Review

Fabrication of MAX-Phase-Based Ceramics by Three-Dimensional Printing

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Abstract

Three-dimensional printing (3DP) is a flexible and cost-effective method for direct digital manufacturing that provides capabilities for creating a wide range of part geometries in a broad variety of materials. Recently, a combined process of 3DP and reactive melt infiltration (RMI) has been applied to fabricate MAX-phase-based ceramics, exhibiting great potential in the fabrication of bulk compounds with complicated shape. This paper briefly summarizes the fabrication of Ti₃AlC₂- and Ti₃SiC₂-based ceramics with the combined process. 3DP facilitates the prior design of a porous preform with specific pore distribution and microstructure, which is beneficial to the control of the volume change of the following reaction in the RMI process, promoting the near-net-shape fabrication of MAX-phase-based ceramics with high flexibility in component geometry.

Keywords: Three-dimensional printing, RMI, Ti₃SiC₂, Ti₃AlC₂.

I. Introduction

MAX phases are thermodynamically stable nanolaminates exhibiting unusual and unique properties. They have combined characteristics of metal and ceramics¹–³ and exhibit attractive properties such as good oxidation resistance, low density, high modulus, good thermal and electrical conductivity, excellent thermal shock resistance and high-temperature strength, and easy machinability⁴–⁶. They are considered promising structural/functional materials for high-temperature applications⁷–⁹. MAX phases have also been successfully used as reinforcement to improve the mechanical properties of intermetallic and ceramic-based composites¹⁰.

Recently, several methods, such as chemical vapor deposition (CVD)¹¹,¹², mechanical alloying (MA)¹³,¹⁴, self-propagating high-temperature synthesis (SHS)¹⁵–¹⁷, hot pressing (HP)¹⁸ and spark plasma sintering (SPS)¹³, have been applied to fabricate MAX-phase-based materials and the properties of these materials studied extensively.

Tzenov et al.¹⁹ fabricated dense Ti₃AlC₂ bulk materials by means of reactive hot isostatic pressing (HIP) of a Ti, C, and Al₄C₃ powder mixture at 70 MPa, 1400 °C for 16 h. Yeh et al.¹⁷ used a 3Ti/1.25C/0.25Al₂C₃ mixture as the starting powder to fabricate Ti₃AlC₂ ceramics by means of SHS, and the increase of the preform density prior to the SHS process led to the increase of Ti₃AlC₂ content from 50.5 to 73.2 wt%. Yanga et al.¹³ combined MA and SPS methods for the fabrication of Ti₃AlC₂-based material. Dense Ti₃AlC₂ was fabricated by means of SPS at 1050 °C for 10–20 min with mechanically alloyed powders from a starting mixture of 3Ti/1.1Al/2C. Han et al.²⁰ fabricated polycrystalline bulk Ti₃AlC₂ with HP from a mixture of TiCₓ (x = 0.6) and Al powder under 25 MPa pressure in the temperature range of 800 to 1600 °C, fully dense and pure Ti₃AlC₂ was synthesized by means of HP above 1400 °C; Vickers hardness up to 6 GPa and flexural strength higher than 900 MPa were measured for the as-fabricated bulk Ti₃AlC₂ samples.

CVD was the early method for the fabrication of dense Ti₃SiC₂ ceramics with high purity¹¹. Deposition is usually conducted at 1300–1600 °C with SiCl₄, TiCl₄, CCl₄ and H₂ as source gases. However, CVD could only be used for the fabrication of thin films and coatings composed of MAX phases. Barsoum et al.² used the HIP method for the fabrication of bulk Ti₃SiC₂ materials. Powder blends of Ti, C and SiC with a molar ratio of 3:2:1 were cold-pressed under 180 MPa. In order to obtain dense materials, the compacted green bodies were then reacted in the HIP device at 1600 °C for 4 h under 45 MPa. Gao et al.²¹ fabricated dense Ti₃SiC₂ bulk material by means of SPS up to 1300 °C. The composition of starting powder mixture was Ti:Si:TiC = 1:1:2 (molar ratio). The as-fabricated Ti₃SiC₂ material contained only 2 wt% TiCₓ impurity. In-situ hot pressing was conducted by Zhou et al.³ using mixed powders with a molar ratio of Ti:Si:C = 0.42:0.23:0.35. Ti₃SiC₂ bulk material with purity of 93 wt% was fabricated at 1550 °C in an atmosphere of flowing Ar. Further study²² shows that using...
a small amount of Al to replace the Si atom in Ti$_3$SiC$_2$ could help reduce the content of remaining TiC and obtain Ti$_3$Si(Al)C$_2$ with high purity. The arc-melting and annealing route was applied by Arunajatesan et al. for the synthesis of bulk Ti$_3$SiC$_2$ from the Ti, Si, and C powders. The bulk sample containing about 2 vol% TiC as the second phase was made from Si-deficient and C-rich starting compositions.

For all these fabrication methods, raw materials with high activity are preferred in order to fabricate dense materials with high purity. The fabrications are always conducted at high temperatures with external force applied for a long reaction time, which restricts the fabrication of bulk samples with complicated shape.

In order to fabricate MAX-phase-based ceramics with complicated shape, Sun et al. combined three-dimensional layered printing (3DP), cold isotactic pressing (CIP) and sintering for the fabrication of Ti$_3$SiC$_2$-powder-based structures. Three-dimensional Ti$_3$SiC$_2$ structures with complex geometry and high density can be fabricated (Fig. 1).

A number of studies have used the combination of 3DP and reactive melt infiltration (RMI) to fabricate dense ceramics with near-net shape. In the RMI process, a molten metal spontaneously penetrates into pores driven by capillary force, and then reacts with the porous preform to form the bulk compounds. During RMI, no external pressure is required, which is beneficial to the near-net-shape fabrication of complex structures and offers high flexibility with regard to component geometry. The combined process has been successfully applied to fabricate SiSiC with complicated shape. Moon et al. firstly used 3DP to prepare porous carbonaceous preforms composed of glassy carbon powders with a particle size of 45–105 μm, using an acetone-based furfuryl resin binder as a printing solution. After liquid silicon infiltration at 1450°C in N$_2$ atmosphere, a SiSiC composite with a coarse SiC grain structure was formed. Travitzky et al. prepared a SiC-C preform from a mixture of SiC and starch-cellulose powders by means of 3D printing and pyrolysis, and then dense SiSiC was obtained with the infiltration of liquid silicon into the SiC-C preform. Dense Ti$_3$AlC$_2$- and Ti$_3$SiC$_2$-based ceramic parts with a complex shape were fabricated with a combined method of RMI and 3DP. By means of the pre-design of the preform with 3DP and control of the subsequent reactions, it is possible to control dimensional changes during RMI. The as-fabricated ceramics can attain good mechanical properties. The following paper will focus on progress in the fabrication of Ti$_3$AlC$_2$- and Ti$_3$SiC$_2$-based ceramics. It consists of three parts: the first part is the structural design of the porous preform, which directly affects the infiltration kinetics of the melt in the RMI process; the second part is the reaction mechanism during the fabrication process; the third part summarizes the microstructure and mechanical properties of different MAX-phase-based ceramics, indicating the relationship between properties and microstructure.

II. Structural Design of the Preform

(1) Infiltration kinetics during the fabrication of max-phase-based ceramics

Metal melt infiltration into a porous solid preform is one of the preferred methods to fabricate ceramic-metal and ceramic matrix composites. A molten metal can penetrate into the pores driven either by an external force (squeeze casting) or by the action of capillary pressure created when the liquid wets the solid. The reaction between the porous ceramic preform and the molten metal may considerably enhance the wettability. Compared to pressure-assisted infiltration techniques such as squeeze casting or gas-pressure infiltration, RMI offers high flexibility with regard to component geometry. With the pre-design of the composition and microstructure of porous preform, bulk materials with complex and near-net shape can be obtained.

The infiltration of Al melt into a TiC/TiO$_2$ preform was initially used to fabricate Ti$_3$AlC$_2$-based ceramics with high crack resistance. After that, works on fabricating Ti$_3$SiC$_2$-based ceramics by means of LSI were
conducted 41 – 45. Lu et al. 44 fabricated Ti3SiC2-based ceramic with the infiltration of silicon melt into a TiC-C preform. Nan et al. 39 fabricated Ti3SiC2-based ceramics with high electrical conductivity and bending strength in a joint process of three-dimensional printing (3DP) and liquid silicon infiltration. Shan et al. 45 synthesized Ti3SiC2 bulk ceramics by infiltrating Ti/TiC preform with silicon melt. Hwang et al. 46 fabricated high-purity Ti3SiC2 compounds with the infiltration of silicon melt into a TiC x (x = 0.67) preform. Ti2SnC was also recently fabricated with the infiltration of Sn melt into TiC0.5 47. Besides the fabrication of bulk compounds, RMI has also been considered as an effective way to introduce Ti3SiC2 into fiber-reinforced ceramic matrix composites such as C/C-SiC and C/SiC composites 6, 8, 42, 48 for optimizing mechanical and tribological performance.

During RMI, the wetting behavior of the melt on the pore surface in the preform is the main factor affecting the infiltration kinetic, which determines the microstructure uniformity and the maximum size of the as-fabricated material. Once metal melt penetrates into the preform, a reaction may occur at the metal melt–ceramic interface. In the course of the reaction, the metal firstly diffuses into the formed solid phases, and then reacts with the ceramic particles, and the infiltration process changes from reaction control to diffusion control. Fast infiltration speed is conducive to the homogeneous microstructure of as-fabricated material, and the infiltration kinetics evaluation should always be considered as a primary factor.

The infiltration kinetics of metal melt into the porous preform is provided by Darcy’s law 49. Infiltration depth, h, was estimated as a function of time, t, (Eq. 1)

\[ h = \left( \frac{\varepsilon_p}{1 - \varepsilon_p} \right) \left( \frac{r k \gamma \cos \theta}{6.25 \mu} \right)^{1/2} \]  

where \( \varepsilon_p \) is the pore volume fraction; \( r \) is the pore radius; \( \mu \) denotes the viscosity of melt; \( k \) is the particle size factor; \( \gamma \) is the surface tension and \( \theta \) is the wetting angle.

During the infiltration process, the pore radius and wetting angle directly affect the infiltration depth in accordance with Eq. (1). Table 1 summarizes the infiltration depth of Al and Si melt into the TiC preform with different pore radius within 60 s at 1000 °C and 1600 °C in vacuum, respectively 10, 40. For pores with same radius, the infiltration rate of Si melt and Al melt varies slightly, which can be attributed to the close viscosity and wetting angle.

Table 1: Infiltration depth of Al and Si melt into a TiC preform with different pore radius in 60 s at 1000 °C and 1600 °C in a vacuum 10, 40.

<table>
<thead>
<tr>
<th>Pore radius (µm)</th>
<th>Infiltration depth (cm)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Al melt</td>
</tr>
<tr>
<td>0.07</td>
<td>2</td>
</tr>
<tr>
<td>1</td>
<td>2.4</td>
</tr>
<tr>
<td>23</td>
<td>14.2</td>
</tr>
<tr>
<td>30</td>
<td>20</td>
</tr>
</tbody>
</table>

Infiltration efficiency could be significantly improved with the use of a preform with bimodal pore size distribution 50. Pores with larger radius are the inter-agglomerate ones, the ones with smaller radius are the intra-agglomerate pores (Fig. 2). The existence of large pores guarantees the rapid infiltration of metal melt.

![Fig. 2: Typical microstructure of a preform with bimodal pore size distribution 42.](image)

(2) Microstructural design of the preform

In order to obtain a preform with the bimodal pore size distribution mentioned in II(1), an indirect 3DP process was used for the structural design. The basic design principles and essential factors are as follows:

3DP is a layered fabrication process in which the sliced 2D profile of a computer model is printed on a fresh layer of powder via deposition of a suitable binder. Successful 2D profiles are then printed on a freshly laid layer of powder until the entire model is completed. The printed binder joins the respective profiles of each layer together. The part is completed upon removal of the unbound powder and suitable post-processing 51.

Successful realization of a specific 3DP process involves not only the printing process itself, but also the formulation of a suitable combination of a powder and binder material system along with process details for printing and post-processing, both of which play major role in determining the characteristics of the as-produced parts. During the building process, there is no phase change of material involved based on the inkjet principle, guaranteeing its high manufacturing speed.

In order to attain a homogeneous and coherent spreading of each layer during the printing procedure, particle or granulate dimension, morphology and flowability are important factors to be considered. Generally, in order to obtain ideal deposition ability, spherical powder particles with a diameter of 20 µm and larger are preferable 52. Spherical powders tend to flow better and have low internal friction. Fine powders (~ 1 µm) tend to agglomerate owing to van der Waal’s forces and moisture effects, but have the potential advantages of increase sinterability, lower surface roughness, smaller minimum features, and thinner layers 53. The application of spray-dried 54 or freeze-dried 10, 39 granules enables the use of fine sinter-
ing-active particles. The higher the pourability of powder, the thinner the layers are built by the leveling roller, and hence the higher the quality of the printed body. It was also demonstrated that the wettability and pourability of powders can be sufficiently improved by means of plasma treatment.

While 3DP may produce porous preforms with a high degree of freedom in terms of geometry and shape, the infiltration of metal melt into the porous preform results in a dense microstructure of the component. Net-shape manufacturing is possible by controlling dimensional changes associated with a displacive reaction between the infiltrated melt and ceramic preform phase. When granules are used as the printing powder, the size of granules decides the thickness of each layer and thus the pore radius of the inter-agglomerate, and the particle size distribution decides the pore radius of the intra-agglomerate. The comparison of a TiC/TiO₂ preform and a TiC preform is summarized in Table 2. The pore size distribution of the two preforms is different, because in the TiC/TiO₂ preform the smaller TiO₂ particles fill the spaces between the large TiC particles and reduce the intra-agglomerate pores fractions.

### Table 2: Comparison of the 3DP process of TiC/TiO₂ preform and the TiC preform

<table>
<thead>
<tr>
<th>TiC/TiO₂ preform</th>
<th>TiC preform</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Starting materials</strong></td>
<td>TiC (1.2 μm): TiO₂ (30 nm) = 63:31 (weight ratio)</td>
</tr>
<tr>
<td><strong>Binder</strong></td>
<td>Dextrin (6 wt%)</td>
</tr>
<tr>
<td><strong>Blending and granulation</strong></td>
<td>Freeze-drying after ball milling</td>
</tr>
<tr>
<td><strong>Printing powder size/shape</strong></td>
<td>~200 μm diameter/spherical</td>
</tr>
<tr>
<td><strong>Porosity of the green part</strong></td>
<td>55 vol%</td>
</tr>
<tr>
<td><strong>Density of the sintered part</strong></td>
<td>1.61 g/cm³</td>
</tr>
<tr>
<td><strong>Pore volume fraction</strong></td>
<td>51 vol% (inter), 14 vol% (intra)</td>
</tr>
</tbody>
</table>

Post-processing of the preform was conducted in flowing N₂ at 800 °C for 2 h, and it was then sintered in flowing Ar at 1400 °C for 0.5 h. The dextrin decomposed into amorphous carbon in the pre-sintered preform. The bending strength of the 3D-printed preform was lower than 5 MPa, and increased to 5 ~ 10 MPa after pre-sintering. The porosity of the preform increased owing to the pyrolysis of the dextrin during post-processing.

### III. Reaction Mechanism

#### (1) Formation mechanism of Ti₃AlC₂ during RMI

According to Fan et al., TiC and Al are thermally stable in the metal melt infiltration temperature range, so an indirect reaction path was designed accordingly. The preform was composed of TiC, TiO₂, and dextrin with a weight ratio of 63:31:6.

\[
2\text{TiO}_2(s) + \text{C}(s) \rightarrow \text{Ti}_2\text{O}_3(s) + \text{CO}(g) \quad (2) \\
\text{Ti}_2\text{O}_3(s) + 8\text{Al}(l) \rightarrow 2\text{TiAl}_3(l) + \text{Al}_2\text{O}_3(s) \quad (3) \\
\text{TiAl}_3(l) + 2\text{TiC}(s) \rightarrow \text{Ti}_3\text{AlC}_2(s) + 2\text{Al}(l) \quad (4)
\]

The first step of the reaction was the generation of TiAl₃. Because the binder material, dextrin, decomposes into amorphous carbon during post-processing, carbon exists in the preform and participates in the reactions. Eq. 3 and Eq. 4 provide a formation path for Ti₃AlC₂. The generated Ti₂O₃ during the reaction process was reported to offer better wetting ability with Al melt than TiO₂ at 1400 °C. TiAl₃ then reacted with TiC to generate Ti₃AlC₂.

#### (2) Formation mechanism of Ti₃SiC₂ during RMI

Fan et al. studied the formation mechanism of Ti₃SiC₂ during the RMI process. The study shows that TiC prefers to react directly with Si to form TiSi₂ and SiC, the generation of Ti₃SiC₂ requires the formation of TiC twins. The carbon existing in the system reacts with TiSi₂ to form Ti₃SiC₂. Eqs. 5 ~ 9 give the reactions that occur during RMI.

\[
\text{TiC}(s) + 3\text{Si}(l) \rightarrow \text{TiSi}_2(l) + \text{SiC}(s) \quad (5) \\
\text{Si}(l) + \text{C}(s) \rightarrow \text{SiC}(s) \quad (6) \\
\text{TiSi}_2(l) + 3\text{C}(s) \rightarrow \text{TiC}(s) + 2\text{SiC}(s) \quad (7) \\
\text{Si}(l) + \text{TiSi}_2(l) \rightarrow \text{Ti-Si}\text{rich}(l) \quad (8) \\
2\text{TiC}(s) + \text{Ti-Si}\text{rich}(l) \rightarrow \text{Ti}_3\text{SiC}_2(s) + 2\text{Si}(l) \quad (9)
\]

### IV. Near-Net-Shape Fabrication during RMI

RMI is a reaction-infiltration competition process, the control between the reaction speed and the infiltration speed is an essential factor for the realization of near-net-shape fabrication. Nan et al. printed TiC preforms with a bimodal pore structure by means of 3DP, and the preforms were then infiltrated with liquid silicon at 1600 ~ 1700 °C, to obtain Ti₃SiC₂-based ceramics with favorable mechanical properties and high conductivity. However, a strong volume shrinkage of 51.6 vol% was observed. Subsequently, Al was then introduced into the infiltration melt, leading to a decrease in the infiltration temperature. The results showed that the volume shrinkage ratio of samples decreased to 6.43 % and 1.17 % with the
Al_{40}Si_{60} and Al_{75}Si_{30} as the infiltration metal melt, respectively. A similar study was also conducted by Wang et al. 

porous TiC preforms were firstly prepared by means of cold pressing, and then Al-Si alloy infiltration was used to fabricate Ti_{3}Si(Al)C_{2}-based ceramic, no apparent volume change occurred during the fabrication process. The study also showed that the existence of Al could promote the nucleation of Ti_{3}SiC_{2}, because it helps reduce the twin boundary energy of TiC.

In contrast, a joint process of three-dimensional printing and reactive melt infiltration realized near-net shape fabrication of a Ti_{3}AlC_{2}-based ceramic gearwheel with complex shapes. Fig. 3 shows the CAD model of a gearwheel (left) and the corresponding printed part after RMI. Compared with the pre-sintered preform, the as-fabricated ceramic shows an expansion of 4% along the in-plane direction and shrinkage of 3.2% along the out-of-plane direction. The total dimensional change is significantly small compared with the sintering of a powder component, which facilitates the fabrication of bulk materials with complex shape.

V. Microstructure and Mechanical Properties of MAX-Phase-Toughened Ceramics Fabricated with RMI

The phase composition of ceramics fabricated by means of RMI includes intermediate products like TiAl_{3} or TiSi_{2}, ceramic particles like Al_{2}O_{3} or SiC and some unreacted TiC and residual melt. Since the mechanical properties of different components vary significantly, the effect of phase distribution on the mechanical properties of as-synthesized ceramic parts is significant.

Figs. 4a and b present the typical microstructure of Ti_{3}SiC_{2}- and Ti_{3}AlC_{2}-based ceramics fabricated with RMI. The bright particles in Fig. 4a are Ti_{3}SiC_{2}, the dark areas are TiSi_{2}, TiC particles could be found embedded in both areas. The phase content of Ti_{3}SiC_{2}, TiSi_{2} and TiC are 45, 25 and 21% respectively. In Fig. 4b the phase with bright contrast is Ti_{3}AlC_{2}, the gray phases are TiAl_{3}, the round dark particles are Al_{2}O_{3}, the matrix with dark contrast is Al, and the small particles with white contrast are TiC. According to image analysis of the SEM micrographs, the fractions of Ti_{3}AlC_{2}, TiAl_{3}, Al_{2}O_{3}, Al, and TiC are approximately 35, 30, 10, 20, and 5 vol%, respectively.

Compared with conventional ceramics, MAX phases are damage-tolerant ceramics owing to their special atom bonding and laminated structure. In the loading process, deformation modes of delamination, laminate fracture, buckling and kink-band occur to promote crack deflection, which absorbs lots of energy and favors the improvement of flexural strength and fracture toughness. As shown in Table 3, the Ti_{3}AlC_{2}-toughened composites attained a four-point bending strength of 320 ± 40 MPa, and a Young’s modulus of 184 ± 24 GPa. The fracture toughness of the as-fabricated sample at 1300 °C is 8.1 ± 1.7 MPa m^{1/2}, the sample fabricated at 1400 °C has an increased K_{IC} of 9.7 ± 0.8 MPa m^{1/2}. The composite with the highest amount (45 vol%) of Ti_{3}SiC_{2} obtained by means of LSI attained a bending strength of 293 ± 17.8 MPa. The composite fabricated with Al_{75}Si_{30} infiltration attained a bending strength of 233 ± 53 MPa and a K_{IC} of 4.6 ± 1.08 MPa m^{1/2}, while the composite obtained with Al_{40}Si_{60} alloy infiltration attained a bending strength of 107 ± 4 MPa and a K_{IC} of 5.9 ± 0.6 MPa m^{1/2}.
According to Table 3, the differences in the mechanical properties of the materials are due to the different phase distribution of materials leading to varied properties. Table 4 presents the basic properties of the components in the as-fabricated ceramics. As can be seen from Table 4, the main phase (Ti$_3$SiC$_2$ and Ti$_3$AlC$_2$), and the intermediate phase (TiSi$_2$ and TiAl$_3$) of the two ceramics exhibit similar properties. The different mechanical properties of different ceramics can be attributed to the different phase distribution. Ti$_3$AlC$_2$ consists of small rod-shaped grains with small grain size, exhibiting dispersive distribution in the ceramics. These crosslink with each other in the matrix, providing high toughening efficiency. The microstructure-crack path interaction in Ti$_3$AlC$_2$ based ceramics is significantly tortuous (Fig. 5), and crack deflection occurs along the weak interface, leading to a fluctuating crack propagation path, which greatly extends the effective crack length and absorbs more fracture energy, leading to an improvement in fracture resistance. However, Ti$_3$SiC$_2$ with large particle size is concentrated in specific areas, thus restricting the toughening areas, and between the Ti$_3$SiC$_2$ particles are the brittle TiSi$_2$ phase and small TiC particles.38, 40.

![Fig. 4: BSE of Ti$_3$SiC$_2$ (a)- and Ti$_3$AlC$_2$ (b)-based ceramics fabricated with RMI 40, 10.](image1)

![Fig. 5: Optical track micrograph of crack propagation of Ti$_3$AlC$_2$ based ceramics fabricated at 1400 °C.](image2)

Table 3: Bending strength and fracture toughness of different ceramics.38, 39.

<table>
<thead>
<tr>
<th></th>
<th>Ti$_3$AlC$_2$</th>
<th>Ti$_3$SiC$_2$</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>1300 °C</td>
<td>1400 °C</td>
</tr>
<tr>
<td>$\sigma_f$ (MPa)</td>
<td>320±40</td>
<td>—</td>
</tr>
<tr>
<td>$K_{1C}$ (MPa·m$^{1/2}$)</td>
<td>8.1±1.7</td>
<td>9.7±0.8</td>
</tr>
</tbody>
</table>

Table 4: Basic properties of the components in the as-fabricated ceramics.2, 10, 24, 37 – 39.

<table>
<thead>
<tr>
<th>Properties</th>
<th>$T_M$ (°C)</th>
<th>$\rho$ (g/cm$^3$)</th>
<th>E (GPa)</th>
<th>CTE ($\times 10^{-6}$/K)</th>
<th>$K_{1C}$ (MPa·m$^{1/2}$)</th>
<th>$\sigma_f$ (MPa)</th>
<th>HV (GPa)</th>
</tr>
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<tbody>
<tr>
<td>Ti$_3$AlC$_2$</td>
<td>1400</td>
<td>4.25</td>
<td>297</td>
<td>9</td>
<td>7.2, 4.6</td>
<td>375, 172</td>
<td>3.5</td>
</tr>
<tr>
<td>Ti$_3$SiC$_2$</td>
<td>1800</td>
<td>4.53</td>
<td>322</td>
<td>8.6, 9.7</td>
<td>8–16</td>
<td>439</td>
<td>4.0</td>
</tr>
<tr>
<td>TiAl$_3$</td>
<td>1350</td>
<td>3.37</td>
<td>216</td>
<td>13</td>
<td>2</td>
<td>162</td>
<td>6</td>
</tr>
<tr>
<td>TiSi$_2$</td>
<td>1540</td>
<td>4.04</td>
<td>250</td>
<td>9</td>
<td>2–3</td>
<td>170</td>
<td>16</td>
</tr>
<tr>
<td>Al$_2$O$_3$</td>
<td>2015</td>
<td>3.97</td>
<td>380</td>
<td>8.6</td>
<td>3.3</td>
<td>350</td>
<td>16</td>
</tr>
<tr>
<td>SiC</td>
<td>2827</td>
<td>3.2</td>
<td>418</td>
<td>4.02</td>
<td>3.5</td>
<td>460</td>
<td>9.2</td>
</tr>
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</table>
VI. Conclusions

MAX phases are promising reinforcements for interpenetrating phase composites. RMI is a promising method for the near-net shape manufacturing of MAX-phase-reinforced composites. During the fabrication process, both raw material and pore distribution of the preform should be pre-designed to facilitate the infiltration of metal melt, in order to control deformation as well as the phase distribution of the final material. The control of the reaction speed is an important factor during the fabrication process for control of the deformation. More studies on control of the reaction speed and the reduction of residual alloys should be conducted.

References

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