FATIGUE CRACK GROWTH STUDIES ON AGED ALLOYS BASED ON AL–CU–MG SYSTEM

A. Somi Reddy

Department of Mechanical Engineering, KITS, Warangal – 506 015, India

Email: arumalle_reddy@yahoo.co.in

ABSTRACT

The addition of 0.4 wt% of silver or cadmium to the alloy Al-4%Cu-0.3%Mg, which has a high Cu:Mg ratio changes the nature, morphology and dispersion of the precipitates that from on age – hardening at medium temperatures such as 150°C – 200°C. The purpose of this paper is to report the effects of these micro structural changes on the tensile and fatigue behavior of the alloys. The tensile properties of the alloys aged at 170°C increased in the order Al-Cu, Al-Cu-Mg, Al-Cu-Mg-Cd and Al-Cu-Mg-Ag. Despite differences in their microstructures and tensile properties, the fatigue performance of the alloys was found to be relatively unaffected. Fatigue deformation was similar in each case and the alloys showed identical fatigue limits (± 70 MPa). It is concluded that microstructure rather than tensile properties has the dominant effect on the fatigue properties of the alloys. Moreover, providing the precipitates are capable of dispersing dislocations generated by fatigue stressing, their precise nature is unimportant so far as the nucleation of fatigue cracks is concerned. Some differences were observed in the fatigue crack growth behaviour between silver – containing alloy and other alloys. In all cases, cracks propagated along grain boundaries and the fact that the particles there were closer in the silver – containing alloy may explain this result.

Keywords: *Aged alloys, fatigue, fatigue crack growth, fatigue limit*

1. INTRODUCTION

Age-hardenable aged aluminium alloys are extensively used for different applications in air craft and automobile industries owing to their high strength to weight ratio. The components fabricated out of these alloys when used for such applications undergo severe fatigue loading conditions. A study on the fatigue behaviour of these alloys appears to be utmost important and hence the concerning literature is briefly presented [1, 2].

Fatigue failures are catastrophic in nature and the process of fatigue failure usually occupies three distinct stages: i) hardening or softening of the material during fatigue stressing or both depending on the initial condition of the material, ii) crack initiation and iii) crack propagation. Fatigue cracks nucleate easily in reasons where the slip is localized. The important localization reasons are persistent slip bands (PSBs), inclusions (hard second phase particles/ matrix interface in commercial materials) and grain boundaries. PSBs are the slip bands, which appear in the same location during fatigue deformation even after removal of these bands by polishing. Nucleation of fatigue cracks in PSBs occurs by the formation of small extrusions and intrusions where the bands emerge at surface of the specimen. During the initial stages of fatigue stressing small cavities may form in these bands perhaps by vacancy agglomeration. These cavities later linkup together to form cracks in these bands. Fatigue behaviour of age-hardenable aluminium alloys is to be understood in general in three different conditions: i) solution treated condition ii) peak hardened condition and iii) over aged condition. Higher fatigue strengths are expected in the solution treated condition for the reason that the solute atoms participate in pinning the moving dislocations.

In the aged condition, the type (coherent or non-coherent), size and distribution of participate particles have significant effect on the fatigue properties. Fine and more uniformly distributed precipitate particles offer relatively less resistance to the motion of dislocations since moving dislocations can easily cut these particles. This allows the dislocations to become concentrated in discrete slip bands. Slip localization promotes early crack nucleation in the fatigue life, by developing localized soft regions. The desirable property of slip dispersal appears to be promoted by larger precipitates or presence of submicron intermetallic compounds that are normally present in commercial aluminium alloys.

In alloys hardened by fine precipitates, the to and fro motion of dislocations in PSBs can cause localized softening of these regions. Such a softening is associated with resolution of the fine precipitates within the PSBs. This occurs due to the fact that the moving dislocations in the PSBs can repeatedly cut the fine particles to such an extent that they are no longer stable, causing resolution.

Fatigue cracks tend to grow in two stages: shear mode and tensile mode. During the early stages of fatigue life, the cracks propagate in active slip planes which are inclined at ± 45° to the tensile axis (Stage-II) During the next stage, crack propagation occurs at 90° to the tensile axis. Stage II crack propagation occupies most of the fatigue life except where the cross slip is difficult (e.g., Nickel based super alloys where fatigue failures are mostly by stage I cracking). The characteristic feature of this mode of propagation is the presence of ductile or brittle striations on the fracture surface.

In agehardeneable alloys regions next to grainboundaries are frequently free of precipitate particles due to vacancy and solute depletion. These regions are known as precipitate free zones (PFZs). Because of the absence of precipitates, the PFZs are softer than the remainder of the grains. As a consequence, when the alloys are deformed, there is a tendency for the strain to be concentrated in the PFZs. This situation occurs during cyclic stressing in fatigue and it has been shown that stress concentrations at grain boundary triple points and early crack nucleation may occur in these. In commercial alloys, the precipitate free zones are narrow and have little effect on fatigue deformation, because cracking tends to be transcrystalline in these materials. Their effect seems to be quite pronounced in high purity alloys, where the zones are relatively wider as compared to commercial alloys. Initiation of fatigue cracks may occur in grain boundaries and hence the nature and dispersion of particles in these boundaries would be expected to influence this process. Similarly, the fracture characteristics of these particles, their size and spacing will influence the rate at which such grain boundary cracks will propagate

2. EXPERIMENTAL DETAILS

The alloys used in the present investigation are prepared from high – purity materials. Melting is carried out in gas fired furnace using a graphite crucible. The inside of the crucible was initially lined with high – purity refractory material (aluminium powder), dried at 150°C to drive off moisture, and finally baked at 950°C. Calculated amounts of aluminium and copper were placed in the crucible and the charge was slowly heated to 690°C, by continuously monitoring the temperature using thermocouple. Thereafter, the melt was heated quickly to 790°C, with proper care to prevent loss of aluminium by oxidation. The bath was thoroughly stirred so as to ensure the complete dissolution of copper in the melt. The fuel is turned off and the temperature of the melt was allowed to fall until it reached 720°C at which stage magnesium addition was prevented by plunging this metal into the melt using a graphite holder. The melt was again stirred, degassed using FESECO degassing tablets, and poured immediately into a rectangular cast iron mould which has been preheated to 150°C. When required silver was added to the melt at 790°C and cadmium additions were made together with magnesium. The cast ingots were homogenized at 450°C for 24hrs. Samples taken from top and bottom of the ingots were analysed for chemical composition of the alloys. The cast ingots were scalped to remove surface defects, hot pressed to give a reduction of 25% in several stages, and hot rolled at 450°C in several stages to 5mm thickness. A sufficient number of hardness samples were cut from the plate and polished for the determination of hardness / time curves. The remaining material was cold rolled to 1.5mm and 1mm thickness to provide specimen for fatigue testing and creep testing respectively. Tensile properties at room temperature were also determined from rolled sheet.Fatigue samples were machined in a Tensil Kut machine using a template as per the sample design shown in Figure. 1. All the test alloys are solution treated in a salt bath at 520°C, cold water quenched and aged in oil baths at 170°C. The samples are quickly transferred to oil baths and approximately one-minute time elapsed before ageing at 170°C.

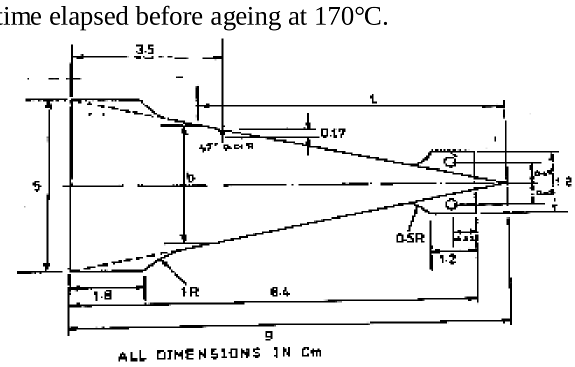

Figure 1- Sample Design for Fatigue Technique

Aged and fatigue deformed structures are observed using JEOL – 100C transmission electron microscope. To obtain these microstructures 3mm diameter discs were punched out from 1mm thick sheet. The discs were mechanically polished on 1200 grade polishing paper to 0.1mm thickness, cleaned in acetone and finally thinned for electron microscopy in a Tenupol jet-polishing machine. The electrolyte was 40% acetic acid, 30% orthoprosphoric acid, 20% nitric acid, and 10% water and the bath operated at 8V at room temperature. The thin foils were examined in JEOL – 100C transmission electron microscope. Fatigue tests were carried out in a cantilever, plane bending machine operating at 24 Hz. In its operation, the wider end of the sample was clamped, while the narrow end was stressed by a power driven connecting rod via a variable eccentric. When the sample failed, the connecting rod activated a micro switch, which cut off the power. The number of cycles was read directly from a counter. The tests were carried out in laboratory conditions and the location of the laboratory was such that the changes in the humidity and temperature were relatively small during the conduct of the fatigue tests. The fatigue tests were carried out on un notched tapered samples at four different stress levels of 140, 115, 90 and 70 MPa to establish the fatigue behavior in the form of S - N curves on log - log coordinates. Notched fatigue samples were used to carryout the fatigue crack propagation studies. All the crack propagation studies on all the alloys were carried out at an alternating stress level of ± 100 MPa. For this, each specimen was polished on one side using progressively finer emery papers and diamond pastes so that the progress of cracks could be observed using optical microscope. To do this, it was necessary to interrupt the tests at regular intervals.

3. RESULTS AND DISCUSSIONS

The S-N curves of different test alloys are given in Figure. 2. Figure. 2a shows the fatigue crack growth behaviour of test alloys. Table 1 gives the tensile properties and the types of precipitates present in different test alloys Figure. 3 gives the typical representative fatigue fracture of test alloys. Crack propagation characteristics of different alloys are compared in Figure. 4 in the form of da / dN vs √a, where a is the crack length. The grain boundary microstructures of ternery alloy and Ag- containing alloy are compared in Figures. 5 & 6.

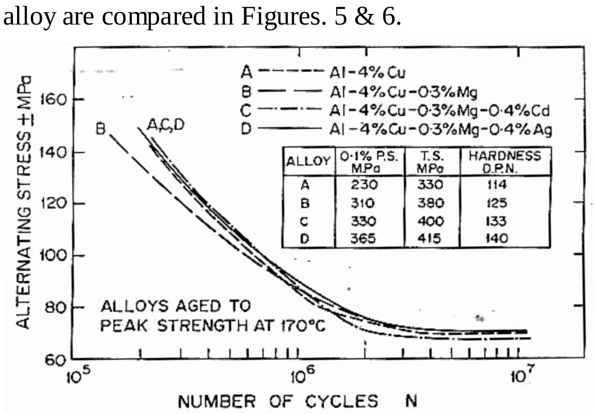

Figure 2 -S-N Curves of Different Test Alloys

Table-1 Properties of alloys aged to peak strength at 170°C

Sl. No.	Alloy	Major Precipitates	Properties			
			Hardness DPN	0.2% P.S MPa	T.S. MPa	Fatigue limit MPa
1.	Al-4% Cu	Θ^1	114	230	330	70
2.	Al-4%Cu-0.3%Mg	S^1, GP (Cu,Mg) zones	125	310	380	70
3.	Al-4%Cu-0.3%Mg -0.4%Cd	Θ^1	133	330	400	70
4.	Al-4%Cu-0.3%Mg -0.4% Ag	Omega, Θ^1	140	365	415	70

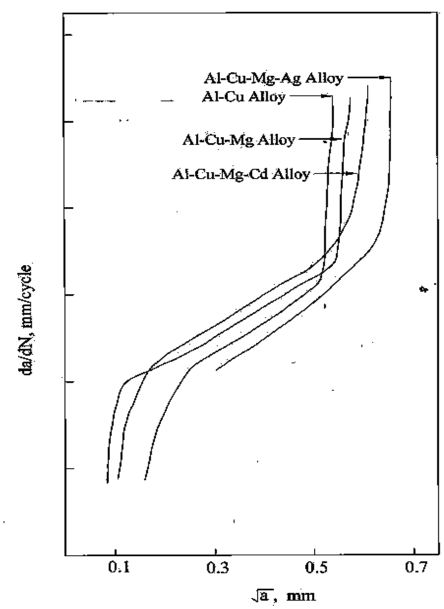

Figure 2a -Fatigue Crack Growth Behaviour of Test Alloys (Ref.2)

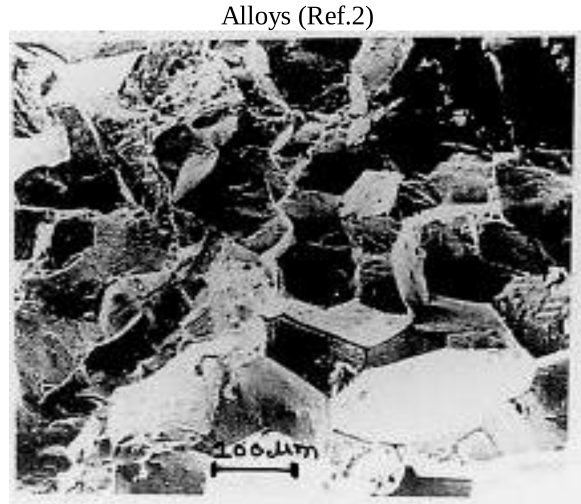

Figure 3- Al-Cu Alloy showing Intergranular Failure. Fatigue Stressed at + 90MPa. No. of Cycles: 9.99×10^5

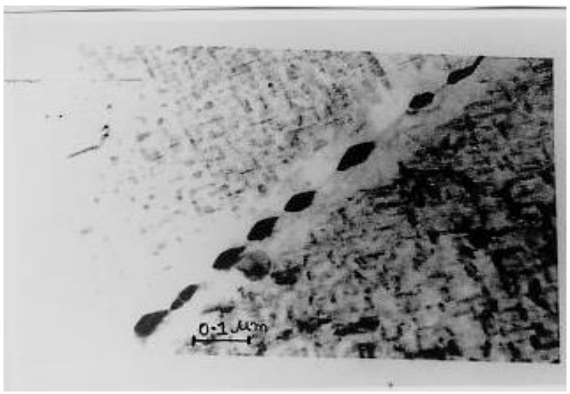

Figure 4 -Grain Boundary Microstructure of the Al-4% Cu-0.3% Mg

Figure 5 -Grain Boundary Microstructure of the Al-4% Cu-0.3% Mg-0.4% Ag

The S – N test that were carried out using smooth specimens stressed in plane bending conditions give a measure of the relative ease of crack nucleation. This follows because, once a crack has been initiated and the cross – section of the specimen reduced propagation proceed rapidly. It is concluded therefore that the relative ease of crack initiation was similar in each alloy despite differences in microstructure and tensile properties.

The explanation for this behaviour resides in observations of the dislocation substructures developed during the fatigue tests. In each case it was noted that the precipitates dispersed and fragmented dislocations rather than being cut by them. As a consequence, the dislocations did not concentrate in preferred slip bands which is known to be undesirable and leads to accelerated crack initiation [3].

Overall, it has been shown that microstructure, rather than level of tensile properties, controls the initiation of fatigue cracks in the four alloys that were studied. Moreover, provided the precipitates can disperse and fragment dislocations, their precise nature, volume fraction and the planes on which they form seem to be unimportant in this regard. The precipitates have behaved rather like a non-deformable dispersoids that are beneficial in improving resistance to fatigue crack initiation in commercial agehardeneable alloys [4].

Observations of rates of crack propagation did reveal some differences between the alloys, notably that the silver-containing alloy generally showed faster rates over a wide range of crack lengths. Since cracks mainly progressed along grain boundaries in the four alloys, it is presumed that differences in microstructure in this region must be important. The only significant difference appeared to be the fact that the grain boundary particles in the silver-containing alloy were more closely spaced than in the other alloys. Such a distribution of particles may have assisted intergranular crack growth. This may happen because cracks may propagate more readily through particles or at particle/matrix interface and can readily link together if the particles are closely spaced.

4. CONCLUSIONS

The following are the observations that can be made on the fatigue behaviour of different test alloys.

1. Closely spaced grain boundary precipitate particles provide faster crack propagation rates during fatigue deformation.
2. As long as the precipitate particles are able to disperse the slip during fatigue deformation the nature of the precipitate particles, their dispersion and the crystallographic planes on which they precipitate is unimportant as per as the nucleation of fatigue cracks is concerned.

5. REFERENCES

1. Arumalla, S.R and Polmear, I.J., 1985, *"Fatigue and Creep Behaviour of Aged Alloys Based on Al-4%Cu-0.3%Mg"*, 7th International Conference on Strength of Metals and Alloys (ICSMA 7)", Montreal, Canada, P453.
2. Arumalla, S.R 1985, *"Fatigue and Creep Behaviour of Aged Alloys Based on Al-4% Cu-0.3%Mg",* M.Engg.Sc. thesis, Monash University, Melbourne, Australia.
3. Boyapati, K. and Polmear, I.J., 1979, *"Fatigue and Microstructure Relationships in some Aged Aluminium Alloys,"* in 'Fatigue of Engineering Materials and Structures,' Pergamon Press, Great Briton, 2, 23 – 33.
4 Lutjering, G., 1980, in "Micromechanisms in Particle Hardened Alloys' (Martin, J.W. Edr), P 142.