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Ultrasonic attenuation study of defects in aluminium under the action of constant stress

G. T. FEI, Z. G. ZHU, P. CUI
Institute of Solid State Physics, Academia Sinica, Hefei 230031, Peoples Republic of China

In attempting to distinguish the interaction between dislocations and that between dislocations and point defects, the variations of the ultrasonic attenuation, $\Delta\alpha$, and strain ε are measured as a function of action time t of different constant tensile stress for 99.999 Al polycrystalline specimens at room temperature. It is found that the ultrasonic attenuation decreases and the strain increases at the beginning, and then reaches a saturation state with increasing time. The $\Delta\alpha$ can be divided into two parts. The initial decrease in $\Delta\alpha$ is due to the interaction between dislocations and point defects and the final equilibrium is caused by an interaction purely between dislocations. When the applied constant tensile stress is lower than 0.9 MPa, there is no dislocation multiplication.

1. Introduction

The interactions between dislocations and that between dislocations and point defects play very important roles in the mechanical properties of materials. Recently, ultrasonic attenuation, $\Delta\alpha$, has been employed to study the interactions of defects and some advanced results have been reported [1–4]. Vincent *et al.* [1] and Zhu *et al.* [2] observed $\Delta\alpha$ under the action of push–pull stress. The $\Delta\alpha$ exhibits a butterfly shape as a function of strain, which is due to the interaction between dislocations at high strain amplitude or to the interaction between dislocations and point defects at low strain amplitude. Fei *et al.* [3] measured $\Delta\alpha$ at an early stage of fatigue. The variation of $\Delta\alpha$ as a function of cycle number at high strain amplitude is different to that at low strain amplitude. Gremaud *et al.* [4] found various shapes of $\Delta\alpha$ under the action of load–unload stress, which were explained as being caused by the interaction between the dislocations and point defects.

From the above review it can be concluded that either an interaction between dislocations or that between the dislocations and point defects can cause the ultrasonic attenuation. This paper deals with the ultrasonic attenuation under the action of a constant stress. The aim of the present investigation is to distinguish the ultrasonic attenuation caused by interactions between dislocations and that caused by interaction between dislocations and point defects.

2. Experimental procedures

Cold-rolled rods of purity 99.999% polycrystalline Al with a diameter of 15 mm produced in the Fushun Aluminium Factory, China were used in the experiment. The shape and size of the specimens are given elsewhere [3]. The experiments were carried out on the fatigue ultrasoundmeter developed by Zhu *et al.* [5] with constant stress control. The load mode is illustrated in Fig. 1. The measurements were performed at room temperature. An emitted ultrasonic pulse with a frequency of 15 MHz is a longitudinal wave propagating parallel to the stress axis. The time of ultrasonic propagation through the specimen is 11 μs. The ultrasonic attenuation $\Delta\alpha$ is defined as the difference between the ultrasonic attenuation value α_t during the action of stress and α_0 the initial state before the experiment, which is calculated using the following formula:

$$\Delta\alpha = \alpha_t - \alpha_0 = \frac{20}{t_{12}} \times \lg \frac{A_0}{A_t} (\mathrm{dB}/\mu\mathrm{s}) \qquad (1)$$

where t_{12} is the time of ultrasonic propagation through the specimen, A_0 the ultrasonic amplitude acquired before the experiment, and A_t the ultrasonic amplitude acquired due to the action of stress. Therefore ultrasonic attenuation only reflects the internal structure change associated with the action of stress.

3. Results

The variation of $\Delta\alpha$ versus action time t at different tensile stress values σ is shown in Fig. 2. It can be seen that $\Delta\alpha$ declines rapidly at the beginning and then generally becomes stable. As is seen from Fig. 2, $\Delta\alpha$ generally decreases to zero when σ is 0.9 MPa. When σ increases, the level of $\Delta\alpha$ increases but the shape is similar to that under a σ of 0.9 MPa. The higher σ is, the longer is t for $\Delta\alpha$ to reach equilibrium. The corresponding change of strain ε is shown in Fig. 3. At the beginning, ε increases with increasing t, and then tends to stable values. When σ increases, the saturation value of ε increases.

Fig. 4 schematically illustrates the basic pattern of curves in Fig. 2. The $\Delta\alpha_1$ indicates the decrease of $\Delta\alpha$

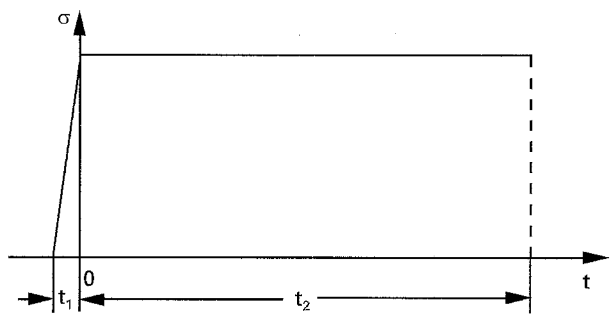

Figure 1 Schematic diagram of loading $t_1 = 2\text{s}$; $t_2 = 2\text{h}$.

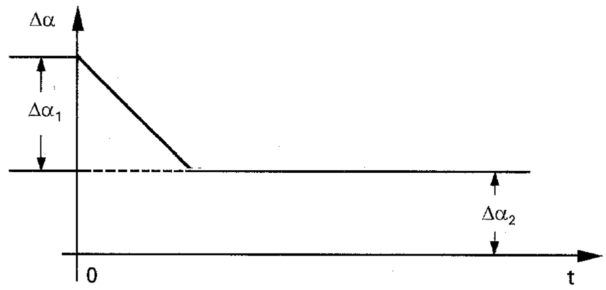

Figure 4 Schematic diagram of the variation of $\Delta\alpha$.

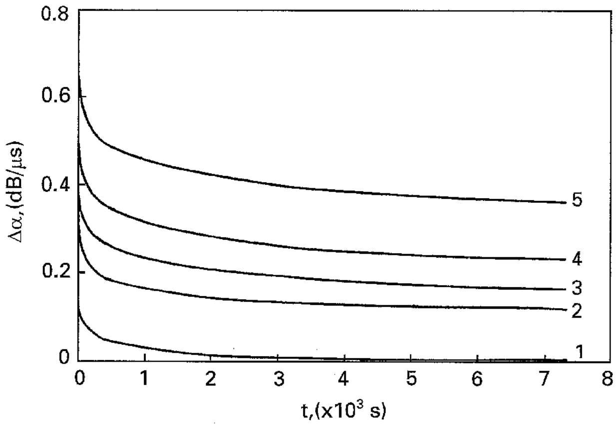

Figure 2 Variation of $\Delta\alpha$ with t under the action of constant tensile stress σ. (1) $\sigma = 0.9$ MPa; (2) $\sigma = 5$ MPa; (3) $\sigma = 10$ MPa; (4) $\sigma = 15$ MPa; (5) $\sigma = 20$ MPa.

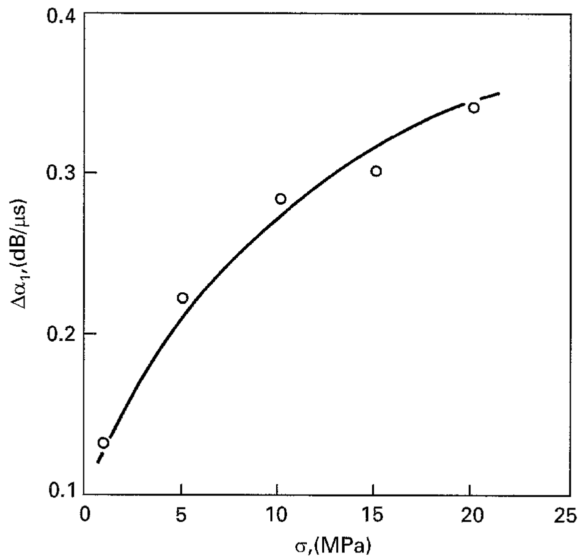

Figure 5 $\Delta\alpha_1$ versus constant tensile stress σ.

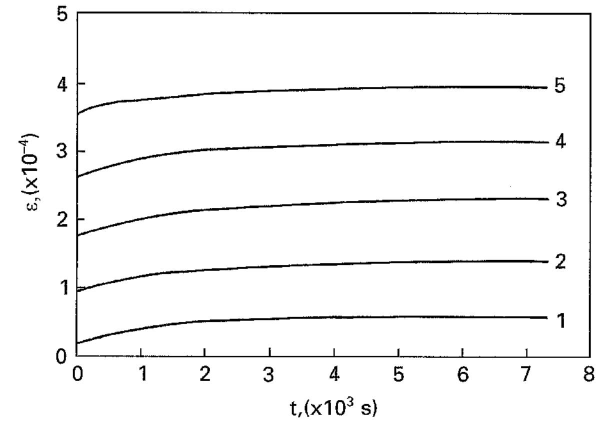

Figure 3 Variation of ε with t under the action of constant tensile stress σ. (1) $\sigma = 0.9$ MPa; (2) $\sigma = 5$ MPa; (3) $\sigma = 10$ MPa; (4) $\sigma = 15$ MPa; (5) $\sigma = 20$ MPa.

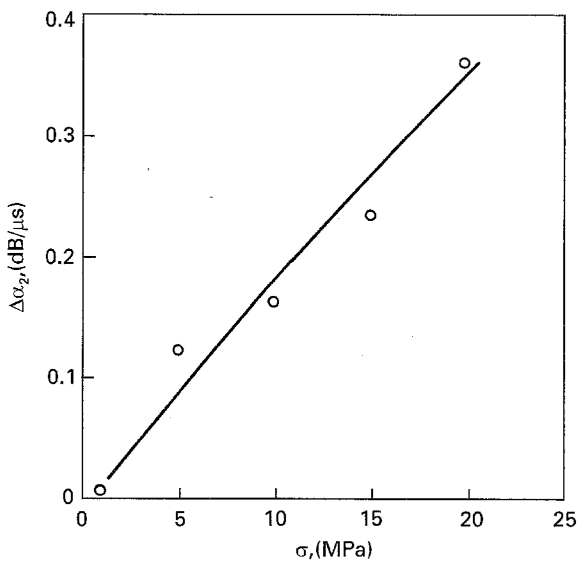

Figure 6 $\Delta\alpha_2$ versus constant tensile stress σ.

from initiation to equilibrium in each curve and the $\Delta\alpha_2$ the equilibrium value. The $\Delta\alpha_1$ as a function of σ is shown in Fig. 5. It can be seen that the $\Delta\alpha_1$ increases with increasing σ, but it is not a straight line. The $\Delta\alpha_2$ seems proportional to σ, as shown in Fig. 6.

4. Discussion

In the experiment, the ultrasonic wave can be used as a probe thereby having no effect on the internal defect configuration of the specimen. It can cause dislocations to vibrate *in–situ*, since the amplitude of an ultrasonic wave is much lower than that of the applied stress [3].

According to the G–L model [6], the relationship between $\Delta\alpha$, the density of dislocation Λ and the length of free dislocation segment L is $\alpha \sim K\Lambda L^4$, where K is the damping coefficient. In the experiment, the original specimen was cold-worked. Before the experiment, the specimen had been kept at ambient temperature for sufficient time to allow the point defects to distribute around the dislocation segments in the stable state as is shown in Fig. 7. The effective pinning of point defects made the free dislocation segments very short.

4.1. About $\Delta\alpha$ at $t=0$

Once the stress is exerted the following processes immediately occur (1) dislocations bow out (Fig. 8a); (2) the pinned dislocations depin from the point defect atmosphere (Fig. 9); (3) when the applied stress exceeds a critical value, F–R sources operate, the dislocation multiplies and produces new dislocations. The above three processes account for the behaviour of $\Delta\alpha$ at $t = 0$, because they all contribute to the lengths of the dislocation segments increasing compared with that before the stress is applied.

4.2. About $\Delta\alpha_1$

When the time increases, the following processes will happen under the action of a constant stress:

(1) Point defects are dragged towards an equilibrium state by dislocation segments (Fig. 8), since the process that the dislocation bow out (Fig. 8a) is not an equilibrium. This process has little contribution to the decrease of $\Delta\alpha$ from $t = 0$, because the L is almost unchanged (one can find it by comparing Fig. 8a and Fig. 8b). Accompanying the increase in time, this process generally tends to stability.

(2) The process that the pinned dislocations depin from the point defect atmosphere is recoverable. When the time increases, point defects will move towards the dislocation under the action of the dislocation stress field (Fig. 9b). As the time increases, the density of point defects at around the dislocation segments increases again. Along with this process the free length of the dislocation segments generally decreases, because the free length of the dislocation segments is inversely proportional to the line density of point defects in the dislocation segments: $L \sim 1/c$ [7], where L is the free length of dislocation segment, c the line density of point defects at dislocation segments. Therefore $\Delta\alpha$ will decrease with point defects moving towards the dislocation. In other words, $\Delta\alpha$ will decrease with increasing time.

The above two process account for the behaviour of the $\Delta\alpha_1$. It causes the $\Delta\alpha$ initially to descend as is shown in Fig. 2. When the applied stress is equal to or lower than 0.9 MPa, only the above two recoverable process happened, so $\Delta\alpha$ can tend to zero.

It can be seen from Fig. 5 that $\Delta\alpha_1$ increases with increasing applied stress. This may result due to the following reason. When the applied stress is high, more dislocation segments depin from the point defect atmosphere, so the difference between L at $t = 0$ and$L$ at the equilibrium after a long time is larger. This accounts for the behaviour shown in Fig. 5.

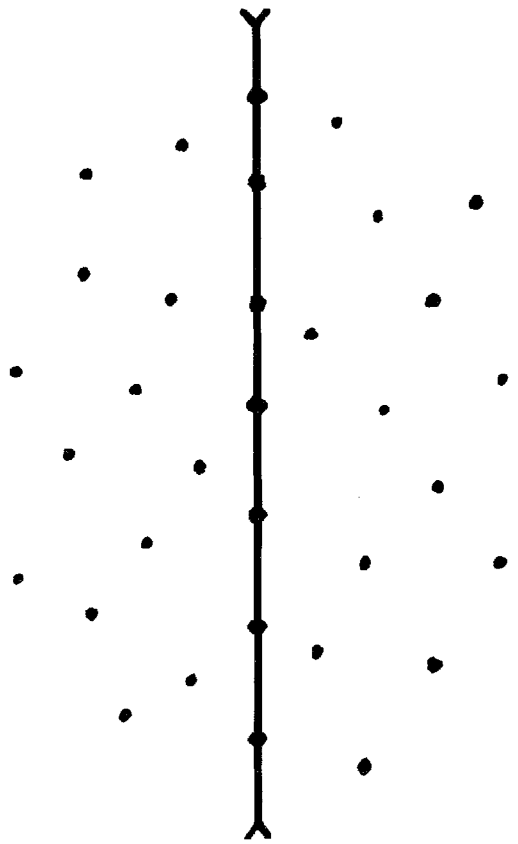

Figure 7 Schematic diagram of point defects distribution prior to loading.

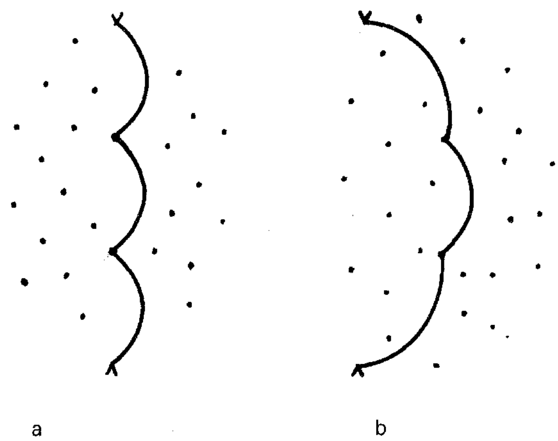

Figure 8 Schematic diagram of point defects movement dragged by dislocation. (a) at the beginning; (b) at the stabilizing.

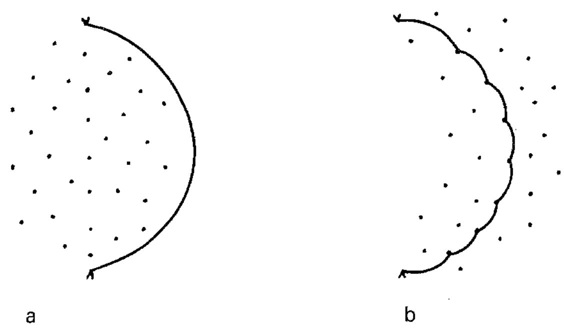

Figure 9 Schematic diagram of dislocation dragged out of point defect atmosphere. (a) at the beginning; (b) at the stabilizing.

4.3. About $\Delta\alpha_2$

When the applied stress is higher than 0.9 MPa, dislocation multiplication occurs. This process is not recoverable, independently of time. The observed behaviour of $\Delta\alpha_2$ results from this process. The larger the applied stress, the bigger is the multiplication. So $\Delta\alpha_2$ increases with increasing applied stress, just as is shown in Fig. 6. This process makes $\Delta\alpha_2$ always present no matter how long the time.

A $\Delta\alpha$ can be caused by the slip of a grain boundary. However it is small since the slip is small at room temperature.

4.4. About creep

It can be seen from Fig. 3 that creep also occurs in polycrystalline pure aluminium at room temperature. Dislocation movement dragging the point defects (Fig. 8) and the movement of point defects towards a dislocation under the action of the dislocation stress field (Fig. 9) can contribute to creep. Both of these processes are controlled by the diffusion of point defects. Since the diffusion of point defects is slow at room temperature, the creep caused by these processes is also small and normally tends to stability with increasing time. The above two reasons result in the creep curves shown in Fig. 2. Grain boundary slip can relax the shear stress in two sides of a grain boundary, so it is possible for a grain boundary to slip although the slip distance is very small.

5. Conclusion

When a constant stress was applied on the Al specimen, the ultrasonic attenuation decreased at the beginning and then reached an equilibrium state with increasing time. The beginning descent is attributed to point defects moving towards the unpinning dislocations. The final equilibrium value is due to the dislocation multiplication, which is not recoverable.

Acknowledgement

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