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Short communication

Effect of sintering time on the bending strength and CTE of SiC/Al–35Si composite

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ABSTRACT

SiC/Al–35Si composite were prepared using a novel addition method of silicon particles. The effects of sintering time on the bending strength and CTE of the composite were investigated. The results showed that the optimal properties could be obtained when sintering time is 4 h. The prolongation of the sintering time increased the size of silicon phase and α-Al phase without solution strengthening of silicon, resulting in the increase of CTE. However, the increase of silicon phase size within a certain scope is beneficial to load transfer due to the formation of less pores and good bond interfaces. The bending strength first increases up to the maximum value and then trends to decrease. Overall, SiC/Al–35Si composite has a good combination of bending strength and CTE.

1. Introduction

As electronic packaging develops in the direction of high thermal conductivity, low coefficient of thermal expansion(CTE), small size and light weight [1]. The CTE describes how the size of an object changes with a change in temperature. Specifically, it measures the fractional change in size per degree change in temperature at a constant pressure. The application of conventional electronic packaging materials, such as Mo–Cu or W–Cu, demonstrate poor welding and machining properties, high densities, and not suited to use in aerospace and aviation fields; Kovar or Invor have appropriate CTEs, but have some shortcomings of low thermal conduction properties, affecting the material intensity and stability [2–4]. Therefore, more and more attentions have been attracted in the new electronic packaging materials to meet the requirements for more hard working conditions.

During the past two decades, a large amount of researches about the use of Al–SiC and Al–Si–SiC composites have been reported. Hyo S. Lee [3] et al. studied the mechanical properties of SiC/Al–Si metal matrix composites fabricated by vacuum hot pressing, they found that the strength was mainly affected by the silicon phase size. Thermal expansion behavior of aluminum matrix composites was reported by N Chawla and Tran Huu Nam [4,5] et al., they found that the thermal expansion behavior was strongly influenced by the presence of voids and internal stress, and then resulted in decreasing the CTE. Interfacial bonding strength of 2014Al matrix composites reinforced with oxidized SiC particles, fabricated by vacuum hot pressing, was reported by Youngman Kim [6] et al., they found that the interaction between matrix and reinforcement was the key factor effecting the mechanical property, such as decreasing the bending strength. Because poor interface combination areas are the sites of damage, which occur at an early stage of the loading process, and then promote the crack propagation up to result in the final fracture. Similar research results were reported by Zhao-hui Zhang, Ege Anil Diler and Xue C et al. [7,8,13]. In general, the thermal expansion properties of materials with different phases will behave in a complex combination method, because of a variety of interacting factors. It was well known that the relaxation of residual stress during the descent temperature process could lead to an elastic plastic deformation zone in the metal matrix around a ceramic phase. Therefore this will reduce the bonding strength and then resulting in devices failure [9]. It can be summarized from the studies mentioned above that the size of silicon phase, interaction and difference of CTE between matrix and reinforcement, presence of voids and internal stress are the main factors influencing CTE and bending strength of composites. For the purpose of decreasing the relaxation of residual stress, the difference of CTE between matrix and reinforcement should be reduced.

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For the SiC particles and hypereutectic Al–35Si alloy powder, the mismatch of CTE is low, leading to dimensional stability during working temperature circulation change. In addition, the CTE and bending strength of SiC/Al–35Si composites, prepared by vacuum hot-pressed sintered technology, have not been reported up to now. Against this backdrop, the objective of this paper is to investigate the effect of sintering time on the CTE and bending strength of composite by controlling silicon phase.

2. Experimental procedure

Al–35Si alloy (Al-35 wt.% Si) fabricated by atomization was used as matrix and SiC particle was used as reinforcement. The silicon existed in the Al–35Si alloy powder with Al–Si supersaturated solid solution in the α-Al phase. Compared with equilibrium solidification, the nucleation and growth of silicon phase were inhibited, due to rapid solidification. Therefore, the precipitation of primary silicon particles has been restrained and the silicon existed in the Al–35Si alloy powder with Al–Si supersaturated solid solution in the α-Al phase, forming a metastable structure or amorphous phases. In the vacuum hot-pressed sintering process, the sintering temperature promotes the precipitation of Si from α-Al phase and favors the growth of precipitated silicon phase as the sintering temperature provides high energy to enhance the mobility of the atoms. The average particle diameter of Al–35Si and SiC particle is 10 μm and 20 μm, respectively. The volume fraction of SiC particles in the composites is 35%. The Al–35Si and SiC particles were ball-mixed in a high-speed planetary mill for 3 h with the rotation speed of 250 rpm. The milling medium was ethanol and agate balls, the weight ratio of ball and powder was 3:1. The composites were achieved by putting powder mixture into the mold and then heated in vacuum diffusion welding furnace with adjusting sintering time from 1 h to 5 h at the sintering temperature of 590 °C, with a heating rate of 12 °C per min, pressure of 2.3×10^{-3} Pa, pressing pressure of 65 MPa.

The microstructure and phases of the composites was characterized by X-ray diffraction (XRD) (Model Bruker D8 Discover) with Cu Kα radiation, transmission electron microscope (TEM) (JEM-2100) and selected area electron diffraction (SAED). The size of silicon phase, fracture surface of the composites was analyzed by scanning electron microscope (SEM) (Hitachi S-4800, Hitachi, Japan). Three-points bending strength tests were carried out via an AG-100KNA test machine on 3 mm × 4 mm × 36 mm specimens with a loading speed of 0.5 mm/min. The specimen surfaces for thermo-mechanical evaluating, Φ5 mm × 50 mm in size, were polished with 1 μm diamond paste. The CTEs were evaluated with thermo-mechanical analysis equipment (Model TMA-50, SHIMADZU, JAPAN). Thermal expansion of the specimens was examined by a linear position transducer during heating and cooling with rates of 3 °C per min.

3. Results and discussion

Fig. 1a shows the SEM microstructure of SiC/Al–35Si composites under a sintering time of 4 h. It can be seen that the SiC particles distribute relatively uniform and macroscopically homogeneous in the matrix. The multiple-scale silicon phases have a variety of shapes, ranging from 0.1 μm size up to 30 μm size smaller than the sizes reported by Refs. [12,15]. It is reported that cracks could pass through silicon particles, and the surface of SiC particles could have a shrinkage due to interfacial reaction products [3,10,12]. Therefore, fracture surface analysis and XRD spectra of the composite were carried out. Fig. 1b exhibits the fracture surface of the composite under a sintering time of 4 h it can be seen that no exposed SiC particles or coarse silicon particles exist in the fracture surface, indicating that no cracks pass through silicon particles, the interfacial bonding in SiC/matrix, Si/matrix are relatively strong, and the size distribution of silicon particle precipitating from hypereutectic Al–35Si alloy in this preparation condition is optimal. Fig. 1c displays the XRD spectra of SiC/Al–35Si composite under a sintering time of 4 h. It is indicated that the phases of the specimen are Al, SiC and Si, as observed from the diffraction peaks of Al, SiC and Si indexed in the spectra. Therefore, we detect that no any other reaction products (Al_4C_3, $MgAl_2O_4$ [10,11]) exist in the composites.

In order to further identify whether there are any other reaction phases in the composites, TEM and SAED were carried out. Fig. 2a demonstrates the SiC/Al interface in the SiC/Al–35Si composite under a sintering time of 4 h. The interface is clear and thin, combining with the analysis of SAED pattern (Fig. 2b), it can be also found that semi-coherent phase boundary exists between SiC and Al, due to lattice mismatch (7.079%) in the standard range of values (5% – 25%) by calculating method. Meanwhile, from Fig. 2b we can obtain that $[0\bar{1}\bar{1}]Al//[0\bar{1}\bar{1}]SiC$, $(\bar{1}11)Al//(\bar{1}11)SiC$, $(200)Al//(200)SiC$. Combined the results of XRD analysis, it is indicated that there are no new reaction phase products, and it has a great bonding strength between Al and SiC particle. And this result shows that why we don't find SiC particles in fracture surface (Fig. 1b).

In order to examine the quality of SiC/Al–35Si composites, bending strength and CTE tests are accomplished. Fig. 3 shows the effects of sintering time on the bending strength (a) and CTE (b) of Al–35Si composites.

From Fig. 3a, it can be seen that, with increasing the sintering time, the bending strength increases at a sintering time range between 1 h and 4 h, and then decreases at a sintering time range between 4 h and 5 h. The bending strength is highest about twice as much as those reported by Refs. [12,15] at a sintering time of 4 h. In general, it is considered that the bending strength depends on the combined action of the size of precipitating silicon particles, reinforcement/matrix interfacial bonding strength and presence of voids [13]. Because the poor combined interface areas and presence of voids are the sites of damage, which occur at an early stage of the loading process, and then promote the crack propagation up to result in the final fracture [14]. It is known that the larger size of precipitating silicon particles would decrease bending strength due to the cracks passing through coarse brittle silicon particles [3,15]. Firstly, larger sized silicon particles have larger interface area with the matrix, and thus endure higher stress concentration. Secondly, the particle fracture strength depends on the instinct flaws within the particle. Since the size of a flaw is controlled by the size of the silicon particle, larger particles are more likely to fracture because they have a statistic probability of containing a flaw that is greater than smaller ones [14]. In this paper, it was known that silicon particles will precipitate from hypereutectic Al–35Si alloy at 590 °C [12]. And the silicon particle size increases with increasing the sintering time. Meanwhile there are more areas of contacted interface between the reinforcement and matrix and less voids existed in the composite. This is due to that softening matrix is beneficial to reinforcements flowing freely, increasing the effective combined areas and wettability of reinforcement/matrix with the increase of sintering time. Therefore, it is known from analyses above that the increase of bending strength is mainly attributed to the increase of the interface combination and less voids at a sintering time range between 1 h and 4 h. When sintering time is 5 h, the decrease of bending strength could be ascribed to the increase of silicon phase size.

From Fig. 3b, it can be seen that, with increasing the sintering time, the CTE of the composite increases. In general, it is considered that the CTE of the MMCs is mainly attributed to the thermal expansion behavior of the matrix and reinforcement [4,9]. In this