ANISOTROPY OF WORK HARDENING IN 5456 ALUMINUM ALLOY①

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ABSTRACT The influence of intermediate recrystallization annealing and final recovery annealing on the strength of 5456 aluminum alloy sheet, which could not be strengthened by heat treatment, was investigated. The corresponding micro-mechanism of that influence on the strength was discussed. The introducing of intermediate recrystallization annealing during cold rolling will change the cold deformation degree, defect structure state and the volume fractions of different texture components of cold rolled sheet, and therefore the strength of work hardening will be influenced directly, and the strength levels as well as the strength anisotropy after the final recovery annealing will also be influenced indirectly. The strength levels in different directions could be adjusted by an optimized combination of intermediate recrystallization annealing and final recovery annealing.

Key words aluminum alloy rolling deformation texture work hardening anisotropy

1 INTRODUCTION

Many aluminum alloys can not be strengthened by heat treatment, therefore the work hardening of cold deformation is applied frequently to insure the required strength[1]. According to the external stresses there usually exist certain defect structure and special texture in aluminum alloy after different deformation processes[2], which will have important influences on the anisotropy of mechanical properties. Concerning the application area of the aluminum alloys the anisotropy should be either avoided[3] or utilized[4]. In this aspect the investigation on deformation texture and structure obviously has practical importance. The 5456 aluminum alloy sheet is used in the present work to analyze the influences of deformation process, deformation texture as well as the defect structure on the alloy strength.

2 EXPERIMENTAL

Starting from 450~ 480 ℃, the commercial 5456 alloy ingot (600 mm thick, 0.16% Si, 0.32% Fe, 0.037% Cu, 0.57% Mn, 4.95% Mg, 0.040% Cr, 0.015% Zn and 0.015% Ti, mass fraction) was hot-rolled down to 3 mm and coiled, the finishing temperature was about 300 ℃. Two samples (A and B) were taken from the hot band and cold-rolled down to 0.7 mm. Sample A was recrystallized at 350 ℃ for 2 h (intermediate annealing). Then the two samples were rolled down to 0.35 mm, and annealed at 205 ℃ for 20 min as final annealing.

Assuming that d_0 is sheet thickness and d is perpendicular distance from the observed position to the sheet center, s value is defined as $s = 2d/d_0$. It is apparent that $s = 1$ expresses sample surface and that $s = 0$ expresses sample center. The {111}, {200}, {220} and {113} incomplete pole figures of the two samples before and after final annealing were measured at different s positions, and the Bunge method was used to calculate the orientation distribution functions (ODF)[5]. The yield strength $\sigma_{0.2}$ and the tensile strength σ_b were measured along the rolling

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direction (RD) and transverse direction (TD). The deformation structure was observed under optical microscope.

3 RESULTS

The microscopic observation shows that the final annealing did not lead to recrystallization of the cold rolled sheet. According to the general annealing treatment of 5456 alloy, it can be deduced that the rolling sheet has experienced a strong recovery process during final annealing.

Fig. 1 gives the φ_2 sections of the ODF calculated for sample A after final annealing, and illustrates a typical rolling texture which has been observed very often in aluminum[6]. The rolling texture can be expressed by a density distribution along the β orientation fiber for different sheet pieces taken from sample A, i. e. A1, A2 and A3 with different s values (Fig. 2(a)), as well as the exact positions of the fibers in orientation space (Fig. 2(b)). It can be seen that the fibers, the corresponding peaks and the fiber positions overlap so perfectly that the deformation must be very homogeneous and leads to such high reproductivity of the rolling texture. The same overlap could also be observed for samples A and B at other experimental stages. Fig. 3

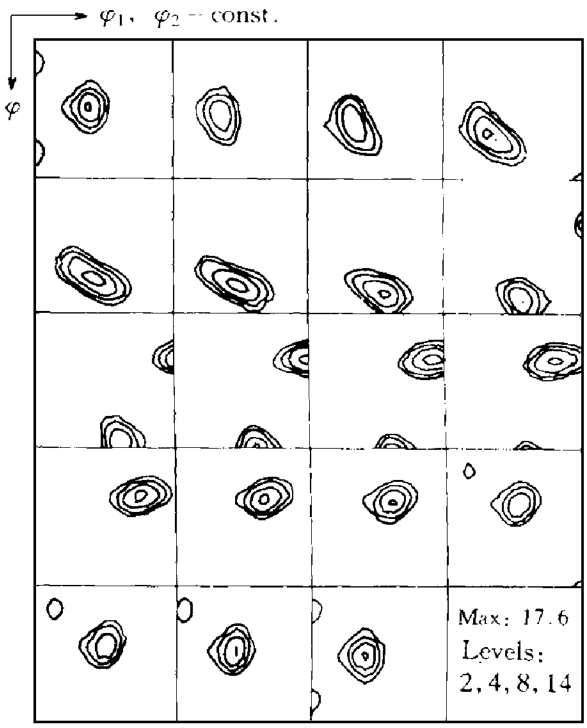

Fig. 1 Texture in sample A after final recovery annealing

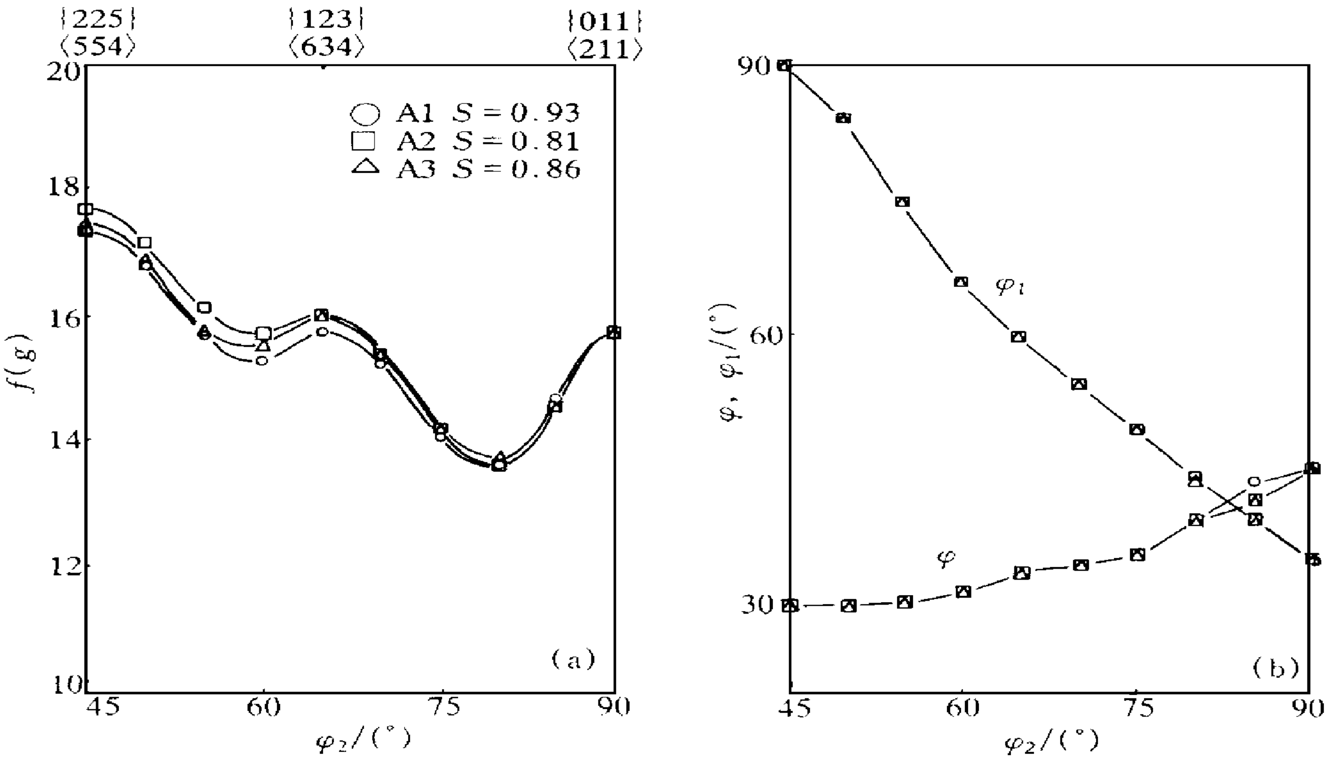

Fig. 2 β-fiber analysis of sample A after the final annealing

(a) —Orientation density distribution along β fiber; (b) —Positions of β fiber in orientation space shows the density distribution on the β orientation fiber for samples A and B before and after the final annealing, which demonstrates certain distribution differences of the orientation densities. Because of the high texture reproductivity shown in Fig. 2, the differences do not concern the measurement accuracy, but are induced by different experimental processes.

It can be observed from the β fiber analysis shown in Fig. 3 that the grain orientations are basically accumulated around {011} ⟨211⟩, {225} ⟨554⟩ and {123} ⟨634⟩ along the β fiber. No principal texture changes took place in samples A and B before and after the final annealing, which is the general case in aluminum sheets[6]. The final annealing results in a density increase along the β fiber for sample A, but not for sample B. The density distributions of samples A and B are also different. The orientation density of sample A is higher around {225} ⟨554⟩, but that of sample B is higher around {011} ⟨211⟩, from which the influence of intermediate annealing could be seen. The method of Lücke *et al*[7] was used to quantitatively calculate the texture components in Fig. 3. Table 1 gives the results. Both of the intermediate annealing and the final annealing have influenced the volume fraction and scattering width of rolling texture components, in which the final annealing has reduced the scattering width of the texture components in sample A. The corresponding increase of orientation density on β fiber (Fig. 3) indicates that the recovery annealing sharpened the texture in sample A, but did not in sample B.

Table 2 gives the experimental data of the tensile test in RD and TD for samples A and B, which are closely relative to the texture and microstructure, from which the influence of the intermediate and the final annealing on mechanical properties and their anisotropy can be observed. The final annealing reduces the strength to certain extent, but the strength drops in RD and TD are different. The yield strength drop is relatively high in RD (Table 2, bottom part). On the other hand the intermediate annealing has also decreased the final strength levels.

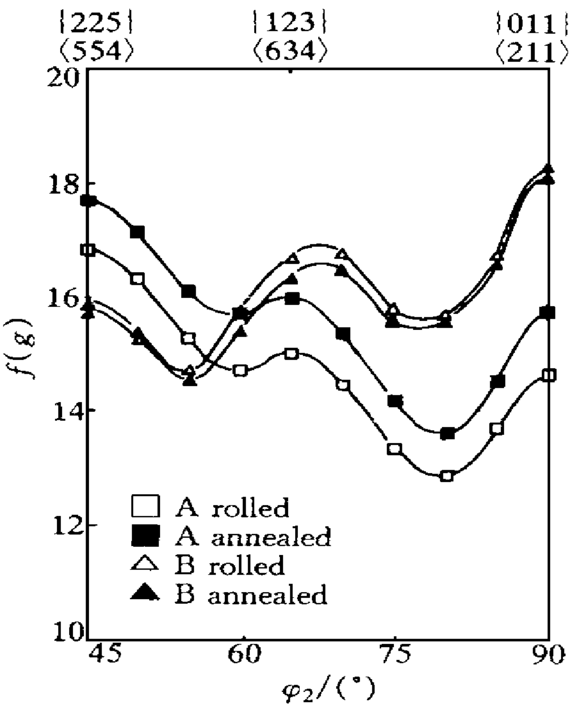

Fig. 3 Textures in samples A and B before and after final recovery annealing

Table 1 Quantitative analyses of texture components before and after final annealing and corresponding orientation factors by tensile test

Quantitative analyses of texture components		{001} ⟨211⟩		{225} ⟨554⟩		{123} ⟨634⟩	
		Volume fraction	Scattering width	Volume fraction	Scattering width	Volume fraction	Scattering width
Rolled	A	23.1%	9.48°	22.6%	8.97°	46.3%	9.43°
	B	27.0%	9.23°	18.5%	8.38°	47.0%	8.99°
Annealed	A	23.3%	9.24°	20.8%	8.45°	46.4%	9.18°
	B	25.1%	8.99°	18.3%	8.32°	46.6%	9.04°
Orientation factors of different textures		RD	TD	RD	TD	RD	TD
		0.41	0.28	0.33	0.41	0.40	0.45

4 DISCUSSION

4.1 Intermediate recrystallization annealing

At 350 ℃, the Mg atoms are basically solved in the matrix of 5456 alloy, so that the recrystallization process can be completed without obstruction of precipitated particles. Sample A has 50% final accumulated rolling reduction, but sample B has 88%. The difference is induced by the intermediate annealing of sample A. Therefore sample B should have experienced higher work hardening and correspondingly have higher yield strength. But Table 2 shows that sample B has lower yield strength than sample A in RD, which should be obviously due to the existence of rolling texture. The volume fractions of {123} ⟨634⟩ component in samples A and B after cold rolling are very similar, but the differences on components {011} ⟨211⟩ and {225} ⟨554⟩ are apparent (see Table 1). The yield strength σ_s in aluminum single crystal is anisotropic, and the equation $\sigma_s = \tau_c / m$ is valid according to the law of critical resolved shear stress, in which τ_c represents the critical resolved shear stress and m is the orientation factor. Table 1 shows the orientation factors of the slip systems which will firstly become active in the grains oriented at {011} ⟨211⟩, {225} ⟨554⟩ or {123} ⟨634⟩ under the tensile stress. Comparing the data of orientation factors, the component {011} ⟨211⟩ has lower tensile deformation resistance in RD, and the component {225} ⟨554⟩ has lower tensile deformation resistance in TD. Sample B has more {011} ⟨211⟩ texture and less {225} ⟨554⟩ texture after rolling (Table 1), and therefore its yield strength in RD is lower than that of sample A. This indicates that the influence of texture components on the yield strength is stronger than that of work hardening.

About 77% rolling reduction had been accumulated in sample A before intermediate annealing, and the corresponding rolling texture components should also be {011} ⟨211⟩, {225} ⟨554⟩ and {123} ⟨634⟩. After intermediate recrystallization the annealing texture components mainly the {124} ⟨211⟩ and certain {001} ⟨100⟩ will replace the rolling texture[8], in which the {124} ⟨211⟩ has very similar orientation to the {123} ⟨634⟩ produced by rolling. Sometimes both are considered identical. Therefore the following cold rolling results in a rapid accumulation of recrystallized grains around {123} ⟨634⟩ and furthermore along the β fiber towards {225} ⟨554⟩[6], which leads to a higher volume fraction of texture component {225} ⟨554⟩ in sample A. Most grains do not have tendency to accumulate around {011} ⟨211⟩, and the final rolling texture component {011} ⟨211⟩ becomes less in sample A (Table 1).

Table 2 Tensile strength before and after final annealing

Strength in different directions		$\sigma_{0.2}$/MPa RD	$\sigma_{0.2}$/MPa TD	σ_b/MPa RD	σ_b/MPa TD
Rolled	A	385	357	419	424
	B	371	371	421	431
Annealed	A	334	337	389	398
	B	335	348	391	401
Strength drop	A	51	20	30	26
	B	36	23	30	30

4.2 Final recovery annealing

The recovery annealing reduced the dislocation density, work hardening effect and therefore the strength levels of deformed aluminum alloy (Table 2). During rolling a metal sheet experiences a drawing stress in RD and a compressve stress in ND (normal direction to sheet surface), and macroscopically becomes longer in RD and thinner in ND. If the rolling reduction is not very high, a part of the slipped dislocations which have induced grain elongation in RD will migrate in a reverse direction on the original slip plane easily during the following recovery annealing, and therefore release the internal elastic stress induced by deformation and reduce the dislocation density. If in this case the tensile test is performed in the RD, the yield strength will become clearly lower than that of the unrecovered. Generally there is no normal strain in TD, and the recovery of the dislocation structure will not induce much yield strength drop. The strength drop in TD observed in Table 2 is mainly due to the decrease of dislocation density in the whole sample and is not so high as that in RD. In sample B the much higher rolling reduction before recovery annealing has resulted in serious dislocation interactions and complicated dislocation configuration. In this case the reserved migration of dislocations on original slip plane becomes rather difficult, and the yield strength drop is not as high as that in sample A after recovery annealing (Table 2), but is more dependent on the reduction of dislocation density in the whole sample.

If a group of dislocations slip on a special slip plane during cold deformation, the positive and negative dislocations will migrate in opposite directions. With increasing deformation degree the positive and negative dislocations will pile up on the slip plane because of certain obstructs, and result in torsion of grain lattice and randomization of grain orientations. When many grains are oriented around a stable orientation for rolling deformation, the randomization of most grains will increase the scattering width of corresponding rolling texture component. At low rolling reduction the reversed slip of dislocations could clearly reduce the lattice torsion, orientation randomization, and therefore reduce the scattering width of the texture components. The corresponding orientation density will be increased simultaneously, which has been really observed in sample A with low accumulated rolling reduction (see Table 1 and Fig. 3). This effect is not well observed in sample B (Fig. 3) for its high accumulated rolling reduction and complicated dislocation configuration.

Because of the reasons mentioned above the recovery annealing has also resulted in the anisotropy changes of yield strength. The yield strength in RD and TD, e. g. in sample A, is anisotropic before recovery annealing, but becomes almost isotropic after the recovery annealing. In sample B the yield strength is roughly isotropic before recovery annealing, but becomes anisotropic after the recovery annealing. Therefore, combining the intermediate annealing and the recovery annealing the volume fraction of texture components can be changed, and the strength in different directions on alloy sheets can be adjusted.

5 SUMMARY

The intermediate recrystallization annealing has reduced the final cold deformation degree of 5456 aluminum alloy sheet, changed the volume fractions of the rolling texture components, and directly influenced the corresponding strength anisotropy. During the final recovery annealing the dislocations can migrate in a reversed direction on the original slip plane, and therefore apparently reduce the yield strength in rolling direction as well as the randomization of grain orientation distribution. By optimizing the combination of intermediate recrystallization annealing and final recovery annealing the final texture and defect structure could be controlled, and therefore the strength levels in different directions of the alloy sheet could be adjusted.

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