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INVESTIGATION OF THE MICROSTRUCTURE AND MECHANICAL PROPERTIES IN MODIFIED 2009 Al/SiC_p COMPOSITES

Qiong Li and Frank E. Wawner
Department of Materials Science and Engineering
University of Virginia
Charlottesville, VA 22903
U.S.A.

ABSTRACT

The objective of this investigation is to modify alloy 2009 (an Al/SiC particulate material produced by Advanced Composite Materials Corporation) by adding Ag to promote the formation of the Ω phase in the material, and to evaluate the material's mechanical properties. The Ω phase is a semi-coherent precipitate that has been shown to be more thermally stable in Al-Cu-Mg alloys than the normally obtained θ' or S' phases. This paper describes the matrix microstructure obtained as a result of various heat treatments, thermal stability of these phases, and tensile properties of the modified composite as a function of the thermal exposure.

INTRODUCTION

Al-Cu-Mg alloys containing a small amount of Ag have been shown to possess superior hardening by aging between 150°C and 250°C [1, 2]. Recent investigations have shown that this hardening is mainly due to the precipitation of the plate-like Ω phase on the {111} Al matrix planes [3]. The Ω phase is semicoherent with the Al matrix and has a tetragonal structure and composition Al₂Cu [4]. Compared with the θ' phase, the low coarsening rate of the Ω phase should enable use of the alloys at elevated temperatures [5]. The purpose of this research is to utilize an alloy strengthened by the Ω phase as a matrix metal to form a composite for high temperature application. The composite with the Ω phase in the matrix could have better thermal stability, ductility, and toughness hence extended applications at elevated temperature.

EXPERIMENTAL

The P/M Al powder and modified 2009 composite (Al-Cu-Mg-Ag/15v/o SiC particles) were supplied by Advanced Composite Materials Corporation (ACMC). A P/M bulk sample was fabricated at the University of Virginia to evaluate the microstructure. The composition of the powder is shown in TABLE I. The sample was hot pressed in a vacuum furnace then hot rolled, solutionized at 525°C for 3 hours, water quenched and aged at 190°C for 5 hours. The composite sample, designated 2009M/SiC_p, was fabricated at ACMC. The composition of the

TABLE I. Compositions of the 2009M powder and 2009M/SiC composite

	Cu wt%	Mg wt%	Ag wt%	Si wt%	Fe wt%
2009 M P.M alloy	3.3	0.76	0.56	0.07	0.08
2009M SiC _p MMC	2.9	0.7	0.47	0.16	0.04

*silicon analysis is for information only and not an acceptance test.

composite varied after manufacture and the actual composition is shown in TABLE I. The as-extruded material was solutionized at 520°C for 3 hours, water quenched and subsequently aged at 190°C for various times. Hardness tests were performed on an Indention hardness machine using a Rockwell scale. Tensile tests were performed on an ATS mechanical test machine. The tensile samples were machined to a cylinder with a diameter of 3.25 mm and a gauge length of 31.75 mm. The mechanical test samples were peak aged at 190°C for 3.5 hours, thermally exposed at different conditions and tested at different temperatures.

The TEM specimens were thinned mechanically to 0.15 mm then electrolytically polished to electron transparency. The TEM experiments were performed at 120kV using a Philips EM 400T.

RESULTS AND DISCUSSION

Hardness Tests

Hardness test results show that the peak aging at 190°C is between 3-6 hours. The sample aged immediately after cold water quench has a higher hardness value than the one aged at 190°C after 24 hours of natural aging. With natural aging the hardness value of the sample increases gradually to near maximum after 24 hours with little increase thereafter. The peak aging time for the composite is faster than the unreinforced alloy by about 3 to 5 hours. Continuing investigations suggest that the slow increase in hardness of the composite aged at ambient temperature corresponds to the internal stress relaxation. The internal stress relaxation may also give rise to the sharp increase of the composite hardness curve at elevated temperature.

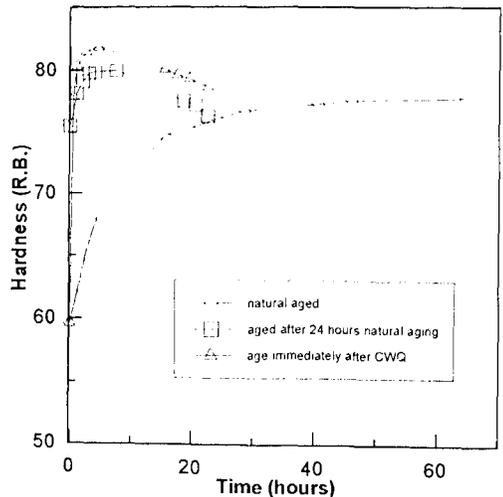


Figure 1 Hardness curves of 2009M/SiC_p composites aged at 190°C and natural aged.

Microstructure

Three phases are observed in the peak aged P/M alloy which was used for the fabrication of the composite. The precipitate phases formed in the P/M alloy are Ω (Al_2Cu on $\{111\}$), S' (Al_2CuMg), and θ' (Al_2Cu). It was shown by Polmear et al. [1, 2] that adding small amounts of Ag to Al-Cu-Mg alloys with a proper Cu/Mg ratio would promote precipitation of the semi-coherent phase. The Ω phase forms on the $\{111\}$ planes in the aluminum matrix as a uniform dispersion of large but very thin hexagonal-shaped plates. In Figure 2, the P/M alloy sample aged at 190°C for 5 hours, with the TEM specimen tilted to the $[001]$ orientation, all three phases can be seen. In the micrograph, the Ω phase is present in considerable amounts and inclined to the $\langle 001 \rangle$ orientation appearing as a plate with some fringes. The S' phase formed on the $\{001\}$ planes with a needle morphology and appears as lines and dots due to viewing the needles end on. The θ' phase formed on the $\{001\}$ planes with a plate morphology and appears as lines. It is very difficult to distinguish the S' and θ' phase from visual appearance. The microstructure of the composite aged at ambient temperature for 8 days shows no formation of precipitate or G.P. zones which suggests that the hardness increase at room temperature is due to the internal stress relaxation (Figure 3). There are no S' reflections observed in the corresponding electron diffraction pattern. These observations are consistent with DSC result from literature. The DSC results obtained by other researchers showed that formation of G.P. zones was suppressed because of the presence of reinforcements [6]. It was observed that dislocations are unevenly distributed in the matrix which is well documented in the literature. Dislocations form preferentially near particles and between closely spaced particles.

TEM observations indicate that the proposed Ω phase does not form in large quantities in the artificially aged 2009M/SiC_p composite. One reason is that the composition of matrix has a lower Cu/Mg ratio than originally proposed. The originally proposed Cu/Mg ratio was 7, however, the aluminum powder received had a ratio of approximately 4.3. and was about 4.1 in the final composite sample. A high Cu/Mg ratio promotes the Ω phase formation in the matrix whereas low Cu/Mg ratio leads to S' phase formation. The high Si content in the composite is another reason. Previous results indicate as little as 0.1 percent Si content could hinder the Ω formation. The composite contained 0.16% Si whereas the P/M sample that contained Ω had only 0.07% Si. A few σ precipitates [7] were observed in the matrix of the composite, but the amount and size of σ phase was very small. The primary precipitate formed in the 2009M/SiC_p composite matrix is the S' phase. Figure 4 shows the general microstructure in the composite matrix. The sample was aged at 190°C for 5 hours immediately after water quench. The micrograph shows the rods of S' precipitates along $\langle 001 \rangle$ orientation which was confirmed by the selected area electron diffraction. It is worth noting that the S' phase in the matrix is small in size, uniformly distributed and of high density. This is different from microstructures previously reported in the literature for Al-Cu-Mg composites [8]. It is possible that the Ag addition reduces the nucleation barrier and interfacial energy and the internal stresses increase the driving force for the S' formation. Therefore the S' precipitates possess a small size and a uniform distribution. The preferential nucleation of S' precipitates on dislocations was also observed. An example of the preferential growth of sheets of precipitates at dislocation sites is illustrated by the micrograph provided in Figure 5 which shows dislocations located around a particle in the matrix of the composite decorated with S' .

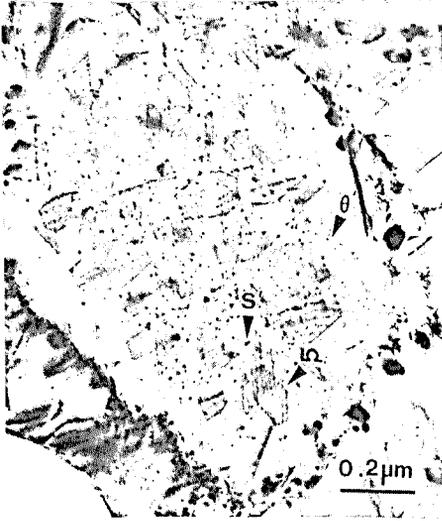


Figure 2 Three phases Ω , s' , and θ' formed in 2009M P/M alloy.

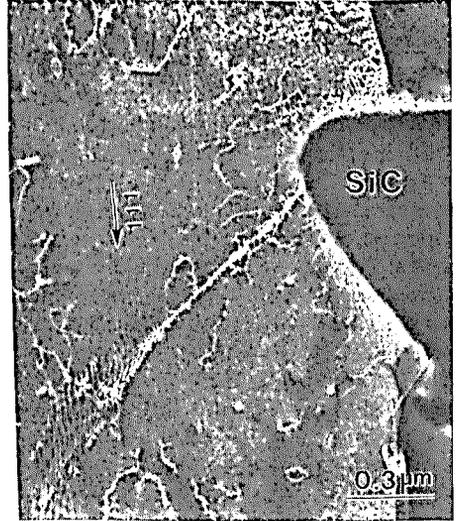


Figure 3 The microstructure of 2009M/SiC_p after natural aging 8 days, no precipitates formed.

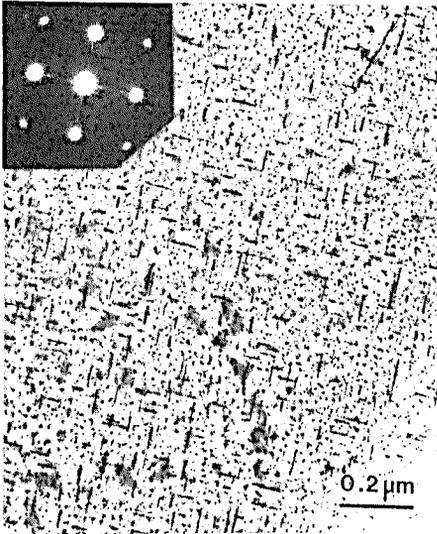


Figure 4. The S' and θ' precipitates uniformly distributed in the 2009M/SiC_p, aged at 190°C for 5 hours.

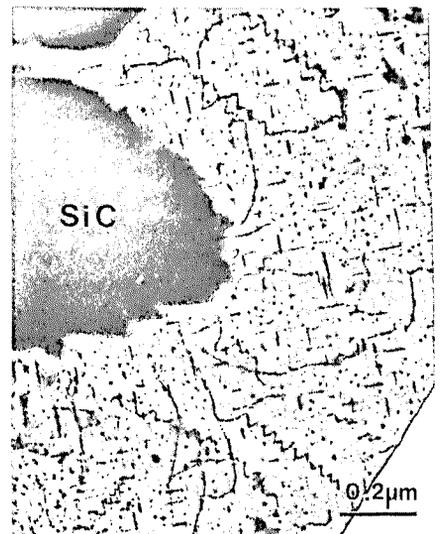


Figure 5. The dislocations around SiC particles decorated by s' precipitates.

Mechanical Properties

The results of tensile tests are listed in the TABLE II. Compared with standard 2009/SiC_p composite (Cu-3.6%, Mg-1.3%, Fe-0.7%, Al-balance and SiC_p-19v/o), the peak aged 2009M/SiC_p composite has a similar tensile strength, yield strength and modulus. The significant difference is the total elongation of the composites, the 2009M/SiC_p composite has a 5.6% total elongation instead of 3.2% for the 2009/SiC_p composite [9]. The natural aged composite possesses a total elongation over 10% (the measurement of the elongation was limited by the range of the extensometer in the experiment). The improvement of the ductility

TABLE II. Tensile Strengths of 2009M/SiC_p Composite

Conditions	UTS (MPa)	ϵ_T (%)	σ_y (MPa)	E (GPa)
P.A. 190°C, 3.5 h	520.6	5.6	424.0	91.0
P.A.+150°C, 125 h	530.2	5.4	451.6	88.3
P.A.+150°C, 500h	510.2	5.1	418.5	87.6
P.A.+500 cycles 150/32 °C	516.4	5.0	424.0	98.6
natural aged * (8 days)	499.9	> 10	304.8	93.8
P.A. tested at 150°C	434.4	4.5	389.6	80.7
P.A. tested at 177°C	395.1	5.0	362.0	72.4

* one tensile test, others are average of 2 tensile tests

suggests that the 2009M/SiC_p composite could have a higher fracture toughness. After thermal exposure 500 hours at 150°C, the tensile properties of the composite do not change

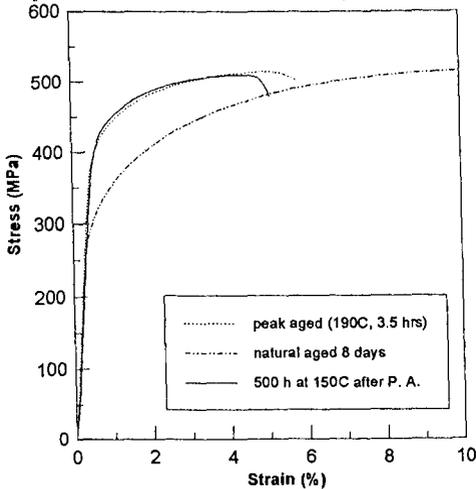


Figure 6 stress-strain curves of 2009M/SiC peak aged and exposed at 150°C for 500 hours after P.A.

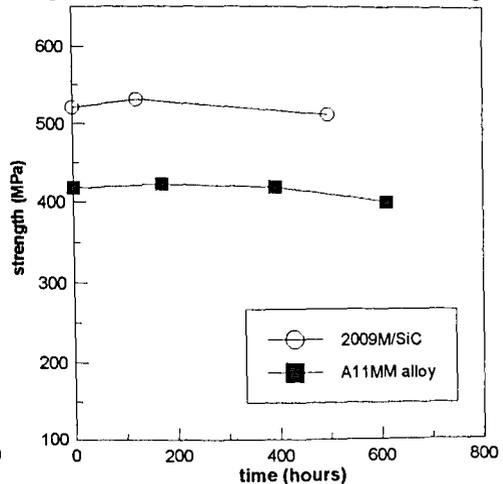


Figure 7 UTS of 2009M/SiC and A11MM I/M alloys as a function of exposure time at 150°C.

significantly. Figure 6 shows the stress-strain curves of the composites peak aged and after an exposure at 150°C for 500 hours. Two composites demonstrate similar stress-strain curves except near the end of curves. After exposure at 150°C for 500 hours, the total elongation of the composite was reduced by about 9% as compared to the peak-aged material. Figure 7 depicts the tensile strength of the composite as a function of time exposed at 150°C. The graph shows a very small drop of the UTS, the trend is the same as that for an experimental alloy (A11MM) which is strengthened by the Ω phase. The UTS of the A11MM alloy was back-calculated from shear strength data. These tensile results suggest that the composite has excellent thermal stability which is desired for aircraft application. Thermal cycling of the composite between 150°C and 32°C for 500 times does not reduce the UTS. Other studies have shown that thermal cycling leads to overaging and strength loss [10]. Elevated temperature tensile tests at 150°C and 177°C were observed to reduce the yield strength by approximately 8 and 15% respectively.

S' Phase Coarsening Behavior

The microstructural evolution at elevated temperature will directly affect the mechanical properties of the composite. A literature search indicated that there is very little S' coarsening data available. One of the reasons is that in most Al-Cu-Mg composite systems, the non-uniformity of the S' phase in the matrix makes it difficult to analyze. The size of S' precipitates as a function of thermal exposure was measured from the TEM micrographs taken in this investigation. This preliminary study shows that S' size is about ten times smaller than that reported in the literature data on S' in other materials, such as 2618, 2124, 2009/SiC, and 2124/SiC, after limited thermal exposure.

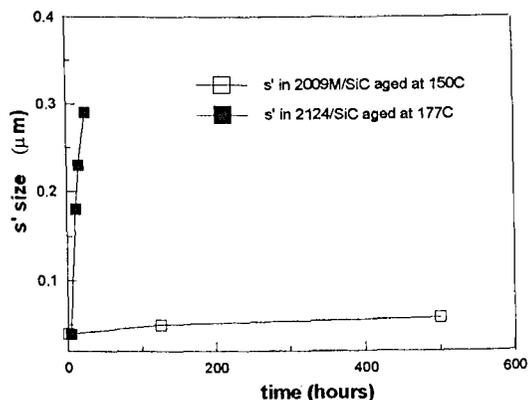


Figure 8 S' phase coarsening behavior in the different composites and temperatures.

Figure 8 shows the coarsening behavior of the S' precipitates in 2009M/15v/o SiC_p composite aged at 190°C for 3.5 hours then exposed at 150°C. This data is compared with S' precipitate coarsening in a 2124/13v/o SiC_w composite aged at 177°C [8]. Although the heat treatment conditions were different for the two composite samples, the trends of the S' phase demonstrate a great difference in the two materials. Figure 9 shows the fine S' size and uniform distribution in the 2009M/SiC_p composite thermally exposed at 150 for 500 hours after peak-aging. The limited coarsening after this thermal treatment is obvious (compare with Figure 4). Figure 10 shows histograms of the size distribution for the peak aged sample and after thermally exposing at 150°C for 500 hours. The mean size only increases from 0.043 μm to 0.059 μm. This compares with an increase of from 0.04 μm to almost 0.3 μm in just 35 hours for the 2124/SiC composites in reference 8.



Figure 9. The s' size and distribution in the 2009M/SiC_p thermally exposed 500 hours at 150°C after peak aging.

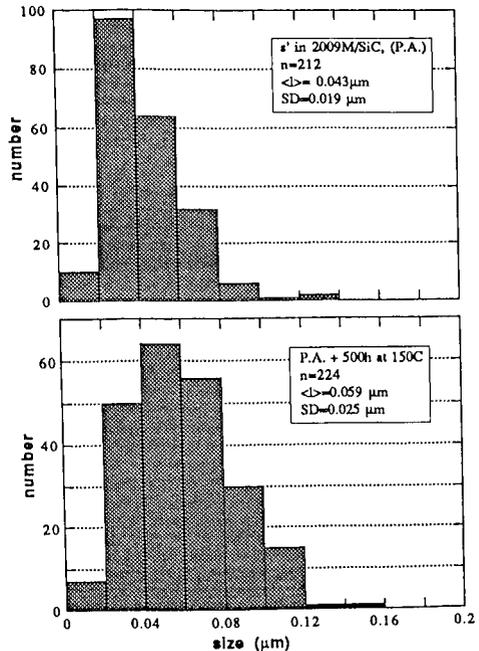


Figure 10. The histogram of the s' size for the 2009M/SiC_p composite.

CONCLUSIONS

The anticipated Ω phase was not obtained in the matrix of the 2009M/SiC composite. It is felt that this is due to the low Cu/Mg ratio and the unreasonably large amount of Si which has been shown to inhibit Ω phase formation.

The matrix microstructure was composed predominately of very small, uniformly distributed S' phase. The S' precipitates exhibited considerable thermal stability in that they showed very little coarsening after 500 hours at 150°C. This is considerably better than literature data on similar composite systems.

The tensile strength, yield strength, and elongation to failure were 521 MPa, 424 MPa and >5% respectively for the peak aged condition and did not decrease appreciably with prolonged thermal exposure at 150°C. Naturally aged samples gave a UTS of 500 MPa, a yield strength of 305 MPa and elongation of >10% after 24 hours.

Elevated temperature tensile tests at 150°C and 177°C gave a reduction in yield strength of 8% and 15% respectively.

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