Residual-stress relaxation mechanism and model description of 5052H32 Al alloy spun ellipsoidal heads during annealing treatment

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Abstract Marginal-restraint mandrel-free spinning is an advanced technology for manufacturing ellipsoidal heads with large diameter-thickness ratios. Nevertheless, the spinning-induced residual stress, which greatly influences the in-service performance of spun heads, should be removed. In this study, the effects of annealing on the residual-stress relaxation behavior of 5052H32 aluminum alloy spun heads were investigated. It is found that the residual stress first rapidly decreases and then remains steady with the increase in annealing time at the tested annealing temperatures. The relaxation of the residual stress becomes increasingly obvious with the increase in annealing temperature. When the annealing temperature is less than 220 °C, there are no obvious changes in grain size. Moreover, the spinning-induced dislocations are consumed by the static recovery behavior, which decreases the residual stress during annealing. When the annealing temperature is approximately 300 °C, the broken grains transform into equiaxed grains. In addition, static recrystallization and recovery behaviors occur simultaneously to promote the relaxation of the residual stress. Considering the different stress relaxation mechanisms, a model based on the Zener-Wert-Avrami equation was established to predict the residual-stress relaxation behavior. Finally, the optimized annealing temperature and time were approximately 300 °C and 30 min, respectively.

Keywords Spinning · Residual stress · Annealing treatment · Microstructure evolution · Relaxation mechanism

1 Introduction

With the rapid development of the aerospace industry, large-scale high-performance curved components are increasingly required. The spinning process is usually employed to manufacture integral curved heads owing to its remarkable flexibility and efficiency [1–5]. Usually, the spinning parameters can be optimized to improve the forming accuracy and microstructures of spun parts [6–10]. However, owing to the dramatic plastic deformation in the spinning process, the generation of residual stress in spun heads is inevitable [11–13]. In particular, tensile residual stress causes deterioration of service performance and life [14, 15]. Therefore, it is essential to study the distribution characteristics and relief strategies of the residual stress in spun components [16].

Many studies on the relaxation strategies of residual stress in alloy components have been conducted. For instance, Hu and Wang [17] and Chang et al. [18] studied the influence of solution treatment on spun parts; they found that the residual stress could be eliminated and the grain size refined. Hui et al. [19] revealed that annealing treatment promoted the formation of equiaxed grains in a Cu-Sn alloy spun part. Tsivoulas et al. [20] found that the residual stress decreased in Cr-Mo-V steel tubes during heat treatment. Yu et al. [21] investigated the influence of spinning parameters on the stress distribution and damage of hemispherical Al alloy parts, and optimized spinning parameters were obtained. Lin et al. [22] studied the effects of staggered spinning parameters on the microstructures

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and residual-stress distribution of spun cylindrical parts. Zhou et al. [23] discovered that the residual relaxation behavior mainly occurred during the initial 20 min of exposure, and the extent of relaxation increased with heating temperature. Other studies have shown that the residual-stress relaxation in the annealing process occurs from 20 min to 40 min [24, 25], and then the residual stress remains approximately stable. In addition, many researchers have established different models to describe residual-stress relaxation behavior. Xie et al. [26] developed a finite-element model to predict residual-stress relaxation behavior. Sembiring et al. [27] provided intelligent models for describing the residual-stress relaxation behavior of superalloys. Jagtap and Chason [28] developed a stress relaxation kinetic model to depict the stress recovery and relaxation behaviors in polycrystalline thin films. Juijerm and Altenberger [29] demonstrated the relaxation behavior of residual stress by the Zener-Wert-Avrami formula. The Zener-Wert-Avrami model is widely used to describe the residual-stress relaxation behavior of aluminum alloys [30, 31] and other alloys [32, 33] accurately in the heat treatment process.

In Refs. [34, 35], the spinning parameters have been optimized to improve the forming accuracy and microstructures of ellipsoidal heads. However, the spinning-induced residual stress still significantly influences the in-service performance of spun heads. Although some research has been conducted on the residual-stress relaxation behavior in alloy components, there have been few efforts to develop residual-stress relaxation strategies of 5052H32 aluminum alloy spun heads. However, this issue is extremely important for the industrial production of thin-walled spun alloy components.

In this work, an attempt was made to reduce the residual stress by annealing treatment, and the effects of annealing treatment on the residual-stress relaxation behavior and microstructures of 5052H32 aluminum alloy spun heads were studied in detail. The residual-stress relaxation mechanisms were also analyzed. A model was developed to describe the residual-stress relaxation features of the spun ellipsoidal head during the annealing treatment. Finally, optimized annealing parameters were obtained.

2 Material and experiments

2.1 Spun heads and annealing scheme

In this study, a 5052H32 aluminum alloy was used, and its chemical composition (mass fraction, %) is shown in Table 1. First, the spun heads were manufactured from 5052H32 aluminum alloy square plates (the initial width and thickness were 400 mm and 1 mm, respectively) by

Table 1 Chemical compositions of 5052H32 Al alloy (mass fraction, %)

Mg	Si	Fe	Cu	Mn	Cr	Zn	Al
2.47	0.05	0.27	0.01	0.05	0.25	0.01	Bal.

marginal-restraint mandrel-free spinning, as shown in Fig. 1a. A detailed spinning process has been reported previously [34, 35]. A two-pass spinning process was used in this study. In the first pass, the forming angle, roller installation angle, feed ratio, roller fillet radius, and spindle speed were approximately 40°, 45°, 1.2 mm/r, 6 mm, and 300 r/min, respectively. In the second pass, the angle between passes was 10°, and the other spinning parameters were the same as those in the first pass. Annealing experiments were then carried out on the spun heads. The following annealing temperatures (140 °C, 220 °C, and 300 °C) and annealing times (10–60 min) were designed, as depicted in Fig. 1b.

2.2 Residual-stress measurement

The basic principle of X-ray diffraction (XRD) is mainly to use the interaction of the X-ray beam and material lattice to measure the lattice distance change of the crystal plane caused by residual stress. The residual stress was measured using the XRD equipment in this study; the measuring system provided concrete residual-stress values. XRD test equipment produced by the Proto Company (Canada) was used to measure the residual stress, as shown in Fig. 2a. Cr target was selected, and the Gaussian method was used as the peak positioning technology. The test voltage was 25 kV, and the tube current was 5 mA. The exposure time was 2–4 s, and the number of exposures was seven. The beta angle was between −20° and 20°, and the Bragg angle was set to 156.31°. In addition, there were four measuring points on the external surface, and they were evenly selected in the steady spinning regions of the spun heads, as shown in Fig. 2b. The residual stresses in the generatrix and circumferential directions were measured three times each.

2.3 Microstructure observations

To investigate the effects of the annealing parameters on the microstructures of the spun heads, electron backscatter diffraction analysis was performed. The specimens were cut from the spun head, ground with waterproof abrasive paper, and polished by double-jet electropolishing with 10% HClO₄ and 90% CH₃OH at −30 °C. Microstructures on the external surfaces of the spun heads were observed.

Fig. 1 Spun heads and annealing scheme **a** the spun heads, **b** the annealing scheme

Fig. 2 Measurement of residual stress **a** XRD test equipment, **b** measuring position

3 Results and discussion

3.1 Effects of annealing parameters on the residual stress

Figure 3a shows the initial residual stress in the generatrix and circumferential directions of the spun heads. The residual stress remained approximately stable in both directions because the spinning deformation behavior was stable in the steady spinning region. However, the residual stress in the generatrix direction was obviously greater than that in the circumferential direction. This is because the roller feeds along the generatrix direction, and the material flow is more intense than that along the circumferential direction. Therefore, the degree of plastic deformation is relatively larger along the generatrix direction. Figures 3b

and c depict the effects of the annealing parameters on the residual stress in the generatrix and circumferential directions, respectively. The residual stress in both directions rapidly decreased in the first 10 min and then slowly decreased. Furthermore, the residual stress remained approximately stable after annealing for 30 min. Meanwhile, the residual-stress relaxation law was consistent in both directions during the annealing treatment and mainly occurred in the early annealing stage.

Figures 3d and e show the effects of annealing temperature on the relaxation percentage of residual stress for annealing times of 10 min and 30 min. The relaxation degree of the residual stress in both directions clearly increases with increasing annealing temperature. As depicted in Fig. 3e, when the annealing temperature was 140 °C, the static recovery behavior was inadequate. There

Fig. 3 Effects of annealing time and temperature on **a** the initial residual stress, **b** the residual stress in the circumferential direction, **c** the residual stress in the generatrix direction, **d** the relaxation percentage of residual stress at 10 min, **e** the relaxation percentage of residual stress at 30 min

were still considerable dislocations in the spun head. The relaxation percentages of the residual stress in the circumferential and generatrix directions were 52.32% and

55.2%, respectively. When the annealing temperature was 220 °C, the relaxation percentages of residual stress in the two directions were 85.18% and 83.98%, respectively,

Fig. 4 Microstructures and grain size of the spun heads **a, b** initial, **c, d** 140 °C, 30 min, **e, f** 220 °C, 30 min

Fig. 5 Misorientation angle distribution of **a** the unannealed spun alloy, **b** the annealed spun alloy at 220 °C for 30 min, **c** the annealed spun alloy at 300 °C for 10 mins, **d** the annealed spun alloy at 300 °C for 60 min

because the raised temperature enhanced the static recovery behavior. Furthermore, when the annealing temperature increased to 300 °C (above the recrystallization temperature of 260 °C), the relaxation percentages of residual stress in the two directions were 92.67 % and 90.80 %, respectively. Static recovery and recrystallization [36, 37] occurred simultaneously during annealing. Therefore, the relaxation degree of the residual stress was relatively adequate at approximately 300 °C. At annealing temperatures of 140 °C and 220 °C, the annealing treatment aggravated the atomic diffusion and dislocation rearrangement, inducing thermal recovery and reduced residual stress. When the annealing temperature was 300 °C, recrystallization and recovery behavior occurred simultaneously to consume dislocations and deformation storage energy further. Subsequently, the residual stress decreased. Therefore, the suitable annealing treatment parameters were approximately 300 °C and 30–60 min.

3.2 Effects of annealing parameters on the microstructures

Figures 4a and b depict the microstructures and grain size distribution of the spun heads before annealing, i.e., the slender microstructures are surrounded by many broken grains. Figures 4c and e show the microstructures of the spun heads annealed for 30 min at 140 °C and 220 °C, respectively. Clearly, there are some slender and broken grains, and the size is distributed between 0.5 μm and 25.5 μm . Furthermore, small grains were dominant. There were no obvious differences in the microstructures of the unannealed and annealed alloys. Figure 5 shows the distribution of the misorientation angles for different annealing parameters. The misorientation angle of the spun alloy annealed for 30 min at 220 °C was almost the same as that of the unannealed spun alloy. Both were dominated by the low-angle grain boundary within 0°–10°. Furthermore, there was no obvious recrystallization at lower

Fig. 6 Microstructures and grain size distribution of the annealed spun alloy at 300 °C for **a, b** 10 min, **c, d** 30 min, **e, f** 60 min

temperatures. Therefore, the relaxation of the residual stress mainly relies on static recovery at lower annealing temperatures.

Figure 6 shows the microstructures and grain size distributions of the spun alloy annealed at 300 °C. When the annealing time was 10 min, as shown in Fig. 5c, a high-angle grain boundary (HAGB) gradually appeared. Meanwhile, as Figs. 6a and b show, there were still some elongated or broken grains, and some fine recrystallized grains were found near the grain boundaries. The proportion of small grains was clearly larger than that of the unannealed spun alloy. Therefore, the static recrystallization and recovery behaviors jointly promote the relaxation of the residual stress. When the annealing time was 30 min, as shown in Figs. 6c and d, static recrystallization behavior further occurred, and the number of recrystallized nuclei gradually grew. Therefore, the number of recrystallized grains increased, and the broken grains gradually disappeared during annealing. When the annealing time was increased to 60 min, as shown in Figs. 6e and f, the elongated or broken grains were completely replaced by equiaxed recrystallization grains. In addition, grain growth was very prominent. As shown in Fig. 5d, for spun alloy annealed at 300 °C for 60 min, the HAGBs were dominant, and the mean misorientation angle was 31.5°. This indicates sufficient recrystallization behavior. Therefore, the residual-stress relaxation behavior completely occurs at elevated annealing temperatures.

3.3 Effects of annealing parameters on the dislocation density

The residual stress obtained by XRD tests can be converted into dislocation density, according to Eqs. (1)–(3) [36, 37].

Fig. 7 Effects of annealing time and temperature on the dislocation density of spun heads

$$D = \frac{\lambda}{\beta_C^f \cos \theta}, \quad (1)$$

$$\varepsilon = \frac{\beta_G^f}{4 \tan \theta}, \quad (2)$$

$$\rho = \frac{2\sqrt{3}\langle \varepsilon^2 \rangle^{1/2}}{|b|D_1}, \quad (3)$$

where D represents microcrystal size, λ the wavelength, β_C^f the integral breadth, θ the Bragg angle. ε represents micro-strain; β_G^f is the integral breadth. ρ represents dislocation density; D_1 is the apparent crystallite or domain size; b represents the Burgers vector, which is 0.268 nm for Al alloys [38].

Figure 7 depicts the effects of the annealing time and temperature on the dislocation density. The dislocation density rapidly decreased within the first 10 min and then remained stable at the tested annealing temperatures. This is because most of the spinning-induced dislocations were consumed by the recovery or recrystallization behavior in the early annealing stage, which caused a rapid decrease in the dislocation density. However, the inherent remaining dislocations were difficult to consume. Therefore, the change in dislocation density was slow at the later annealing stage. This was consistent with the relaxation law of residual stress, which was discussed in Sects. 3.1 and 3.2. In addition, the dislocation density decreased with increasing annealing temperature. Meanwhile, the dislocation was further consumed by the static recrystallization behavior at an annealing temperature of 300 °C. Therefore, the dislocation density clearly decreased.

Figure 8 shows the kernel average misorientation (KAM) maps and the distribution of the local misorientation angle in the spun alloy annealed at 300 °C. A large KAM value indicates high dislocation density [38]. Figures 8a and b showed that the local misorientation angle was densely distributed in the range 1°–3.5° for the unannealed spun alloy, i.e., there was a high dislocation density. After annealing for 30 min (see Figs. 8c and d), the local misorientation angle was mainly distributed in the range 0.1°–0.5°, but there was still a small amount of high local misorientation angle. This was because the majority of the dislocation density was consumed by the static recovery and recrystallization behaviors. Therefore, the dislocation density decreased as annealing progressed. When the annealing time was 60 min (see Figs. 8e and f), the dislocation density was further reduced. The local orientation angle was mainly distributed in the range 0.1°–0.8°, i.e., the high dislocation density generated during spinning was almost eliminated. Hence, the change in the dislocation density agrees well with that of the residual stress.

Fig. 8 KAM maps and distribution of local misorientation angle in annealed alloy at 300 °C for **a, b** the unannealed alloy, **c, d** 30 min, **e, f** 60 min, **g** the color label of local misorientation

4 Residual-stress relaxation model

4.1 Establishment of the residual-stress relaxation model

The residual-stress relaxation can be modeled using the Zener-Wert-Avrami equation [29]

$$\frac{\sigma_1}{\sigma_0} = \exp(-(At)^m), \quad (4)$$

where σ_1 and σ_0 are the residual stresses of the annealed and unannealed alloys, respectively, m a material constant related to the release mechanism of residual stress, t the annealing time, and A a function related to the material and annealing temperature, i.e.,

$$A = B \exp\left(-\frac{\Delta H}{kT}\right), \quad (5)$$

where B is a material constant and can be evaluated as $3.06 \times 10^{12} \text{ min}^{-1}$ for aluminum alloy [29], k Boltzmann's constant, which is $8.617343 \times 10^{-5} \text{ eV/K}$, T the annealing temperature, and ΔH the activation enthalpy during the release process of the residual stress. Equation (6) is obtained by the logarithmic transformation of Eq. (4). The relationship between $\lg(-\ln(\sigma_1/\sigma_0))$ and $\lg(t)$ is linear at a given temperature.

$$\lg\left(-\ln\left(\frac{\sigma_1}{\sigma_0}\right)\right) = m \lg A + m \lg(t). \quad (6)$$

Figure 9 shows the distribution of $\lg(-\ln(\sigma_1/\sigma_0))$ and $\lg(t)$ at different annealing temperatures. The relationship between $\lg(-\ln(\sigma_1/\sigma_0))$ and $\lg(t)$ in the circumferential and generatrix directions of the spun heads can be well fitted by straight lines. This indicates that the relaxation behavior of the residual stress during the annealing process satisfies the Zener-Wert-Avrami equation. The slope of the fitting line represents the value of the material constant

m in Eq. (6), and the activation enthalpy ΔH is obtained by regression analysis in combination with Eq. (5), as shown in Tables 2 and 3.

The aforementioned analysis demonstrates that the values of m in the circumferential and generatrix directions at 300 °C are larger than those at 140 °C and 220 °C. This shows that the relaxation mechanism of the residual stress at 300 °C is different from that at 140 °C and 220 °C. Furthermore, this conclusion is consistent with previous experimental findings. Therefore, according to the different relaxation mechanisms of residual stress, static recovery, and recrystallization mechanisms, the model can be revised in the following format. Here, m and ΔH are the average values in both directions.

$$\begin{cases} \sigma_1 = \sigma_0 \exp\left(-(Bt)^m \exp\left(-\frac{n}{(T + 273.15)}\right)\right), \\ m = 0.20959, n = 2961.18 \quad (T \leq 260^\circ\text{C}), \\ m = 0.27404, n = 4499.83 \quad (T > 260^\circ\text{C}). \end{cases} \quad (7)$$

Table 2 Values of m and ΔH of circumferential residual stress

Temperature/°C	m	$\Delta H/\text{eV}$
140	0.245	1.20
220	0.192	1.23
300	0.276	1.40

Table 3 Values of m and ΔH of generatrix residual stress

Temperature/°C	m	$\Delta H/\text{eV}$
140	0.193	1.19
220	0.209	1.25
300	0.272	1.43

Fig. 9 Distribution of $\lg(-\ln(\sigma_1/\sigma_0))$ and $\lg(t)$ at different annealing temperatures in the **a** circumferential direction **b** generatrix direction (t/min)

Fig. 10 Comparisons of the predicted and experimental residual stresses **a** in the circumferential direction, **b** in the generatrix direction, **c** correlation coefficient

4.2 Verification of the residual-stress relaxation model

To verify the accuracy of the established model, the residual stresses predicted by the model were compared with the experimental values, as shown in Figs. 10a and b. The prediction results at the tested annealing parameters agree well with the experimental results. Meanwhile, the correlation coefficient R between the predicted and experimental residual stresses can be calculated using Eq. (8), where E_i and P_i are the experimental and predicted residual stresses, respectively, and $\bar{E}$ and $\bar{P}$ the mean values of E_i and P_i , respectively. The experimental and predicted residual stresses are shown in Fig. 10c. The value of the correlation coefficient R is 0.986 6, and the absolute

average error is 4.37 MPa. This indicates that the established model has a high accuracy in predicting the residual-stress relaxation in the 5052H32 aluminum alloy spun heads during annealing treatment.

$$R = \frac{\sum_{i=1}^N (E_i - \bar{E})(P_i - \bar{P})}{\sqrt{\sum_{i=1}^N (E_i - \bar{E})^2 \sum_{i=1}^N (P_i - \bar{P})^2}} \quad (8)$$

5 Conclusions

The effects of the annealing parameters on the residual-stress relaxation behavior and microstructures were examined in detail. The residual-stress relaxation

mechanisms were also analyzed. A model was developed to describe the residual-stress relaxation features of the spun ellipsoidal heads during the annealing treatment. Some conclusions can be drawn as follows.

- (i) Residual-stress relaxation mainly occurs during the early annealing stage. Increasing the annealing temperature decreases the residual stress. The change in the dislocation density is similar to that of the residual stress.
- (ii) When the annealing temperature is less than 220 °C, there is no obvious difference in the microstructures of the unannealed and annealed alloys. The residual stress decreases owing to the static recovery behavior. When the annealing temperature is approximately 300 °C, static recrystallization and recovery occur simultaneously to reduce the residual stress further.
- (iii) The correlation coefficient of the developed model is 0.986 5. The model can well describe the residual-stress relaxation features of spun ellipsoidal heads during annealing treatment. The optimized annealing temperature and time are approximately 300 °C and 30 min, respectively.

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