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Ablation-resistance of combustion synthesized
TiB₂-Cu cermet

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**Abstract:** TiB₂-Cu ceramic-metal composite was prepared by the combustion synthesis method from elemental titanium, boron and copper powders. The synthesized product consists of TiB₂ and copper two phases. The microstructure of composite is homogeneous and TiB₂ ceramic particles integrate with each other and skeleton structure forms in the same way as the widely used W/Cu alloy. With the addition of copper metal, the strength, fracture toughness, thermal expansion coefficient and thermal conductivity of TiB₂ are improved. The thermal shock and ablation resistance of the TiB₂-Cu composite was investigated by heating it using a plasma torch arc heater. Fatal breakup took place in the monolithic TiB₂ ceramic once the plasma arc flow faced the surface of the ceramic, whereas, no cracks were found on the ablation surface of the TiB₂-Cu ceramic-metal composite. The fraction of mass loss of the composite was 4.09 %, which was close to that of the traditional W/Cu alloy. The volatilization of the metal binder and mechanical erosion was the main mechanisms of the ablation. A model of ablation processing of the composite was given.

**Key words:** TiB₂-Cu ceramic-metal composite, combustion synthesis, microstructure, thermal shock resistance, ablation mechanism

**I. Introduction**

Copper infiltrated tungsten (Cu/W alloy) is widely used in aircraft propulsion systems and space thermal protection systems because of its high strength at elevated temperature, high resistance to thermal shock and ablation resistance [1]. However, the disadvantage of a Cu/W alloy is the high density (>16g/cm³) that limits its application in some advanced structure, such as rocket motor, where a light weight is preferred [2]. Therefore, development of new composites with light weight has been attracting more attentions.
Ceramic matrix composites are potential candidates for application in thermal structure for their high melting point, light weight, and excellent thermal stability. As well known, the atomic bonding in these materials ranges from purely ionic to totally covalent. On grain boundaries, contiguous grains can maintain their original stoichiometric ratio and orientation, thus forming a ‘hard’ type of bonding of ceramic interface, which may lead to stress concentration and nucleation of microcracks. Also, grain boundaries can become the tunnels of rapid crack propagation. Further improvement on fracture toughness of ceramic matrix composites is much required. Recently, there is growing interest in development of ceramic/metal composites in which ceramic matrices are reinforced and toughed by introducing metallic bonding. Theoretically, incorporation of particulates into a ceramic matrix may improve the resistance to crack propagation via different mechanisms, such as crack blunting, shielding and crack deflection due to particle pull-out. A lot of work on mechanical properties of metallic particle toughened ceramic-matrix composites has been carried out [3,4]. The effects of thermal cycling on mechanical properties have also been investigated [5]. However it has been rarely reported on thermal shock and ablation resistance of ceramic/metal composites under a severe thermal condition.

In the present work, fabrication and microstructure control of TiB₂/Cu ceramic-metal composite were based on the transpiration-cooled mechanism of a W/Cu alloy. TiB₂ ceramic was used to replace tungsten due to its low density (4.52g/cm³), high melting point (3253 K), and high thermal conductivity as compared with most of other ceramics. Previous work [6] showed that no chemical reaction between ceramic and
metal took place and no subsequent brittle phases formed even at high temperature when copper (Cu) binder was added into TiB$_2$ ceramic, not like other metal binders such as Fe, Co or Ni [7]. Therefore, Cu was selected as metal binder for the TiB$_2$ ceramic. Combustion synthesis, also named self-propagating high temperature synthesis (SHS) was used to fabricate the TiB$_2$-Cu composites. SHS process is designed to use the extreme reaction heat generated during the formation of refractory materials. The advantages of SHS technique include low energy requirement, relative simplicity of the process and equipment, higher purity of products and low cost [8]. In this work, the microstructure, thermal shock resistance and ablation behavior and mechanism of the TiB$_2$-Cu composite prepared by SHS were investigated.

II. Experimental Procedure

The precursors used in this work were commercial grade titanium powder (size 40-70 μm, ≈99.5%Ti, 0.3%O, 0.04%H, 0.07%Fe), amorphous boron (≈1 μm, ≈96%B, 1.8%Mg, 1.5%O, 0.24%Fe) and copper powder (70-100 μm, ≈99.7%Cu, 0.015%Fe, 0.05%H2O). The Ti, B, and Cu powders were weighted corresponding to the following reaction:

$$\text{Ti} + 2 \text{B} + 40 \text{wt\% Cu} \rightarrow \text{TiB}_2 + 40 \text{wt\% Cu}$$  \hspace{1cm} (1)

The Ti, B, and Cu powder reactants were dry-mixed in a stainless steel vial. Then, the mixture was cold-pressed un-axially into stainless steel dies to prepare cylindrical green compacts. The diameter of the compact was 55 mm and the pressure used was 33 MPa. The cold-pressed sample was put into a special SHS reactor with the details
illustrated in our previous paper [9]. The synthesis of the TiB₂-Cu composite was conducted through self-propagating exothermic reaction combined with pseudo hot isostatic pressing.

The microstructure of the TiB₂-Cu composite was characterized using scanning electron microscopy (SEM, JSM-5610LV) and transmission electron microscopy (TEM, Philips CM12/ STEM). The flexural strength was measured by a three-point bending test with a crosshead speed of 0.5 mm/min. The dimension of the three-point bend specimen is 30×3×4 mm with a span of a 20mm. Load-displacement curves were recorded by attaching strain gauges to the tensile surfaces of the specimens. Fracture toughness was evaluated using a single-edge notched beam test with a span of a 16 mm and a crosshead speed of 0.05 mm/min using the test bars (2×4×20 mm) with a notch of 2 mm in depth and 0.2 mm in radius. Five specimens are tested for each mechanical property. The density was measured by the Archimedes water-immersion method. Thermal expansion was measured from room temperature to 1000°C with a commercial thermomechanical analyzer. The specimen dimensions are 3×4×10 mm. Thermal diffusivity and specific heat were measured using a laser-flash technique and the specimen size was ø10× (1.5-2) mm and ø8×35 mm, respectively. Three specimens were tested for each thermalphysical properties.

In order to evaluate the resistance to thermal shock and ablation, specimens of TiB₂, TiB₂-20Cu, TiB₂-30Cu, TiB₂-40Cu and W/Cu alloy with 30 mm in diameter and 10 mm in height were prepared using electron-discharge machining followed by surface grinding and polishing. The ablation experiment was conducted in the plasma arc.
torch heater (PCS250, China). The highest temperature of plasma arc was above 5000 °C. The schematic drawing of the plasma arc heater is shown in Fig.1. Nitrogen gas was chosen as the working gas and the ablation was lasted for 20 seconds. The detailed testing conditions are shown in Table 1. The temperature of sample surface was measured using a pyrophotometer with the maximum limit of 3000 K. The mass loss was calculated by weighing the specimen before and after the ablation by using an electric balance with minimum range is 0.1 g. The microstructure and fracture characteristics of the TiB₂/Cu composites before and after ablation were examined using SEM. For the sake of comparison, the ablation behavior of a traditional W/Cu alloy used widely under severe thermal condition was also tested under the same procedure.

III. Results and discussion

3.1 Phase structure and microstructures

During the process of combustion synthesis, titanium, boron and copper may interact to form some possible products following chemical reactions.

\[ Ti + 2B \rightarrow TiB_2 \]  \hspace{1cm} (2)

\[ Ti + B \rightarrow TiB \]  \hspace{1cm} (3)

\[ Ti + Cu \rightarrow TiCu \]  \hspace{1cm} (4)

In order to determine the reaction direction and the possible phases in the synthesized product, reaction free energy of equations (2), (3) and (4) were theoretically calculated according to the thermodynamics data [10]. The calculated reaction free
energy of TiB₂ is the lowest among three possible reactions. In the other words, TiB₂ is the stable phase in the Ti-B-Cu system. The calculated reaction free energy of TiCu is negative but smaller and therefore it is not likely to be formed during reaction. Even if TiCu is formed, it can further react with boron to form TiB₂ phase according to the following reaction.

\[
\text{TiCu} + 2\text{B} \rightarrow \text{TiB}_2 + \text{Cu}
\]  

(5)

Fig.2 shows the X-ray diffraction pattern of TiB₂/Cu ceramic-metal composite. It can be seen that TiB₂ and copper exist but TiCu intermetallic phase was not identified in the products. It is consistent with the thermodynamics calculation. With the propagation of combustion wave, TiCu interphase that may form during the combustion synthesis, will be reacted with boron to produce TiB₂ and copper.

Fig.3 shows the SEM images of the typical microstructure of TiB₂-Cu composite prepared via the SHS method. As shown in Fig.3(a), the microstructure is homogeneous and just a few pores are observed. TiB₂ ceramic particles integrate each other and a skeleton structure forms, as similar as the structure of a W/Cu alloy [2]. The dark binder phases between TiB₂ particles were confirmed by EDX analysis as copper. The relative density of the composite measured using Archimedes method was 96.1 %. At high magnification, the morphology of TiB₂ ceramic is in block-like shape and near equivalent axis-like morphology [Fig.3(b)]. TiB₂ has a C32 hexagonal structure \((a = 3.033 \text{ Å}, \ b = 3.06 \text{ Å} \text{ and } c = 4.56 \text{ Å})\), characterized by the alternate stacking of Ti planes and the graphite-like B network along the \(c\)-direction [11]. The slowest-growing planes in a TiB₂ structure are of the basal \((0001)\) and prism\(\{1\overline{1} 00\}\).
families and the fastest-growing planes are \(\{11\overline{2}0\}\) families, suggesting that TiB\(_2\) should have a plate-like or fairly equiaxed growth morphology. The size of the TiB\(_2\) particles was fine (1-6 \(\mu\)m) because the cooling rate was so rapid that the particles had no time to grow up.

The TEM image of the interface between TiB\(_2\) and Cu is shown in Fig.4. The interface between Cu and TiB\(_2\) is in a good quality, the bonding between them is very strong, and no further chemical reaction between TiB\(_2\) and Cu takes place. Therefore no subsequent brittle phases forms.

### 3.2 Thermal shock resistance

The macro-morphologies before and after ablation for the TiB\(_2\)-Cu composite and W/Cu are shown in Fig.5. For the traditional W/Cu alloy, no cracks are found (Fig.5(a)), but a fatal breakup is associated with the monolithic TiB\(_2\) ceramic when it is heated by the plasma arc (Fig.5(b)). An ablation hole appears on the ablation surface of the TiB\(_2\)-Cu composite as shown in Fig.5(c). No cracks were observed on the ablation surface of the composite in the ablation process and after cooling down from high temperature. It means that the properties of thermal shock resistance and resistance to thermal shock and ablation of the TiB\(_2\)-Cu composite are excellent.

The surface temperature of the composite increased dramatically in a very short time when the sample was heated by the plasma torch arc. It only took about one second to heat up to 3000 K. The large thermal stress could occur because of the great temperature gradient between the surface and the interior of the sample. However, no
cracks were observed on the ablation surface of the composite. The result revealed
that the thermal shock resistance of the composite was good when the temperature
increased dramatically. With the heat transfer towards the inside of the sample, the
temperatures of surface and inside of the specimen tend to identical. The specimen
was eroded and ablated by the high-rate, high-temperature and high-pressure plasma
spray. In this study, the power of the arc heater was immediately turned off after 20
seconds heating and the thermal stress could be very large while the specimen cooled
down. If the thermally induced stresses reached a critical value, cracks could nucleate
at defects and propagate in a direction perpendicular to the maximum normal stress
[12]. After long time ablation, the thermal shock resistance and the maximum tension
stress of a specimen can be decreased inevitably. Crack can be easier to form after the
ablation than before or at the beginning of the ablation. However, no fracture took
place for both the traditional W/Cu alloy and TiB₂-Cu composite.

To further evaluate the thermal stress induced crack initiation and propagation in the
TiB₂/Cu composites, three thermal shock resistance parameters, i.e. thermal stress
fracture resistance parameter (R), thermal stress damage resistance parameter (R IV),
and thermal stress crack stability parameter (R st) were investigated. These parameters
are widely used to evaluate thermal shock resistance of ceramics [13], and are defined
as follows:

\[
R = \frac{\sigma(1-\nu)}{\alpha E} \quad (6)
\]

\[
R_{IV} = \frac{(K_{IC}/\alpha)^2}{(1-\nu)} \quad (7)
\]

\[
R_{st} = \left[\gamma/(\alpha^2E)\right]^{1/2} \quad (8)
\]
where $\sigma$ is the tensile strength of the material, $\nu$ is Poisson’s ratio, $\alpha$ is the thermal expansion coefficient, $E$ is the elastic modulus, $\gamma$ is the fracture surface energy, and $K_{IC}$ is the fracture toughness. $R$ represents a critical temperature difference, $\Delta T_c$, to which a body can be subjected without the initiation of fracture under steady state heat flow or severe transient thermal conditions [13]. $R^{IV}$ indicates the resistance to catastrophic crack propagation of ceramics under a critical temperature difference, $\Delta T_c$ [14]. $R_{st}$ represents the resistance to crack repropagation under the critical temperature difference, $\Delta T_c$.

Table 2 gives the mechanical and thermophysical properties of monolithic TiB$_2$, TiB$_2$/Cu composites and W/Cu alloy [2]. With addition of Cu, the strength, fracture toughness, thermal expansion coefficient and thermal conductivity of TiB$_2$ are increased, but the elastic modulus is decreased. W/Cu alloy has higher fracture toughness and thermal conductivity than that of TiB$_2$/Cu composite. The thermal shock parameters are also calculated and given in Table 3. With the addition of Cu, $R$, $R_{IV}$ and $R_{st}$ of TiB$_2$ are greatly increased by 86, 64 and 143%, respectively. This further confirmed that the addition of Cu has a beneficial effect on the thermal shock resistance of TiB$_2$. R for TiB$_2$/Cu composite is close to the W/Cu alloy. Therefore, both TiB$_2$/Cu and W/Cu have a similar resistance to thermal induced stress fracture. The thermal stress damage parameter ($R^{IV}$) and thermal stress crack stability parameter ($R_{st}$) of TiB$_2$/Cu are much less than that of W/Cu alloy. Further improvement of $R^{IV}$ and $R_{st}$ of the TiB$_2$/Cu composite is necessary.
3.3 Properties and mechanism of ablation resistance

3.3.1 Ablation behaviour

The results of ablation test for TiB$_2$-Cu composite and W/Cu alloy are shown in Table 4. The fraction of mass loss of the composite is 4.09 %, which is slightly greater than that of the traditional W/Cu alloy (2.30 %). Although the mass loss of the TiB$_2$-20Cu and TiB$_2$-30Cu, the fraction of mass loss of the composite was less than that of TiB$_2$-40Cu, cracks appeared in the surfaces of these cermts after thermal shock. It indicates that the ablation resistance of the TiB$_2$-40Cu composite is superior.

As shown in Fig.5(c), there is a deep ablation hole in the TiB$_2$-40Cu sample, while only a few shallow holes in the W/Cu sample. TiB$_2$-40Cu composite was sectioned symmetrically along the axial direction so that the composition from the ablation surface to the bottom of sample could be detected by spectrum analysis. Three locations, i.e., center of the ablation hole, edge of the ablation hole and area between ablation hole and the edge of the sample were selected for the spectrum analysis. The analysis results are shown in Table 5. It can be seen that Cu content decreases from the bottom of the sample to the ablation surface. During the ablation, Cu volatilization in a sample surface may take place due to the high temperature (>3000°C), and Cu in the middle section can be melted. Cu can migrate from inside to the ablation surface via the capillary force because of composition gradient.

The scanning electron micrographs of the ablation surface of the composite are shown in Fig.6. The morphology of the hole center (Fig.6 (a)) is very different from that before ablation (Fig.3). The irregular plate structure was analyzed by energy
dispersive X-ray (EDX). Only Ti element was detected but no Cu was found. Therefore, all the metal binders were volatilized at a temperature of more than 3000°C and only TiB₂ ceramic phase was kept in the hole center. Melting and flowage of ceramic particles were responsible for the change of the morphology.

Fig.6 (b) shows the heavy erosion by the gas flow at the edge of the hole. Sector-like patterns formed due to a strongly scouring of hot gas flow with high temperature and velocity. Based on EDX analysis, the white phase was rich in Ti and O elements as a result of oxidation reaction between TiB₂ and O. The possible oxides are Ti-O compound and B₂O₃. However, no B₂O₃ was found because it is liable to volatilizing due to its low boiling point. It should be mentioned that there was no oxidation reaction of TiB₂ during the ablation as the working heat gas flow was N₂. The oxidation reaction only took place after the ablation due to the exposure of the red-hot TiB₂ in air.

Fig.6(c) shows the abrasive erosion morphology of the interspaces between the edge of the hole and the edge of the sample. Obviously, the erosion was less serious than that at the edge of the hole due to a far distance to the center of the gas flow. The EDX analysis showed the existence of Cu, indicating that the temperature experienced by this area was below the boiling point of Cu.

Fig.6(d) shows the ablation morphology at the specimen edge, which is different from that at other locations (Fig.6 (a)-(c)). The morphology is relatively featureless because it is much far from the plasma arc and not scour by hot gas flow. Also, aggregation and grain growth of ceramic is observed in Fig.6(d). The EDX analysis indicated the
existence of TiB$_2$ and the highest content of Cu in this region due to the relatively less affection by the hot gas flow.

3.3.2 **Ablation mechanisms**

The change of ablation morphology from the center of hole to the edge of specimen indicated that the ablation mechanism was volatilization of metal binder and mechanical erosion. A model of the ablation process of the composite is illustrated in Fig. 7.

In Fig. 7 (a), TiB$_2$ ceramic particles join tightly with Cu metal binder before the ablation. When the plasma arc heats the surface, as shown in Fig. 7 (b), Cu begins to melt and can be volatilized due to the high temperature, and no any changes take place for TiB$_2$ ceramic particles. This is just similar to the transpiration-cooling of a W/Cu alloy, as illustrated in Fig. 8. When the surface of the composite is heated, Cu begins to melt and the heat absorbed by transformation from solid to liquid of Cu can contribute to cooling the skeleton of the composite. Temperature of the composite increases with time and bulk expansion of Cu can take place, and then liquid Cu will assemble and flow to the surface along the capillaries to form surface films which can act as heat insulation layers. Consequently, Cu begins to boil and vaporize due to continuous heating. Then, gas, gas-liquid and liquid multi-layers cooling structure can co-exist in the composite. With continuous heating, gas-liquid interface trends to moving from the surface to inside, Cu will diffuse into the gas layer as liquid or gas state and absorb heat. Cu is continuously volatilized when more heat is replenished by
hot gas flow, and few of TiB₂ particles begin to melt if the temperature is higher than its melting point, shown in Fig.7(c). Fig.7(d) shows that large amount of Cu is volatilized and partly erosive TiB₂ ceramic particles emerge out. The interfaces between TiB₂ particles and Cu binders become flabby and TiB₂ particles lack the protection and fixation of Cu metal binder. As shown in Fig.7(e), no-supported TiB₂ particles are easy to flake away under the combined effect of scouring and shearing induced by the hot gas flow. Finally, the ablation hole is created due to metal volatilization and mechanical erosion.

IV. Conclusions

The quasi-static consolidation in uniaxial compression of combustion synthesized TiB₂-40wt%Cu ceramic-metal composite from Ti, B and Cu powders was investigated. The identified phases of the synthesized product consisted of only TiB₂ and Cu phases without other phases. SEM observation showed that the TiB₂ particles were fine and uniformly distributed in the composite. TiB₂ ceramic particles integrated with each other and the skeleton structure formed in the same way as the W/Cu alloy. With the addition of copper, the strength, fracture toughness, thermal expansion coefficient and thermal conductivity of TiB₂ are increased. Also, the thermal stress fracture resistance parameter, R, the thermal stress damage parameter, R_{IV}, and thermal stress crack stability parameter, R_{st} of TiB₂ are greatly improved with the addition of Cu. The thermal shock resistance and the ablation resistance of TiB₂ ceramic-metal composite were improved with the Cu metal addition. No cracks were found on the ablation
surface of the TiB$_2$-Cu ceramic-metal composite. The fraction of mass loss of the composite was 4.09 %, which was close to that of the traditional W/Cu alloy. The ablation mechanism is believed to be volatilization of the metal binder plus mechanical erosion.

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Reference

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Fig.8 Evaporation mechanism of copper
Table 1 Experimental conditions of plasma ablation

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<table>
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<tr>
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</thead>
<tbody>
<tr>
<td>Electrical current of arc, A</td>
<td>550±10</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Voltage of arc, V</td>
<td>185±5</td>
<td></td>
<td></td>
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<tr>
<td>Power of plasma arc heater, kW</td>
<td>~95</td>
<td></td>
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<tr>
<td>N$_2$ gas Pressure, kPa</td>
<td>490</td>
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<tr>
<td>N$_2$ gas flux, g/s</td>
<td></td>
<td>4.4</td>
<td></td>
</tr>
<tr>
<td>Diameter of nozzle, mm</td>
<td></td>
<td>8</td>
<td></td>
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<tr>
<td>Distance from sample surface to nozzle, mm</td>
<td>10±0.2</td>
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<td>Heat flux density of plasma arc, kW/m$^2$</td>
<td>25121±126</td>
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Table 2 Mechanical and thermo-physical properties of TiB$_2$-Cu composite

<table>
<thead>
<tr>
<th>Materials</th>
<th>TiB$_2$</th>
<th>TiB$_2$-Cu</th>
<th>W/Cu</th>
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<tbody>
<tr>
<td>Density, g/cm$^3$</td>
<td>4.52</td>
<td>5.43</td>
<td>17.2</td>
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<tr>
<td>Flexural strength, MPa</td>
<td>424.8</td>
<td>584</td>
<td>483</td>
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<tr>
<td>Fracture toughness, M Pa·m$^{1/2}$</td>
<td>4.71</td>
<td>8.32</td>
<td>14.72</td>
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<tr>
<td>Elastic modulus, GPa</td>
<td>500</td>
<td>344</td>
<td>312</td>
</tr>
<tr>
<td>Heat conductivity (0-1000°C), W/m·K</td>
<td>24.28</td>
<td>89.5</td>
<td>135</td>
</tr>
<tr>
<td>Thermal expansion coefficient (0-1000°C), ×10$^{-6}$/K</td>
<td>8.1</td>
<td>8.57</td>
<td>5.6</td>
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Table 3 Thermal shock resistance parameters of Cu/TiB₂ composite and W/Cu alloy

<table>
<thead>
<tr>
<th></th>
<th>TiB₂</th>
<th>TiB₂/Cu composite</th>
<th>W/Cu alloy</th>
</tr>
</thead>
<tbody>
<tr>
<td>R = σ(1−ν)/αE, K</td>
<td>106(1−ν)</td>
<td>198(1−ν)</td>
<td>276(1−ν)</td>
</tr>
<tr>
<td>R_{IV} = (K_{IC}/σ)^2/(1−ν), μm</td>
<td>122.9/(1−ν)</td>
<td>201.6/(1−ν)</td>
<td>924.2/(1−ν)</td>
</tr>
<tr>
<td>R_{st} = [γ/α²E]^{1/2}, μm/²K</td>
<td>822.5</td>
<td>1995.6</td>
<td>5954.2</td>
</tr>
</tbody>
</table>

Flexural but not tensile strengths were used to calculate the R and R_{IV}. In the calculation of R_{st}, γ was converted according to the Irwin equation: \( K_{IC}² = 2γE \)

Table 4 Results of ablation tests

<table>
<thead>
<tr>
<th>Materials</th>
<th>Mass before ablation g</th>
<th>Ablation time s</th>
<th>Mass after ablation g</th>
<th>Mass loss g</th>
<th>Fraction of mass loss %</th>
</tr>
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<tbody>
<tr>
<td>TiB₂-20Cu</td>
<td>35.9</td>
<td>20</td>
<td>35.4</td>
<td>0.5</td>
<td>1.39</td>
</tr>
<tr>
<td>TiB₂-30Cu</td>
<td>36.3</td>
<td>15</td>
<td>36.1</td>
<td>0.2</td>
<td>0.55</td>
</tr>
<tr>
<td>TiB₂-40Cu</td>
<td>41.6</td>
<td>20</td>
<td>39.9</td>
<td>1.7</td>
<td>4.09</td>
</tr>
<tr>
<td>W/Cu alloy</td>
<td>117.1</td>
<td>20</td>
<td>114.4</td>
<td>2.7</td>
<td>2.31</td>
</tr>
</tbody>
</table>

Table 5 Quantitative analysis of copper content on cross section after ablation test (wt%)
Fig. 1 Skeleton drawing of plasma arc
Fig. 2 XRD pattern of TiB$_2$-Cu composite
Fig. 3 SEM micrographs of TiB$_2$-Cu composite  
(a) low magnification of TiB$_2$-Cu (b) high magnification of TiB$_2$-Cu composite
Fig. 4 interface between TiB₂ and Cu of TiB₂-Cu composite

1) before ablation of W-Cu  2) after ablation of W-Cu

1) before ablation of TiB₂ ceramic  2) after ablation of TiB₂ ceramic

1) before ablation of TiB₂-40Cu  2) after ablation of TiB₂-40Cu

Fig. 5 Macro-morphologies of the composites pre- and post-test
Fig. 6 Microstructure of TiB$_2$-40Cu composite after ablation

a) center of ablation hole  
b) edge of ablation hole

c) interspace of two edges  
d) edge of the sample
Fig. 7 Formation mechanism of ablation hole with evaporation of metal binder

- a) Cu, TiB₂, Cu
- b) Cu gas phase layer
- c) Cu gas-liquid phase layer
- d) Cu liquid phase layer
- e) Cu matrix
Fig. 8 Evaporation mechanism of copper