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FEASIBILITY OF $(IMC)_d$-REINFORCED METAL MATRIX COMPOSITES

submitted by

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ABSTRACT

A series of metal matrix composites were produced by co-extruding TiB_2-reinforced near-γ titanium aluminide intermetallic composites (IMC) within matrices of two commercial titanium alloy compositions. The components of the composites were produced or obtained in powder form, homogeneously blended, and deformation-processed at 1038°C in an attempt to evolve an aligned metal matrix composite microstructure via the principles of commensurate deformation. The IMC's, which were ultimately blended within the metallic matrix, are characterized by in-situ-synthesized TiB_2; the particulate is highly faceted, single crystalline, and has an average diameter of approximately 0.4 μm. The results of this feasibility study indicate that while the (IMC)-reinforced composites can be successfully produced by extrusion, the extent of deformation in the titanium aluminide intermetallic composite component is much less than that which would be predicted if it deformed equally with the titanium matrix phase. While the expected effect of IMC strength (as varied by the percentage of TiB_2 synthesized within the IMC) on the ability to deform the IMC at high temperatures was not observed, the extent to which the IMC deforms was found to be enhanced by higher flow stresses in the titanium matrix phase. Recommendations based upon the experiences of the present work include creating a more deformable IMC-reinforcing phase by extruding at a higher homologous temperature, e.g., by alternatively utilizing the trialuminide Al_3Ti as a matrix for the IMC-reinforcement in the titanium matrices.

BACKGROUND, PERSONNEL, AND ACKNOWLEDGEMENTS

This report summarizes the results of a study to examine the feasibility of producing metal matrix composites containing intermetallic matrix composite (IMC) reinforcing entities. Towards this end, the Office of Naval Research (ONR) has provided funds which have assisted in the purchase of the required materials. Additional funding for this project, including that for the principal investigator (S.L. Kampe) and two graduate students participating on this project (R.J. Martin and T. Pete), have been provided by Virginia Polytechnic Institute and State University and The United Nations Educational Program for Southern Africa (UNEPSA).

INTRODUCTION

Intermetallic matrix composites have been the subject of considerable development over the past several years. Their development has been motivated by the excellent perceived performance of these materials, especially at high temperatures. Unfortunately, despite extensive efforts directed at several promising candidate intermetallic matrix systems, successful incorporation of these materials is still limited by difficulties associated with processing and the design challenges consequent of their limited ambient temperature ductility, toughness, and damage tolerance. At temperatures corresponding to the ambitiously-high implementation goals of these materials, strength and resistance to creep of the intermetallic matrices have been less than originally expected. For the most part, these materials continue to exhibit the challenges of "ceramic-like" behavior at low temperatures, and the mechanical/structural instabilities of metals at high temperatures.

In a reversal to the conventional approach to the development of intermetallics and IMC's, this exploratory study seeks to examine a composite processing methodology which exploits and benefits from the above "dichotomous" behavior. It is suggested that IMC's offer promise as *reinforcing entities* within metallic matrices. The premise is based upon results which indicate that discontinuously-reinforced IMC's are capable of exhibiting extremely high (ceramic-like) strengths while also showing the ability to exhibit extensive deformation and process-deformability at high temperatures. The former is important in the establishment of metal matrix composite properties at temperatures considered modest by intermetallic standards. The latter can be exploited to assist processability and microstructural design, and is the primary subject of the present study.

Table I illustrates compressive properties for a series of near-γ titanium aluminide matrices reinforced with high loadings (i.e., $\approx$ 40-50 volume percent (v%)) of discontinuous titanium diboride (TiB_2) particles [1]. Also shown for comparison are properties of the two structural ceramics, Al_2O_3 and SiC, that are most frequently mentioned as candidate reinforcement to metallic matrices. As shown, the room temperature properties of the TiB_2-reinforced titanium aluminides exhibit strength and elastic modulus values of a magnitude comparable to that of either structural ceramic.

TABLE I.

Example properties of TiB_2-reinforced near-γ titanium aluminides as compared to candidate structural ceramic reinforcing materials

Material	$\sigma_{fracture,\ compression}$ (GPa)	Elastic Modulus (GPa)	$\sigma_{fracture,\ tension}$ (GPa)	Hardness (1 kg Knoop)
TiAl + 30 v% TiB_2	2100[2]	278[2]	725[2]	86[2]
TiAl + 40 v% TiB_2[1]	2344[3]	290[3]	700[2]	90[3]
TiAl + 50 v% TiB_2[1]	2620[3]	305[3]	400[2]	95[3]
TiAl + 60 v% TiB_2	2900[2]	330[2]	250[2]	100[2]
Al_2O_3 (AD-94)	2103[4]	303[4]	193[4]	10.7[4]
SiC	2500[5]	393[5]	307[5]	24.5[5]

Notes: 1 XD® Ti-45Al + TiB2 (HIP'd)
2 Measured
3 Estimated, based on extrapolation of measured data
4 Reference [2]
5 Reference [3]

Unlike the Al_2O_3 and SiC, however, the titanium aluminide composite is capable of significant plastic deformation at elevated temperatures. Numerous investigations of the structure/property relationships of monolithic near-γ titanium aluminide composition agree that the transition from brittle to ductile behavior occurs abruptly at temperatures of approximately 650°C, varying slightly (±≈25°C maximum) with specific alloy composition. The addition of TiB_2 to the titanium aluminide matrix elevates the flow stress at temperatures below where the ductile-to-brittle transition (DBTT) occurs, and, depending upon the strain rate employed, generally at temperatures above the DBTT, as well. The temperature at which the DBTT occurs remains generally unchanged with TiB_2 reinforcement.

A strategy by which the above-summarized mechanical behavior can be exploited is illustrated schematically in Fig. 1. If dispersed within a metallic matrix and subsequently extruded at elevated temperatures, it is conceivably that the two phases (i.e., the metallic matrix and the IMC) will deform commensurably to create an aligned, discontinuously-reinforced composite, that is, an $(IMC)_d$-reinforced metal matrix composite. The benefits of such a composite proces-

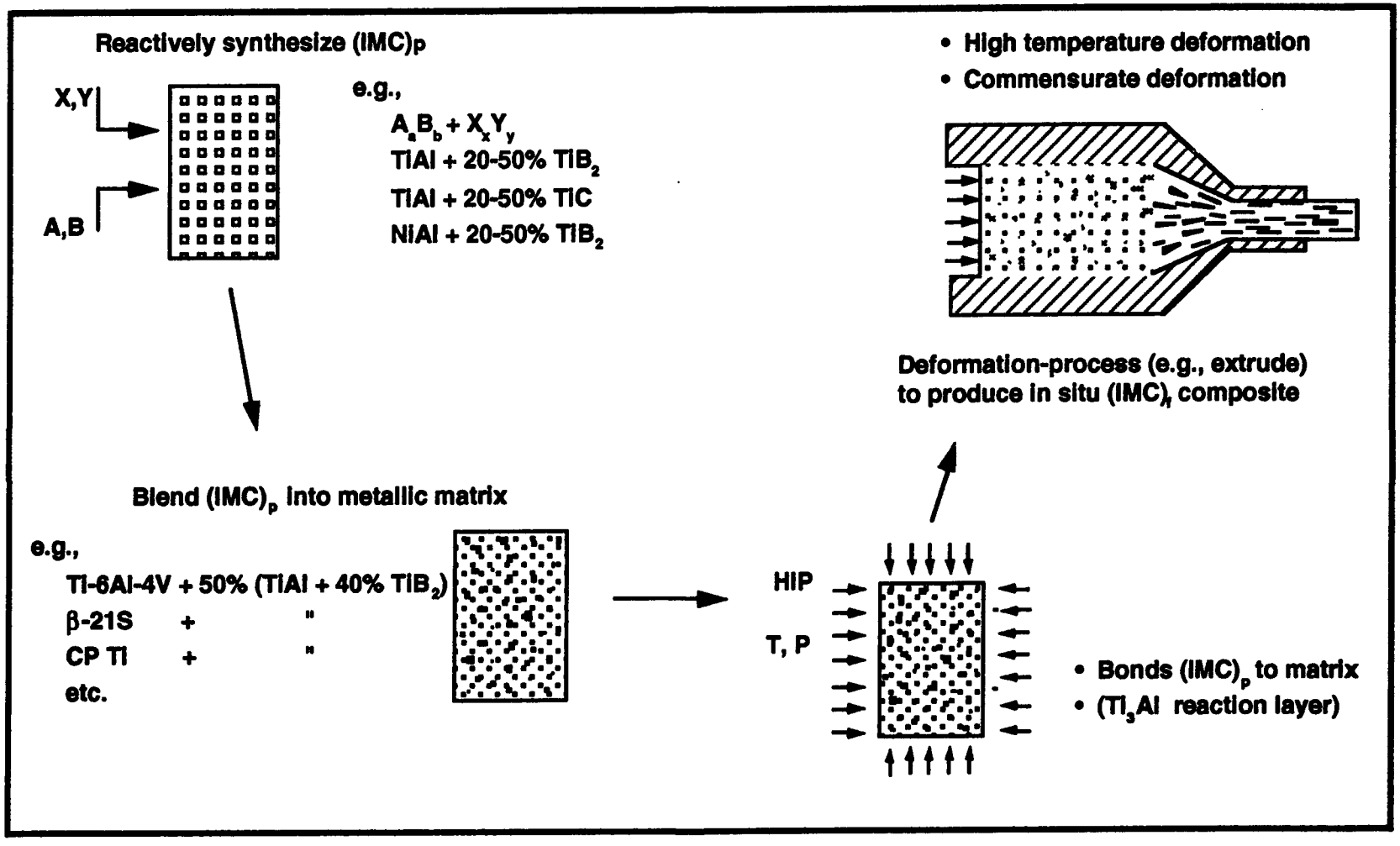


Figure 1. Overview of the processing utilized to create an IMC-reinforced metal matrix composite

sing strategy are numerous. For example, the approach would allow one the potential capability to:

- create reinforcement of varying strengths by varying the volume percentage of TiB_2 within the IMC with little change to the nature of the interface between the IMC and the metallic matrix
- create composites of varying strengths by varying not only the strength of the IMC, but the volume percentage of IMC within the metallic matrix
- create composites with a broad range of microstructures, based upon
  - the extent to which a composite variant is conducive to commensurate deformation ("tunable" by the above), or the degree to which the deformation is imposed.
  - the resulting spacing and aspect ratio of the IMC reinforcement, as influenced by the initial size of the IMC powder and the extent to which the composite is deformed
- create a thermodynamically stable composite microstructure by the ability to independently select an intermetallic matrix which is chemically stable within a desired metallic matrix, since the strength of the IMC is established separately by its level of its reinforcement.

The above outlined approach relies upon the technology of powder-based extrusion of multiphase materials. The successful co-extrusion of multiphase materials depends in a complex way on the relative flow characteristics of the component phases. Conditions required to co-extrude are, at best, intuitive and not understood from a fundamental point of view [6].

The purpose of the presently reported work is to establish the feasibility of such an approach by conducting a series of extrusion trials utilizing selected combinations which independently incorporate differing matrix properties, IMC properties, and percentage of IMC incorporated loading within the metallic matrix. Due to the limited scope of resources available to this project, the trials have been used only as a means to explore the influence of the wide range of microstructural and processing variables which are available for later processing and microstructural optimization.

In this short study, a variance of matrix properties will be effected through the use of either commercially-pure titanium (C.P. Ti) or the higher-temperature capable Ti-6Al-4V (weight percentages) alloy. IMC reinforcement properties are varied by incorporating either 20, 30, or 40 volume percent (v%) TiB_2 within a Ti-47Al (atomic percent) near-γ titanium aluminide matrix produced using reactive synthesis techniques, thus creating a range of low to high strength reinforcements, respectively. The IMC reinforcement has been blended within the selected titanium matrices at levels of 20, 30 and 50 v%.

EXPERIMENTAL PROCEDURES

Elemental commercially-pure titanium (C.P. Ti) was procured as both a matrix variant and as a synthesis constituent for the TiB_2-reinforced Ti-47Al IMC. The hydride/dehydride powder was obtained in the size ranges of either -325 mesh ($\leq$ 45 μm diameter) or -100 mesh ($\leq$ 150 μm diameter) from Micron Metals, Incorporated (Salt Lake City, UT). Elemental boron powder was obtained in crystalline form (99.5 purity) from CERAC, Inc. (Milwaukee, WI) in a size range of -100 mesh. Elemental aluminum powder (-100 mesh, 99.97% purity) was also obtained from CERAC. Spherical, Ti-6Al-4V -35 mesh powder was obtained from Nuclear Metals, Inc. (Concord, MA). The Ti-6Al-4V powder was produced using the Plasma Rotating Electrode Process (PREP).

The TiB_2-reinforced near-γ titanium aluminide IMC was produced using reactive synthesis techniques. Synthesis was carried out utilizing facilities at either Virginia Polytechnic Institute and State University (VPI&SU) or at Martin Marietta Laboratories (MML - Baltimore, MD). Elemental powders were blended to target proportions and mechanically pressed into ≈ 101 x 101 x 6 mm wafers (MML) or 20 mm dia. x 25 mm compacts (VPI&SU). The green compacts were subsequently synthesized into IMC's by way of a self-sustaining, exothermic reaction initiated by induction heating in an evacuated (MML and VPI&SU) or inert-gas-filled (VPI&SU) vacuum chamber. TiB_2 of nominal volume percentages of 20, 30, and 50% were produced in the Ti-47Al near-γ titanium aluminide matrix.

Following reaction synthesis, it is necessary to reduce the size of the resulting IMC sponge into powder form. Initial size reduction was effected via jaw crushing followed by milling in a pulverizer, disk mill, and/or by mechanical attrition. The latter two methods were performed with assistance from XFORM, Inc. (Cohes, NY). Sizing of the milled IMC powders was performed using standard screening techniques.

To produce the multilithic metal matrix composites, the IMC reinforcement variants produced were blended with proper proportions of either the C.P. Ti or Ti-6Al-4V.

Extrusion cans were produced by welding nose and tail sections to 15 cm lengths of 73 mm outside diameter ("Schedule 40") C.P. Ti tubing. Welding of the titanium cans was performed at Martin Marietta Laboratories. The blended composite powder was loaded into the can and hermetically sealed under a vacuum of approximately 10^{-3} atm.

Extrusion was performed at 1038°C by personnel of Universal Energy Systems, Inc. (UES; WPAFB, OH) using Wright Research and Development Laboratory (WRDL) facilities. The billet was held at temperature for 2 hours, removed from the furnace, and immediately extruded resulting in an cross-sectional area reduction ratio of 14:1. The total time elapsed from the moment the billet was removed from the furnace to completion of extrusions was less than six seconds. The extruded billet was slow-cooled (≈0.1°C/s) to room temperature in vermiculite.

All metallographic observation, both prior to and following extrusion was performed by either optical microscopy on polished and lightly etched (H_2O-2v% HF) sections, or scanning electron microscopy (SEM). TiB_2 particle sizes were measured by two methods: by "manually" measuring major and minor diameters off micrographs using a hand-held magnifying (7X) eyepiece, or by real-time computer-driven image analysis during SEM. TiB_2 particle diameters are reported as the equivalent diameter, that is, the diameter of a sphere of equal volume to the ellipsoid shape measured. Post extrusion microstructures were examined in both longitudinal and transverse section orientation.

RESULTS AND DISCUSSION

A. Synthesis of IMC reinforcement

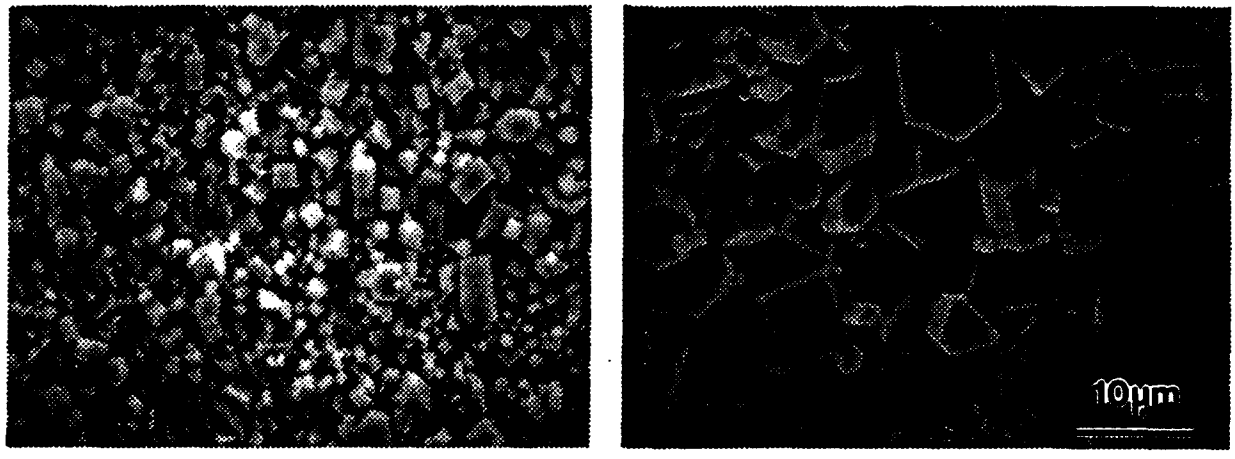
Figure 2 shows the as-synthesized TiB_2 particulate which reside within the Ti-47Al near-γ titanium aluminide intermetallic matrix. The range of sizes which are formed in-situ within the matrix can be seen in Fig. 2a, and the characteristic faceted shape of the particulate is more clearly shown in Fig. 2b and suggests that each TiB_2 is single crystalline, consistent with previously reported results [4].

The size distribution of TiB_2 which has evolved is quantitatively shown in Fig. 3. Average equivalent particle diameter was found to be 0.719 (Fig. 3a) and 0.712 μm for distributions determined via manual- and computer-driven image-analyses, respectively. The diameters measured in the present study are considerably less than those reported in the literature for similarly-synthesized composites. For example, the results of Bryant et al. [5] as well as those of Kampe et al. [6] have previously shown the average particle diameter of TiB_2 in near-γ titanium aluminide to be consistent at approximately 2.0 μm. The scale of the particulate in the present study is particularly significant because these sizes are such that the spacing between particles in

a metallic or intermetallic matrix are sufficiently small to expect that the extremely potent Orowan-type strengthening mechanisms can become operative.

B. IMC Comminution and Blending

Following synthesis, the IMC sponge was reduced via pulverization, disk milling, and attritor milling. Whereas pulverization has been demonstrated in previous efforts [1] to be exclusively satisfactory to achieve comminution to the goal -100 mesh sizes, the present study required additional processing. In particular, the IMC sponge appeared to be more resistant to size reduction, i.e., less brittle and exhibiting higher resistance to breakage. The increased resistance to fracture of the IMC may be a consequence of the much smaller TiB_2 particulate sizes in the present study, i.e., a tougher composite has been synthesized. As a result, disk milling was required to reduce the sponge. The goal of the initial milling was to reduce the size of the IMC sponge to sizes corresponding to -100 mesh sieving. For one of the two variants (Ti-47Al + 20 v% TiB_2), further milling was performed in an attritor mill to increase the proportion of -100 mesh powder. Table II illustrates the size distribution resulting from the disk+attritor milling as applied to the Ti-47Al + 20 v% TiB_2 IMC variant, and for disk milling only as applied to the Ti-47Al + 30 v% TiB_2 IMC.



a. b.

Figure 2. As-synthesized IMC's: a) Ti-47Al + 30 v% TiB_2 IMC as viewed from a polished and deep-etched planar surface, and b) Ti-47Al + 50 v% TiB2, viewed from within a pore of an as-synthesis-reacted sponge particle (note lamellar morphology of the two-phase matrix, 'A').

Table II.

The initial distribution of powder sizes, relative to the sizing goal of -100 mesh, following milling of the as-reacted IMC sponge.

Ti-47Al + 20 v% TiB_2				
	Mass In (gm)	-100 mesh (gms)	-60 / +100 mesh (gms)	+60 mesh (gms)
Disk Mill	1018	300 (29.5%)	237 (23.3%)	481 (47.2%)
Attritor	718	179 (24.9%)	429 (59.7%)	110 (15.3%)
Total Disk+Attritor	1018	479 (47.1%)	429 (42.1%)	110 (10.8%)
Ti-47Al-30 v% TiB_2				
Disk Mill	1060	258 (24.3%)	229 (21.6%)	573 (54.1%)

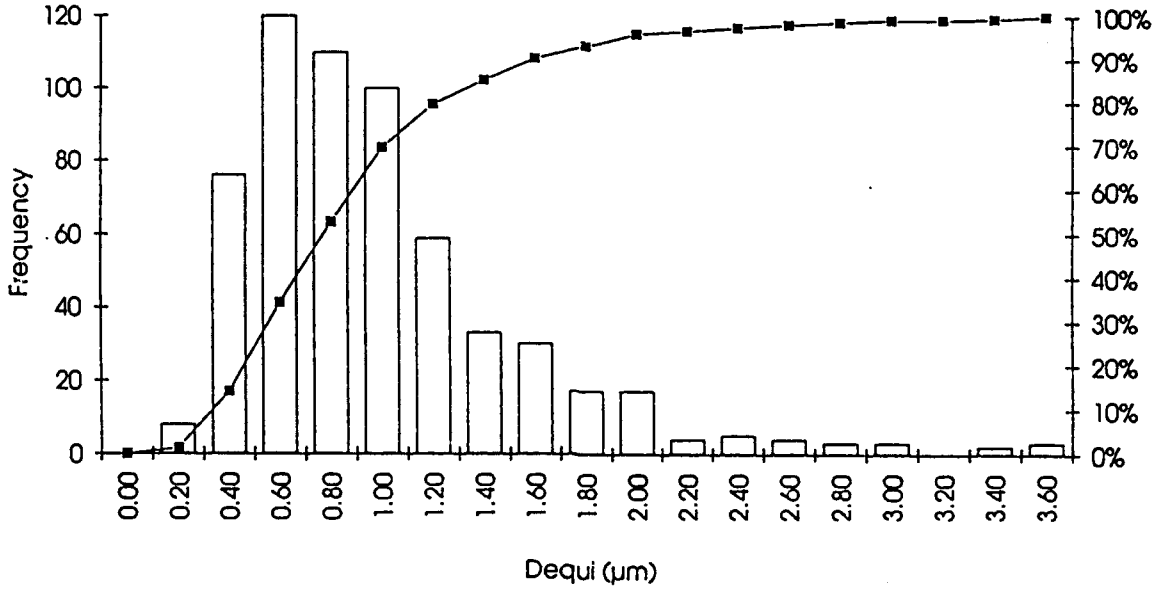
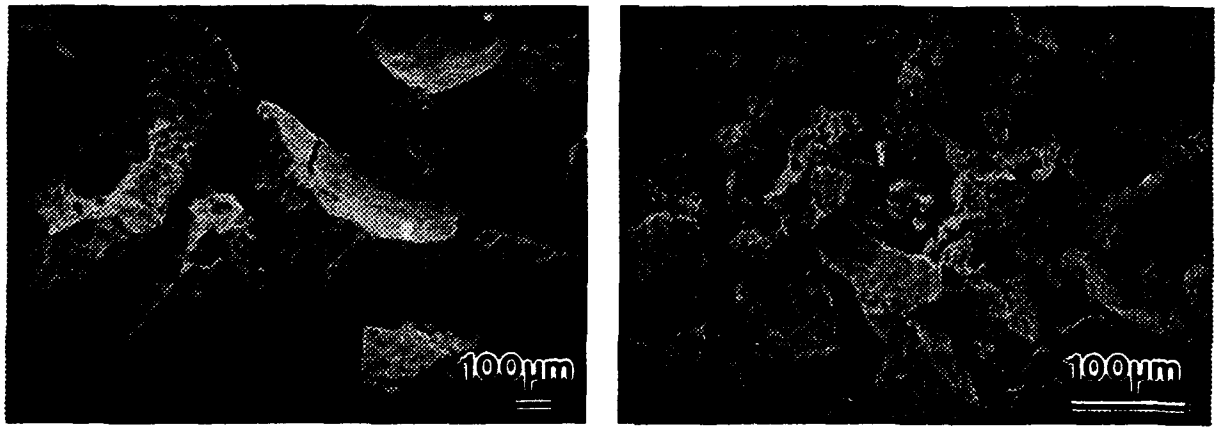


Figure 3. Size distribution histograms and cumulative percent for TiB_2 within the Ti-47Al matrix:; shown is the distribution of average equivalent diameters as measured using image analysis.

The results from Table II indicate that initial milling in the disk mill results in a yield of approximately 25-30 weight percent of -100 mesh powder, with the 20 v% TiB_2 IMC sponge surprisingly exhibiting slightly higher size-reduction efficiency. Comparable weight percentages were obtained for size ranges corresponding to -60/+100 mesh sieving. As shown for the 20 v% TiB_2 variant, however, attritor milling can further reduce the coarse size ranges such that -100 mesh sizing yields approximately 50 weight percent, -60/+100 yields approximately 40 weight percent, with the unusable +60 mesh sponge accounting for only approximately 11 weight percent of the total, initial mass milled.

A more specific description of the actual size ranges comprising the results of Table II are shown in Table III for the Ti-47Al+30 v% TiB_2 IMC processed by disk milling only, and for the Ti-47Al+20 v% TiB_2 processed by disk milling plus mechanical attrition. Figure 4 shows the size and shape of the resulting as-comminutively processed IMC powder.

Table IV summarizes the nominal compositions of the composites which were subsequently blended in the form of powder for extrusion processing. To examine the effect of IMC-reinforcement spacing within the metal matrix, C.P. Ti matrix powder was blended with Ti-47Al + 20 v% TiB_2 IMC of two size ranges, that is, C.P. Ti + 20 v% (-50 / +120 mesh) IMC and C.P.



a. b.

Figure 4. The size and shape of the as-comminutively reduced Ti-47Al + 30 v% TiB_2 IMC powder: a) +50 mesh size, and b) -200 mesh size (note difference in magnification).

TABLE III.

Specific size distribution (in weight percent) of the Ti-47Al + 30 v% TiB_2 (disk milled) and the Ti-47Al + 20 v% TiB_2 (disk-plus-attritor milled) IMC's

Sieving Range	Ti-47Al+30v% TiB_2 (disk milling only) Weight %	Ti-47Al + 20 v% TiB_2 (disk-plus-attritor milling) Weight %
+50	43.8	22.2
-50 / +120	36.4	49.2
-120 / +200	10.1	11.1
-200 / +325	5.4	7.3
-325	4.3	10.2

+ 20 v% (-200 / +325 mesh) IMC. To examine the role of the percentage of the IMC which could be extruded within the metallic matrix, Ti-6Al-4V powder was blended with both 20 and 40 v% Ti-47Al + 30 v% TiB_2 IMC. In the latter instance, the size of the IMC powder corresponds to the entire size distribution summarized in Table III. Micrographs of the C.P. Ti and Ti-6Al-4V utilized for the matrix constituent are shown in Fig. 5.

C. Extrusion

Following milling of the IMC, the resulting powder was blended with powders of the matrix materials - either C.P. Ti or Ti-6Al-4V. In the former instance, the two IMC powder size distributions were separately blended within -325 mesh C.P. Ti matrix powder such that the influence of IMC-reinforcement spacing could be evaluated.

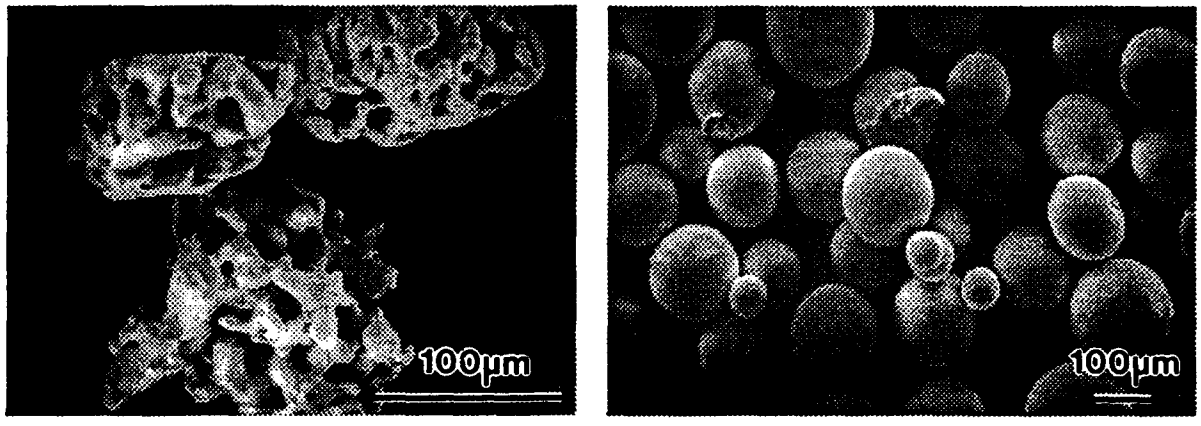
Figure 6 is a macroscopic photograph showing sections of the (IMC)-reinforced metal matrix composite extrusions produced. All extrusions were successful from the perspective of exhibiting stable and continuous flow through the extrusion dies, i.e., no extrusion-based instabilities such as spalling, macroscopic cracking, or splitting were observed.

Microstructures of the as-extruded composites are shown in Figures 7 and 8. The micrographs reveal that the majority of the deformation imposed to the composite powder blend during extrusion was accommodated by the C.P. Ti or the Ti-6Al-4V matrix component. That is, the

TABLE IV.

Summary of the as-blended nominal compositions and powder characteristics of the IMC-reinforced metal matrix composites prepared for extrusion.

METAL MATRIX		IMC REINFORCEMENT		
Composition	Size (sieve range)	Composition	Volume %	Size (sieve range)
C.P. Ti	-325	Ti-47Al + 20 v% TiB_2	20	-50 / +120
C.P. Ti	-325	Ti-47Al + 20 v% TiB_2	20	-200 / +325
Ti-6Al-4V	-60	Ti-47Al + 30 v% TiB_2	20	± 50 (all)
Ti-6Al-4V	-60	Ti-47Al + 30 v% TiB_2	40	± 50 (all)



a. b.

Figure 5. The size and shape of the powder used as the matrix constituent in the IMC-reinforced composites: a) C.P. Ti produced via hydride/dehydride (-325 mesh), and b) Ti-6Al-4V produced via Plasma Rotating Electrode Process (PREP, -60 mesh).

IMC-reinforcing particles exhibited less deformation than would be predicted if deformation was fully commensurate with the matrix. Figure 7 shows that for the C.P. Ti matrix material, an essentially fully dense composite has been produced but with essentially no discernible deformation exhibited by the IMC component. The Ti-6Al-4V matrix composite, however, appears capable of transferring sufficient load during the deformation to lead to elongated IMC reinforcing particles. However, evidence of cracking within the IMC due to the incompatibility of strain is noted, as is the presence of porosity associated with the irregular shape of the initial IMC particles and their limited ability to deform, evidence of porosity within and around the IMC reinforcement is evident.

An experience-based summary and a review of the available theory, albeit primarily phenomenological in origin, associated with the direct extrusion of powder into bulk form has been presented by Roberts and Ferguson [1]. The authors further discuss guidelines associated with the co-extrusion of dissimilar constituents, including issues associated with the selection of

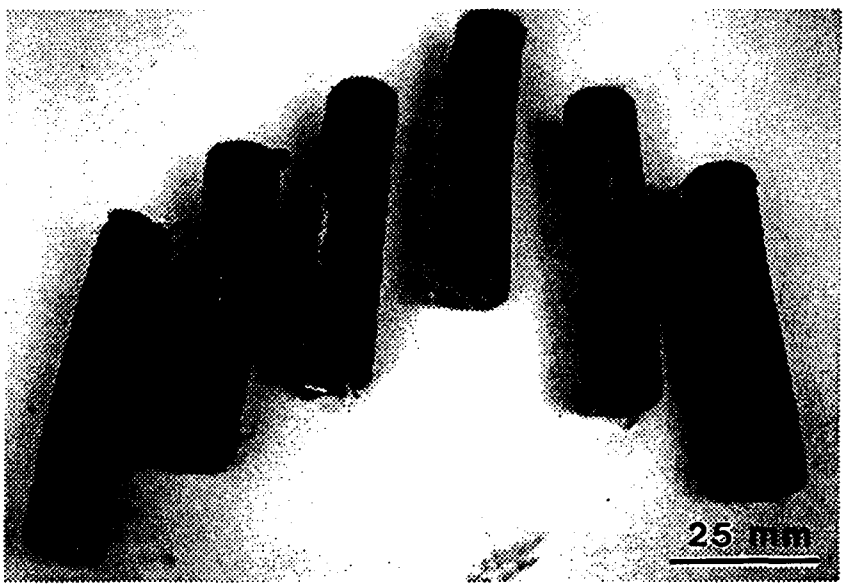


Figure 6. A macroscopic photo of sections of the co-extruded metal matrix+IMC-reinforced composites. No evidence of external cracking or gross mechanical incompatibility is evident.

the container material in which the loose powder is packed, evacuated, and eventually extruded. With several examples of exception, the following guidelines are recommended:

- thermodynamic (chemical) compatibility between the constituents and the can material
- similar flow stress behavior between the constituents and the container, including both yield and strain hardening behavior.

The effective quantification of the boundaries by which these guidelines become influential and important is beyond the current understanding of the process of direct extrusion of powder.

In the present work, the several observations are pertinent and provide insight into the requirements which are necessary to produce an in-situ IMC-reinforced composite via deformation processing by co-extrusion techniques. First, the magnitude of the flow stress of the

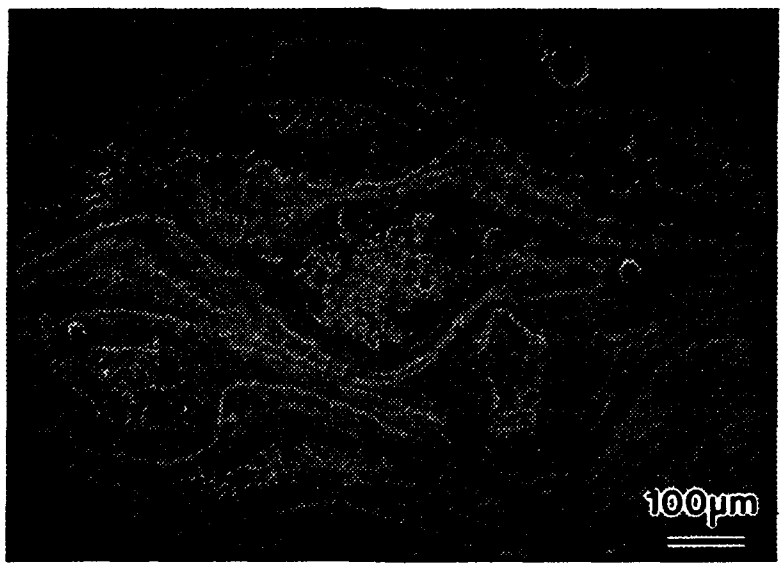
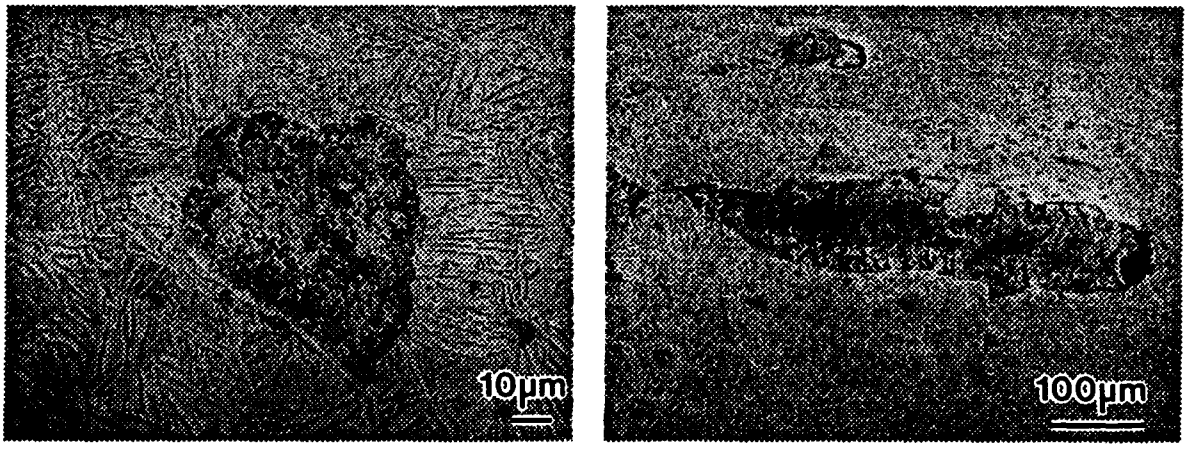


Figure 7. A micrograph of a longitudinal section of the C.P. Ti + 20 v% IMC matrix composite produced via co-extrusion deformation processing. Essentially all of the extrusion derived deformation has been accommodated by the lower-flow-strength C.P. Ti matrix component.

matrix constituent appears to have a very strong influence on the ability to commensurably deform the reinforcing IMC entity. Specifically, the Ti-6Al-4V matrix composite shown in Fig. 8, presumably is characterized by a higher flow stress at the extrusion temperature. The higher flow stress in the continuous phase during extrusion will hence be capable of transferring a higher proportion of the imposed strain-based load to the dispersed IMC phase. The C.P. Ti matrix material appears totally incapable of sustaining a load of sufficient magnitude to deform the IMC phase. However, it is noted that the C.P. Ti container did appear to extrude commensurably with the enclosed Ti-6Al-4V powder.

Quantified values of the relative flow stresses of the matrix materials and the IMC reinforcements at the temperatures of extrusion are not generally available, and obtaining actual values were beyond the means and scope of the present effort. These observations were suprising in that their relative values were originally assumed to be approximately equal due to the fact that the temperature of extrusion was well beyond the practical use temperature of each of these materials; that is, extrapolation of yield data to the temperatures of extrusion indicate that the flow stress of all the materials incorporated rapidly decline and converge. Nonetheless, as illustrated by Roberts and Ferguson, knowledge of the flow behavior of the consolidated or cast forms of the powder constituents often does correlate well to the actual direct powder co-



a. b.

Figure 8. The as-extruded microstructure of the Ti-6Al-4V + 30 v% IMC composite produced via co-extrusion deformation processing: a) transverse section, and b) longitudinal section. In b), while some deformation as been transferred to the IMC-reinforcement, evidence of both cracking due to strain incompatibility and porosity is present.

extrudability, since powder forms have been shown to exhibit better deformability relative to bulk counterparts.

The Ti-6Al-4V matrix material has also been shown to result in a higher proportion of residual porosity in the as-extruded microstructure (Fig.). This is compatible with the presumption that the flow stress of the Ti-6Al-4V alloy is indeed higher than that of the C.P. Ti at these temperatures of extrusion, and/or that the extrusion ratio employed was insufficient to lead to full consolidation of the spherical (more-difficult-to-consolidate) powder.

SUMMARY

A series of metal matrix composites have been produced by the co-ext of metallic alloy (matrix) and intermetallic matrix composite (IMC, reinforcement) der. While the composites were shown, significantly, to be successfully processable and extrudable by this method, the desired commensurably-deformed microstructure was not generally observed. The lack of extrusion-derived deformation in the IMC-reinforcing component of the metal matrix composite is believed to be a consequence of the large differential in the flow strengths between the components of the composite. When the flow stress of the matrix is extremely low (as for C.P. Ti matrix) no discernible deformation in the IMC-reinforcement is observed, and the as-extruded composite is free of porosity and is essentially fully dense. When the flow stress of the matrix materials is elevated (as for Ti-6Al-4V), the IMC-reinforcement is observed to deform but the microstructure contains evidence of residual porosity and strain-incompatibility cracking, both being associated with the presence of the difficult-to-deform IMC component.

Synthesis of the IMC component by reaction techniques lead to composites with extremely fine faceted TiB_2 particulate, exhibiting an average diameter of approximately 0.7 μm compared to 2.0 μm as reported in previous studies. The finer dispersion lead to improved properties of the IMC as manifested by the difficulty associated with comminutively reducing the powder size.

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RECOMMENDATIONS FOR FUTURE WORK

In the short term, continued study and understanding of the results described above will be performed. An additional extrusion is currently scheduled which incorporates a larger extrusion ratio, i.e., a more severe reduction in area which translates into higher deformation-derived stresses, better potential consolidation efficiency, and smaller spacings between IMC-reinforcing particles.

Tensile testing will be performed on the composites produced. Tensile and fracture properties will be related to the strength and distribution of the IMC-reinforcement within the composite microstructure. Sufficient data will be available to attempt a correlation between composite strength and a relatively wide range of reinforcement spacings as effected through selective IMC powder sizing and extrusion ratio.

In the longer term, it important to note the significance of being able to successfully co-extrude the blended powder composites described above, with no evidence of the macroscopic extrusion instabilities which frequently plague the processing of advanced materials in this manner. However, the desired in-situ evolution of the composite microstructure remains to be developed and optimized. It is believed that the successful development of this approach to in-situ composite processing can lead to tremendous material design capability and design.

To obtain the desired microstructures, it is suggested that the role of the relative magnitudes of the flow behavior of the individual components of the composite be understood and modeled. In this regard, a study of the strength and flow behavior of the matrix material and the numerous variants offered by the IMC-based reinforcement approach be performed.

The results of the present study lead to the intuitive conclusion that better matching between the flow stress of the matrix material and the IMC-reinforcement be attained. Since the higher strength Ti-6Al-4V matrix exhibited increased resistance to full densification as indicated by the presence of porosity in these composites, a more fruitful approach might involve attempting to decrease the flow behavior of the IMC at the temperatures of extrusion. An example where this could accomplished, with little sacrifice in performance, would involve modifying the matrix of the IMC such that it is extruded at a higher homologous temperature (i.e., T/T_{mp}), where T would be the temperature of extrusion and T_{mp} the melting point of the IMC. As applied to the current system, this could be easily and effectively accomplished by creating IMC's from a Al_3Ti

(trialuminide) matrix, rather than from the higher-melting TiAl-based intermetallic utilized presently.

An understanding of the processing and physics associated with the development of the fine dispersion of TiB_2 within the IMC is important to facilitate the exploitation of reaction synthesis techniques as a means to produce effective and technologically-viable in-situ composites. The scale of the TiB_2 formed in the present study is extremely significant since brings the microstructure into the realm where the potent Orowan strengthening may be operative and effective.