Effects of the aging state and tensile strength on the fatigue properties of 6A01 aluminum alloy

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Abstract

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Abstract

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Keywords:6A01 Al alloy; Aging state; Tensile strength; Fatigue strength

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Introduction

Low weight and high strength are the main basis for the selection of materials for high-speed trains. 6XXX series Al alloys have been widely used in the body structural parts due to their low density, medium strength, good-forming ability and excellent welding performance [1-4]. However, fatigue cracks usually happen in the Al alloy components because of the alternating load during long-term service [5-10]. It is estimated that 80%˜90% failures of engineering components may result from fatigue fracture [6,9], therefore, it is necessary to optimize the fatigue performance of the 6XXX series Al alloy to improve their fatigue damage resistance.

The fatigue performance of Al alloys is generally affected by two kinds of factors: i.e external factors such as service environment [11-14] and loading mode [12,15-17], and intrinsic factors including grain size, inclusion size and microstructure homogeneity [17-20]. These internal defects often lead to the initiation of fatigue cracks and control their high-cycle fatigue (HCF) properties [21]. However, no matter what kind of defects, the initiation of fatigue cracks is always caused by the localization of plastic deformation under cyclic stress lower than the nominal yield strength of the alloy. Therefore, how to inhibit strain localization during cyclic loading is the key factor to improve the fatigue performance.

For heat-treatable strengthening Al alloys, it is also found the aging state affects the fatigue property through influencing the strain localization. Zhang *et al.* [22] found that at the underaged (UA) state the fatigue life is higher than that of peak-aging (PA) for the high-strength Al alloys. The reason is that during fatigue loading, plastic deformation becomes localized in the soft precipitate-free zone (PFZ) of the PA state, which provides conditions for the initiation of fatigue cracks. Besides, Li *et al* . [23] found that at the UA state the fatigue life is twice that of the overaged (OA) state at the same stress amplitude for 2E12 Al alloy, which was also attributed to the high slip reversibility of dislocations and strain uniformity at the UA state. However, Leng *et al* . [24] found that the OA state of 7075 Al alloy with higher yield strength and elongation can exhibit higher fatigue strength relative to the UA state. They attributed the reason to the high fatigue cracking resistance of the UA state because of the high strength.

From the research results above, it is still puzzled that there is not clear rule for the effect of aging state on the fatigue properties of Al alloys. Apart from the effect of the aging state, the tensile strength itself may also have a great effect on the fatigue strength. This is because normally, increasing the tensile strength can improve the fatigue cracking resistance [25,26], therefore, both the aging state and tensile strength should have significant effects on the fatigue properties of Al alloys.

For further evaluating the coupling effect of tensile strength and aging state on the fatigue properties, in this study, we employed the low-strength 6A01 Al alloy with three aging states (UA, PA and OA). Then the HCF tests were carried out on the 6A01 Al alloy with different tensile strengths. Finally, the influencing mechanisms of the tensile strength and aging state on the fatigue properties were deeply analyzed by investigating the microstructure evolutions at different aging states.

Experimental procedures

In this study, the industrial 6A01 Al alloy was selected as a rolling plate. The chemical composition of the 6A01 Al alloy was 0.48 pct. Si, 0.59 pct. Mg, 0.14 pct. Fe, 0.14 pct. Cr, 0.05 pct. Cu, 0.05 pct. Zn and balanced Al. To revel the influence of the aging states on the fatigue properties, we carried out heat treatment experiments. The specific heat treatment process is shown in Fig.1.

Hardness tests were administrated by a Vickers hardness tester employing a loading force of five hundred gf and a habitation time of 10 s. Seven indentations were tested for every specimen, and therefore the norm of the center 5 indentations was adopted. The size of tensile and fatigue specimens of 6A01Al alloy are shown in Fig. 2 and Fig. 3. Tensile tests were performed on an Instron 5982 universal mechanical testing machine with a strain rate of $10^{-3}s^{-1}$ at room temperature (RT). The tension- tension HCF tests were carried out on a GPS20 testing machine with a frequency of 100 Hz and load ratio R = 0 controlled by the stress amplitude.

The grain size of the 6A01 Al alloy was observed by optical microscope (OM). Besides, the fatigue fracture and precipitated phase of 6A01 Al alloy were observed by scanning electron microscope (SEM) and transmission electron microscope (TEM), respectively.

OM observations were conducted on a BX53M instrumentation, automatically ground and polished, and treated with an anodic membrane in the solution (1.1 g H_3BO_3 + 3 ml HF+ 97 ml H_2O) for 75 s at an applied voltage of 25 V. The sample observed by TEM were punched into 3 mm diameter disks from slices, and then mechanically thinned to a thickness of 50 μm. Then TEM samples were prepared by Struers-Tenupol-5 twinjet electro polishing device. with polishing solution of 10% $HClO_4$ and 90% C_2H_5OH, operated at -25 °C and 35 V. TEM examinations were performed an FEI Tecnai F20 transmission electron microscope with a working voltage of 200 kV.

Experimental results

Hardness variation

To quickly determine the proper aging states, we systematically studied the hardness variation with aging time of the 6A01 Al alloys at 175 °C, and the result is shown in Fig. 4. It can be seen that the highest hardness appears at an aging time of 8 h. Besides, the hardness values are nearly the same when the aging times are 1 h and 72 h, which are much lower than that of 8 h. Therefore, we choose the aging times of 1 h, 8 h and 72 h as the UA, PA and OA states for 6A01 Al alloys, respectively, as marked in Fig. 4.

On the other hand, it could be seen that the hardness values the aged 6A01 Al alloy show two peaks and a valley in between, which can be owing to the microstructure evolution with the aging time. It is generally explained that the occurrence of the first peak may be attributed to the high density of the GP zones. With the increase of aging time, the GP zones gradually transform into the β'' phase, and the second peak starts to appear due to the accumulation of the β'' phase. Therefore, the valley of the hardness value should result from the transition of the two strengthening mechanisms.

Microstructures and precipitates

The microstructures of the 6A01 Al alloy at different aging states are shown in Fig. 5. It is seen that the grains are elongated along the rolling direction for the three aging states, which is a typical microstructure of Al alloys formed by hot extrusion [27, 28]. Besides, with the increase of aging time, the grain size remains unchanged. Although the aging treatment does not affect the grain size, it still has a great influence on the size and volume fraction of the precipitated phases, which will be elaborated in the following paragraphs.

Figs. 6(a)-3(c) shows the TEM images of the matrix precipitates for the UA, PA and OA states, severally. It is apparent that the precipitated phases are equally distributed within the matrix. Besides, with the rise of aging time, the sizes of precipitated phases step by step increase, while the density firstly increases and so decreases.

After solution aging treatment of the Al-Mg-Si alloys, it was found that the precipitation sequence of the strengthening phase during the aging process followed: SSSS - GP (I, II) zones - metastable β''- b (Mg_2Si) [29-33]. Therefore, the precipitated phases are mainly GP zones at the UA state, as shown in Fig. 6(a). The GP zones are coherent with the Al matrix. Besides, a large amount of needle-like β'' (Mg_2Si) phases appeared at the PA state, as shown in Fig. 6(b). The needle-like β'' phases are semi-coherent with the Al matrix and play the main role in strengthening. The β equilibrium phase precipitated at the OA state, as shown in Fig. 6(c). The β phases are incoherent with the Al matrix. Fig. 6(d) summarizes the average size of precipitated phases at different aging states. The precipitated phases size of the UA, PA and OA states are 34 nm, 45 nm and 100 nm, respectively.

Tensile and fatigue properties

Fig. 7(a) displays the tensile stress-strain curves of the 6A01 Al alloy at different aging states. It is apparent that the tensile and yield strengths are the highest for the PA state, and those of the UA and OA states are very close to each other. The tensile strength of the UA, PA and OA states are 282 MPa, 324 MPa and 286 MPa, respectively.

The relationship between uniform elongation and tensile strength of the 6A01 Al alloy is shown in Fig. 7(b). It is found that the UA state shows a better matching of strength and plasticity than that at the OA state. According to the previous studies [34-36], the materials with high strength-plasticity matching will yield a higher fatigue strength. Therefore, it is expectant that the UA state should show better fatigue properties than the OA state.

The S-N curves of the 6A01 Al alloy at different aging states are shown in Fig. 8, in which the fatigue strength was calculated by the staircase method. From Fig. 8, it is obvious that the fatigue lives of the PA state are higher than those of the UA and OA states, besides, those of UA are higher than OA. As the fatigue strength can intuitively reflect the fatigue performance, we summarized the relationship between the aging time and fatigue strength, as shown in the Fig. 8(b). It can be seen that with the increase of aging time, the fatigue strength firstly increases and then decreases, and the PA state displays the highest fatigue strength.

Fatigue fractographies

The fatigue fracture surfaces of the three states are shown in Fig. 9. It may be seen that the fatigue fracture morphologies are basically the same for the three aging states, and the fatigue cracks mainly initiated on the free surface of the specimens. Therefore, we can infer that the fatigue damage mechanism should be similar for the different aging states. For Al alloys, the fatigue sources usually originate from inclusions, porosity and heterogeneous microstructure [37-42]. However, compared with cast Al alloy, there are almost no inclusions and porosity in wrought aluminum alloy. It is also true through the observation of fatigue fractures. Therefore, it will be inferred that the fatigue cracking ought to be mainly caused by the heterogeneous microstructure or surface stress concentration.

Discussions

Influence of tensile strength on fatigue strength

From the experimental results above, it can be concluded that the PA state displays both the highest tensile and fatigue strength. Besides, the UA state with lower tensile strength but higher uniform elongation relative to the OA state shows higher fatigue strength. These results indicate that both the strength and elongation have important effects on the fatigue properties, as claimed by the previous studies [35,43]. In view that the tensile strength under the true stress-strain coordinate (named as the true tensile strength) can reflect the tensile strength and uniform elongation simultaneously, we summarize the relationship between the fatigue strength and the true tensile strength, as shown in Fig. 10(a). It is apparent that increasing the true tensile strength can effectively improve the fatigue strength. The main reason is that the true tensile strength is a comprehensive reflection of both the yield strength and the work-hardening capacity, both of which affect the fatigue damage significantly. In general, the yield strength affects the plastic deformation resistance, while the work-hardening capacity influences the strain homogeneity during cyclic deformation [44,45].

Fig. 10(b) shows the work-hardening curves and the true stress-strain curves of 6A01 Al alloy at the three aging states, which indicates that the UA state has higher strain-hardening ability that the OA state. Therefore, it is consistent with the fact that the UA state exhibits higher fatigue properties than the OA state. These experimental results also clarify the fact that the effect of tensile strength on the fatigue strength is greater than that of aging states for the low-strength Al alloys.

Besides, it can be found that the tensile strength of the 6A01 Al alloy at the UA state is similar with the OA state, but the fatigue strength at the UA is higher than that at the OA state. The reason may be explained that the elongation at the UA state is higher than that at the OA state, so the slip distance of dislocations is increased, inducing the reduced plastic deformation inhomogeneity and the inhibited fatigue crack initiation.

According to the study of Pang *et al.* , the relationship between the fatigue strength and tensile strength of AISI 4340 high-strength steel displays a parabolic form on the whole [46]. However, our results between fatigue strength and tensile strength still show a monotonous relationship. In comparison with the high-strength steel, titanium alloy and hard Al alloy, the present 6A01 Al alloy belongs to the low-strength material. Although the tensile strength of the 6A01 Al alloy at the PA state is the highest among the three aging states, the tensile strength may have not yet reached the peak of the parabola. In other words, the tensile strength and fatigue strength are still in the positive correlation stage. Therefore, the fatigue strength of the 6A01 Al alloy can be further improved by increasing the tensile strength.

Influence of grain boundary precipitates on the fatigue properties

For materials with low strength and high plasticity, grain boundary (GB) is that the most common initiation site for fatigue cracks [47]. Previous studies on the HCF properties of wrought alloys also indicate that the localization of plastic deformation around GB is the main reason for fatigue crack initiation [22,48]. Therefore, we investigated the microstructures of the GBs for various aging states. Fig. 11 shows the GB precipitates (GBP) and the PFZ at different aging states. It is quite clear from Figs. 10(a)-10(c) that the spacing of GBP and the width of PFZ firstly increase and then decrease with the increase of aging time, and they are the largest at the PA state. According to our previous study [48], increasing the space of the GBP ought to facilitate the penetration of dislocations across GBs. Besides, increasing the dimension of PFZ also benefits for the evacuation of dislocations from GBs, which also relieves the pilling-up of dislocations on GBs. Both of the on top of factors might decrease the fatigue damage along GB and improve the fatigue properties finally.

On the other hand, the dense chain-like GBP in the UA and OA states may reduce the binding energy of GBs, so that the pilling-up of dislocations is easy to prompt cracks at the GBs. However, the GBP with intermittent distribution in the PA state has little reduction on binding energy of GBs, so that the GBs are more tolerant to the dislocation impingement. This may be also the reason why the PA state has the highest fatigue strength in the 6A01 Al alloy.

Conclusions

Through studying the fatigue properties of the 6A01 Al alloy at different aging states, the following conclusions can be drawn.

1. Among the three aging states, the PA state displays both the highest strength and the highest fatigue strength. This indicates that the effect of the tensile strength on the fatigue strength is greater than that at the aging state for the low-strength Al alloy.
2. For the 6A01 Al alloy at the PA state, the discontinuous distribution of the GBP and the increased width of the PFZ also contribute to the improved fatigue property.
3. The 6A01 alloy at the UA state has higher strain hardening ability, better plasticity and higher true tensile strength that the OA state.

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