Deformation and fracture behavior of an Al–Li alloy 8090

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The effect of microstructure on the deformation mode and fracture behavior of age-hardened alloys has been the topic of many investigations. The deformation behavior could be modified by varying the aging condition. In the underaged (UA) condition, precipitate shearing tends to lead to an inhomogeneous distribution of slip, whereas, in the overaged (OA) condition, precipitate bypass by dislocations tends to homogenize the slip distribution [1–5]. Varying the aging condition also alters the grain boundary structure. If aging is continued, precipitate free zone (PFZ) adjacent to the grain boundaries widens and the grain boundary precipitates coarsen. This effect, together with changes of slip distribution, influences the tensile properties, fatigue properties, and fracture behavior of the alloys.

In commercial age-hardened Al alloys, Mn, Cr, or Zr are usually added to form dispersoids. These dispersoids are incoherent with the matrix and can not be sheared by dislocations. Dispersoids bypassing would promote homogeneous distribution of slip. Dispersoids can also inhibit recrystallization so as to improve the toughness and stress corrosion cracking resistance of the alloys.

At present, Al–Li-based alloys are used in industry. The addition of Li can increase the specific strength and specific stiffness and decrease the weight of the structures. The microstructure and mechanical properties of the alloys have been studied extensively [6]. However, the effect of slip distribution on the fracture behavior is not yet available. A commercial Al–Li–Cu–Mg–Zr alloy was used in this work to study the influences of microstructure on slip distribution and fracture behavior.

The alloy 8090 (Al–2.43Li–1.17Cu–0.74Mg–0.12Zr–0.09Fe–0.09Si–0.03Ti, all wt.%) used was from a 50-mm thick hot rolled plate in T651 state. All the specimens were cut along the rolling direction. The alloy was solution treated at 530 °C for 30 min, quenched in water at room temperature and aged immediately.

Tensile tests were performed at an initial strain rate of 3.3×10^{-4} s^{-1}. Tension–tension fatigue tests were performed sinusoidally at 60 Hz, using a stress ratio of 0.25, in laboratory air.

Tensile and fatigue fractures were examined in a S-570 scanning electron microscope (SEM). The microstructure of the alloy was examined in a CM-12 electron transmission microscope (TEM).

According to the results of the tensile tests, two aging conditions have been selected to carry out fatigue tests and microstructural examination. The aging condition, tensile and fatigue properties of the alloys are given in Table I. It can be seen that, with an increase of tensile strength, the fatigue strength increases, and the fatigue strength/tensile strength ratio also increases. The UA alloy shows the higher tensile and fatigue strengths as well as fatigue strength/tensile strength ratios.

The microstructures of the alloy have been examined using TEM. The UA alloy contains a fine dispersion of δ' particles. For the OA alloy, δ' particle coarsening has taken place coupled with the formation of other precipitates of S and T_1 phases, which are heterogeneously distributed and preferentially precipitated on the subgrain boundaries and dislocations. Grain boundary precipitates and precipitate free zone (PFZ) adjacent to the grain boundaries can be also seen for the OA alloy. In the UA and OA conditions, Al_3Zr dispersoids exist and not be affected by the aging. The size of the δ' particle has been measured. The diameters of the δ' particle are 10 nm in the UA condition and 42 nm in the OA condition.

Thin foils were taken from the deformed (about 6%) tensile specimens. The UA alloy shows heterogeneous deformation with planar slip bans formed mainly by the activation of one slip system (Fig. 1a). The OA alloy shows homogeneous deformation with tangled dislocations, no pronounced slip bands could be seen (Fig. 1b).

The UA alloy showed dimpled tensile fracture. The OA alloy exhibited mixed rupture with dimples and a small amount of intergranular fracture.

Fatigue fractures are different from the tensile fractures. Almost all of the fatigue fracture in the UA alloy was faceted. The surfaces of the facets were parallel to the maximum shear stress direction (Fig. 2a). In the OA alloy, fatigue fracture was transgranular, perpendicular to the tensile axis, with clear and regular striations on the surface (Fig. 2b).

From the TEM results, it can be seen that the deformation mode of the alloy is controlled by the interaction of dislocations and precipitates. In the UA condition, the dislocations cut precipitates, the precipitates shearing promoted heterogeneous deformation and produced planar slip bands. For the OA alloy, the precipitates could not be sheared by dislocations, and deformation was homogeneous. The dislocations were bowed out around the precipitates and no slip bands were formed. The Al_3Zr dispersoids could not control the deformation mode in this alloy. For the other age-hardend Al alloy, the deformation mode is usually controlled by dispersoids [7, 8].

The UA alloy showed dimpled tensile fracture. The OA alloy exhibited mixed rupture with dimples and intergranular fracture. The difference in fracture mode could be accounted for the grain boundary structure.

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TABLE I Mechanical properties of the alloy in different aging conditions

Sample	Aging condition	YS (MPa)	UTS (MPa)	El (%)	S (MP)	S/UTS
UA	190 °C, 1 hr	270	420	14.1	220	0.52
OA	150 °C, 4 hr + 230 °C, 16 hr	230	380	7.2	180	0.47

Notes. YS: yield strength; UTS: ultimate tensile strength; El: elongation; S: fatigue strength at the number of cycles to failure $N_f = 5 \times 10^6$.

The fatigue fractures are different from the tensile fracture. Almost all the fracture is faceted in the UA alloy, which indicates the first stage mode of fatigue crack propagation. For the commercial age-hardened Al alloys which contained dispersoids, this situation could be only seen in the Al–Li alloys. The fracture is perpendicular to the tensile axis and with striations on the surface for the OA alloy, which indicates that the fatigue crack propagated in the second stage mode.

Pelloux and Bowles and Broek [9, 10] proposed different models to account for the formation of facets and striations. If, during one cycle, only one slip system was activated, planar slip bands would be formed in front of the crack and the fatigue crack would propagate along these bands. This is the first stage mode of the fatigue crack propagation. Fatigue cracks propagated in this way would lead to the formation of faceted fracture. Decreasing stress intensity ΔK and increasing tendency for heterogeneous strain distribution will favor deformation by fewer slip bands and slip systems. With increasing crack length, stress and stress intensity ΔK increased; the homogeneity of deformation also increased. If this homogeneity led to the activation of two or more slip systems in front of the crack, the fatigue fracture would propagate on the second stage mode and the striations would be formed on the fracture surface. Thus, the key factor affecting the mode of fatigue fracture is the activation of the slip systems in front of the fatigue crack condition.

Present results could be also explained by these models. The UA alloy contained fine δ' precipitates, which could be sheared by dislocations. The alloy exhibited a strong tendency towards heterogeneous deformation compared to conventional age-hardened Al alloys. The deformed structure of the UA alloy showed parallel slip bands, which indicated the activations of only one slip system. The faceted fracture was formed

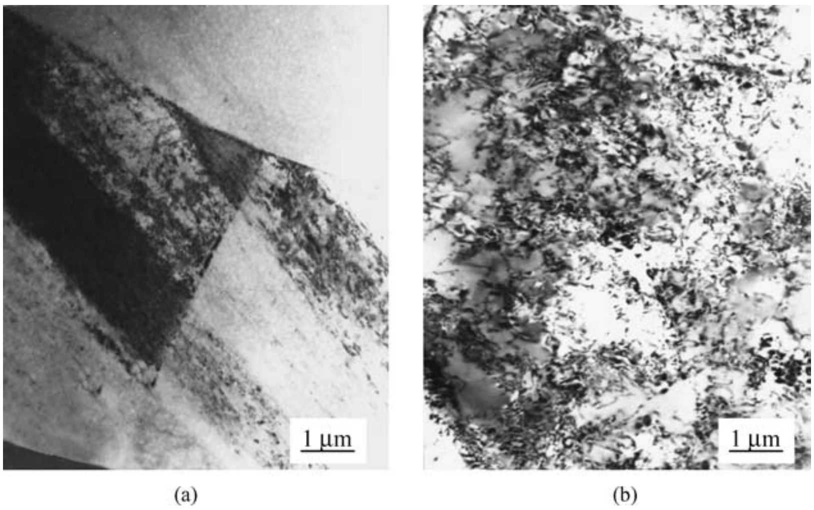

Figure 1 Slip behavior of the alloys in different aging conditions (a) UA condition; (b) OA condition.

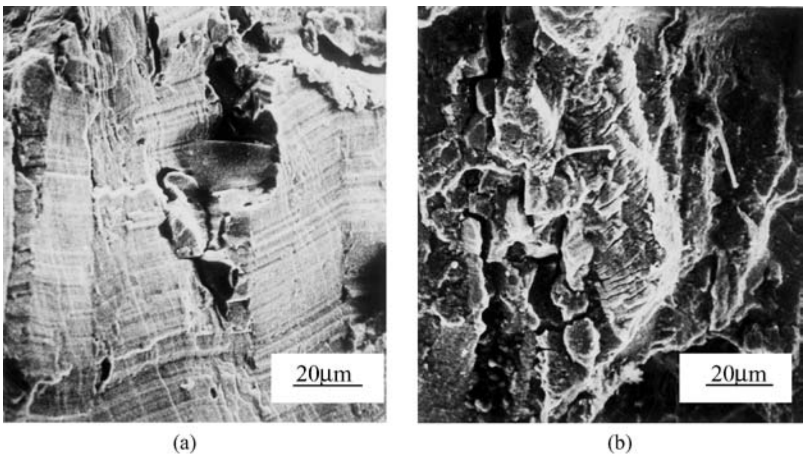

Figure 2 Fatigue fractures of the alloys in different aging conditions (a) UA condition; (b) OA condition.

when the fatigue crack propagated along these bands. With increased ΔK, the deformation heterogeneity decreased. If this deformation homogeneity could lead to the activation of two or more slip systems in front of the cracks, which altered the fracture mode, then striations would be formed on the fracture surface. It is the usual situation for conventional aluminum alloys [2, 7, 11, 12]. However, if this homogeneity could not promote activation of two slip systems, the crack would still propagate on the slip bands belonging to one slip system and the fracture was be still faceted. It is the situation for the UA alloy in this paper. The δ' precipitates showed a strong tendency towards deformation softening effect. The OA alloy exhibited homogeneous deformation by dislocations bypassing the precipitates, multiple slip systems were activated. So the alloy showed striations on the fatigue fracture surface.

This study provides following conclusions:

1. The aging condition affects the deformation behavior of the alloy. The UA alloy exhibited heterogeneous deformation with planar slip bands. The OA alloy deforms homogeneously with tangled dislocations.

2. Fatigue fracture mode is controlled by deformation behavior. Fatigue crack propagated on the first stage mode and the fracture was faceted in the UA alloy. Whereas the fatigue crack propagated on the second stage mode with the striations on the fracture surface in the OA alloy.

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