



Microstructure and interface thermal stability of C/Mo double-coated SiC fiber reinforced γ -TiAl matrix composites

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Received 5 May 2015; accepted 13 October 2015

Abstract: C/Mo duplex coating interfacially modified SiC fiber-reinforced γ -TiAl matrix composite ($\text{SiC}_f/\text{C}/\text{Mo}/\gamma\text{-TiAl}$) was prepared by foil–fiber–foil method to investigate its interfacial modification effect. $\text{SiC}_f/\text{C}/\text{TiAl}$ composites were also prepared under the same processing condition for comparison. Both kinds of the composites were thermally exposed in vacuum at 800 and 900 °C for different durations in order to study thermal stability of the interfacial zone. With the aids of scanning electron microscope (SEM) and energy dispersive spectrometer (EDS), the interfacial microstructures of the composites were investigated. The results reveal that, although adding the Mo coating, the interfacial reaction product of the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite is the same with that of the $\text{SiC}_f/\text{C}/\text{TiAl}$ composite, which is $\text{TiC}/\text{Ti}_2\text{AlC}$ between the coating and the matrix. However, C/Mo duplex coating is more efficient in hindering interfacial reaction than C single coating at 900 °C and below. In addition, a new layer of interfacial reaction product was found between Ti_2AlC and the matrix after 900 °C, 200 h thermal exposure, which is rich in V and close to the chemical composition of B_2 phase.

Key words: Mo coating; TiAl alloy; SiC fiber; titanium matrix composite; interfacial reaction; thermal stability

1 Introduction

For high-temperature structural materials used in the fields of aeronautics and astronautics, high working temperature, high specific strength, high specific modulus and good oxidation resistance are crucial indices. Concerning these merits above, titanium aluminide intermetallics based on γ -TiAl have drawn growing attention [1]. In order to have better mechanical properties, continuous fiber reinforced γ -TiAl composites have been investigated by many researchers. Different fibers have been studied, such as Mo fiber, Ti fiber, Al_2O_3 fiber, TiNb fiber and CVD-SiC fiber [2–6]. Among them, only CVD-SiC fiber has the highest strength and modulus as well as the lowest density, so SiC fiber-reinforced γ -TiAl matrix composites have been paid much attention to.

However, the TiAl–SiC system is not a chemical equilibrium system, as fiber/matrix interfacial reaction would occur to form reaction products during the

fabrication and utilization of the composites at elevated temperatures. It was reported that the interfacial reaction layer of $\text{SiC}_f/\text{Ti-43Al-9V}$ composite is $\text{TiC}/\text{Ti}_2\text{AlC}/(\text{Ti,V})_5(\text{Si,Al})_3$ [7] and that of $\text{SiC}_f/\text{Ti-48Al}$ composite is $\text{TiC}/\text{Ti}_2\text{AlC}+\text{Ti}_3\text{AlC}/(\text{Ti,V})_5(\text{Si,Al})_3$ [8]. Generally, these brittle reaction products are detrimental to mechanical properties of the composites. One of the effective approaches of slowing down interfacial reaction is the application of a barrier coating. For example, C coating, B_4C coating and Gd/Gd B_x duplex coating have been utilized [9,10]. Most of these coatings show good performance on retarding interfacial reaction to some extent. However, for a coating, it is not sufficient to consider only its ability to reduce interfacial reaction, moreover, its thermal expansion compatibility is also very important.

In this work, C/Mo duplex coating was chosen to modify the interfacial zone of the SiC fiber reinforced γ -TiAl matrix composite ($\text{SiC}_f/\gamma\text{-TiAl}$). The idea is based on the following five considerations: 1) Coefficients of thermal expansion of the two coatings (C: $10\times 10^{-6} \text{ }^\circ\text{C}^{-1}$,

Foundation item: Projects (51201134, 51271147) supported by the National Natural Science Foundation of China; Project (2015JM5181) supported by the Natural Science Foundation of Shaanxi Province, China; Project (115-QP-2014) supported by the Research Fund of the State Key Laboratory of Solidification Processing (NWPU), China; Project (3102014JCQ01023) supported by the Fundamental Research Funds for the Central Universities, China

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DOI: 10.1016/S1003-6326(16)64198-5

Mo: $(5.8\text{--}6.2)\times 10^{-6} \text{ }^{\circ}\text{C}^{-1}$) [11] coordinate with that of SiC fiber ($4.5\times 10^{-6} \text{ }^{\circ}\text{C}^{-1}$) and the γ -TiAl matrix ($(12\text{--}14)\times 10^{-6} \text{ }^{\circ}\text{C}^{-1}$) [12,13]; 2) Mo coating is a ductile coating, and can relieve thermal residual stresses; 3) Both of the coatings can reduce the interfacial reaction to some extent [14–17]; 4) Mo is a β -stabilizer of titanium alloys. Once Mo atoms diffuse into TiAl matrix, the ductility of the matrix can be improved [18]; 5) C and Mo display good chemical stability with each other [19]. Our previous study of SiC_f/C/Mo/Ti6Al4V composite also indicates that the two coatings can exist harmoniously below 900 °C [16].

In order to evaluate the modification effect of C/Mo duplex coating on SiC fiber-reinforced γ -TiAl matrix composite, single C coating with the same thickness was used for comparison. The interfacial microstructures and thermal stability of the SiC_f/C/TiAl and SiC_f/C/Mo/TiAl composites were comparatively studied under the same conditions.

2 Experimental

Ti-43Al-9V foils were utilized as matrix of the composites, which mainly contain γ phase as well as some α_2 and B_2 phases. The initial microstructure of the foils was near lamellar structure [20]. The reinforcement SiC fiber (100–120 μm in diameter) applied in this work was fabricated in China by chemical vapor deposition (CVD). The fiber has a tungsten core with a diameter of about 12 μm . As shown in Fig. 1, two kinds of fiber coatings were prepared: 3.3 μm -thick C coating, and 2 μm -thick C coating + 1.3 μm -thick Mo coating, respectively. C coating was deposited by chemical vapor deposition (CVD) while Mo coating was deposited by magnetron sputtering. The composites were fabricated by foil-fiber-foil (FFF) method plus vacuum hot-pressing (VHP) under 1150 °C, 150 MPa, 40 min.

After the preparation of the composites, part of the composites were cut into small pieces for thermal exposure treatment in vacuum. The vacuum thermal exposure conditions include (800 °C, 100 h), (800 °C, 200 h), (900 °C, 100 h), (900 °C, 200 h) and (900 °C, 220 h), respectively. After that, metallographic specimens were prepared by conventional preparation methods of metallographic samples. Microstructures and chemical composition of the as-prepared and thermally-exposed composites were then studied by an SUPRA 55 SEM and an Oxford INCA EDS, respectively.

3 Results and discussion

3.1 Interfacial microstructure of as-prepared composites

Figures 2(a) and (b) show lower magnification

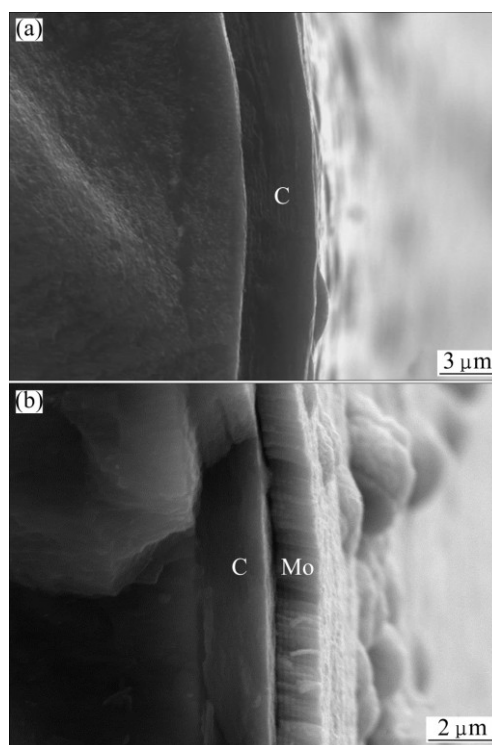


Fig. 1 Coating on surface of SiC fiber: (a) C single coating; (b) C/Mo duplex coating

micrographs of the as-prepared SiC_f/C/TiAl and SiC_f/C/Mo/TiAl composites, respectively. The black layer around the SiC fiber is C coating. The white layer in Fig. 2(b) next to the C coating is Mo coating. It can be seen that, although the two composites have the same total coating thicknesses (3.3 μm) before processing, C/Mo duplex coating performs better on blocking the interfacial reaction than C single coating. This good result is reflected on the thickness of interfacial reaction zone. Figures 2(c) and (d) are higher magnification micrographs of the interfacial zones of the two composites. One can further see that the interfacial reaction layer in the SiC_f/C/TiAl composite is distinctly thicker than that in the SiC_f/C/Mo/TiAl composite.

The interfacial microstructure of SiC_f/C/Ti-43Al-9V composite fabricated by FFF method plus VHP under 1150 °C, 150 MPa, 40 min has been investigated in Refs. [7] and [21]. Their results reveal that the interfacial phase sequence is SiC/C coating/TiC/Ti₂AlC/ γ -TiAl from fiber side to the matrix. The only difference between ZHANG's study [7,21] and this work is the thickness of C coating, which are 1 μm and 3.3 μm , respectively. The thicker the C coating is, the longer duration it can block the chemical reaction between SiC and the matrix. So, the interfacial phase sequence of the as-prepared SiC_f/C/TiAl composite in this work should be also C coating/TiC/Ti₂AlC.

In order to study the element distribution

characteristics and interfacial phase sequence of the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite, EDS line analysis as well as backscattered electron micrograph (SEM-BSE) was carried out in the interfacial zone of the as-prepared composite. The analyzed positions and corresponding results are shown in Fig. 3. According to the distribution and ratio of the elements, the interfacial zone can be divided into four layers, which are marked as 1 to 4 from fiber to the matrix in Fig. 3(b). Layers 1 and 2 are C coating and Mo coating with average thicknesses of $1.77\ \mu\text{m}$ and $1.20\ \mu\text{m}$, respectively. Layers 3 and 4 are interfacial reaction zones, which mainly contain Ti, Al, V and C, almost without Si and Mo elements. In layer 3,

the mole ratio of Ti:C is close to 1:1, so it can be inferred that layer 3 should be TiC. As for layer 4, it consists of Ti, Al, C and V. According to the ratio of the elements and the phase equilibrium between TiC and $\gamma\text{-TiAl}$ [22], the layer 4 may be Ti_2AlC . Moreover, the difference between the $\text{SiC}_f/\text{C}/\text{TiAl}$ and the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composites is the existence of Mo coating. But according to the results above, Mo atoms hardly diffuse into the interfacial reaction layer. So, it is reasonable to believe that the interfacial product sequence of the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite is $\text{TiC}/\text{Ti}_2\text{AlC}$ from Mo coating side to the matrix, which is the same with that of the $\text{SiC}_f/\text{C}/\text{TiAl}$ composite.

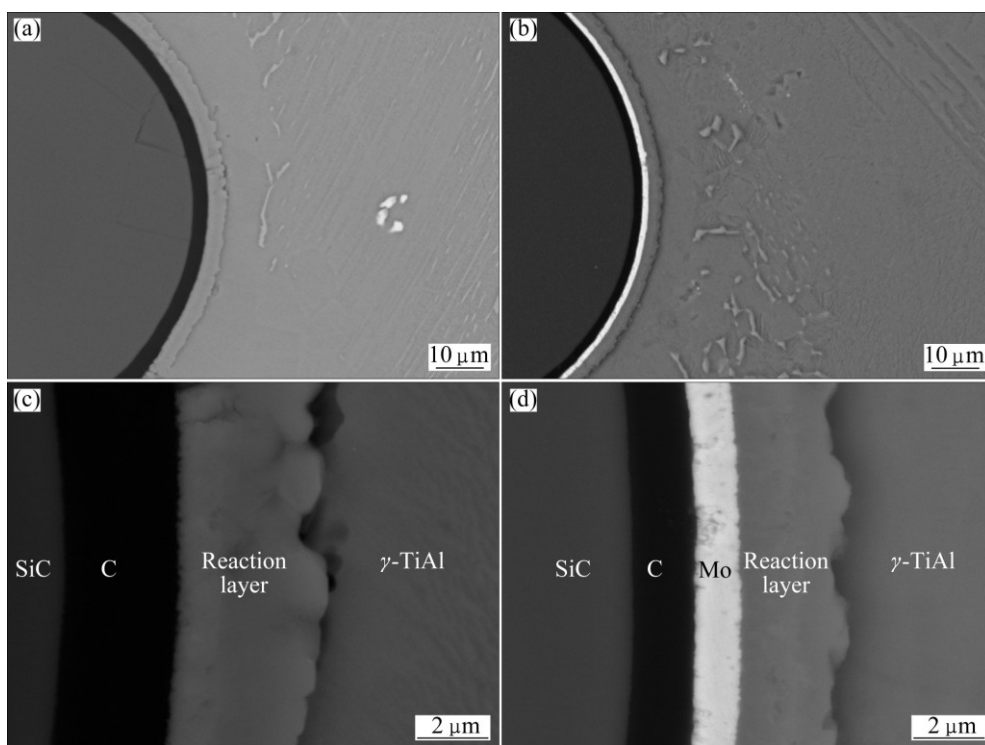


Fig. 2 Lower magnification images (a, b) and higher magnification images (c, d) of interface of as-prepared $\text{SiC}_f/\text{C}/\text{TiAl}$ (a, c) and $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ (b, d) composites

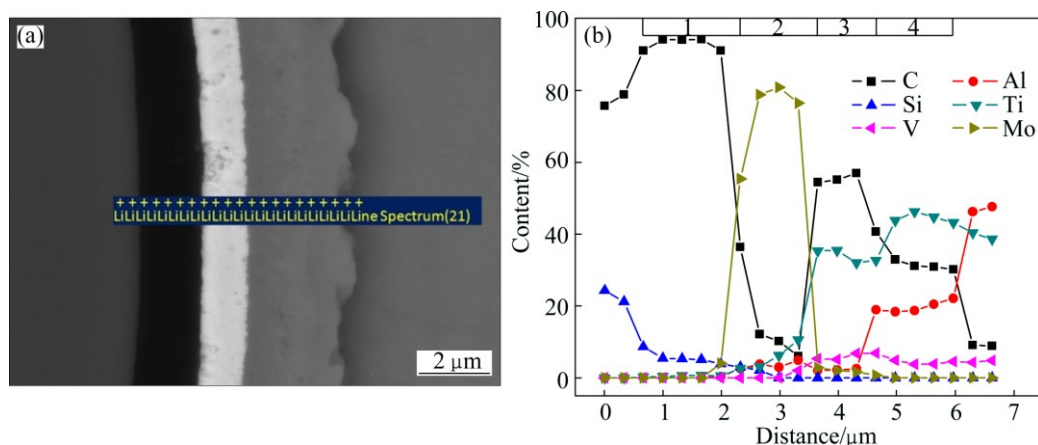


Fig. 3 Position of EDS line analysis on interface of as-prepared $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite (a) and element distribution curves along interfacial zone (b)

The reaction $\text{Ti}+\text{C}\rightarrow\text{TiC}$ is likely to happen first in TiAl–SiC system, which has been proved by many studies [7,9]. The initially formed TiC layer has characteristic of fine grains due to a high nucleation rate. With the concentration decrease of reagent, the nucleation rate also decreases [23]. Then, coarse TiC grain layer forms. Since the diffusion velocity of C atoms is higher than those of other atoms, the growth of interfacial reaction layer is mainly based on depletion of the matrix [17]. Al atoms do not join the first-step reaction, so the matrix next to TiC is Al-rich TiAl. Some studies imply that the enrichment of Al in TiAl matrix is beneficial for the formation of Ti–Al–C ternary compounds in order to reach the phase equilibrium [24]. Moreover, based on Ti–Al–C ternary phase diagram [22], Ti_2AlC and Ti_3AlC are the equilibrium phases for $\text{TiC}/\gamma\text{-TiAl}$ and $\text{TiC}/\alpha_2\text{-Ti}_3\text{Al}$, respectively. Thus, Ti_2AlC layer forms.

V and Mo atoms do not participate in the formation of TiC and Ti_2AlC . Here, it is necessary to interpret the behaviors of V and Mo atoms during the interfacial reaction. As we know, VC and TiC are both of NaCl structure type (FCC), whose lattice constants are 0.4159 and 0.4329 nm, respectively. So, there is no doubt that V can substitute Ti to some degree. Moreover, TiC always has some Ti vacancies, which is also beneficial to the dissolution of V into TiC. In addition, Ti_2AlC and V_2AlC are also of the same Cr_2AlC structure type (HCP), and the lattice constants are $a=0.3056$ nm, $c=1.3623$ nm and

$a=0.2909$ nm, $c=1.3127$ nm, respectively [25]. So the same goes for V in Ti_2AlC . As shown in Fig. 3(b) above, there exist some V in layer 3 (TiC layer) and layer 4 (Ti_2AlC layer).

Figure 3(b) shows that Mo element does not diffuse into the interfacial reaction layer or into the TiAl matrix. While in our previous study of the SiC/C/Mo/Ti6Al4V composite, Mo coating had an affected zone of about 7.9 μm in thickness [16]. The difference between the two composites can be explained in the light of different crystal structures. The Ti–6Al–4V matrix is solid solution with typical metallic bond. Mo element is an isomorphous one of $\beta\text{-Ti}$. So, although Mo atoms do not participate in the interfacial reaction, they can diffuse and dissolve into the matrix largely. While $\gamma\text{-TiAl}$ matrix is intermetallics, whose bonding characteristics is a mixture of covalent bond and metallic bond. In addition, the diffusion of Mo atoms in TiAl is via the substitution of Ti [26], which is hard to occur. As a result, Mo atoms hardly diffuse and dissolve into the $\gamma\text{-TiAl}$ matrix.

3.2 Composites after vacuum thermal exposure

In order to study thermal stability of the interface layers, vacuum thermal exposure treatments were carried out to the two kinds of composites. Figure 4 shows SEM-BSE photographs of the SiC_f/C/TiAl and SiC_f/C/Mo/TiAl composites after 800 °C, 100 h and 800 °C, 200 h thermal exposure, respectively. Compared with the as-prepared composites, no obvious change occurs

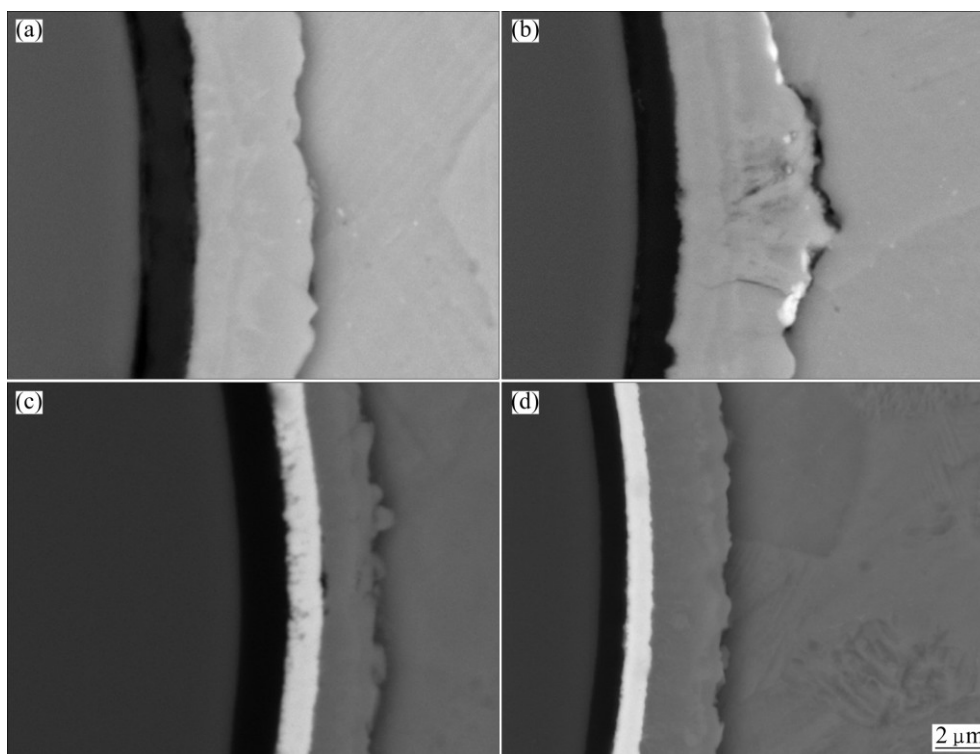


Fig. 4 Interfacial reaction zones of SiC_f/C/TiAl (a, b) and SiC_f/C/Mo/TiAl (c, d) composites: (a, c) After 800 °C, 100 h thermal exposure; (b, d) After 800 °C, 200 h thermal exposure

other than the thickening of interfacial reaction layers and the consumption of coatings. But it can be seen that the thickening of the reaction layer in the $\text{SiC}_f/\text{C}/\text{TiAl}$ composite is faster than that in the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite. The similarity and dissimilarity between the two composites indicate a high stability of C single coating and C/Mo duplex coatings at 800 °C, and C/Mo duplex coating is more efficient than C single coating.

As the consumption of Ti atoms is larger than that of Al atoms in the interfacial reaction, the matrix adjacent to the interfacial reaction layer is Al-rich γ -TiAl in the as-prepared composites [21]. Figure 5 and Table 1 show the EDS composition analysis locations and corresponding results in the $\text{SiC}_f/\text{C}/\text{TiAl}$ and $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composites after 800 °C, 200 h thermal exposure, respectively. It is seen that the matrix adjacent to the interfacial reaction layer of the $\text{SiC}_f/\text{C}/\text{TiAl}$ composite is still rich in Al, while in the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composites, this area has nearly reached homogenous composition with the rest of the matrix.

The matrix adjacent to the interfacial reaction layer is influenced by two processes. One is interfacial reaction, the other is annealing treatment. During the vacuum thermal exposure, the interfacial reaction continues, so the accumulation of Al atoms goes on. Meanwhile, this area also belongs to the matrix. Vacuum thermal exposure to the matrix means a composition homogenization process. If the accumulation velocity of Al atoms exceeds the velocity of composition

Table 1 EDS point analysis results of matrices corresponding to Fig. 5 (mole fraction, %)

Point No.	Corresponding to Fig. 5(a)			Corresponding to Fig. 5(b)		
	Al	Ti	V	Al	Ti	V
Spectrum 1	52.16	39.53	8.31	48.47	43.94	7.59
Spectrum 2	51.96	39.38	8.66	48.58	43.18	8.24
Spectrum 3	52.37	38.95	8.68	49.20	42.68	8.12
Spectrum 4	51.95	39.89	8.16	48.87	43.15	7.98
Spectrum 5	52.24	38.89	8.87	47.66	44.06	8.28
Spectrum 6	45.75	45.82	8.43	46.16	45.15	8.69
Spectrum 7	46.38	45.39	8.23	47.47	45.09	7.44
Spectrum 8	46.42	45.84	7.74	46.67	44.94	8.39
Spectrum 9	45.79	45.67	8.54	47.68	43.91	8.41
Spectrum 10	46.65	45.06	8.29	46.86	45.08	8.06

homogenization process, this area remains rich in Al. If a balance has been achieved between the two velocities, this area would become homogeneous with the rest of the matrix. So, the difference between the two composites indicates that the accumulation velocity of Al atoms in the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composites is lower, which proves that C/Mo duplex coating is more efficient to hinder the interfacial reaction. This result is consistent with the conclusion we have got by comparing the thicknesses of the interfacial reaction layers in the as-prepared composites.

Figure 6 shows the interfacial zones of the $\text{SiC}_f/\text{C}/\text{TiAl}$ and $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composites after 900 °C, 100 h and 900 °C, 200 h thermal exposure, respectively. Just like the situations at 800 °C, the two interfacial layers show good stability, while C/Mo duplex coating still performs better. Especially after 900 °C, 200 h thermal exposure, the C coating in the $\text{SiC}_f/\text{C}/\text{TiAl}$ composite has reduced from initial thickness of 3.3 μm to less than 1 μm , while the C coating in the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite retains a thickness of 1.3 μm , whose primary thickness is just 2 μm .

In addition, it can be seen obviously from Figs. 6(b) and (e) that a new reaction layer appears in both of the composites after 900 °C, 200 h thermal exposure. The difference is that, in the $\text{SiC}_f/\text{C}/\text{TiAl}$ composite, the new layer is approximately circle shape (shown in Fig. 6(c)), while in the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composites, it shows a sawtooth profile (shown in Fig. 6(f)). In order to verify the new phase, EDS line analysis as well as SEM-BSE imaging was carried out in the interfacial zone, as shown in Fig. 7. In comparison with the as-prepared composites, the layers 3 and 4 are still TiC and Ti_2AlC , respectively, which are speculated by the EDS results. The diffusion of Mo atoms is still very limited, which can be proved by the EDS results in Fig. 7(d) and Spectrum 1 in Table 2.

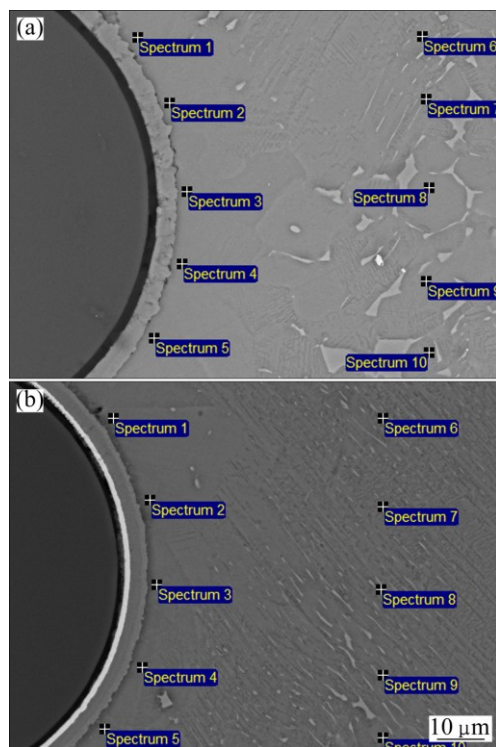


Fig. 5 EDS analysis position of matrix in composites after 800 °C, 200 h thermal exposure: (a) $\text{SiC}_f/\text{C}/\text{TiAl}$; (b) $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$

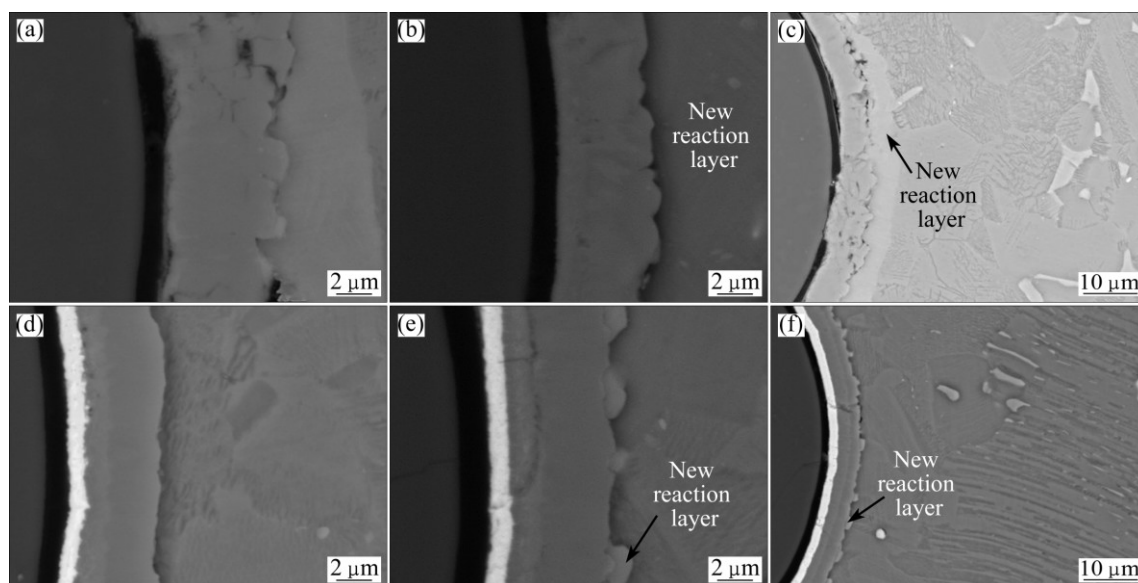


Fig. 6 Interfacial reaction zones of $\text{SiC}_f/\text{C}/\text{TiAl}$ (a, b, c) and $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composites (d, e, f): (a, d) After 900 °C, 100 h thermal exposure; (b, e) After 900 °C, 200 h thermal exposure; (c, f) Low magnification images showing interfacial reaction zone of $\text{SiC}_f/\text{C}/\text{TiAl}$ and $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composites after 900 °C, 200 h thermal exposure, respectively

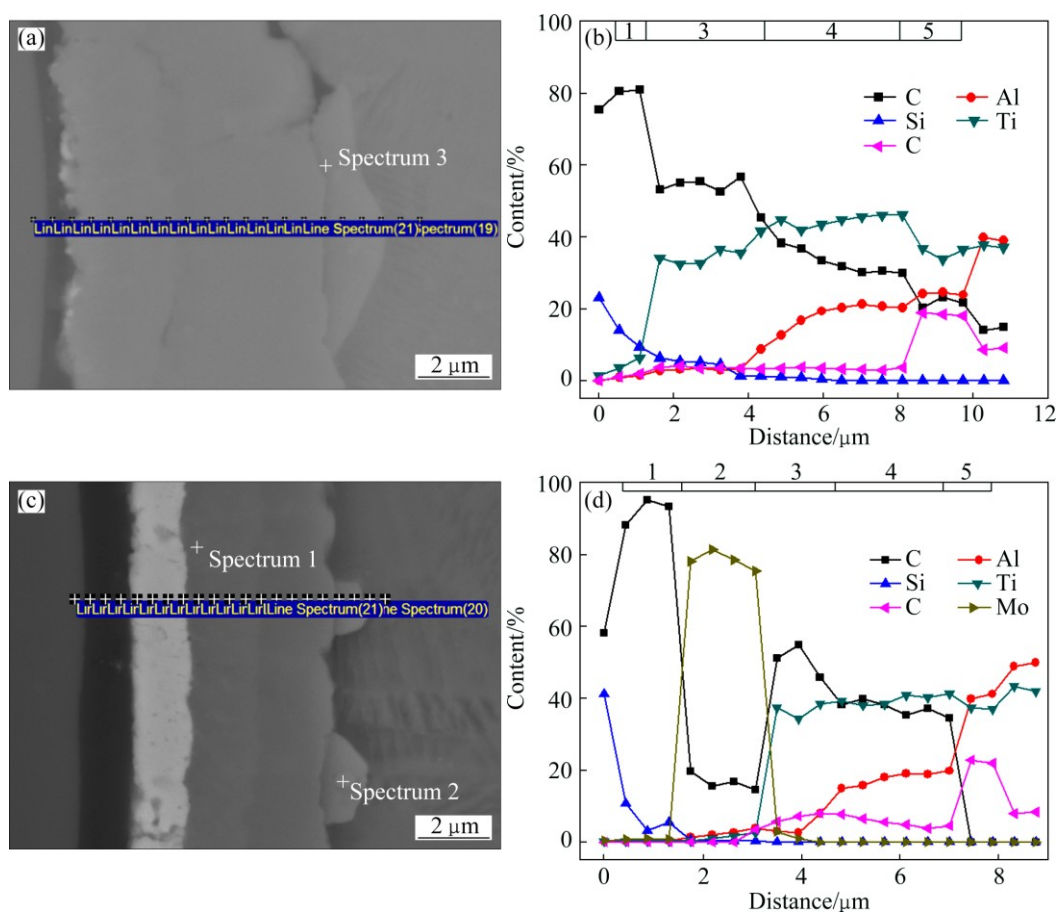


Fig. 7 Interfacial zones of $\text{SiC}_f/\text{C}/\text{TiAl}$ (a) and $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composites (b) after 900 °C, 200 h thermal exposure and element line analysis results (b, d) corresponding to (a) and (c), respectively

Whether in the $\text{SiC}_f/\text{C}/\text{TiAl}$ composite or in the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite, the newly formed layer (hereafter denoted as layer 5) is rich in V. Layer 5, which

is circle-shaped in the $\text{SiC}_f/\text{C}/\text{TiAl}$ composite, has a V content of about 20%. Similarly, layer 5 with the sawtooth profile in the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite

contains approximately 22% V, as listed in Table 2. Besides, Fig. 8 indicates that there exists a concentration gradient of V between the sawtooth-shaped layer 5 and the matrix nearby along the hoop direction. So, one can deduce that the final morphology of layer 5 in $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite is also circular after the diffusion of V atoms. In order to verify this presumption, a longer vacuum thermal exposure was carried out.

Table 2 EDS point analysis results of interfacial zone shown in Figs. 7(a) and (c) (mole fraction, %)

Point	Ti	Al	V	C	Mo
Spectrum 1	37.31	2.95	5.70	51.18	2.86
Spectrum 2	37.16	40.50	22.34	–	–
Spectrum 3	44.99	35.26	19.75	–	–

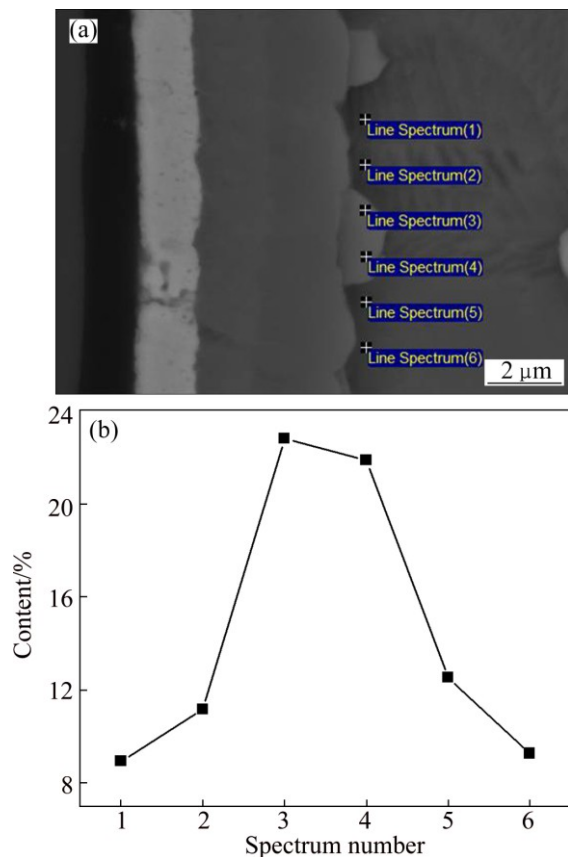


Fig. 8 EDS analysis positions of V-rich layer in composite (a) and corresponding distribution curve of V element (b)

Figure 9 shows the morphology of the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite after 900 °C, 220 h vacuum thermal exposure. It can be distinctly seen that the layer 5 has become circular by the diffusion of V atoms along the hoop direction. This fact verifies the presumption mentioned above. As shown by spectra 1 and 2 in Table 3, the V content of layer 5 does not decrease obviously accompanying with the growth of layer 5, which infers that the matrix or interfacial layer may

provide V element during the growth course. Just like that the interfacial reaction in the $\text{SiC}_f/\text{C}/\text{TiAl}$ composite is severer than that in the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite, layer 5 in the $\text{SiC}_f/\text{C}/\text{TiAl}$ composite appears earlier and grows bigger (see Fig. 6(c) and Fig. 9) than in the $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite. So, the interfacial reaction layer has a greater influence on the formation and growth of layer 5.

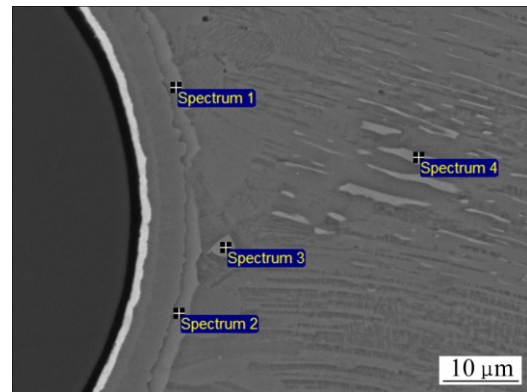


Fig. 9 Interfacial reaction zone of $\text{SiC}_f/\text{C}/\text{Mo}/\text{TiAl}$ composite after 900 °C, 220 h thermal exposure

Table 3 EDS point analysis results of positions shown in Fig. 9 (mole fraction, %)

Point	Ti	Al	V
Spectrum 1	42.86	35.82	21.32
Spectrum 2	41.91	37.04	21.05
Spectrum 3	36.70	39.85	23.45
Spectrum 4	35.12	40.48	24.40

As B_2 phase in the matrix is also rich in V, it is necessary to make a componential comparison between layer 5 and B_2 phase. As shown in Table 3, the V content of layer 5 approximates to that of B_2 phase (Spectra 3 and 4), only the Ti and Al contents having small differences. So, layer 5 should be B_2 phase. The formation process of layer 5 should be related to the precipitation of V in $(\text{Ti},\text{V})\text{C}$ and $(\text{Ti},\text{V})_2\text{AlC}$.

3.3 Interfacial reaction kinetics

As brittle interfacial reaction layer may become potential position for the initiation and growth of cracks which could lower mechanical properties of composites, it is necessary to investigate the growth kinetics of interfacial reaction zone. Figure 10 shows the thicknesses of interfacial reaction layers vs the square roots of vacuum thermal exposure durations under different temperatures. It can be seen that there exists a linear relationship between the two parameters, which suggests that the interfacial reaction is a diffusion-controlled process [27].

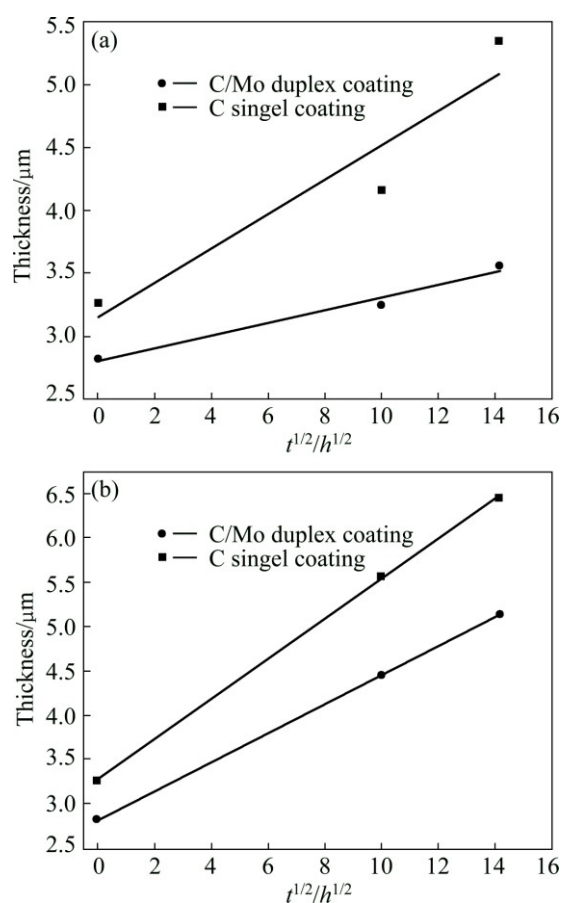


Fig. 10 Interfacial reaction zone thickness as function of square root of time: (a) At 800 °C; (b) At 900 °C

Therefore, the reaction kinetics can be described by the following equation:

$$X = kt^{1/2} + b_0 \quad (1)$$

where X is the thickness of the reaction layer, t is duration, k is the reaction rate constant, and b_0 is the original thickness of the reaction layer.

The reaction rate constants of SiC_f/C/TiAl composite are 0.137 at 800 °C and 0.227 at 900 °C, respectively, while those of the SiC_f/C/Mo/TiAl composite are 0.0506 at 800 °C and 0.164 at 900 °C, respectively. The result shows again that C/Mo duplex coating is more efficient in hindering the interfacial reaction than C single coating.

4 Conclusions

1) The interfacial reaction products of the as-prepared SiC_f/C/Mo/TiAl composite are TiC/Ti₂AlC from Mo coating side to the matrix, which is the same as that of the as-prepared SiC_f/C/TiAl composite. However, the interfacial reaction layer thickness of the former is thinner than that of the latter, which indicates C/Mo duplex coating is more efficient in hindering the

interfacial reaction than C single coating.

2) Mo atoms hardly diffuse and dissolve into the γ -TiAl matrix, either in the as-prepared or in the thermally-exposed composites. As the bond characteristic of TiAl intermetallics is a mixture of covalent bond and metallic bond, the substitution of Ti atoms with Mo atoms is difficult to occur.

3) Vacuum thermal exposure of the SiC_f/C/TiAl and SiC_f/C/Mo/TiAl composites shows that C/Mo duplex coating has a better thermal stability at elevated temperatures up to 900 °C. Meanwhile, with the protection of Mo coating, C coating consumes slower. After the thermal exposure of 900 °C, 200 h, a V-rich product appears, which might be B₂ phase.

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C/Mo 双涂层 SiC 纤维增强 γ -TiAl 基复合材料的显微组织及界面热稳定性

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摘 要: 采用箔-纤维-箔法制备了 C/Mo 双涂层界面改性 SiC 纤维增强 γ -TiAl 基复合材料(SiC/C/Mo/ γ -TiAl), 并研究其界面改性效果。为了对比研究, 在相同工艺下制备了 SiC/C/ γ -TiAl 复合材料。对两种复合材料在 800 °C 和 900 °C 下进行了不同时间的真空热暴露处理, 用以研究界面区域的热稳定性。采用扫描电子显微镜和能谱仪分析了复合材料界面的显微组织。结果表明, 尽管增加了 Mo 涂层, 但是 SiC/C/Mo/ γ -TiAl 复合材料与 SiC/C/ γ -TiAl 复合材料的界面反应产物相一致, 均为介于涂层和基体之间的 TiC/Ti₂AlC。但是, 在 900 °C 及其以下温度, C/Mo 双涂层比 C 单涂层能更好地阻碍界面反应。此外, 900 °C, 200 h 热暴露后, 在 Ti₂AlC 和基体之间发现了一层新的界面反应产物, 该产物富 V, 与 B₂ 相的化学成分接近。

关键词: Mo 涂层; TiAl 合金; SiC 纤维; 钛基复合材料; 界面反应; 热稳定性

(Edited by Yun-bin HE)