Preparation and thermophysical properties of graphite flake-carbon fiber coreinforced copper matrix composites

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Keywords: Cu, composite, modeling, thermal conductivity

Abstract

Graphite flake-carbon fiber core reinforced copper matrix composites were prepared by vacuum hot pressing technology. The carbon fibers were dispersed ultrasonically in alcohol and then mixed with graphite flake and alloys powder (Zr and Cu) for hot pressing sintering. The effects of the carbon fiber content on the microstructure, bending strength and thermal conductivity of the composites were investigated. The results show that the interface of the composites is well bonded. When the volume fraction of carbon fiber is 1%–3%, the carbon fiber can be uniformly dispersed in the matrix, and the bending strength of the composites can be improved effectively. When the volume fraction of carbon fiber is 2%, the bending strength reaches a maximum of 152 MPa, which is an increase of 60% compared with that of the composites without carbon fiber. However, an excessive addition of carbon fiber (4% or more) leads to an uneven distribution of carbon fiber, and the bending strength of the composites decreases. When the volume fraction of carbon fiber is 2%, the thermal conductivity of the composite is 597 W·m⁻¹·K⁻¹. The acoustic mismatch model (AMM) associated with the Digimat MF module is able to predict the thermal conductivity of the anisotropic multiphase composites.

1. Introduction

With the integration and miniaturization of high-power semiconductor electronic components, the thermal conductivity and thermal expansion coefficient performance requirements of heat dissipation materials have improved significantly. Therefore, in recent years, a series of excellent heat dissipation materials have been developed [1–4]. It is extremely important to design and develop packaging materials for electronic devices by compounding reinforcement with a metal matrix to obtain metal matrix packaging materials with good thermal conductivity and adjustable thermal expansion coefficients [5–7]. Copper has high thermal conductivity and good processing performance [8, 9]. Graphite flake has ultrahigh thermal conductivity and a negative thermal expansion coefficient in the basal plane. The graphite flake/copper composites has a high thermal conductivity of 600 W·m⁻¹·K⁻¹, good machinability and relatively low density. Therefore, it has become a hot issue in today’s research and the best candidate for the planar heat dissipation solution of electronic components [10–12].

Although the use of graphite flake/copper composites has made good progress in thermal conductivity and thermal expansion performance, the mechanical properties of graphite/copper composites are very poor due to the poor mechanical properties of graphite flakes. Researchers [13–16] developed two methods to improve the mechanical properties, namely, reinforcement surface modification and matrix alloying. Bai [13] improved the mechanical properties of graphite flake/copper composites from 50 MPa to 80 MPa by coating the surface of the flakes with a SiC layer. Zhang [14] reported the graphite flake/Cu-Zr composites have an approximately 90 MPa of bending strength by matrix alloying method. If the mechanical properties of the composites can be further improved, the range of applications will be more extensive.

Introducing hybrid reinforcement in a matrix can make use of advantages of reinforcements in order to obtain composites with good comprehensive properties. The literature [17] noted that adding fibers can effectively enhance the mechanical properties of composites. Therefore, it is a good idea to improve the
mechanical properties of graphite flake/copper composites by adding polyacrylonitrile-shortened carbon fibers. When the high thermal conductivity of graphite flake combined with the high mechanical properties of fiber, the composites with excellent properties will be obtained. However, the thermal conductivity of the polyacrylonitrile fiber is only approximately 140 W·m⁻¹·K⁻¹, and the thermal conductivity of the carbon fiber/copper composites decreases with increasing fiber content. Therefore, it is necessary to study the thermophysical properties of multiphase hybrid graphite composites to achieve the unity of high thermal conductivity and high strength.

In this study, graphite flake-carbon fiber coreinforced copper matrix composites were prepared by vacuum hot pressing technology. The composites have both high thermal conductivity and high bending strength. The effect of the addition of carbon fibers on the thermophysical properties of graphite flake/copper matrix composites was studied. The strengthening mechanism of carbon fiber to mechanical properties was analyzed and the importance of interfacial bonding was illustrated. The thermal conductivity of the anisotropic reinforced copper matrix composites was analyzed by using an acoustic mismatch model (AMM) associated with the Digimat MF module. By analyzing the influence of fiber content on the properties of composites, the optimal added content of fiber was obtained.

2. Material and methods

2.1. Raw materials
The flake graphite used was more than 99.9% pure, with an average diameter of 400 μm and thickness of 20 μm (morphologies are shown in figure 1(a)). The fiber had an average length of 1 mm and diameter of 7 μm (morphologies are shown in figure 1(a)). The composition of the matrix alloys powder prepared by ball-milling were Cu-98 wt% and Zr-2 wt%, with powder sizes of 50 um and 10 um, respectively (morphologies are shown in figures 1(a) and (d)).

2.2. Fabrication of composites and characterization
Figure 2 shows the sample preparation process and the sampling location of the sample performance test. Firstly, the fibers were mixed with alcohol in a beaker. Subsequently, the flake graphite and alloys powder were added into a beaker and continued to mix. In order to study the effect of fiber content about the thermal and mechanical properties, the alloy powder always accounts for 50 percent of the total volume and the volume ratio of fiber to graphite flake was 1:49, 2:48, 3:47 and 4:46. Then, the mixed powder was loaded into the mould after drying. The samples were prepared by vacuum hot-pressing sintering. The vacuum, pressure and sintering temperature was 1 Pa, 30 MPa, and 980 °C, respectively.
The Archimedes method was used to measure the density of the sample. The results showed that all samples had a density of more than 99%. Element distribution and microstructure structure of composites were analyzed by Energy Dispersion Spectra attached to the FE-SEM (TESCAN MIRA3). X-ray diffraction (XRD, RIGAKUTTR3) was employed to analyze the phase compositions of the composites. The thickness and micro morphology of the carbide were analyzed with FIB (FEI scios). Transmission electron microscopy (TEM, JEM-2800) was used to investigate the interfacial structure of composites. The bending strength of the sample in Z direction (the same direction of hot pressing) was tested by three-point bending method with sample size of 20 × 5 × 4 mm. The thermal diffusivity coefficient (α) of the composites in X-Y plane (normal plane in the direction of hot pressing) was measured by the laser flash method with sample of Φ 25.4 mm × 1 mm. The thermal conductivity (λ) of the composites was calculated according to the following formula:

\[ \lambda = \alpha \times c_p \times \rho \]  

The value of the composite specific heat capacity (C_p) was the sum of the specific heat capacity multiplied by the mass fraction of each component. The density (ρ) was calculated by the Archimedes method.

3. Results and discussion

3.1. Microstructure characterization

Figure 3 shows the SEM morphology of composites with different carbon fiber content. As the content of carbon fiber increased, there were an increasing number of black dots. Therefore, the black dot in the picture is carbon fiber, and the black strip phase is graphite flake. There are no pores, indicating that the obtained composites possessing a highly densified assembly. This is consistent with the situation of the graphite/copper composites prepared by Cui [16] by the hot pressing method, which has no obvious pores.

When the volume fraction of carbon fiber is 1%–3% (figures 3(a)–(c)), carbon fibers are uniformly distributed in the copper matrix. When the fiber volume fraction reaches 4% (figure 3(d)), an obvious fiber aggregation region (in the red box) appears. It is consistent with the phenomenon described in the literature [18], which is the aggregation phenomenon caused by an excessive addition of fibers.

The aggregation of fibers occurs in the mixing process and is related to the content of fibers. In the mixing process of carbon fiber/copper composites [18], the carbon fibers gradually change from an aggregation state to a dispersed state through stirring. With increasing carbon fiber content, when the carbon fiber volume fraction reaches a certain threshold (30% to 40%, even higher), it is very difficult to change the state for carbon fibers because excessive carbon fibers will block each other in the stirring process. Therefore, aggregation of fibers will occur in composites with a high volume of carbon fibers. However, in the graphite flake-carbon fiber coreinforced copper matrix composites, the blocking effect of graphite flakes is further strengthened. The

Figure 2. Schematic of the preparation process of composite and the test of performance.
Carbon fibers are blocked by the adjacent graphite flake. This means that the graphite flakes, regarding as a partition, prevent the carbon fibers from changing from an aggregated state to a dispersed state in the stirring process, which leads to the aggregation of fibers in composites with a low volume of carbon fiber.

### 3.2. Phase composition and interfacial structure of composites

Figure 4 shows the XRD pattern of composites. As shown in the figure, peaks of Cu, graphite and ZrC are detected, indicating that Zr reacts with carbon to generate the corresponding carbides, namely, ZrC. At the same time, the peak strength of the graphite (002) surface is very high, which indicates that the graphite flake has a preferred orientation due to the hot pressing process.

Figure 5 shows the energy spectrum at the interface of the composites. As shown in the figure, there is an obvious transition layer between the carbon and copper matrix. The interface of the transition layer and the carbon and copper matrix is well bonded without obvious pores, and the thickness of the transition layer is approximately 0.36 μm. The energy spectrum shows that the content of Zr in this region is high, and there is an
obvious presence of C because the Zr element located in the matrix spontaneously diffuses from the matrix to the interface at high temperatures \cite{14}. Combined with the ZrC detected by the XRD pattern, these results indicate that Zr reacts with carbon to form the corresponding carbides.

To further determine the interfacial structure and its crystal structure, FIB was used to cut the interfacial bonding part of the sample, as shown in figure 6. This part was characterized by TEM, and a TEM image of the interface joint was obtained. Figure 7(a) shows that the interface is divided into three regions: copper matrix, ZrC and graphite. The gap between the ZrC and graphite regions is caused by the breakdown of the graphite because it is too thin. The interface between the matrix and ZrC was observed by HRTEM, as shown in figure 7(b). The interface between the two was well bonded without pores, and the ZrC in the sample could be further determined by the lattice spacing under HRTEM. HRTEM observation of the bonding interface between carbide and graphite is shown in figure 7(c). It can also be found that both interfaces are well bonded and have no pores. It indicates that the ZrC generated can improve the bonding between the matrix and the reinforcement phase.
3.3. Mechanical properties of composites and strengthening mechanism

Figure 8 shows the bending strength of composites with different carbon fiber content. As shown in the figure, the bending strength of composites with a 2% volume fraction of carbon fiber increases from 89 MPa to 152 MPa compared with composites without carbon fiber. This indicates that the addition of carbon fiber plays an important role in improving the bending strength of composites. The key strengthening mechanisms of fiber reported in composites [19] can be summarized as follows: 1. shear lag theory. 2. interface debonding. The difference between this two mechanisms is that the former is based on the assumption of perfect interface, while the latter believes that the interface is not perfect. In the former, the broken fiber absorbs energy, while in the latter energy was consumed primarily for the interface debonding.

In order to study the strengthening mechanism of carbon fiber in this paper, we observed the fracture of the composites, as shown in figure 9. It shows the presence of two morphologies of carbon fiber at the fracture,
namely, the fractured carbon fiber embedded in the matrix (figure 9(a)) and the carbon fiber completely pulled out of the matrix (figure 9(b)). It is reasonable to believe that the pulled fiber and the fractured fiber have great differences in fracture models, which can be divided into two fracture models, namely Model 1 (shear lag model) \[20\] and Model 2 (interface debonding) \[21\], as shown in figure 10.

Model 1 is a simple shear lag model, which can be used to explain the fractured carbon fiber. In the simple shear lag model, the applied load is transferred to the carbon fiber through the matrix. At this time, the carbon fiber is stressed. When the load exceeds the critical value, cracks pass through the carbon fiber, which can be verified from the broken fiber. However, the composites interface will affect the transfer of stress \[22\]. It is well known that the internal stress will cause the relative sliding between the fiber and the matrix during the transfer process. Therefore, the interface debonding of the matrix and carbon fiber should be considered. Model 2 is interface debonding, which can be used to explain the pulled carbon fiber. In the interface debonding theory, the interface debonding occurs first, and cracks will expand along the interface, which can be verified from the pullout fiber.

We can observe the two fracture models both in the composite. The reason is that energies of interface debonding between matrix and fiber exceed the fracture energies of the fiber in some area. Whereas some areas does not. When the energy of interface debonding exceeds the fracture energy of fiber, fiber fractured preferentially, which is Model 1; otherwise, it is Model 2. Remarkably, the energy of interface debonding...
Table 1. Mechanical properties of copper/Flake composites with 50 Vol% flake.

| Sample                                      | Bending strength/(MPa) |
|---------------------------------------------|------------------------|
| Copper/Flake composites (with 2 wt% Ti)     | 80                     |
| Copper/Flake composites (with 2 wt% Zr)     | 95                     |
| Copper/Flake composites (with 2 wt% Cr)     | 93                     |
| Copper/Flake composites (with 1 wt% Mo)     | 75                     |
| Copper/Flake composites (with 2 wt% Zr and 2 Vol% fiber)* | 152                   |

* In this study

depends on the length of the interface. The literature [23] reported that the finite element method was used to simulate single fiber pull-out experiment. It has been found that shorter free length, defined as the portion of the fiber between the material and the position at which the fiber is pulled from, cause fibers to have higher pullout forces than fibers with longer free lengths. It means that longer interfaces will absorb more energy [21]. Therefore, the length of the interface will determine the fracture model.

As shown in figure 10, when the shear plane was located in the middle of carbon fiber, the interfacial length between fiber and copper matrix is very long. On the contrary, if the shear plane was located at ends of the fiber, the interfacial length between fiber and copper matrix is very short. Therefore, the position of the shear plane determines the fracture model of the fiber. The fracture model of the former tends to be Model 1 and the latter tends to be Model 2.

In the case of Model 1, when the composite was subjected to load, carbon fiber can effectively bear the load, and the cracks pass mainly through the carbon fiber. Since the elastic modulus of carbon fiber is much larger than that of graphite flake, the fractured energy of fiber is much larger than that of graphite flake. In composites design, carbon fiber was used to replace graphite flake. So the energy required for fracture of composites was greatly increased. Therefore, the bending strength of composites was greatly improved.

In the case of Model 2, the energy was consumed primarily for the interface debonding. The literature [14] reported that the cracks directly pass through the graphite flake, after Zr was added. It proves that the energy of interface bonding is more than the fractured energy of graphite flake. As a result, when the graphite flake in the composites was replaced by carbon fiber, the mechanical properties of the composites were greatly improved.

In order to show the reinforcement effect of fibers in this paper, other literatures [11, 14–16] on the mechanical properties of composites enhanced by alloying are shown in table 1. Due to the poor mechanical properties of graphite flake, the mechanical properties of graphite flake by alloying method can hardly exceed 100 MPa. The mechanical properties were significantly improved by adding carbon fiber after alloying. Compared with the composites with adding Zr only, the mechanical properties of graphite flake–carbon fiber core reinforced copper matrix composites increased by 60%.

Notably, when the volume fraction of carbon fiber reaches 4%, the bending strength begins to decrease obviously. Combined with the aggregation in the morphology of composites with 4% carbon fibers (figure 3(d)), we investigate the fracture of aggregation, as shown figure 9(c). It shows that a large number of fibers are aggregated without a copper matrix. The reason is that the pores, originating from aggregation, make it difficult for copper to fill in the vacuum hot pressing process. As the result, pores from aggregation in the composites become the source of cracks, and this leads to a significant decrease in the bending strength of the composites. On the other hand, the aggregated carbon fiber can’t react with Zr in the copper powder. The mechanical interface bonding is much lower than the interface bonding after carbide generation. Therefore, the carbon fibers in the aggregation state are easily debonding when the load is applied to the composites. In the sum, excess fiber that leads to pores cannot provide the effect of enhancing the mechanical properties. And simultaneously, it creates a crack source. Therefore, only a small amount of carbon fiber can be added to enhance the bending strength of composites. The mechanical properties of the composites could be effectively enhanced by the addition of 1 vol% to 2 vol% carbon fibers.

3.4. Thermal conductivity and theoretical prediction models

The thermal conductivity is a key parameter of heat dissipation materials and is related to the intrinsic thermal conductivity, spatial arrangement and interfacial bonding degree of the reinforcement. Graphite flakes can obviously improve the thermal conductivity of the composites, while carbon fibers have the opposite effect. Therefore, it is important to deeply understand the effect of the thermal conductivity of composites when they are mixed. However, to date, no theoretical model has directly predicted the thermal conductivity of composites due to their anisotropy.

The Digimat software can predict the constitutive behavior of heterogeneous and anisotropic materials such as polymer matrix composites (PMCs), rubber matrix composites (RMCs) and metal matrix composites
(MMCs). The MF module of the Digimat software is a multiphase nonlinear material constitutive prediction tool based on Eshelby inclusion theory and mean field homogenization method. It can be used to realize fully coupled nonlinear analysis and predict the thermal conductivity of multiphase composites.

In this study, the MF module in the Digimat software was used to construct a copper matrix composite with graphite flakes and carbon fibers as the reinforcement phase, and the thermal conductivity of the composites in the X-Y direction was simulated and predicted. The orientation of the reinforcing phase in the composite has a certain influence on its thermal conductivity. The orientation angle of most of the graphite flakes in composites was below 10° by observation of SEM, and the study [4] previously pointed out that the influence on thermal conductivity was less than 5% when the flake deflection angle was less than 10°. Therefore, the model was simplified in this study: all graphite flakes were assumed to be perfectly oriented, which meant that all graphite flakes were parallel to the X-Y plane, and the fibers were randomly distributed in two dimensions on the X-Y plane. The specific steps of model are as follows: 1 The module of thermal analysis was selected. 2 The intrinsic parameters of materials were set. (specific settings were discussed below) 3 Phases were generated by selecting the material previously set. 4 The state of the phase in the composites was set by selecting the specific topology (the intrinsic properties of the material), volume fraction(content in the composites) and orientation (mentioned above). The copper was set as the matrix. The graphite flake and carbon fiber were set as the inclusion. 5 The solver for actuarial calculation (Digimat-MF) was selected. Finally, the predicted thermal conductivity values of the composites with different fiber volume fractions in all directions were obtained. In this study, we mainly discuss the thermal conductivity in the dominant direction of the composites (X-Y plane).

Since carbide is generated at the interface in the actual process, the carrier will lose at the interface in the process of energy transfer, which indicates interfacial thermal resistance at the interface. The interfacial thermal resistance \( R \) can be regarded as a layer of thermal barrier coating wrapped on the surface of the graphite flake and fiber [24]. This has a certain influence on the intrinsic thermal conductivity of graphite. To make the simulation more realistic, it is necessary to modify the intrinsic thermal conductivity for the effective thermal conductivity. According to the series method [14], the relationship among the effective thermal conductivity, intrinsic thermal conductivity and interfacial thermal resistance of graphite can be obtained as follows:

\[
K_{eff}^{XY} = \frac{K_{XY}^{eff}}{1 + \frac{2K_{XY}^{eff}t}{D}}
\]

\[
K_{eff}^{ZZ} = \frac{K_{Z}^{eff}}{1 + \frac{2K_{Z}^{eff}t}{t}}
\]

where, \( K_{XY}^{eff} \) is the effective radial thermal conductivity of the reinforcement phase, and \( K_{Z}^{eff} \) is the intrinsic radial thermal conductivity of the reinforcement phase. \( K_{XY}^{eff} \) is the effective axial thermal conductivity of the reinforcement phase, and \( K_{Z}^{eff} \) is the intrinsic thermal conductivity of the reinforcement phase. \( D \) is the diameter of the reinforcing phase; \( t \) is the thickness of the reinforcing phase; \( R \) is interfacial thermal resistance.

The interfacial thermal resistance \( R \) of the composites is composed of three parts: the interfacial thermal resistance \( R_1 \) between the carbon matrix and the carbide, the interfacial thermal resistance \( R_2 \) of the carbide itself, and the interfacial thermal resistance \( R_3 \) between the carbide and the copper matrix. Therefore, the interfacial thermal resistance \( R \) of the composites is calculated according to the following formula:

\[
R = R_1 + R_2 + R_3
\]

where, \( R_1 \) is the interfacial thermal resistance between carbon matrix and carbide; \( R_2 \) is the thermal resistance of the carbide itself; \( R_3 \) is the interfacial thermal resistance between the carbide and the copper matrix.

The thermal resistance \( R_2 \) of the carbides can be calculated according to the following formula:

\[
R_2 = \frac{t_{layer}}{K_{layer}}
\]

where, \( t_{layer} \) is its own thickness; \( K_{layer} \) is the thermal conductivity.

The above \( R_1 \) and \( R_3 \) can be calculated according to the phonon mismatch model (AMM), as follows:

\[
R = \frac{2(\rho_m \nu_m + \rho_{re} \nu_{re})^2}{C_m \cdot \rho_m \cdot \nu_m \cdot \rho_{re} \cdot \nu_{re}}
\]

where, \( C_m \) is the heat capacity of the matrix material; \( \rho_m \) is the density of the matrix material and \( \rho_{re} \) is the density of the reinforcing phase. \( \nu_m \) is the average phonon velocity of the matrix material, and \( \nu_{re} \) is the average phonon velocity of the reinforcement phase.

The average phonon velocity \( \nu_m \) of the isotropic copper matrix is calculated according to the following formula:
where, $\nu_m$ is the average phonon velocity of copper matrix; $\nu_l$ is the longitudinal phonon velocity; $\nu_t$ is the transverse phonon velocity.

For anisotropic graphite flake and carbon fiber, the average phonon velocity is $\nu_{re}$, calculated according to the following formula:

$$\frac{3}{\nu_{re}^2} = \frac{1}{\nu_l^2} + \frac{2}{\nu_t^2}$$  \hspace{1cm} (7)

where, $\nu_{re}$ is the average phonon velocity of the enhanced phase; $\nu_l$ is the longitudinal phonon velocity; $\nu_t$ is the transverse phonon velocity; $\nu_o$ is the phonon velocity in the normal direction of the plane.

All data in the above formula refer to Table 2. Other data are as follows: thickness of the carbide layer is 0.36 $\mu$m, thickness of the graphite flake is 15 $\mu$m and diameter is 400 $\mu$m. The length of the carbon fiber is 1000 $\mu$m, and the diameter is 7 $\mu$m. By substituting all above data into the formula, the effective thermal conductivity of graphite and carbon fiber can be revised to 877 W m$^{-1}$ K$^{-1}$ and 129 W m$^{-1}$ K$^{-1}$, respectively. The measured thermal conductivity of pure copper after sintering is 345 W m$^{-1}$ K$^{-1}$ as the matrix thermal conductivity. Substituting all revised data into the MF software, the revised forecast value is finally obtained.

Figure 11 shows the influence of the fiber content on the thermal conductivity of graphite-carbon fiber coreinforced copper-matrix composites. As shown in the figure, the disparity of the predicted value obtained by using the MF module is less than 8%, which means that the acoustic mismatch model (AMM) associated with the digit MF module effectively predicts the thermal conductivity of the multiphase composites. With regard to the disparity, there are several reasons for the difference between the theoretical value and the experimental value: (1) The distribution of carbon fiber is not an ideal uniform distribution, which will be pores. In addition, pores greatly reduce the thermal conductivity. (2) The orientation calculated by the simulation model is the perfect orientation. Through microstructure photos of the composites, it can be found that there is a certain gap between the orientation of a few graphite flakes and carbon fibers and the perfect orientation. Thus, the experimental data are smaller than the simulated data. (3) The thickness of the carbide in the model is 0.36 $\mu$m, but the thickness of the carbide at the interface is slightly different in some areas. This leads to a change in the interface thermal resistance, resulting in a deviation between the experimental value and the theoretical value.

Figure 11 shows that the thermal conductivity gradually decreases with increasing carbon fiber content. The reason is that the intrinsic thermal conductivity of carbon fiber is only 130 W m$^{-1}$ K$^{-1}$, while the intrinsic thermal conductivity of graphite flakes is 1000 W m$^{-1}$ K$^{-1}$. When the content of carbon fibers increases, the content of graphite flakes decreases oppositely, so the thermal conductivity of the composites decreases gradually. Moreover, the specific surface area of carbon fibers is larger than that of graphite flakes, and more carbon fibers represent more interfaces. Electrons dominate the heat conduction of the metal matrix, phonons and electrons bear the heat transfer of graphite, and the interface of the composites causes scattering of electrons and phonons at the interface. Therefore, the destruction of matrix continuity caused by more interfaces is another reason for the decrease in the thermal conductivity of the composites.

According to the above conclusions, the mechanical and thermal properties of the composites decrease with an increase in fibers at a high volume fraction. Excess fiber causes aggregation, which leads to a decrease in the mechanical properties. At the same time, fiber itself leads to a decrease in the thermal properties of composites. Therefore, the volume fraction of carbon fiber should be controlled within a certain range to effectively improve the comprehensive properties of composites. A small amount of carbon fiber can improve the mechanical properties but has little effect on the thermal properties. For example, when the volume fraction of carbon fiber is 2%, the bending strength and thermal conductivity of the composites are 152 MPa and 597 W m$^{-1}$ K$^{-1}$, respectively, and the composites have good comprehensive properties.

4. Conclusion

In this work, graphite flake-carbon fiber coreinforced copper matrix composites were prepared and the influence of fiber content on composite materials was studied. A theoretical prediction model for multiphase composite materials was proposed. The key findings are summarized as follows:

(1) The addition of carbon fiber improved the bending strength of the composites. When the volume fraction of carbon fibers was 2%, the maximum bending strength of the composites was 152 MPa, which was 60% higher than that of the graphite/copper composites without carbon fibers.

(2) Excessive fiber addition (4 vol% or more) will lead to fiber aggregation, and fiber aggregation will degrade the mechanical properties of composites.
Table 2. Parameters used in theoretical calculation of AMM model [25–30].

| Phase        | Density/(kg·m$^{-3}$) | Thermal conductivity/(W·m$^{-1}$·K$^{-1}$) | Specific heat/(J·Kg$^{-1}$·K$^{-1}$) | Phonon velocity/(m·s$^{-1}$) |
|--------------|------------------------|--------------------------------------------|-------------------------------------|-------------------------------|
| Cu           | 8960                   | 390                                        | 385                                 | 4910 Longitudinal 2500 Transversal |
| ZrC          | 6730                   | 21                                         | 36.4                                | 8228 Longitudinal 5026 Transversal |
| Graphite flake| 2260                   | 1000 Transversal 10 Longitudinal           | 710                                 | 22160 Longitudinal 14660 Transversal 4140 Out-plane |
| Fiber        | 1800                   | 130 Longitudinal 10 Transversal            | 710                                 | 22160 Longitudinal 14660 Transversal 4140 Out-plane |
(3) The addition of carbon fiber has little effect on the thermal conductivity of the composites. When the volume fraction of carbon fiber added to the composites was 2%, the thermal conductivity of the composites was 597 W m$^{-1}$ K$^{-1}$.

(4) The AMM model combined with MF software can effectively predict changes in the thermal conductivity of multiphase core reinforced copper matrix composites and can guide the design of multiphase composites.

Acknowledgments

This work was supported by the projects of the National Science Foundation of China (No. 51704113), Innovation and Entrepreneurship Training program for College students (202110534027), Scientific Research Fund of Hunan Provincial Education Department (21B0468) and the Natural Science Foundation of Hunan Province (No. 2018JJ3172). Thanks to Hunan MAG New Material Technology Co. Ltd for providing testing services to this study.

Data availability statement

The data that support the findings of this study are available upon reasonable request from the authors.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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