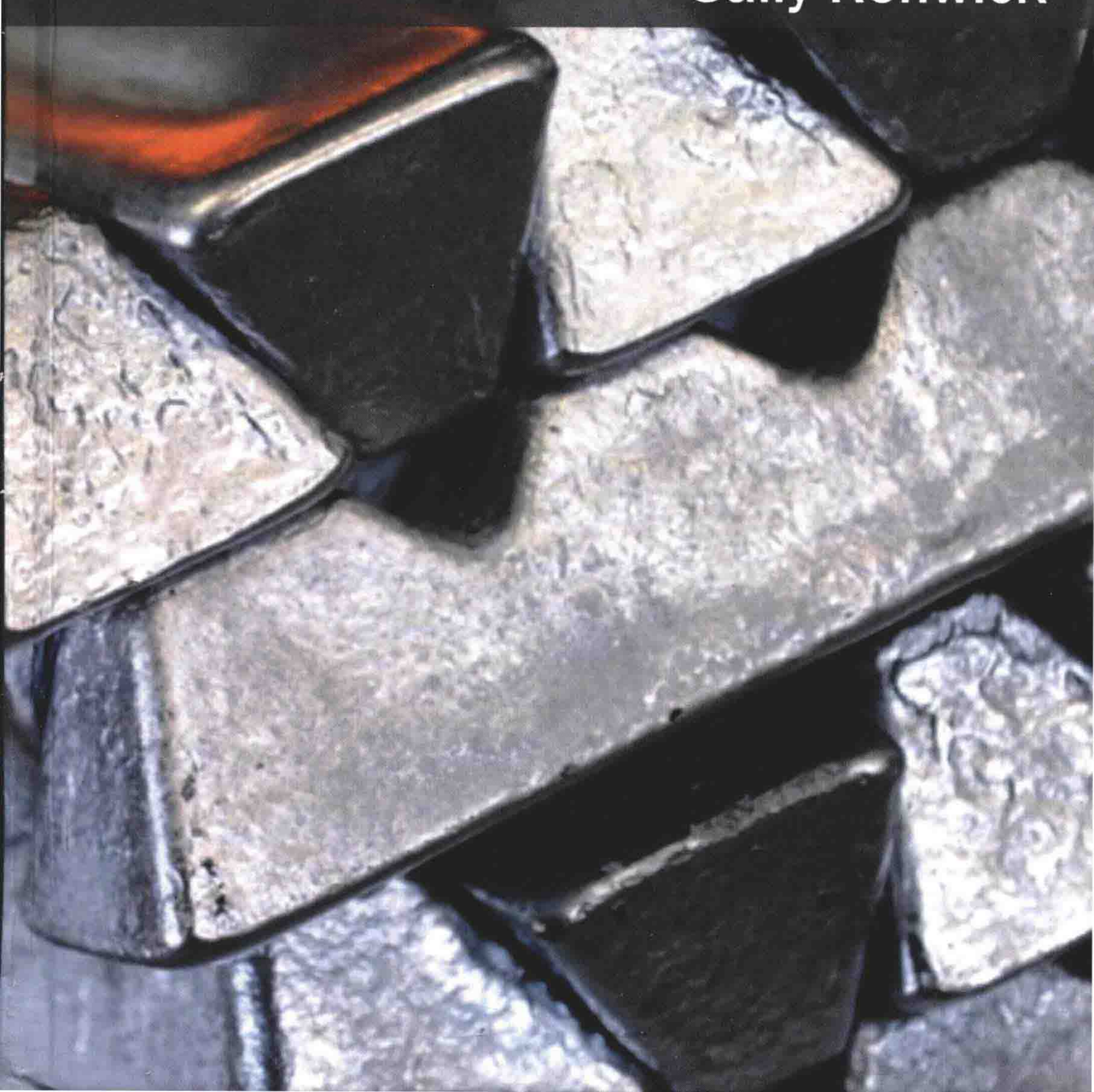
Deterioration, Structures and Analysis of ALUMINIUM ALLOYS

Sally Renwick



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Edited by **Sally Renwick**

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Preface

There has been a breakthrough in the production of aluminium alloys. New procedures of welding, casting, forming and surface modification have emerged to advance structural integrity of aluminium alloys. The book covers major topics grouped under two sections namely, "Corrosion and Mechanical Damage on Aluminium Alloys" and "Micro-Nano Structures and Image Analysis". This book aims to serve the needs of a broad spectrum of professionals ranging from academic to industrial communities by providing latest information. It also serves the purpose of assisting technocrats, entrepreneurs and other individuals interested in the application and production of aluminium alloys.

Various studies have approached the subject by analyzing it with a single perspective, but the present book provides diverse methodologies and techniques to address this field. This book contains theories and applications needed for understanding the subject from different perspectives. The aim is to keep the readers informed about the progresses in the field; therefore, the contributions were carefully examined to compile novel researches by specialists from across the globe.

Indeed, the job of the editor is the most crucial and challenging in compiling all chapters into a single book. In the end, I would extend my sincere thanks to the chapter authors for their profound work. I am also thankful for the support provided by my family and colleagues during the compilation of this book.

Editor

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Part 1

Corrosion and Mechanical Damage of Aluminium Alloys

1

Deformation Characteristics of Aluminium Composites for Structural Applications

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1. Introduction

Silicon carbide (SiC) particulate-reinforced aluminium matrix composites (AMC) are attractive engineering materials for a variety of structural applications, due to their superior strength, stiffness, low cycle fatigue and corrosion fatigue behaviour, creep and wear resistance, compared to the aluminium monolithic alloys. An important feature of the microstructure in the Al/SiC composite system is the increased amount of thermal residual stresses, compared to unreinforced alloys, which are developed due to mismatch in thermal expansion coefficients of matrix and reinforcement phases. The introduction of the reinforcement plays a key role in both the mechanical and thermal ageing behaviour of the composite material. Micro-compositional changes which occur during the thermo-mechanical forming process of these materials can cause substantial changes in mechanical properties, such as ductility, fracture toughness and stress corrosion resistance.

The satisfactory performance of aluminium matrix composites depends critically on their integrity, the heart of which is the quality of the matrix/particle reinforcement interface. The nature of the interface depends in turn on the processing of the AMC component. At the micro-level, the development of local concentration gradients around the reinforcement can be very different to the nominal conditions. The latter is due to the aluminium alloy matrix attempt to deform during processing. This plays a crucial role in the micro-structural events of segregation and precipitation at the matrix-reinforcement interface.

The strength of particulate-reinforced composites also depends on the size of the particles, interparticle spacing, and the volume fraction of the reinforcement [1]. The microstructure and mechanical properties of these materials can be altered by thermo-mechanical treatment as well as by varying the reinforcement volume fraction. The strengthening of monolithic metallic material is carried out by alloying and supersaturating, to an extent, that on suitable heat treatment the excess alloying additions precipitates out (ageing). To study the deformation behaviour of precipitate hardened alloy or particulate reinforced metal matrix composites the interaction of dislocation with the reinforcing particles is much more dependent on the particle size, spacing and density than on the composition [2]. Furthermore, when a particle is introduced in a matrix, an additional barrier to the movement of dislocation is created and the dislocation must behave either by cutting through the particles or by taking a path around the obstacles [3].

At present, the relationship between the strength properties of metal matrix composites and the details of the thermo-mechanical forming processes is not well understood. The kinetics of precipitation in the solid state has been the subject of much attention. Early work on growth kinetics has been developed for the grain boundary case [4] and for intragranular precipitation [5]. These approaches have been integrated to produce a unified description of the inter- and intra-granular nucleation and growth mechanisms [6, 7]. More recently, successful attempts have been made to combine models of precipitate growth at interfaces with concurrently occurring segregation in aluminium alloys [8]. Studies of the relation between interfacial cohesive strength and structure have only recently become possible. This is due to of remarkable advances in physical examination techniques allowing direct viewing of interface structure and improved theoretical treatments of grain boundary structure.

The ability of the strengthening precipitates to support the matrix relies on the properties of the major alloying additions involved in the formation of these precipitates. The development of precipitates in Al-based alloys can be well characterised through heat treatment processing. Heat treatment affects the matrix properties and consequently the strain hardening of the composite. Furthermore, the distribution and concentration of these precipitates greatly affect the properties of the material where homogenous distribution of small precipitates provides the optimum results.

The role of the reinforcement is crucial in the microdeformation behaviour. The addition of SiC to aluminium alloy increases the strength and results in high internal stresses, in addition to the ones caused by the strengthening precipitates. Furthermore, the SiC reinforced particles are not affected by the heat treatment process. A great deal of attention has been recently devoted to understanding the strengthening mechanisms in metal matrix composites, which are distinguished by a large particulate volume fraction and relatively large diameter. Another important matter in understanding and modelling the strength of particulate MMCs is to consider the effect of particle shape, size and clustering [9-11], as well as the effects of clustering of reinforcement on the macroscopic behaviour and the effects of segregation to the SiC/Al interfaces [12]. Important role also play the effects of casting condition and subsequent swaging on the microstructure, clustering, and properties of Al/SiC composites [13].

Aluminium honeycomb sandwich panel constructions have been successfully applied as strength members of satellites and aircraft structures and also in passenger coaches of high-speed trains such as the TGV in France and the Shinkansen in Japan [14]. However, the cost of producing the all welded honeycomb structure has been a key factor for not using this technology on mass production rate. Recent developments in manufacturing methods have given rise to a range of commercially viable metallic foams, one being Alulight. In comparison to aluminium honeycomb core construction, metallic foams show isotropic properties and exhibit non linear mechanical deformation behaviour. The metallic foams have the potential to be used at elevated temperatures up to 200°C [15]. They also have superior impact energy absorption and improved strength and weight savings. However, the successful implementation of both aluminium honeycomb and metallic foam sandwich panels for aerospace and transportation applications is dependent upon an understanding of their mechanical properties including their resistance to fatigue crack growth and the resistance of aluminium alloys to environmentally induced cracking or stress corrosion cracking.

This chapter discusses first the relationship between the interfacial strength with the thermo-mechanical deformation process and the resulting macroscopic mechanical behaviour of particle-reinforced aluminium matrix composites. Micro-compositional changes which occur during the thermo-mechanical processing of these materials can cause substantial changes in mechanical properties such as ductility, fracture toughness, or stress corrosion resistance. A mico-mechanistic model will be presented for predicting the interfacial fracture strength in AMCs in the presence of magnesium segregation. Finally, the use of powerful nondestructive evaluation tools, such as infrared thermography, will be discussed to evaluate the state of stresses at the crack tip and to monitor fatigue crack growth in particle-reinforced aluminium alloy matrix composites.

In the second part of the chapter the structural integrity of Aluminium Honeycomb (HC) sandwich panels is compared with the new core material concept of aluminium foams. Aluminium Honeycomb sandwich panels are used to reduce weight whilst improving the compressive strength of the structure with the aerospace industry being one of the prime users of HC sandwich panels for structural applications. The cost of producing all welded HC structures has been the key factor for not using this technology on a mass production basis. An alternative to the aluminium honeycomb (HC) sandwich panels is the metallic foam sandwich panel, which has been gaining interest in the same field. These foams are anisotropic, exhibit non-linear mechanical behaviour, and they have the potential for use at temperatures up to 200°C. They have superior impact energy absorption, and improved strength and weight savings. The lower weight as compared to conventional solid wrought aluminium alloys will mean a reduction in fuel consumption thus providing economical savings.

This chapter attempts to investigate whether aluminium honeycomb sandwich panels, with their homogenous hexagonal core can be successfully replaced by metallic foam sandwich panels, which have an inhomogeneous core. A successful replacement would improve the confidence of manufacturers in the exploitation of this new material in replacing traditional materials. Current levels of understanding of cyclic stressing in metallic foam sandwich panels is limited and models of long term understanding of this aspect of failure are very important for both aerospace and automotive sectors. Burman et al [16] suggests that fundamental fatigue models and concepts proven to work for metals can be applied to metallic foam sandwich panels. A study by Shipsha et al [17] investigated experimentally both metallic foam and other cellular foams, using compact tension specimens. Shipsha's et al research is extremely interesting and implies that a sandwich panel should be considered whole and not two separate entities. Banhart and Brinkers has shown that it is very difficult to detect the features leading to fatigue failure in metallic foams due to the metallic foam being already full of micro cracks [18]. However, Olurin [19] investigation suggest that the fatigue crack growth mechanism of Alulight and Alporas foam is of sequential failure of cell faces ahead of crack tip. The main conclusion is that for a given ΔK, the fatigue crack propagation rate, da/dN decreases with increasing density and for a given stress intensity, the fatigue crack propagation rate increases when the mean stress is increased.

Current levels of understanding of cyclic stressing in aluminium foams is limited and models of long term understanding of this aspect of failure are important for both aerospace and automotive sectors. This is particularly important for low-density foam and honeycomb materials which despite thin ligament thickness, have good properties in compression. A method of analysis is proposed to predict life expectancy of aluminium honeycomb and metallic foam sandwich panels.

2. SiC-particulate reinforced aluminium matrix composites

2.1 Materials

Aluminium - silicon - magnesium alloys (A359) are important materials in many industrial applications, including aerospace and automotive applications. The alloys from the Al-Si-Mg system are the most widely used in the foundry industry thanks to their good castability and high strength to weight ratio. Materials based on A359 matrix reinforced with varying amounts of silicon carbide particles are discussed in this chapter.
Four types of material are used: 1) Ingot as received A359/40%SiC, with an average particle size of 19±1 micron, 2) Ingot as received A359/25%SiC, with an average particle size of 17±1 micron, 3) Hot rolled as received A359/31%SiC with an average particle size of 17±1 micron and 4) Cast alloy as received A359/30%SiC with particles of F400grit, with an average particle sizes of 17±1 micron. Table 1, contains the details of the chemical composition of the matrix alloy as well as the amount of silicon carbide particles in the metal matrix composites.

TYPES	Si	Mg	Mn	Cu	Fe	Zn	SiC
INGOT A359	9.5	0.5	0.1	0.2	0.2	0.1	40
INGOT A359	9.5	0.5	0.1	0.2	0.2	0.1	25
CAST A359	9.5	0.5	0.1	0.2	0.2	0.1	30
ROLLE D A359	9.5	0.5	0.1	0.2	0.2	0.1	31

Table 1. Types and composition of the material

The microstructure of such materials consists of a major phase, aluminium or silicon and the eutectic mixture of these two elements. In this system, each element plays a role in the material's overall behaviour. In particular, Si improves the fluidity of Al and also Si particles are hard and improve the wear resistance of Al. By adding Mg, Al - Si alloy become age hardenable through the precipitation of Mg_2Si particulates.

2.2 Heat treatment

Properties in particulate-reinforced aluminium matrix composites are primarily dictated by the uniformity of the second-phase dispersion in the matrix. The distribution is controlled by solidification and can be later modified during secondary processing. In particular, due to the addition of magnesium in the A359 alloy, the mechanical properties of this material can be greatly improved by heat treatment process. There are many different heat treatment sequences and each one can modify the microstructural behaviour as desired [20]. Precipitation heat treatments generally are low temperature, long-term processes. Temperatures range from 110°C to 195°C for 5 to 48 hours. The selection of the time temperature cycles for precipitation heat treatment should receive careful consideration. Larger precipitate particulates result from longer times and higher temperatures. On the other hand, the desired number of larger particles formed in the material in relation to their interparticle spacing is a crucial factor for optimising the strengthening behaviour of the composite. The objective is to select the heat treatment cycle that produces the most favourable precipitate size and distribution pattern. However, the cycle used for optimising one property, e.g. tensile strength, is usually different from the one required to optimise a different property, e.g. yield strength, corrosion resistance.

Heat treatment of composites though has an additional aspect to consider, the particles introduced in the matrix. These particles may alter the alloy's surface characteristics and increase the surface energies [21].

The heat treatments were performed in Carbolite RHF 1200 furnaces with thermocouples attached, ensuring constant temperature inside the furnace. There were two different heat treatments used in the experiments, T6 and modified-T6 (HT-1) [21, 22].

The T6 heat treatment consists of the following steps: solution heat treatment, quench and age hardening (Fig. 1).

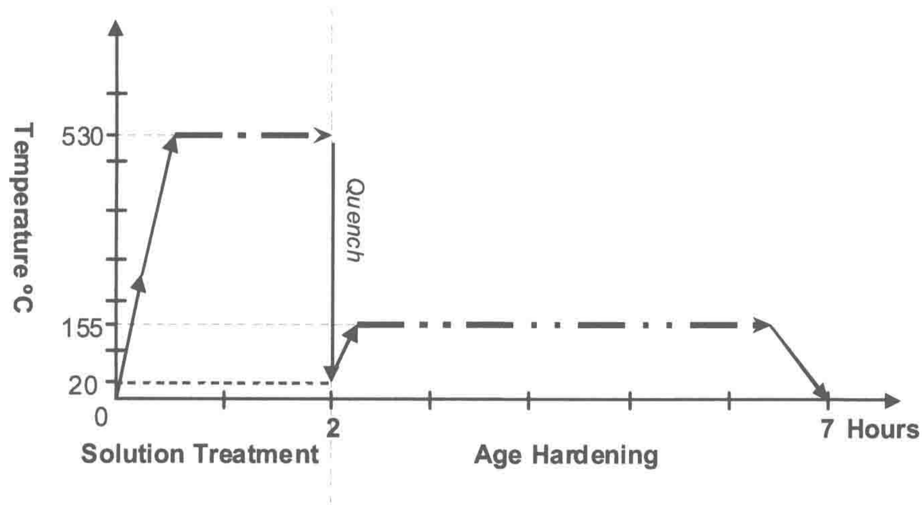

Fig. 1. T6 Heat treatment diagram showing the stages of the solution treatment for 2 hours and artificial ageing for 5h

In the solution heat treatment, the alloys have been heated to a temperature just below the initial melting point of the alloy for 2 hours at 530±5°C where all the solute atoms are allowed to dissolve to form a single phase solid solution. Magnesium is highly reactive with Silicon at this temperature and precipitation of Mg_2Si is expected to be formed. The alloys were then quenched to room temperature. In age hardening, the alloys were heated to an intermediate temperature of 155°C for 5 hours where nucleation and growth of the β' phase. The desired β phase Mg_2Si precipitated at that temperature and then cooled at room temperature conditions. The precipitate phase nucleates within the grains at grain boundaries and in areas close to the matrix-reinforcement interface, as uniformly dispersed particles. The holding time plays a key role in promoting precipitation and growth which results in higher mechanical deformation response of the composite.

The second heat treatment process was the modified-T6 (HT-1) heat treatment, where in the solution treatment the alloys have been heated to a temperature lower than the T6 heat treatment, at 450±5°C for 1 hour, and then quenched in water. Subsequently the alloys were heated to an intermediate temperature of 170±5°C for 24 hours in the age hardened stage and then cooled in air (Fig. 2).

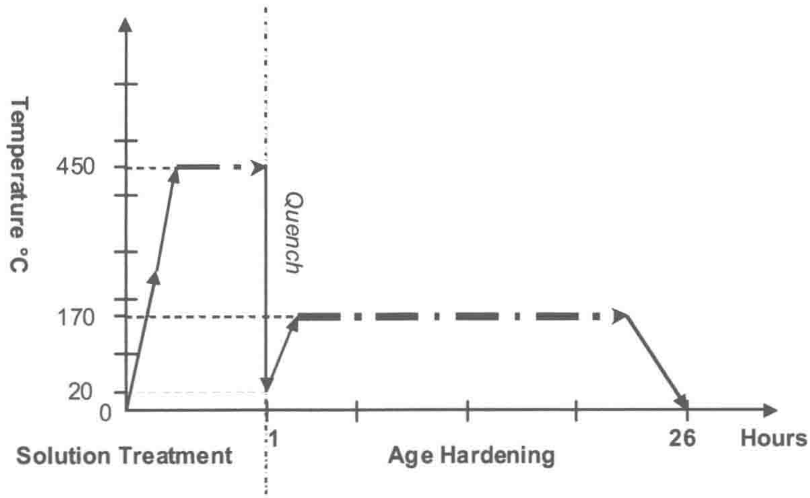

Fig. 2. Modified T6 (HT-1) showing stages of solution treatment for 1 hour and artificial ageing for 24h

In both heat treatments undesired formation of phases, like the Al_4C_3, is a possibility and selection of the solution treatment as well as the age hardening processes should be carefully considered. Temperature and time control, therefore, is extremely important during heat treatment. If the melt temperature of SiC/Al composite materials rises above a critical value, Al_4C_3 is formed increasing the viscosity of the molten material, which can result in severe loss of corrosion resistance and degradation of mechanical properties in the cast composite; excessive formation of Al_4C_3 makes the melt unsuitable for casting. In the A359/SiC composite high silicon percentage added in excess aids to the formation of some oxides (SiO_2) around the SiC reinforcement, something that retards the formation of Al_4C_3, since such oxides prevent the dissolution of SiC particles [22].

2.3 Metallographic examination

In order to analyse the microstructure, a series of sample preparation exercises were carried out, consisted of the cutting, mounting, grinding and polishing of the samples. The microstructures were investigated by SEM, EDAX, and XRD to determine the Al/SiC area percentage, size and count of particulates.

The microstructures of the examined MMCs in the as received condition have four distinct micro phases as clearly marked on the image micrograph, which are as follows: the aluminium matrix, the SiC particles, the eutectic region of aluminium and silicon and the Mg phase (Fig. 3). The distribution of SiC particles was found to be more or less uniform, however, instances of particle free zones and particle clustered zones were observed.

Matrix-reinforcement interfaces were identified by using high magnification Nano-SEM. In the as received hot rolled images the Al Matrix/SiC reinforcement interface is clearly identified (Fig. 4). These interfaces attain properties coming from both individual phases of constituents and facilitate the strengthening behaviour of the composite material.

In the modified T6 (HT-1) condition the microsturucture of the cast 30% SiC has the same phases as in the as received state, plus one rod-shape phase (Fig.5) along the matrix and at the matrix-reinforcement interface has been identified to be Mg_2Si precipitates in an early