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Heat Treatment Effect on Spray Formed Al/SiC Composite

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Abstract: The aim of this investigation was to study the effect of a series of heat treatments on spray formed metal matrix composites. The composite material was produced by spray forming AA 7475 aluminium alloy and co-depositing silicon carbide particles at 20 % volume fraction. The microstructure was characterised in the as received and heat-treated solution, aged and overaged conditions by optical and scanning electron microscopy. The effect of the heat treatments on the mechanical properties was evaluated by Brinell hardness and tensile testing. The results indicate that solution at 520 °C for two hours was ineffective for homogenising the matrix microstructure. At higher solution temperature 570 °C for two hours most of the precipitates dissolved. The peak hardness was achieved only after ageing at 150 °C for ten hours. These observations indicated that the composite precipitation kinetics was slower, when compared to data reported elsewhere in metal matrix composites obtained by other techniques and in similarly non-reinforced material. The fractography analysis indicates that the tensile specimens rupture occurred mainly through the Al/SiC interface.

Introduction

The composite material is characterised by its inhomogeneities, containing at least two distinct phases, where one is continuous being denominated the matrix and the other is usually called the reinforcement. The reinforcement can be in the shape of whiskers, particles, short fibres and continuous fibres[1]. The MMC's mechanical properties can be controlled by the reinforcement type, its volumetric fraction, matrix chemical composition and microstructure, the interaction between the non deformable reinforcement particles and the matrix, the formation of highly complex dislocation substructures in the matrix, the matrix residual stresses and precipitation caused by processing conditions such as hot extrusion and heat treatments[2].

A feasible technology for the obtention of metal matrix composites - MMC is based on the co-deposition of the reinforcement during the spray forming of the matrix alloy. The main objective of spray forming is to obtain semi-finished product shapes with high density and simultaneously obtaining a fine, uniform grain-size microstructure without macro-segregation. This process allows the obtaining of high performance alloys, highly alloyed metals; *in situ* alloyed materials and metal matrix composites[3-6].

Experimental

Material. The material used in this work was a spray formed (Osprey Process) aluminium AA 7475 (nominal composition in w. %: 6 % Zn, 2.1 % Mg, 1.4 % Cu, Al balance) reinforced with silicon carbide particles (20 % volume fraction). A limited amount of material was supplied as 15 mm diameter extruded bars.

Solution treatment. The material was solution treated in two distinct conditions. Part of the samples were solution treated at 520 ± 5 °C for 2 h and the other was solution treated at higher temperature, 570 ± 5 °C for 2 h. The samples were quenched in water at room temperature, aiming to achieve maximum supersaturation of alloying elements in preparation for subsequent ageing.

Ageing and overaging. After solution treatment, some samples were aged in a thermostatic oil bath at 120 ± 2 °C for 24 h. The samples were cooled in water until room temperature. After the solution and ageing heat treatments, samples were overaged at 150 ± 2 °C for 10 h followed by cooling in water at room temperature. A second overaging heat treatment was performed at 250 ± 5 °C for 10 h after the previous overaging.

Microstructural characterisation . Samples for metallographic examination were cut either by a diamond wheel or by spark machining. Samples taken from the material, before and after heat treatments, were mounted in thermoplastic material and subsequently ground using silicon carbide paper down to grit 800. Then, the samples were mechanically lapped to 1/4 μm diamond finish. Colloidal silica suspension was finally used to polish the surface of the samples. Fractured tensile test specimens on the longitudinal section were analysed in the SEM.

An extensive microstructural characterisation of this Al/SiC composite in the as received and heat-treated conditions, supported by transmission electron microscopy was undertaken and it was published elsewhere[7].

Hardness and tensile testing. Brinell hardness measurements (load 62.5 kg and steel ball 2.5 mm) were carried out in the as received composite condition and after solution, ageing and overaging heat treatments. Tensile tests measurements were performed on the material in the as received and heat-treated conditions. The tensile testing were done according to DIN 50125 [8].

Results and discussion

Microstructural characterisation. The optical micrograph, Fig. 1a, shows the overall microstructure of the as received composite material in the transverse direction. The composite material showed a hydrostatic density slight inferior, 2.861 g/cm^3 than the theoretical density 2.883 g/cm^3. This difference was mainly due to porosity associated to SiC agglomerations. In these agglomerations, there was a lack of matrix penetration during the spray forming. The SiC particles were loose inside the agglomerations and were removed during polishing. It was not possible do discern any interfacial reaction between the SiC particles and the matrix by means of optical microscopy. Fig. 1b, an optical micrograph of the longitudinal section of the composite material, shows the SiC particles alignment due to the hot extrusion process.

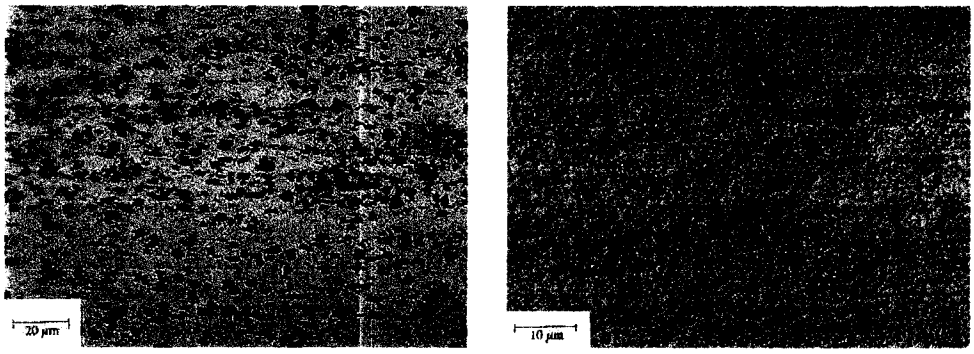

Fig. 1. Optical micrographs of the spray formed composite material. **a)** Cross-section of the composite showing the matrix, the SiC particles and some agglomerations. **b)** Longitudinal direction of the composite showing SiC particles alignment due to the hot extrusion process.

Mechanical strength

Brinell hardness. The results from the Brinell hardness measurements are shown in Fig. 2. The as received composite showed a higher hardness when compared to the solution treated material. This was probably due to the hot extrusion process and to some natural ageing at room temperature. The solution treatment at 520 ± 5 °C for 24 h caused a partial solution of the equilibrium precipitates, putting solute atoms into solid solution in preparation for subsequent ageing. Hence, this allowed the MMC softening by lessening the interference to slip by precipitates on crystallographic planes, and, by changing the second phase particles size. As for the solution treatment at higher temperature 570 ± 5 °C for 2 h, the hardness decreased further indicating a better solubilization, subsequent decreasing the interaction of moving dislocations with precipitates.

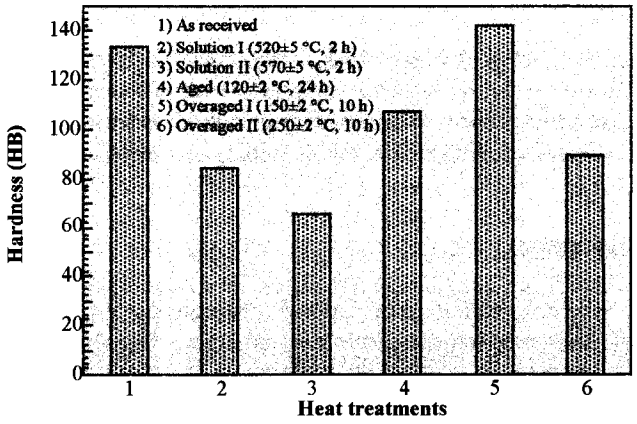

Fig. 2. Brinell hardness variation according to the heat treatments used for the spray formed composite material.

As expected, the aged composite showed higher hardness than the solution treated material. In this heat treatment condition, the formation of very fine and evenly distributed precipitates in the matrix also called Guinier-Preston zones (GPZ) started. This fine and coherent precipitation was an impediment to dislocation movement thus, increasing hardness.

The hardness value for the overaged composite material at 150 ± 5 °C for 10 h, denoted as condition overaged I in Fig. 2, was higher than the aged samples. This hardness rise is a circumstantial evidence of the low precipitation kinetics observed for this composite material. In addition, a maximum age hardening would occur with the presence of critical dispersion of GPZ or an intermediate precipitate, or both. After the peak condition, the hardness at some stage should be reduced. This was attained in the second overaging heat treatment 250 ± 5 °C for 10 h, denoted as condition overaged II. This overaged condition led to continuous precipitates coarsening as expense of its number.

Tensile testing. The mechanical strength of an alloy is closely related to the tension required to move a dislocation across its slip plane. This stress will be higher as the barriers against dislocation movement build up. The mechanical strength variation of the studied composite could be explained by a high dislocation density generated by thermal expansion difference between matrix and reinforcement and by the precipitation arrangements due to heat treatments. Fig. 3 shows the results obtained of the yield strength at 0.2 % deformation and the ultimate tensile strength, for the composite in the as received condition, solution heat treated at 520 °C for 2 hours, aged at 120 °C for 24 hours and overaged at 150 °C for 10 hours. It should be mentioned that the yield strength data for the overaged composite was not available due to problems occurred during the tensile tests.

It can be seen in Fig. 3 that the yield and ultimate tensile strength of the solution treated composite, had a small decrease in relation to the as receive composite. This was due to, as was observed by transmission electron microscopy[7], that the solution heat treatment at 520 °C for 2 hours was not enough to solubilize all the precipitates and intermetallics already existing in the as received composite. Therefore, this solution treatment caused a small decrease in the composite strength by partially dissolving second phase particles that acted as barriers to the dislocation movement.

The yield and ultimate tensile strength for the aged 120 °C for 24 hours, composite were higher than the as received material. This was due to a higher number of precipitates and to the presence of Guinier-Preston zones (GPZ) in the aluminium matrix.

For the overaged composite material, 150 °C for 10 hours, besides the heat treatment produced a hardness increase, see Fig. 2, it does not materialised as an improvement in the ultimate tensile strength. As matter of fact, there was a small decrease in the tensile strength compared to the aged composite. This was probably due to precipitates coarsening and solute depletion at grain boundaries.

There was a narrow change in ductility of the solution treated composite, 520 °C for 2 hours, in relation to the as received material, see Fig. 4. This ductility increase was due to the uniform strain since there was a small reduction of area. For the composite aged at 120 °C for 24 hours and overaged at 150 °C for 10 hours, the increase in the mechanical strength and in the hardness, caused a decrease in the ductility when compared to the composite material in both, as received and solution treated condition.

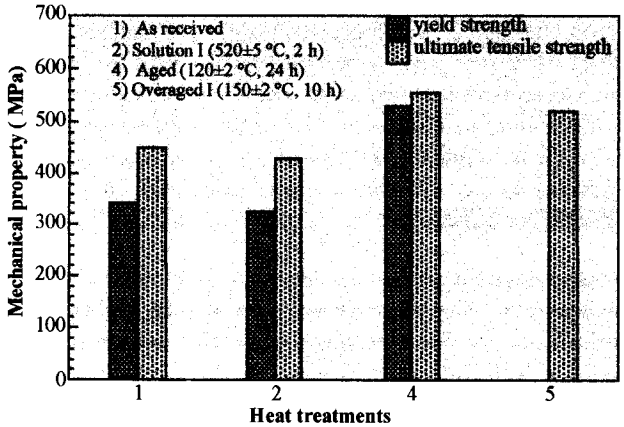

Fig. 3. Variation of yield and ultimate tensile strength of the composite material in function of the heat treatments.

Fractography. Fig. 5 shows the micrographs obtained by SEM, from the longitudinal direction of tensile tested composite specimens in the as received, solution treated 520 °C for 2 hours, aged 120 °C for 24 hours and overaged 150 °C for 10 hours, respectively. It is observed that the fracture occurred at the Al/SiC interface and that there was many cracked SiC particles. Many publications have reported the Al/SiC interface decohesion occurring during tensile tests[9-11]. According to Lloyd[10], the interface decohesion during tensile testing, occurs when the stress concentration exceeds the particles/matrix bonding strength. This interfacial decohesion can also be attributed to the presence of fragile oxides at the interface Al/SiC, such as MgO and Mg_2Si that have precipitated during ageing and overaging heat treatments. All the micrographs show fractured SiC particles. According to literature[11], this tendency of the SiC particles to crack increases with its volumetric fraction and with the increase in the maximum stress during plastic deformation. For uniaxial tensile testing, the reinforcement cracks usually have a normal orientation to the load axis.

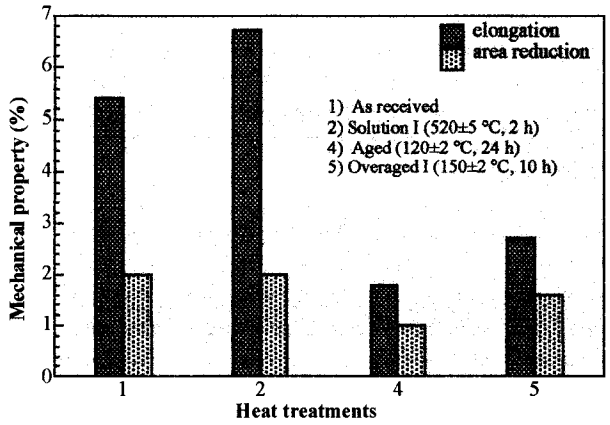

Fig. 4. Change of the elongation and the reduction of area at the fracture of the composite material in function of the heat treatments.

It is noticeable in Fig. 5 that, the larger SiC particles are more cracked than the small particles. Lewandowski and Lui [9], reported that this was mainly due to a higher probability to

find critical size flaws in larger SiC particles. According to Davidson[12] these defects may already exists before the composite was made. They could be originated during the SiC comminution or during the composite processing.

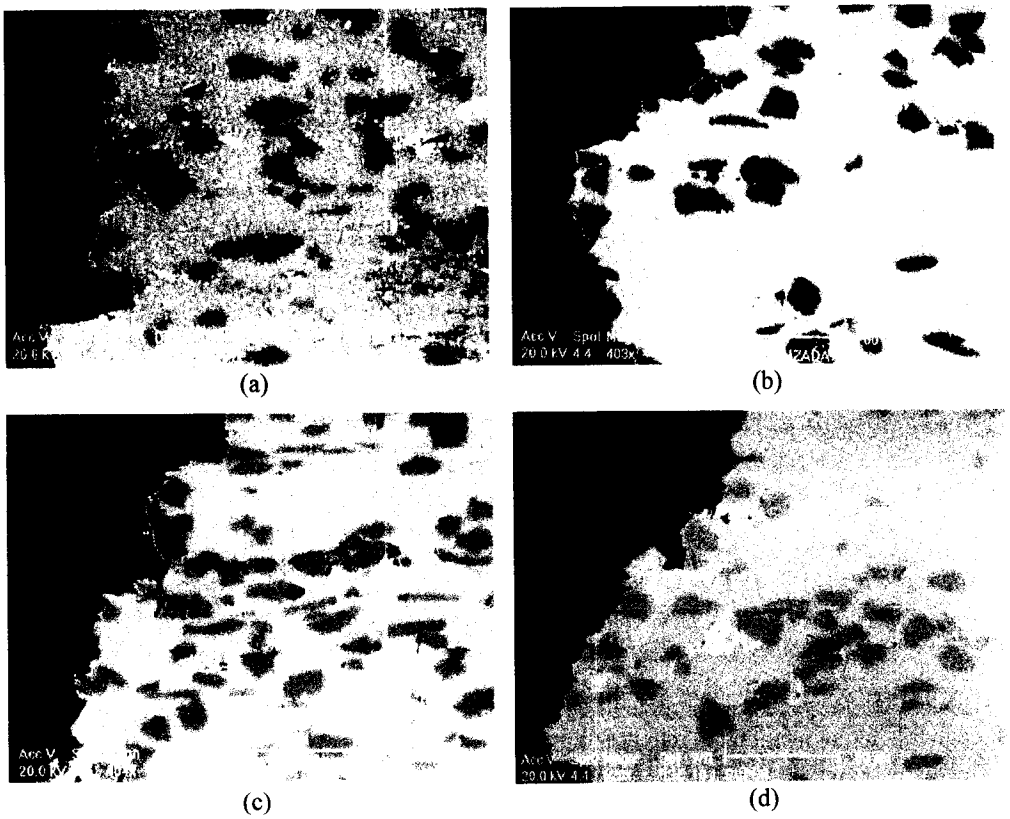

Fig. 5. SEM micrographs in the longitudinal direction of the tensile test specimens. a) As received. b) Solution 520 °C for 2 h. c) Aged 120 °C for 24 h. d) Overaged 150 °C for 10 h.

Conclusions

The composite peak hardness was attained in the 150 ± 2 °C for 10 h ageing condition. This fact demonstrated a composite slow precipitation kinetics when compared to results from other works in similar material, subjected to similar heat treatments. This was probably due to precipitation of the Mg_2Si phase at the Al/SiC interface.

The yield and ultimate tensile strength were higher for the composite material aged at 120 °C for 24 hours. This was likely to be caused by the presence of a larger number of precipitates and to Guinier-Preston zones that interfere with the dislocation movement.

The fractography evaluation showed that the fracture was mainly through the interface Al/SiC and that the majority of the larger SiC particles were cracked.

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