAN and B. SPRISSLER, Martin Marietta Laboratories Rept. MML TR-87-66C, ONR, 1987.
81. K. S. KUMAR Martin Marietta Laboratories Rept., 1989.
82. L. CHRISTODOULOU, P. A. PARRISH and C. R. CROWE, *Mater. Res. Soc. Symp. Proc.* **120** (1988) 29.
83. D. LEWIS, "Metal Matrix Composites; Processing and Interface," edited by R. K. Everett and R. J. Arsenault (Academic Press, NY, 1991) p. 121.
84. M. J. KOCZAK and K. S. KUMAR, U. S. Patent 4, 808 372 (1989).
85. M. S. NEWKIRK, A. W. URQUHART, H. R. ZWICKER and E. BREVAL, *J. Mater. Res.* **1** (1986) 81.
86. V. SHTESSEL, S. SAMPATH and M. KOCZAK, "*In-Situ* Composites," edited by M. Sing and D. Lewis (The Minerals, Metals and Materials Society, 1994) p. 37.
87. T. S. SRIVATSAN, R. ANNIGERI and A. PRAKASH, *Composites Part A* **28A** (1997) 377.
88. A. G. METCALFE, in "Composite Materials," Vol. 1 Interfaces in Metal Matrix Composites (Academic Press, New York, 1974).
89. R. K. GALGALI, H. S. RAY and A. K. CHAKRABARTI, *Mater. Sci & Tech.* **14** (1998) 810.
90. M. K. MAL and S. E. TARKAN, *Heat Treating* June (1971).
91. S. TSUREKAWA, S. MATSUBARA, H. KURISHITA and H. YOSHINAGA, *Mater. Trans. JIM* **33** (1991) 821.
92. S. SEETHARAMU, M. N. SRINIVASAN and P. D. MANGALGIRI, *J. Indian Inst. Sci.* **63A** (1981) 213.
93. M. H. LORETTO and D. G. KONITZER, *Metall. Trans.* **21A** (1990) 1579.
94. M. S. THOMPSON and V. C. NARDONE, *Mater. Sci. Engng.* **144A** (1991) 121.
95. H. LAGACE and D. J. LLOYD, *Can. Metall. Quart.* **28** (1989) 145.
96. Private communication, Edelstahl Witten-Krefeld GmbH, Germany.
97. O. N. DOGAN, J. A. HAWK, J. H. TYLCZAK, R. D. WILSON and R. D. GOVIER, *Wear* **225–229** (1999) 758.
98. J. V. WOOD, K. DINSDALE, P. DAVIES and J. L. F. KELLIE, *Mat. Sci. and Tech.* **11** (1995) 1315.
99. J. D. ESHELBY, *Proc. Roy. Soc.* **A241** (1957) 376.
100. M. R. PIGGOT, "Load-Bearing Fiber Composites" (Pergamon Press, Oxford, 1980).
101. V. C. NARDONE and K. M. PREWO, *Scripta. Metall.* **20** (1986) 43.
102. M. F. ASHBY, "Strengthening Methods in Crystals," edited by A. Kelly and R. B. Nicholson (Elsevier, Amsterdam, 1971) p. 184.
103. R. J. ARSENAULT and N. SHI, *Mater. Sci. Engng.* **81** (1986) 175.
104. R. M. AIKIN, JR., JOM. **49**(8) (1997) 35.
105. F. J. HUMPHREYS, H. BASU and M. R. DJAZEB, in Proceeding of 12th Riso International Symposium on Metallurgy and Materials Science, Roskilde, 1991, edited by N. Hansen (Riso National Laboratory, Denmark, 1991) p. 51.
106. P. M. KELLY, *Int. Met. Rev.* **18** (1973) 31.
107. S. V. KAMAT, J. P HIRTH and R. MEHRABIN, *Acta Metall.* **37** (1989) 2395.
108. N. SHI and R. J. ARSENAULT, *Scripta. Metall. Mater.* **28** (1993) 623.
109. *Idem., Acta Metall. Mater.* **39** (1991) 2255.
110. V. TVERGAARD, *ibid.* **38** (1990) 185.
111. R. K. GALGALI, H. S. RAY and A. K. CHAKRABARTI, *Mater., Sci & Tech.* **14** (1998) 1189.
112. M. F. ASHBY, in 4th Institute of Materials Conf. Metal Matrix Composites (Design and innovation, Institute of Materials, London, 1993).
113. C. C. DEGNAN, J. V. WOOD, in Proceeding ICCM-10, Whistler, B. C., Canada, August 1995.
114. R. K. GALGALI, H. S. RAY and A. K. CHAKRABARTI, *Indian Found. Jour.* **44** (1998) TP1.
115. Private communication, Read Advanced Materials, U.S.A.

Received 31 March 2000
and accepted 21 March 2001