EFFECT OF DEEP ROLLING ON THE FATIGUE BEHAVIOR OF UNDER-, PEAK- AND OVER-AGED AA6110 AT ROOM TEMPERATURE

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Abstract

Cylindrical specimens of the age-hardened aluminium wrought alloy AA6110 (Al-Mg-Si-Cu) in different aging treatments (under-, peak- and over-aged) were deep rolled and cyclically deformed under stress control. The cyclic deformation behavior and s/n-curves of deep rolled and polished AA6110 in under-, peak- and over-aged conditions have been investigated at room temperature. The properties of deep rolled regions, residual stresses and work hardening effects were characterized by X-ray diffraction methods. Depth profiles of residual stresses, full width at half maximum (FWHM) values of the X-ray diffraction peaks and microhardness near the surface of the deep rolled condition are presented. Residual stress relaxation at the surface was investigated during the fatigue test. Results of quasistatic tension tests of the non-surface treated specimens are given to characterize the heat treatment state.

1. Introduction

Aluminium alloys exhibit many advantageous properties. Excellent corrosion resistance in most environments, reflectivity, high strength and stiffness to weight ratio, good formability, weldability and recycling potential make them ideal candidates to replace heavier materials (steel or copper) [1-3]. Recently higher strength Al-Mg-Si-Cu aluminium alloys i.e. AA6013, AA6110 or AA6111 were developed especially for applications in the automotive industry. It is generally accepted that the precipitation sequence in Al-Mg-Si alloys initiates with spherical clusters (GP zones), metastable phases, β'' and β' leading to the equilibrium β phase Mg_2Si. The precipitation in copper-containing Al-Mg-Si aluminium alloys is characterized by the quaternary phase, Q ($Al_5Cu_2Mg_8Si_7$) and its metastable precursor, Q'. This phase is supposed to contribute significantly to mechanical properties of these alloys [4, 5].

A very attractive and simple method to enhance fatigue properties are mechanical surface treatments i.e. shot peening, deep rolling or laser shock peening which are often performed on light alloys i.e. aluminium [6, 7], magnesium [6, 8] and titanium alloys [9, 10]. Typically, mechanical surface treatments lead to near-surface compressive residual stresses and increased dislocation density (cold work) which serve to retard or inhibit surface fatigue crack initiation as well as fatigue crack growth [11, 12].

The main purpose of this work is to investigate effects of deep rolling on the fatigue behavior of AA6110 in different aging conditions, namely under-, peak- and over-aged. Cyclic deformation- and s/n-curves of non- and mechanically surface treated (deep rolled) AA6110 at room temperature are discussed. In addition, near-surface residual stress-, work-hardening and hardness-depth-profiles as well as residual stress profiles of the fatigued state are presented.

2. Materials and Experimental procedure

The aluminium wrought alloy AA6110 was delivered as extruded bar with a diameter of 34 mm. The chemical composition of this alloy is 0.86 Si, 0.19 Fe, 0.45 Cu, 0.46 Mn, 0.78 Mg, 0.17 Cr, 0.02 Zn, 0.01 Ti and Al balance (all values in wt%). Cylindrical specimens with a diameter of 7 mm and a gauge length of 15 mm were prepared. The loading direction during fatigue investigations corresponds to the extrusion direction of the bar. The specimens were solution heat treated at 525°C for 30 minutes followed by water quenching to room temperature. Quenched specimens were aged immediately at 160°C for 1 hour, 12 hours and 1 week, which were called respectively as under-, peak- and over-aged conditions. The non-mechanically surface treated specimens were electrolytically polished in the gauge length leading to a material removal of 100 µm before testing to avoid any influence of machining or heat treatment. For deep rolling, a hydraulic rolling device with a 6.6 mm spherical rolling element and a pressure of 100 bars was applied at room temperature.

Quasistatic tension tests were carried out using a mechanical testing machine at a strain rate $(d\varepsilon/dt) = 10^{-3}\ s^{-1}$. Tension-compression fatigue tests were conducted with a servohydaulical testing device under stress control without mean stress (R = -1) and with a test frequency of 5 Hz. Strain was measured using capacitative extensometers.

Residual stress depth profiles were determined by successive electrolytical material removal using the classical $\sin^2\Psi$-method with Cu Kα radiation at the {333}-planes and $\frac{1}{2} s_2 = 19.77\times10^{-5}\ mm^2/N$ as elastic constant. Near surface work hardening was characterized by the full width at half maximum (FWHM) values of the X-ray diffraction peaks.

3. Results and Discussion

The important effect of precipitates is that the aluminium alloy is hardened and strengthened. Results of the hardness and quasistatic tension tests of non-surface treated specimens can be seen in Table 1. Increased hardness and 0.2% yield strength values were observed after the aging treatment.

Table 1. Hardness and tensile properties of AA6110 in different aging treatments.

Condition	Aging parameter	Hardness [HV5]	$\sigma_{0.2}$ [MPa]	UTS [MPa]	Elongation [%]
SSSS*	-	84	155	302	33
under-aged	160°C, 1 hour	125	292	400	28
peak-aged	160°C, 12 hours	139	425	455	22
over-aged	160°C, 1 week	120	393	413	24

*SSSS is super saturated solid solution (as quenched)

Fig. 1 shows non-statistically evaluated s/n-curves of polished specimens of under-, peak- and over-aged AA6110 at room temperature. The low cycle fatigue performance seems to be crack initiation rather than propagation controlled and hardness of materials is one of the important properties for the low cycle fatigue. Table 1 shows quite similar hardnesses of the under-, peak- and over-aged condition, therefore no significant differences in fatigue lifetime between polished under-, peak- and over-aged AA6110 at room temperature are seen in Fig. 1. However, the cyclic deformation behavior was different because of the different size and structure of precipitates in the matrix. Fig. 2 shows cyclic deformation curves of the polished under-, peak- and over-aged conditions at an applied stress amplitude of 400 MPa. The difference between the cyclic deformation curves is associated with dislocation-precipitate and dislocation-dislocation interaction during cyclic deformation.

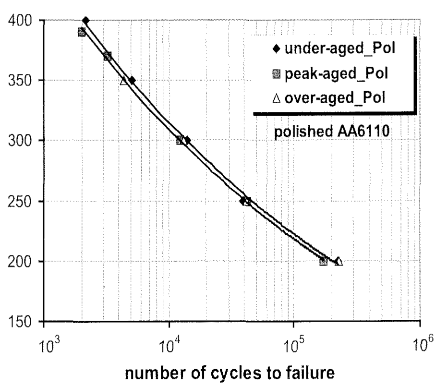

Fig. 1. Non-statistically evaluated s/n curves of polished specimens in different aging treatments.

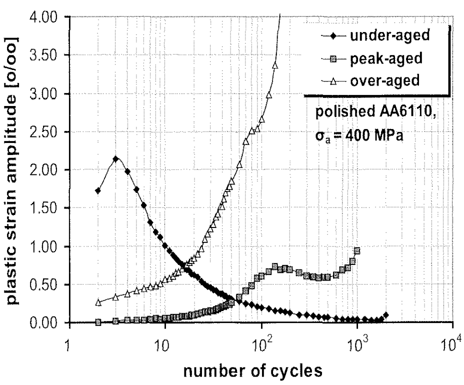

Fig. 2. Cyclic deformation curves of polished AA6110 in different aging treatments at an applied stress amplitude of 400 MPa.

In Fig. 2, cyclic hardening was observed in the under-aged condition due to the presence of GP zones, increasing dislocation densities and dislocation-dislocation interactions during cyclic deformation. The major precipitates in the peak-aged AA6110 alloy are ordered and coherent with the aluminium matrix. During cyclic deformation, the to-and-fro motion of dislocations through the ordered precipitates causes a mechanical local disordering or scrambling of the atoms in the precipitates. The structure of the precipitates becomes disordered and degraded. The hardening due to ordering contribution is lost, therefore cyclic softening was the result for the peak-aged condition [13, 14]. From results of the quasistatic tensile test (in Table 1), the over-aged condition showed a low work-hardening rate ($d\sigma/d\varepsilon$). It indicated that in the over-aged condition the microstructure comprised partially coherent and incoherent precipitations. Dislocations could move continuously to pass through the small partially coherent precipitates and work-hardening was comparatively small. Dislocation could move also to-and-fro through the small partially coherent precipitates during cyclic deformation. Any ordering contribution to hardening of over-aged condition was lost and cyclic softening occurred.

After deep rolling, compressive residual stresses as well as increased full width at half maximum values of X-ray diffraction peaks were observed at the surface and in near-surface regions up to approximately 0.7 mm. All residual stresses were measured in axial direction. Fig. 3 shows depth-profiles of the compressive residual stress and work hardening state after deep rolling in under-, peak- and over-aged condition. Maximum compressive residual stresses of -380, -295 and -292 MPa were measured at a depth of 40, 20 and 20 µm of the deep rolled under-, peak- and over-aged AA6110, respectively. Deep rolling led also to increased hardness at the surface and in near-surface regions as shown in Fig. 4. All effects from deep rolling as mentioned resulted in fatigue lifetime enhancement of under-, peak- and over-aged AA6110 at low and medium applied stress amplitude (in Fig. 5) due to lower plastic strain amplitude (in Fig. 6). High applied stress amplitudes caused instability of near-surface residual stresses and of the work hardening state therefore the deep rolling was ineffective in the low cycle fatigue regime [7].

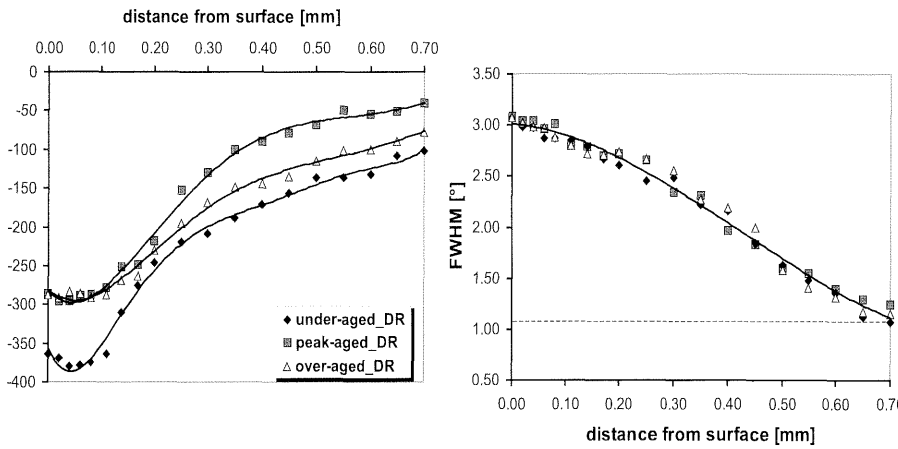

Fig. 3. Residual stress- (left) and FWHM-depth-profiles (right) of deep rolled under-, peak- and over-aged AA6110.

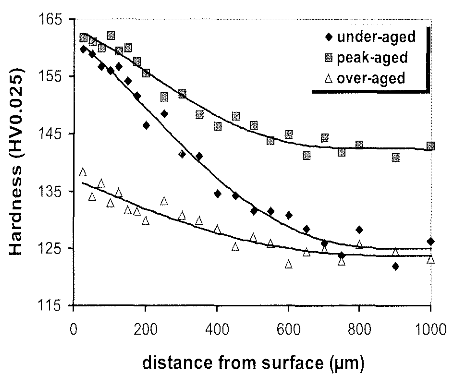

Fig. 4. Hardness-depth-profiles of deep rolled under-, peak- and over-aged AA6110.

Macroscopic residual stresses as well as work hardening can relax during cyclic deformation with increasing applied stress amplitude and number of cycles [15]. Fig. 7 shows relaxation of compressive macro and micro (FWHM-value) residual stress at the surface of the deep rolled under-, peak- and over-aged condition during stress-controlled fatigue tests at an applied stress amplitude of 350 MPa.

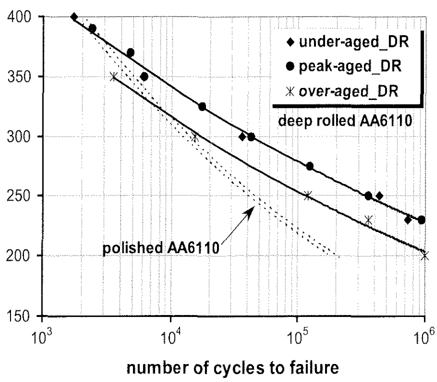

Fig. 5. Non-statistically evaluated s/n curves of deep rolled specimens in different aging treatments (compared with s/n curves of polished specimens).

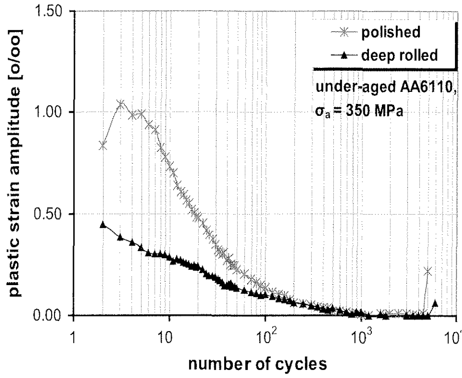

Fig. 6. Cyclic deformation curves of polished and deep rolled under-aged AA6110 at an applied stress amplitude of 350 MPa.

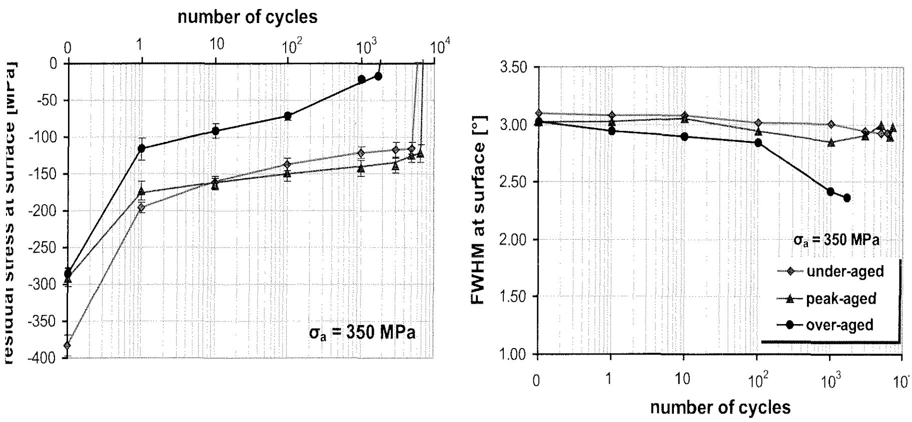

Fig. 7. Residual stress (left) and FWHM (right) relaxation during stress controlled fatigue of deep rolled under-, peak- and over-aged AA6110 at an applied stress amplitude of 350 MPa at room temperature.

In Fig. 7, the compressive residual stresses of the deep rolled under-aged condition relaxed more than in the deep rolled peak-aged condition, especially in the first cycle of fatigue test. Although initial compressive residual stresses of the deep rolled under-aged condition before fatigue were higher when compared with the deep rolled peak-aged AA6110 (see Fig. 3), the higher compressive residual stresses of the deep rolled under-aged condition were not beneficial because they relaxed almost in the first cycle of the fatigue test. Compressive residual stresses of the deep rolled over-aged condition relaxed also more than in the deep rolled peak- and under-aged condition leading to lower fatigue lives, although initial compressive residual stresses of the deep rolled over-aged condition before fatigue were not different from the deep rolled peak-aged condition. Moreover, instability of FWHM-values at the surface could be seen in the over-aged condition at an applied stress amplitude of 350 MPa and deep rolling could not enhance the fatigue life of the over-aged condition if the applied stress amplitude was higher than about 325 MPa (see Fig. 5).

4. Conclusions

The investigated AA6110 aluminium alloy shows cyclic hardening in the under-aged condition and cyclic softening in the peak- and over-aged conditions in stress-controlled fatigue tests at room temperature. Deep rolling induced compressive residual stresses and work hardening states into surface regions which result in fatigue lifetime enhancement due to a reduction of the plastic strain amplitude. The effectiveness of deep rolling was governed by the cyclic stability of near-surface work hardening which is known to retard crack initiation. Deep rolling had no beneficial effect on the fatigue life if instability of near-surface work hardening occurred during fatigue test.

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