Tensile properties and thermal expansion behaviors of continuous molybdenum fiber reinforced aluminum matrix composites

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Abstract

The tensile properties and thermal expansion behaviors of continuous molybdenum fiber reinforced aluminum matrix composites (Mo/Al) have been studied. The Mo/Al composites containing different volume percents of Mo fibers were processed by diffusion bonding. The strengths of unidirectional Mo/Al composites were close to the rule-of-mixtures. The strengths of 0°/90° dual-directional composites increased with fiber content, while those of 45°/135° composites remained relatively low. The coefficients of thermal expansion (CTEs) of the composites decreased as the fiber content increased, close to the values of Mo fibers. With increasing temperature, the CTEs of unidirectional composites increased, while those of dual-directional composites decreased due to large accumulated thermal stresses. The CTEs of 45°/135° composites were lower than those of 0°/90° composites because of contraction effect. At temperatures above 250 °C, the CTEs of the dual-directional composites gradually increased due to matrix yielding and interfacial decohesion.

Keywords: A. Metal matrix composites; B. Mechanical properties; B. Thermomechanical properties

1. Introduction

Since the 1960s, metal matrix composites (MMCs) of high strength, high stiffness, low density, good elevated-temperature stability, and good wear resistance have been widely developed [1–3]. Among them, continuous fiber reinforced MMCs combining the low coefficients of thermal expansion (CTEs) of the fibrous reinforcements and the high thermal conductivities of the metallic matrices have attracted much attention for the application to aerospace industry [3–8]. Especially, continuous metallic fibers, like molybdenum (Mo), stainless steel, and tungsten, etc., have the advantages of high strength, good ductility, large flexibility, and thus easiness to weave [3,9–13]. The most successful and reliable process for the fabrication of continuous metallic fiber reinforced MMCs is diffusion bonding of alternative layers of wound fibers and matrix metal foils [3,10–13]. The process yields a higher fabrication rate, better controllability, higher repeatability, and better product quality than conventional powder metallurgy or melt-casting, and is thus also applied in this study to produce continuous Mo fiber reinforced aluminum (Al) matrix composites.

After the fabrication and heat treatment processes of MMCs, large residual thermal stresses are formed due to the differences in the CTEs of fibers and matrices [3,14–16]. The residual stress, \( \sigma_0 \), can be generally estimated from the Young’s modulus of the matrix, \( E_m \), the CTEs of the matrix and the fiber, \( \alpha_m \) and \( \alpha_f \), respectively, and the temperature difference, \( \Delta T \), using the following equation [3,14–16]:

\[
\sigma_0 = \frac{E_m(\alpha_m - \alpha_f)\Delta T}{1 + v_m},
\]

where \( v_m \) is the Poisson ratio of the matrix. The residual stresses affect the properties of MMCs, including the mechanical strengths and strains, electrical and thermal...
conductivities, and even subsequent thermal expansion behaviors [3,6,14–17]. For the applications of MMCs to aerospace and electronic packaging industries, the thermal expansion behaviors of the MMCs are an essentially important subject and need to be clarified. Especially, Al matrix composites reinforced by continuous Mo fibers (Mo/Al) are expected to possess a high strength, low density, and low thermal expansion coefficient, and thus have a high potential for the application to aircraft industry [3]. Therefore, this work reported herein focuses on the investigation of the thermal expansion behaviors of Mo/Al composites, accompanied with the studies on the microstructure characterization and mechanical properties of the composites.

Among the studies of the thermal expansion behaviors of MMCs, those on unidirectional fiber reinforced MMCs have been more intensively performed [3,14–20], while few regarding dual-directional fiber strengthened MMCs have been reported [3,6,15,16]. However, isotropic thermal properties of the MMCs to be obtained by the addition of dual-directional fibers are rather necessary for the applications of the MMCs. More complicated thermal expansion behaviors of the MMCs reinforced by dual-directional fibers attributed to the complex deformation constraints from different directions need to be clarified. Therefore in this study, the thermal expansion behaviors of Al matrix composites reinforced by dual-directional Mo fibers are also emphasized.

2. Experimental procedures

Pure Al foils (AA 1100) of 100 μm thick and Mo fibers with a diameter of 150 μm were selected as the raw materials to produce Mo/Al composites. The Al foils were firstly placed on a framework of an automatic winding machine, and the Mo fibers were then wound on the Al foils. The spacing between the fibers was controlled by the rotation speed of the framework and the movement of the fibers. Alternate layers of the properly spaced Mo fibers and Al foils were stacked and then diffusion bonded at 500 °C, 100 MPa in a vacuum of 10⁻² Torr for 10 min to obtain the Mo/Al composites. The volume percentages of the Mo fibers in the composites were adjusted by changing the spacing of the Mo fibers and the number of Al foils between each stacking layer of fibers. For unidirectional Mo/Al composites, the winding direction of Mo fibers was fixed the same; while for dual-directional Mo/Al composites, the winding direction was changed for 90° between each layer of the fibers. The microstructures of the Mo/Al composites were examined using an optical microscope (OM) and a scanning electron microscope (SEM, JEOL JSM-5410). The chemical compositions of the composites at the interface between Mo fibers and Al matrix were analyzed by an electron probe X-ray microanalyzer (EPMA, JEOL JXA-8800M).

For the measurements of the tensile properties and thermal expansion behaviors of both unidirectional and dual-directional Mo/Al composites, specimens of different directions (0°/90° parallel or perpendicular to fibers and 45°/135° oblique to fibers, as schematically defined in Fig. 1) were cut from the diffusion-bonded composites. The longitudinal direction (L) was defined as along the long axis of the specimens and the transverse direction (T) along the short axis. Tensile tests of the Mo/Al composites were conducted using an Instron testing machine. Before tests, the composite specimens were machined to a gauge size of 3 mm in width, 1 mm in thickness, and 25 mm in length, and their surfaces were polished to 600 mesh. The fracture surfaces of the composites after tensile tests were observed by SEM. For investigating the thermal expansion behaviors, the Mo/Al composite specimens were cut and polished to a size of about 10 × 5 mm². Seiko SS/5200 Thermal Analysis-320 TMA was used to measure the thermal expansions of the specimens from room temperature to 550 °C at a heating rate of 10 °C/min for three thermal cycles. During measurements, a small load of 10 g was applied onto the specimens, and the sample chamber was purged with nitrogen. The CTEs of the composites were calculated by differentiating the measured thermal expansions of the specimens. For comparison, the tensile properties and thermal expansion behaviors of the raw materials, i.e., as-received Mo fibers and the Al plates that were obtained by stacking and hot pressing several 1100 Al foils under the same diffusion bonding conditions, were also measured.

3. Results and discussion

3.1. Microstructure and interface characterizations of Mo/Al composites

Fig. 2 shows the microstructures of dual-directional Mo/Al composites reinforced with different contents of Mo fibers. Good interface joining between the Mo fibers and the Al foils and between two Al foils was successfully achieved by diffusion bonding without the existence of any pores or seams caused by insufficient flow of the Al matrix. Similar performance was also obtained for the Mo/Al composites reinforced by unidirectional fibers (pictures
not shown here). However, some of the distribution of the Mo fibers was not as uniform as expected, attributed to the local bending of the as-received Mo fibers and the movement of some Mo fibers caused by the flow of Al matrix during hot pressing. Furthermore, in some composites containing low volume percentages of Mo fibers (5 and 15 vol%), fiber necking and even fracture were found during hot pressing as shown in Fig. 2(d). The plastic flow of the Al foils under the compressive stresses of hot pressing induced large shear stresses acting on the fibers especially in the composites with low fiber contents and much easily resulting in the necking and fracture of the fibers. For those fibers around which other fibers of perpendicular direction located, the fracture of the fibers occurred much preferentially due to the extra compressive/shear stresses coming from the perpendicular fibers.

No obvious interface reaction between Mo fibers and Al matrix was found under present processing conditions of Mo/Al composites as shown in Fig. 2. However, it was expected from Mo/Al binary phase diagram that a severe interface reaction under the typical processing conditions around 500–600 °C for a longer time than one hour [21]. A porous and non-uniform interfacial reaction layer would then detrimentally affect the mechanical properties of the composites [22]. Thus, the interface microstructures after treatment under present processing temperature of 500 °C and a high temperature of 625 °C for 8 h were characterized to realize the formation of any improper interface reaction under the typical processing conditions around 500–600 °C. Fig. 3(a) shows the microstructure of the diffusion-bonded Mo/Al composite after heat treatment at 625 °C in air atmosphere for 8 h. An obvious but non-uniform, discontinuous interfacial reaction layer was formed due to the inhibition of Mo/Al reaction by some localized oxide layers remaining on the surface of the as-received Mo fiber. In comparison under the pressure of diffusion bonding, 100 MPa, at 500 °C in vacuum for the same 8 h, a uniform interface reaction was obtained as shown in Fig. 3(b). Because the residual oxide layers were broken or removed from the surface of the Mo fibers by the flow of Al matrix, a thorough contact was achieved between the fiber and the matrix without any barriers. Fig. 3(c), the EPMA line scanning of Mo element, indicated three possible intermetallic compounds formed at the interface, including $\text{Al}_5\text{Mo}_3$, $\text{Al}_3\text{Mo}$, $\text{Al}_{13}\text{Mo}$ sequentially away from the Mo fiber, as listed in Table 1.

Fig. 2. Microstructures of dual-directional Mo/Al composites reinforced with different contents of Mo fibers: (a) 5 vol%, (b) 25 vol%, and (c) 45 vol%; (d) necking and fracture of Mo fiber.
3.2. Mechanical properties of Mo/Al composites

Table 2 listed the tensile properties of as-received Mo fiber, hot-pressed 1100 Al plate, and 15 vol% unidirectional Mo/Al composite tested in longitudinal and transverse directions. The Mo fiber had a high yield strength (YS) of about 820 MPa and an ultimate tensile strength (UTS) of 1000 MPa, whereas the Al plate had very low strength less than 100 MPa. However, both the Mo fiber and Al plate exhibited very small strength deviations less than 2% and had large elongations of about 10% and 30%, respectively. As shown in Fig. 4(a), ductile necking and fracture of the Mo fiber was observed. On the surface of the fiber, the trace of wire drawing remained, associated with some fragments of Mo oxides which had formed during fiber fabrication. The Mo/Al composite reinforced with only 15 vol% unidirectional Mo fibers exhibited a high strength of 182 MPa in longitudinal direction, rather close to the value simply predicted by the rule of mixtures (ROM), because the strong strengthening effect provided by the continuous fibers [3]. The composite also presented a very high ductility of about 20% elongation. Fiber neck-

Table 1
Mo and Al element contents and possible compounds at the interface of Mo/Al composites after diffusion bonding at 100 MPa, 500 °C for 8 h measured by EPMA line scanning

<table>
<thead>
<tr>
<th>Region</th>
<th>Mo (at.%)</th>
<th>Al (at.%)</th>
<th>Possible compound</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>28.1</td>
<td>71.9</td>
<td>Al₃Mo₃</td>
</tr>
<tr>
<td>B</td>
<td>17.8</td>
<td>82.2</td>
<td>Al₅Mo</td>
</tr>
<tr>
<td>C</td>
<td>7.3</td>
<td>92.7</td>
<td>Al₁₂Mo</td>
</tr>
<tr>
<td>D</td>
<td>0</td>
<td>100</td>
<td>Al</td>
</tr>
</tbody>
</table>

Table 2
Tensile properties of Mo fiber, hot-pressed 1100 Al plate, and 15 vol% unidirectional Mo/Al composite tested in longitudinal and transverse directions

<table>
<thead>
<tr>
<th>Specimen</th>
<th>UTS (MPa)</th>
<th>YS (MPa)</th>
<th>Elongation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mo fiber</td>
<td>1000</td>
<td>826</td>
<td>10</td>
</tr>
<tr>
<td>1100 Al plate</td>
<td>78</td>
<td>44</td>
<td>29</td>
</tr>
<tr>
<td>Mo/Al (longitudinal)</td>
<td>182</td>
<td>146</td>
<td>19</td>
</tr>
<tr>
<td>Mo/Al (transverse)</td>
<td>35</td>
<td>34</td>
<td>1</td>
</tr>
</tbody>
</table>

Fig. 3. Microstructures of Mo/Al composites after (a) heat treatment at 625 °C in air atmosphere for 8 h and (b) diffusion bonding at 100 MPa, 500 °C for 8 h, and (c) EPMA line scanning of Mo element.
ing and pullout shown in Fig. 4(b), accompanied by interface decohesion, were the main features of the fracture surface of the composite tested in longitudinal direction. However, the strength of 15 vol% unidirectional Mo/Al composite in transverse direction was only about 35 MPa, even lower than the strength of 1100 Al plate, implying a weak interface between Mo fibers and Al matrix. Lateral cracks formed and rapidly propagated along the fiber/matrix interface without the constraint of fibers thus led to the early fracture of the composite with a low strength and small elongation of only 1%. As shown in Fig. 4(c), interface decohesion between fiber/foil and between foil/foil was obviously seen on the fracture surface of the composite tested in transverse direction. Some oxide layers existing on the surfaces of the Mo fibers and Al foils, like those shown in Fig. 4(a), might be the main reason for the weak interface bonding. From the following equation [23], the interfacial strength $\tau$ between the fiber and the matrix can be obtained as

$$\tau = \frac{P}{\pi d \ell},$$

(2)

where $P$ is the maximum load applied onto the fiber, 17.7 N, and $d$ is the diameter of the fiber, 150 $\mu$m. The average pullout length of the fiber, $\ell$, was measured under a scanning electron microscope as about 1.5 mm. The interfacial strength was then estimated as low as only about 25 MPa, well corresponding with the easy interface decohesion shown in both Figs. 4(b) and (c).

Fig. 5(a) shows the tensile strengths of 0°/90° dual-directional Mo/Al composites increased linearly with the content of Mo fibers according to the strengthening effect provided by the fibers. However as compared to the longitudinal strength of 15 vol% unidirectional Mo/Al composite listed in Table 2, the strengthening effect of the 0°/90° dual-directional fibers was lower due to the interface weakening in 90° transverse direction. Simply by averaging the strengths of the 15 vol% unidirectional Mo/Al composite in longitudinal and transverse directions, the strength of the 15 vol% 0°/90° dual-directional composite was calculated as about 110 MPa, rather close to the value experimentally measured, 115 MPa. All the measured strengths of the dual-directional Mo/Al composites also exhibited very small deviations, except that with the addition of 45 vol% fibers due to the influent flow of Al matrix around the large amount of fibers. Similar to the fractographies of the unidirectional composites, the fracture surfaces of the dual-directional composites shown in Fig. 6 were com-
posed of both fiber necking/pullout (for 0° longitudinal fibers) and interface decohesion (for 90° transverse fibers).

However, the strengths of 45°/135° dual-directional Mo/Al composites shown in Fig. 5(b) remained relatively low as compared to those of 0°/90° composites. The yield strengths of the composites slightly increased with increasing fiber content, and the ultimate tensile strengths of the composites only reached a maximum value of 147 MPa with the addition of 25 vol% Mo fibers. Some of the strengths were even lower than that of 1100 Al plate. Similar to the unidirectional Mo/Al composite tested in transverse direction, weak interfaces between the fibers and the matrix were responsible for the low strengths of the 45°/135° dual-directional Mo/Al composites. During tensile tests, the maximum shear stresses acted on the specimens in 45°/135° directions, where the weak interfaces...
located, resulting in the easy interface decohesion and low mechanical strengths. Thus, due to the largest amount of interface existing in the 45 vol% Mo/Al composite, the ultimate tensile strength of this composite dropped to only 109 MPa in spite of the strengthening effect of the fibers.

3.3. Thermal expansion behaviors of unidirectional Mo/Al composites

The thermal expansion behaviors of composites simultaneously change with temperature because of thermal stresses induced mutual constraints from reinforcements and matrices due to the differences in the CTEs of the reinforcements and the matrices [15,16,24,25]. Fig. 7 shows the CTE curves of Mo fiber, 1100 Al plate, and Mo/Al composite experimentally measured and theoretically predicted in longitudinal and transverse directions from 100 to 550 °C. As expected, the measured CTEs of the Mo fiber, 1100 Al plate, and Mo/Al composite in both longitudinal and transverse directions regularly increased as the temperature increased, but however the CTE values of the composite in these two directions were very different from each other. Under the constraint by the continuous Mo fibers of low CTEs, the CTEs of the composite in the longitudinal direction were very small, almost equal to the values of the fiber, whereas the composite freely expanded in the transverse direction without the constraint of the fibers and then exhibited relatively large CTEs close to the values of the Al plate [16,24,25].

Without consideration of stress transfer at interfaces, the CTEs of fibrous reinforced composites can be simply predicted by the rule of mixtures, i.e., the upper limit suggested by Schapery, as the following equation [24]:

\[ \alpha_C = \alpha_f V_f + \alpha_m V_m \]  

(3)

in which \( \alpha_C \), \( \alpha_f \), and \( \alpha_m \) are the CTEs of the composites, fibers, and matrices, respectively, and \( V_f \) and \( V_m \) are the volume fractions of the fibers and matrices, respectively. However for more precise predictions in real cases, the stress interaction at the interfaces may not be neglected and the rule of mixtures is thus no longer valid. For fibers with lower CTEs, they will restrict the expansion of the matrix of larger CTEs during heating, and the practical CTEs of the composites will then deviate from the values expected by rule of mixtures. Schapery gave more accurate predictions for the CTEs of fibrous reinforced composites in two directions under the consideration of stress balance, associated with the elastic moduli of the fibers and matrices, \( E_f \) and \( E_m \), respectively [25]. The CTEs of the composites in longitudinal direction, \( \alpha_{CL} \), and in transverse direction, \( \alpha_{CT} \), are written as the following equations, respectively:

\[ \alpha_{CL} = \frac{\alpha_f E_f V_f + \alpha_m E_m V_m}{E_f V_f + E_m V_m}, \]  

(4)

\[ \alpha_{CT} = (1 + \nu_f)\alpha_f V_f + (1 + \nu_m)\alpha_m V_m - \alpha_{CL} V, \]  

(5)

where \( \nu_f \) and \( \nu_m \) denote the Poisson ratios of the fibers and the matrices, respectively, and \( V = V_f + V_m \). Because no constraint from continuous fibers existed in the transverse direction, the theoretic CTEs of the composites mainly relate to the Poisson ratios and volume fractions of the fibers and the matrices.

However, the measured CTEs of 45 vol% unidirectional Mo/Al composite in both longitudinal and transverse directions were still different from those theoretically predicted by Schapery’s theory as plotted in Fig. 7. The CTEs of the composite in transverse direction were higher than the predicted values, whereas those in longitudinal direction were lower. This result was similar to the study on the thermal expansion behavior of unidirectional graphite fiber reinforced 6061 Al composite reported by Mitra et al. [6]. During heating, tensile stresses acted on the Mo fibers of small CTEs, and in contrary compressive stresses on the Al matrix of large CTEs along the longitudinal direction, together effectively constructing a stress balance and mutual constraints, and theoretically achieving the predicted CTEs. However under the high compressive stresses, free deformation of the matrix was restricted in the longitudinal direction, and thus the matrix would expand additionally in the transverse direction converted through Poisson ratio due to the low strength of the composite and small constraints in this direction. The additional expansions of the matrix in the transverse direction resulted in the high CTEs of the composite close to the values of the Al matrix in this direction and then reduced the tendency to expand in the longitudinal direction, lowering the CTEs close to the values of the Mo fibers. Furthermore, the deformation of the matrix released the accumulated thermal stresses and accordingly achieved the stably increased CTEs in both longitudinal and transverse directions with increasing temperature.

3.4. Thermal expansion behaviors of dual-directional Mo/Al composites

The thermal expansion behaviors of dual-directional Mo/Al composites were much more complicated than...
those of unidirectional composites because less stress release occurred in so-called “transverse” direction due to more constraints coming from the “dual-directional” fibers. Thus, the tremendous accumulation of thermal stresses during heating spontaneously affected the subsequent thermal expansion behaviors of the composites. Fig. 8 shows the CTE curves of 0°/90° and 45°/135° dual-directional Mof/Al composites reinforced with different contents of Mo fibers measured in longitudinal direction (along the long axis of specimen). The CTEs of these composites generally decreased with increasing the volume percentages of Mo fibers because of a larger constraint effect provided by more fibers of low CTEs as expected by Schapery [24,25]. However, comparing with the CTEs of the unidirectional composites, the CTEs of the dual-directional composites did not stably increase with temperature but varied for large ranges. The CTEs of the dual-directional composites decreased to the lowest values with increasing temperature to about 250–300 °C where the CTEs of the composites reinforced with more than 25 vol% Mo fibers even reached as low as the values of Mo fibers. After that, the CTEs regularly rose as the temperature increased.

The variations in the CTE curves of composites under the influences of residual and accumulated thermal stresses have been reported [6,15,16]. At the temperatures around 100 °C, the residual tensile stresses acting on the Al matrix, which had been formed during cooling from the processing temperatures of Mof/Al composites, were compensated by the contrarily accumulated compressive stresses during heating from room temperature to 100 °C. Under complete stress release, the CTEs of the dual-directional Mo/Al composites then stabilized at representative values predicted by Schapery [24,25]. With continuously increasing temperature, compressive stresses progressively accumulated on the Al matrix and gradually suppressed the expansions of the matrix. However unlike the unidirectional composites, there was no longer stress release or additional expansion of the matrix in the transverse direction of the dual-directional composites because both longitudinal and transverse directions of the composites were reinforced by continuous Mo fibers. Therefore, under effect of large accumulated compressive stresses acting on the matrix, the CTEs of the composites decreased to very low values with increasing temperature to about 250–300 °C. With more addition of the Mo fibers, the lowest points appeared at lower temperatures because more obvious stress accumulation by more fibers facilitated the lowest CTEs under smaller temperature differences.

It was particularly found that the CTEs of 25 and 35 vol% 45°/135° dual-directional Mo/Al composites shown in Fig. 8(b) were much smaller than those of 0°/90° composites and even reached negative values at about 250–350 °C. As shown in Fig. 5, the extremely low yield strengths of the 45°/135° Mo/Al composites relative to the ultimate tensile strengths of the composites and to the yield strengths of the 0°/90° composites might be responsible for the abnormal “shrinkage” behaviors. During heating, the expansions of the 45°/135° composites along fibers (0°/90° directions) resulted in contraction stresses oblique to fibers (45°/135° directions) due to Poisson effect. The large contraction stresses then caused the yielding of the composites of low yield strengths in 45°/135° directions and completely restricted the expansions of the composites in these directions, then even reaching negative CTEs measured. However for 45 vol% 45°/135° dual-directional Mo/Al composites, because the expansions of the composites in 0°/90° directions were rather small, the contraction effect acting in 45°/135° directions was not so obvious that the CTEs of the composites were not as small as those of the 25 and 35 vol% composites.

With continuously increased temperature over 250–300 °C, contrarily, the CTEs of both 0°/90° and 45°/135° dual-directional Mo/Al composites shown in Fig. 8 gradually increased because the expansion constraints on the matrix were relaxed through matrix yielding and interface decohesion at high temperatures [15,16]. The increases in the CTEs were more clearly presented in Fig. 9, the CTE variations of 15 and 35 vol% 0°/90° Mo/Al composites measured in longitudinal direction during three thermal cycles. The CTE curves of the 15 vol% Mo/Al composite in the second and third cycles shown in Fig. 9(a) were even

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**Fig. 8.** CTE curves of (a) 0°/90° and (b) 45°/135° dual-directional Mo/Al composites reinforced with different contents of Mo fibers and measured in longitudinal direction.
higher than that in the first cycle due to the occurrence of some severe interface decohesion after the first cycle as shown in Fig. 10(a). However, although matrix yielding was also expected and the resultant CTE increase also observed in the 35 vol% composite at temperatures above 250–300 °C, no severe interface decohesion occurred after thermal cycles as shown in Fig. 10(b). Thus, the CTE curves of the composite in three thermal cycles shown in Fig. 9(b) almost matched to each other.

The thermal expansion behaviors of dual-directional Mo/Al composites also varied with the geometry of specimens besides the contents of Mo fibers and the measurement directions (0°/90° or 45°/135°). As shown in Fig. 11(a), the CTEs of 35 vol% 0°/90° Mo/Al composite

![Fig. 9. CTE curves of (a) 15 vol% and (b) 35 vol% 0°/90° dual-directional Mo/Al composites measured in longitudinal direction during three thermal cycles.](image9)

![Fig. 10. Microstructures of (a) 15 vol% and (b) 35 vol% 0°/90° dual-directional Mo/Al composites after first thermal cycle.](image10)

![Fig. 11. CTE curves of 35 vol% (a) 0°/90° and (b) 45°/135° dual-directional Mo/Al composites measured in longitudinal and transverse directions.](image11)
measured in transverse direction (along the short axis of the specimen) were obviously smaller than those in longitudinal direction (along the long axis). As aforementioned, the large expansions of the 0°/90° composite along the long axis resulted in a contraction effect and then suppressed the expansions of the composite along the short axis, thus leading to the smaller CTEs measured in the transverse direction. However for 35 vol% 45°/135° composite, the contraction effect resulting from the 0°/90° directions equally acted on the composite in the 45°/135° directions, contributing the same constraints and similar thermal expansion behaviors whether along the long or short axis. Thus, both the CTE values of the 45°/135° composite measured in these two directions were similar only with small differences below 10^{-6}/°C as shown in Fig. 11(b).

4. Conclusions

The Mo/Al composites reinforced with different contents of unidirectional and dual-directional Mo fibers were successfully processed by diffusion bonding at 500 °C, 100 MPa in a vacuum of 10^{-2} Torr for 10 min. The mechanical strengths of unidirectional Mo/Al composites were close to theoretically predicted values. The strengths of 0°/90° dual-directional composites increased with fiber content, while those of 45°/135° composites remained relatively low. The CTEs of the unidirectional Mo/Al composites regularly increased with increasing temperature but obviously deviated from those theoretically predicted due to the expansion constraints of the matrix by the continuous Mo fibers. The CTEs of the dual-directional Mo/Al composites decreased as the fiber content increased. With increasing temperature to about 250 °C, the CTEs of the dual-directional composites decreased close to the values of Mo fibers because of the expansion constraints caused by large accumulated thermal stresses. The CTEs of the 45°/135° composites with more than 25 vol% Mo fibers were relatively lower than those of 0°/90° composites because a contraction effect acted on the composites in these directions. At temperatures above 250 °C, the CTEs of the dual-directional composites gradually increased, attributed to matrix yielding and interfacial decohesion, especially obvious for 15 vol% composites.

References