TENSILE, FATIGUE AND IMPACT PROPERTIES OF THE
AA 2618/Al₂O₃P COMPOSITES

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Abstract

In this paper, the tensile properties (at RT and 200°C), high cycle fatigue and impact behaviour of the 2618 aluminium alloy reinforced with 5% wt. of Al₂O₃ particles (AA 2618/Al₂O₃p/5P) and 10% wt. of Al₂O₃p particles (AA 2618/Al₂O₃p/10P) were studied. The Composites were prepared by two-step mixing of stir casting method followed by forging. A significant increase in tensile strength of the MMCs, with respect to the unreinforced alloys was evidenced by the tensile test while the elongation to fracture decreased. Tensile property at 200°C is higher than the unreinforced counterpart. The high cycle fatigue tests showed that fatigue life is comparable to the monolithic alloy. The SEM analysis of the fracture surfaces showed that the presence of secondary precipitates played a significant role in the mechanism of failure in composites by promoting void nucleation at the particle-matrix interfaces, interfacial decohesion and also the failure of the particles. The impact strength is reduced with respect to the unreinforced alloy.

Keywords: Metal matrix composite, Mechanical properties, Fatigue, impact behavior, Forging

Introduction

Metal-Matrix Composites (MMCs) are used increasingly for different applications because of their higher strength and stiffness and their improved strength to weight ratio compared to the unreinforced alloys. Both short fiber and particulate reinforced MMCs are used in particular for high temperature applications [1]. Anisotropy is often increased over that of unreinforced alloys because of a heterogeneous reinforcement distribution [2-4]. Potential uses of these materials are numerous in industries and they include such areas of application as aerospace (satellite struts), defense (electronic instrument racks), automotive (drive shafts and brake discs), sports goods (golf clubs and mountain bicycle frames) and marine (yacht fittings) [5,6].

Aluminium metal matrix composites (Al MMCs) are being considered as a group of new advanced materials for its light weight, high strength, high specific modulus, low co-efficient of thermal expansion and good wear resistance properties. Combinations of these properties are not available in a conventional material [7]. Some of the early work on MMCs focused mainly on continuous fiber reinforcement. However, high cost of fibers, complex fabrication techniques and limited fabricability restricted their use to those applications where the end could justify the means [8,5]. This opened the way for the development of low cost discontinuously reinforced MMCs such as particle reinforced MMCs [9,10]. The lower cost is related to the more simple production techniques and also to the possibility of using conventional metal working processes such as extrusion, forging, rolling and even super plastic forming [11]. The improvements in properties with respect to the unreinforced alloys strongly depends on the intrinsic properties of both matrix and reinforcement (size, volume fraction and distribution) as well as on the mutual interaction between the constituents at the interfaces, but a correlation between microstructural features and mechanical behaviour is not easy. A major potential problem of particle-reinforced composites is a non-uniform microstructure often resulting from the manufacturing process, which can lead to the presence of clusters of particle or regions without the reinforcement [12,13]. This intrinsic
material inhomogeneity can give a wide scatter in strength and ductility [14] and also in the fatigue behaviour [15].

Among the variety of manufacturing processes available for discontinuous metal matrix composites, stir-casting route is generally accepted as a particularly promising route. Its advantage lies in its simplicity, flexibility and applicability to large quantity production. It is the most economical of all the available routes for metal matrix composite production [16]. When the SiC\(_p\) particles were added into the molten alloys, they were floating on the surface even though high specific density is due to high surface tension and poor wetting between the particles and melt. Gas layers might be the main factor for the poor wettability. But, in a semisolid state mixing (first mixing), gas layer break can happen due to the collision between primary \(\alpha\) Al nuclei and particle. In the second mixing, uniform particle distribution is achieved by the stirrer at above the liquidus temperature [17]. In order to implement MMCs on a wider scale in the modern industry significant effort is still required to understand the relationship between processing and in service behaviour. In particular, the forging route would provide a significant improvement in the elongation properties of the material, thereby making it less susceptible to fracture with respect to its cast counterpart. At the same time, the forging process imparts localized damage in some areas of the work piece, with detrimental effect on properties [18].

While several studies have focused on understanding the nature and influence of the reinforcement particles on matrix microstructure, complex relationships do exist between the fatigue properties and fracture characteristics of a discontinuous particulate-reinforced aluminium alloy based MMC. These include: a) the intrinsic properties of the matrix (composition, aging condition and microstructure) b) intrinsic properties of the particulate reinforcement phase (composition, morphology, size and volume fraction), c) the influence of secondary processing on microstructure and d) the influence of test parameters such as nature, type and magnitude of loading and extent of cyclic plasticity [19]. Presence of discontinuous particulate reinforcement in a ductile aluminium alloy metal matrix will alter the precipitation kinetics of the material during heat treatment and thus the microstructure of the composites compared with the unreinforced monolithic counterpart. A change in intrinsic microstructural features will exert an appreciable influence on mechanical response and resultant fracture behaviour [20]. Although, tensile properties and fatigue properties of particle reinforced aluminium alloy matrix composites have been examined deeply and well known but less information about toughness and impact behaviour of the composite, is available.

The objective of this paper is to understand the effect of Al\(_2\)O\(_3\) particle reinforcement on tensile, high cycle fatigue, impact and fracture behaviour of a AA 2618/Al\(_2\)O\(_3p\) composites. The tensile deformation, high cycle fatigue response, impact resistance and final fracture behaviour of the composites are discussed with respect to the concurrent and mutually interactive influences of composite microstructural effects, matrix deformation characteristics, test temperature and nature of loading.

Experimental Technique

Material System

In this study, aluminium alloy 2618 with the theoretical density of 2760 kg/m\(^3\) is used as the matrix material and Al\(_2\)O\(_3p\) (Alumina particles) of size 5-10\(\mu\)m (Average 7\(\mu\)m size) with density of 3960 kg/m\(^3\) is used as the reinforcement. The particle sizes of Al\(_2\)O\(_3\) were determined using Scanning Electron Microscope (SEM). The chemical composition of 2618 aluminium alloy is tabulated in Table-1. Test specimens were made with 5% and 10% weight represented as 5P and 10P respectively.

Processing of MMC

Initially 2618 Al alloy was charged into the crucible and heated to about 750\(^\circ\)C till the entire alloy in the crucible was melted. The Al\(_2\)O\(_3\) particles were preheated to 400\(^\circ\)C for 10 min and brought to room temperature before incorporation into the melt. The steel mould of size 150x120x30 mm\(^3\) was used for the preparation of cast blanks. The mould also preheated to 550\(^\circ\)C for 10 mins to obtain uniform solidification. After the molten metal was fully melted, degassing tablet was added to reduce the porosity. The melt was transferred into another crucible which was separately heated to maintain the required temperature. The temperature of the melt was reduced to 650\(^\circ\)C (Solidus - liquidus state) to incorporate reinforcement. The preheated Al\(_2\)O\(_3\) particles were added at the rate

| Table-1 : Chemical Composition of 2618 Al. Alloy Matrix (% Weight) |
|------------------------|--------|--------|--------|--------|--------|--------|--------|
| Si  | Mn  | Cu  | Fe   | Mg    | Ni   | Ti   | Ti-Zr |
| 0.2 | 0.2 | 2.3 | 1.1  | 1.5   | 1.1  | 0.2  | 0.25  |
| Zn  | Al  | Balance |
| 0.15 |
of 20 gm/min with manual stirring. Incorporation of the reinforcement particles within a semisolid alloy is claimed to be advantageous because the reinforcement entraps the semi solid alloy thus avoiding particle agglomeration and settling. After the mixing of particle in the semisolid state, the temperature was increased to 720°C. The stirrer was lowered into the melt slowly to stir the molten metal at the speed of 500 rpm. The temperature was also monitored simultaneously during stirring. The mixture was poured into the mould. The maximum duration of mixing was 10 min. The clearance of the stirrer from the bottom of the crucible was approximately 10 mm with the melt depth of 100 mm.

The Cast MMC billets were forged to break the cast structure to get uniform grains for the improvement of properties. The forging operation was carried out with the hammer. For each operation, temperature of 450°C was maintained for the duration of 2 hrs. The forged MMCs were then subjected to heat treatment cycle (maintained at 525°C for 2 hrs followed by water quenching and precipitate hardening at 175°C for 10 hrs). Specimens were prepared from the forged billet after heat treatment for testing.

Specimen Preparation

The specimens were cut from the forged billet to the specification outlined in ASTM E8, ASTM E21, ASTM E23 and BS 3512 to carry out tensile, hot tensile, Izod impact and high cycle fatigue (HCF) tests respectively. The length to diameter ratio of the mechanical test specimen was chosen, so as to minimize buckling during fully reversed (Tension-Compression) stress amplitude controlled deformation. To minimize the effects of surface irregularities, the test specimen surface was prepared by mechanically polishing the gauge section using progressively finer grade of Silicon carbide impregnated emery paper and then finish polished using 0.5 μm alumina powder suspended in distilled water so as to obtain a mirror like finish that is free of all circumferential scratches and surface machining marks.

Microstructure Characterization

Metallographic samples are cut from the forge billet mounted on bakelite with 400 and 600 grit SiCp impregnated emery surface using copious amount of water as lubricant. The polishing operation was carried on Mecapol P230 programmable polishing machine with the speed of 20 to 600 rpm. Fine polishing to near mirror like finish was achieved using lavigated alumina powder suspended in distilled water. Reinforcement morphology and its distribution in the MMC along with other intrinsic microstructural features were identified by examining the samples in an optical microscope (Nikon EPIPHOT-TME inverted microscope) with metal power image analyzer.

Mechanical Testing

The tensile and hot tensile tests were carried out on a TIRA 2820S universal material testing with a modular test system. The testing machine is suitable for examining the strength and deformation behaviour of materials during tension, compression and bending tests upto 20 kN. For the measurement of force, strain gauge load cells are provided with the measuring cycle less than 22 milliseconds. It has the capacity of operating in the speed range of 0.01 to 1000 mm/min. The tests were conducted in controlled laboratory air environment at ambient temperature (27°C). Tensile tests were conducted in accordance with procedures explained in ASTM E8 at the strain rate of 0.033 /s and hot tensile tests were conducted in the same machine in accordance with procedures outlined in ASTM E21 at strain rate of 0.033 /s. The maximum temperature capabilities to perform hot tensile test is 1100 Celsius. The test temperature chosen was 200°C and corresponds to the maximum temperature of the end product application. Izod impact test was carried out on single notch cylindrical specimens as per procedure outlined in ASTM E23. The high cycle fatigue tests were conducted at 50 Hz and at a stress ratio (R= \( \sigma_{\text{minimum}}/\sigma_{\text{maximum}} \)) of -1.0. The number of cycles to cause complete failure or separation is taken as fatigue life.

Hardness of the MMC samples was measured in the EMCO hardness tester. Hardness tester consists of the machine column or stand, the vertically adjustable test anvil, the base mount and the control system as well as an optical measurement. The test anvil reveals a diameter of 90 mm and accommodates work pieces having a weight upto 50 kg.

Failure Analysis

Fracture surfaces of the tensile, cyclically deformed and impact samples were examined in a Scanning Electron Microscope (SEM) to: a) determine the macroscopic fracture mode and b) characterize the microscopic mechanisms governing fracture. The distinction between macroscopic mode and microscopic fracture mechanism
Fractographic examinations were carried out by means of the Jeol (JSM-840A) Scanning Electron Microscope (SEM). For surface preparation, fine coat was applied on the fracture surface using ion sputtering device (JFC-1100E).

**Results and Discussions**

**Microstructure**

Figure 1 (a and b) is the optical micrograph showing the microstructure of the AA2618 / Al₂O₃p/ 10P composite. Fig.1(a) and 1(b) indicates the non-etched and etched microstructure of composites respectively. The microstructural characterization is carried out by optical microscopy. In both the figures, the Al₂O₃ particles are of non-uniform size irregularly shaped and uniformly dispersed in the alloy matrix. The particle alignment is towards the forging direction. Intermetallic compounds dispersed in the matrix and near the interfaces were also observed in SEM. Wettability can be defined as the ability of a liquid to spread on a solid surface. It also describes the extent of intimate contact between a liquid and a solid. Successful incorporation of solid ceramic particles into casting requires that the melt should wet the solid ceramic phase. A weight based measurement was carried out by filtering all the slag left out in the process with different sizes of sieves. It is observed that 90% of the alumina particles were utilized for wetting. Hence, good wettability was achieved in the semi-solid mixing of stir casting method.

The 2618 aluminium alloy is age hardenable and can be strengthened through heat treatments, a fine dispersion of precipitates is giving rise to obstacles for dislocation motion. Normally, high fraction of the second-phase precipitates (Al₃CuMg) is located both at the grain boundaries and inside the grains. The presence of stable intermetallic particles such as aluminides (Al₄FeNi) favours the grain size control and inhibits the dislocation motion [18]. Normally aluminium alloys do not exhibit dynamic recrystallisation (DRX) because of their low hot worked dislocation densities, but nucleation and growth of new grains can indeed be induced in the presence of sufficient quantities of hard second phase, such as composites reinforced with small particles. These appear to raise the local dislocation densities and lattice curvatures above the critical levels needed for the initiation and propagation of DRX. Many authors have widely demonstrated that hot restoration mechanisms are favoured by the addition of Ceramic reinforcements to aluminium alloy and in particular promoted the initiation of DRX during hot deformation by increasing dislocation density in the matrix [22, 23].

**Density and Hardness**

The experimental density of the composites was obtained by the Archimedeans method of weighing small pieces cut from the composite billet, first in air and then in water. The porosities of the produced composites can be evaluated from the difference between the expected and the observed density of each sample. The experimental density of the AA 2618/Al₂O₃p/5P and AA 2618/Al₂O₃p/10P composite is 2780 kg/m³ and 2810 kg/m³ respectively. The density of the composites increases with increasing % weight.

Hardness of the MMC samples was measured in the EMCO hardness tester after polishing to 1μm finish. The brinell hardness of the samples was measured using a ball diameter of 1mm with 10 kg load to obtain an indentation, which would be representative of the macrostructure of the material. The hardness of the 2618/Al₂O₃p/5P and AA 2618/Al₂O₃p/10P composites is 129 HB and 143 HB respectively. The hardness of composites increases with increasing % weight of particle addition.

**Tensile Properties and Fracture**

The tensile properties of the AA 2618/ Al₂O₃p composites at two temperatures (RT and 200°C) with particle size of 5 - 10μm are summarized in Table-2. The results are the mean values of the three tests.

At room temperature (27°C), the composite materials have revealed limited ductility in terms of elongation to failure. The stress-strain diagram of the AA 2618/ Al₂O₃p composites is shown in Fig.2. The particulate reinforced composites exhibits increase in the tensile strength (UTS and 0.2% PS) and decrease in the % elongation to failure with respective to the unreinforced alloy. The increase in the tensile strength is approximately 2.5% and decrease in % elongation is 69% at room temperature condition. With the increase in test temperature to 200°C, the tensile strength (UTS and 0.2% PS) increased to 25% - 32% with respect to the unreinforced alloy. The increase in % strength of MMCs at room temperature is less compared to the test temperature at 200°C. At the same time, tensile strength also increases with the increasing % weight of the
particle. But, the increase in 0.2% PS and UTS is marginal. The % elongation is decreased upto 70% with respect to the unreinforced alloy at room temperature. The % elongation is further reduced at 200°C test temperature compared to the room temperature.

The % elongation is decreased upto 78% with respect to the unreinforced alloy at 200°C. It is observed that the ductility of the AA 2618/Al₂O₃p composites decreased at higher temperature [19]. Hence, AA 2618/Al₂O₃p composites is a good candidate for the high temperature application. It is observed that 0.2% PS and UTS increases to 25% and 32% respectively when the test temperature was increased to 200°C. The ultimate tensile strength and 0.2% PS of the MMC at 200°C is less than the value at RT.

Tensile fracture surface of AA 2618/Al₂O₃p/5P composite at room temperature is shown in Fig.3(a). The fracture surface in Fig.3(a) reveals that most of the portion has facets and ridges and dimple structure in the small portion. Microcracks are also noticed in the fracture surface. The fractured particle also noticed in the surface with particle dispersion. Tensile fracture surface of AA 2618/Al₂O₃p/10P composites at room temperature is shown in Fig.3(b). Microcracks and microvoids are noticed in the fracture surface. The tensile fracture is of quasicleavage type. Tensile fracture surface of the AA 2618/Al₂O₃p/10P composite at elevated temperature (200°C) with low magnification is shown in Fig.4(a). It contains facets and ridges with dimple structure, which reveals quasicleavage nature. It also contains micropores without particles.

Tensile fracture surface of the AA 2618/Al₂O₃p/10P composites at elevated temperature (200°C) with high magnification is shown in Fig.4(b). It contains facets and ridges with dimple structure, which indicates quasicleavage features. In addition, micro-cracks and cracks in the particles also noticed.

At both room temperature and elevated temperature (200°C) the AA 2618/Al₂O₃p composite exhibited limited ductility on a macroscopic scale, with fracture occurring on a plane normal to the far-field tensile stress axis. However, examination of the tensile fracture surfaces at higher magnification revealed the features of reminiscent of locally ductile and brittle mechanisms. It is fairly well documented that the fracture of reinforced monolithic alloys occurs by events of microscopic void nucleation and growth. An essential requirement for void nucleation either at a reinforcing particle and/or other coarse second phase particles in the microstructure is the development of a critical normal stress across the particulate - matrix interfaces [19]. In the MMC, the nucleation of cavities and void is favoured by the concurrent and mutually interactive influences of 1) cracking of the hard, intrinsically brittle and elastically deforming inclusions Al₂O₃p and 2) decohesion at interfaces between the hard and brittle reinforcing Al₂O₃ particulate and the soft and ductile Aluminium alloy matrix.

The mechanisms of fracture are evident in the tensile fracture surfaces of Fig.3(a) where fractured particles surrounded by ductile regions with fine near featureless non-circular dimples called tear ridges and decohesion at particle - matrix interfaces are present. Large voids and dimples are caused by fracture and decohesion of particles, while the small ductile dimples can be attributed to the constraints in plastic flow of the aluminium matrix or to the reduction of strains induced by the particle cracking which lead to the formation of tear ridges in the composites. This can be explained with the high local plastic constraints induced by the larger sized particles and by the clusters. At high temperature, the local stresses are not large enough to crack the particles, since matrix flow stress

<table>
<thead>
<tr>
<th>Sl. No.</th>
<th>Material</th>
<th>Test Temperature (deg C)</th>
<th>0.2 % Proof Strength (MPa)</th>
<th>UTS (MPa)</th>
<th>Elongation (%)</th>
</tr>
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<tbody>
<tr>
<td>1</td>
<td>AA 2618</td>
<td>RT</td>
<td>340</td>
<td>410</td>
<td>10</td>
</tr>
<tr>
<td>2</td>
<td>AA 2618</td>
<td>200</td>
<td>270</td>
<td>300</td>
<td>12</td>
</tr>
<tr>
<td>3</td>
<td>AA 2618 / Al₂O₃p / 5P</td>
<td>RT</td>
<td>364</td>
<td>420</td>
<td>3.1</td>
</tr>
<tr>
<td>4</td>
<td>AA 2618 / Al₂O₃p / 5P</td>
<td>200</td>
<td>325</td>
<td>382</td>
<td>2.7</td>
</tr>
<tr>
<td>5</td>
<td>AA 2618 / Al₂O₃p / 10P</td>
<td>RT</td>
<td>370</td>
<td>427</td>
<td>3.0</td>
</tr>
<tr>
<td>6</td>
<td>AA 2618 / Al₂O₃p / 10P</td>
<td>200</td>
<td>338</td>
<td>396</td>
<td>2.64</td>
</tr>
</tbody>
</table>

Table-2: Tensile Properties of AA 2618 / Al₂O₃p Composites
decreases; however, the normal stress can promote interfacial decohesion and voids nucleation [25]. Voids nucleation was concentrated at the interface of both the particles and matrix, where there is a high matrix strain and at clusters of particles. As shown in Figs. 2(a) and 6(d), the fracture surfaces exhibited many minute dimples in the matrix regions around the particles and the voids are formed due to the particle decohesion.

High Cycle Fatigue

The results of the high cycle fatigue carried out on toroidal specimens of AA 2618/Al₂O₃ composites are listed in the Table-3. Three samples were tested for fatigue test. The fatigue life scatters with respect to % weight of alumina particles also noticed. No inference can be obtained from the percentage of particle addition. The Al₂O₃ particle reinforced aluminium alloy is at least equal to that of the unreinforced alloy matrix. The scatter in the values is due to the intrinsic microstructural inhomogeneity of these materials as regard to both particle size and distribution. The fracture surfaces of the cyclically deformed fatigue specimens are indicated in the Fig.5. Fig 5(a) shows that the fracture surface of the HCF tested at 200 MPa (Cyclic load) of AA 2618/Al₂O₃/5P composite with crack initiation at ‘A’. Fig. 5(b) shows the fracture surface of AA 2618/Al₂O₃/10P composite cyclically deformed at the stress of 200 MPa with the resultant fatigue life (N_f = 160000 cycles). It contains the fracture surface with striations in the stable crack growth region and randomly distributed micropores. The SEM fractograph (Fig.5(c)) shows the cyclically deformed SEM fractograph of AA 2618/Al₂O₃/10P composite with striations.

Impact Behaviour

The results of the Izod-impact test carried out on V-notched specimens of AA 2618/Al₂O₃p composites are listed in Table-4. The results showed that the impact behaviour of AA 2618/Al₂O₃ composite was significantly reduced by the presence of Al₂O₃ particles. The lower impact strength of AA 2618/Al₂O₃p composites is attributed to the presence of brittle Al₂O₃ particles which may act as stress concentration areas. This argument is in agreement with the literature [21]. Fig.6 indicates the SEM fractograph of impact tested specimens of AA 2618/Al₂O₃p/5P composites. It indicates the fracture surface with microcracks and particles of high magnification. It also contains micropores and particle separation/decohesion from the matrix alloy. Fig. 6(a) and (d) shows the dimple structure, ridges and facets in the fracture surface of the impact tested specimens. It also contains microcracks. From the fracture surface, it is evident that material has failed when exposed to the plastic region. Fig.6(d) indicates dimple structure, ridges and facets with high magnification. In the present investigation, the tensile properties (at room temperature and at 200°C), high cycle fatigue behaviour and impact behaviour of AA 2618/Al₂O₃p composites were studied.

The flow behaviour response of particulate reinforced MMCs is mainly governed by two processes: (i) the first one involves load transfer from the matrix to the particles, leading to an increase in flow stress at the interfaces between the ductile matrix and the quasi-rigid particles; (ii) the second one involves the development of microstructural damage producing particle cracking or decohesion of the particle-matrix interface. Particle cracking occurs when the composite material is not able to dissipate the energy produced by the deformation through metallurgical transformation such as dynamic recovery or recrystallisation [24]. This process in turn, is possible only when the particle-matrix interface is largely undamaged. On the other hand, particle cracking could lead to the macroscopic crack initiation up to the fracture. The same could be said for crack initiation at the interface between particles and matrix.

Conclusions

- The microstructure of AA 2618/Al₂O₃p composites revealed uniform distribution of Al₂O₃p particles and

### Table-3 : Fatigue Properties of AA 2618 / Al₂O₃p MMC System

<table>
<thead>
<tr>
<th>Sl. No.</th>
<th>Material</th>
<th>No. of Cycles 200 MPa / R = -1.0</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>AA 2618</td>
<td>1.5 x 10⁵</td>
</tr>
<tr>
<td>2</td>
<td>AA 2618 / Al₂O₃p / 5P</td>
<td>1.4 to 1.6 x 10⁵</td>
</tr>
<tr>
<td>3</td>
<td>AA 2618 / Al₂O₃p / 10P</td>
<td>0.36 to 1.6 x 10⁵</td>
</tr>
</tbody>
</table>

### Table-4 : Impact Strength of AA 2618 / Al₂O₃p Composite

<table>
<thead>
<tr>
<th>Sl. No.</th>
<th>Material</th>
<th>Impact Strength (Nm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>AA 2618</td>
<td>5.5 to 6.5</td>
</tr>
<tr>
<td>2</td>
<td>AA 2618 / Al₂O₃p / 5P</td>
<td>4 to 5.5</td>
</tr>
<tr>
<td>3</td>
<td>AA 2618 / Al₂O₃p / 10P</td>
<td>4 to 5.5</td>
</tr>
</tbody>
</table>
good wettability achieved in semi-solid mixing of stir cast method.

- The tensile tests showed an increase in tensile strength and a decrease in the elongation to failure for the MMC than the unreinforced alloys. The tensile ductility was strongly affected by the material inhomogeneity.

- The tensile strength of the composites decreases with increasing temperature. The ductility in terms of elongation to failure also decreases with increasing temperature.

- Stress amplitude fatigue life response revealed scatter in the fatigue life, with respect to the unreinforced alloy. These results also reported in literature for particle-reinforced composites, which are generally related to the intrinsic microstructural inhomogeneity of these materials as regard to both particle size and distribution. The constraint in deformation induced in the metal matrix by the Al2O3 particle reinforcement coupled with local stress concentration effects and matrix reinforcement particle interfaces promote failure, the joint influences at particle cracking and fast fracture through the matrix.

- The presence of hard, brittle Al2O3 particles in a ductile matrix of AA 2618 decreases the impact strength.

Acknowledgements

The authors wish to provide gratitude to HAL (Hindustan Aeronautics Limited), Bangalore, India, for carrying out the tests pertaining to this work.

References


**Fig.1 Optical Micrograph Illustrating a) Non-etched b) Etched Microstructure of AA2618 / Al₂O₃p / 10P Composites**
Fig. 2 Stress-Strain Diagram of AA2618 / Al₂O₃p / 10P Composite a) at Room Temperature b) at 200°C

Fig. 3 Scanning Electron Micrographs of Tensile Fracture Surface Contains a) Cracked Particles and Microcracks of AA2618 / Al₂O₃p / 5P b) Ridges and Facets of AA2618 / Al₂O₃p / 10P Composites at Room Temperature

Fig. 4 Scanning Electron Micrographs of Tensile Fracture Surface Contains a) Facets, Ridges and Dimple Structures b) Microcracks and Cleavages of AA2618 / Al₂O₃p / 10P Composite at 200 Celsius
Fig. 5 Scanning Electron Micrographs showing HCF Fracture Surface Features of AA2618 / Al2O3p Composites Deformed at Cyclic Stress Amplitude of 200 MPa, Nf = 160000 Cycles a) Overall Morphology and Crack Initiation at A of AA2618 / Al2O3p / 5P Composite b) and c) Striations of AA2618 / Al2O3p / 10P

Fig. 6 Scanning Electron Micrographs showing the Impact Fracture Surface Features of AA2618 / Al2O3p Composites showing a) Macrocracks and Microcracks b) Decohesion, Micropores and Cracked particles c) Dimple Structure d) Facets and Ridges with Dimple Structures