Internal Residual Stress Measurements of Tensile-Deformed Aluminum Single Crystals Using Synchrotron Radiation*

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Abstract

The changes in the microstructures associated with plastic deformation and strain/stress were measured using the strain scanning method with hard synchrotron radiation. Aluminum single crystals tensile-deformed along the $\langle 111 \rangle$ direction exhibit a macroscopically uniform multiple-slip without deformation bands. Plastic deformation has an insignificant effect on lattice spacing, and is likely responsible for the relatively uniform distribution of internal stress with a small absolute value. With regard to the crystallographic orientation, the rocking curves in the as-annealed sample differed with position, strongly suggesting that mosaic structures exist in a single crystal. In the 8% tensile-deformed sample, mosaic blocks are divided into smaller size. The increase in the number of mosaic blocks with deformation causes broadening the rocking curve and FWHM of the profile.

Key words: Aluminum Single Crystal, Tensile Deformation, Synchrotron Radiation, Strain, Stress

1. Introduction

This study analyzes the tensile-deformation of aluminum single crystals along the $\langle 111 \rangle$ direction using synchrotron radiation. Herein changes in microstructures associated with plastic deformation and strain/stress distributions in single crystals are reported.

Plastic deformation and subsequent recrystallization induced by annealing have long been studied because they control the texture of polycrystalline materials. The most fundamental process of plastic deformation is a dislocation slip. In real crystalline materials, plastic deformation not only induces dislocation slips, but also gives rise to complex phenomena such as interactions between dislocations, multiplication and pile-up of dislocations⁽¹⁾.

Due to a reduction in correlated factors, an experimental study using single crystals should elucidate the true nature of such complex phenomena. Tensile deformation is better suited for single crystals than rolling because the shear stress acting on each slip system is

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easily calculated. In addition, selecting the appropriate tensile direction can control the deformation nature of single crystals. For example, in face-center cubic (FCC) crystals, macroscopically uniform deformation occurs without non-uniformly deformed regions when tensile-deformation is along the $\langle 111 \rangle$, $\langle 112 \rangle$, or $\langle 001 \rangle$ direction⁽²⁾⁻⁽⁵⁾. Kashihara et al. reported the tensile deformation and recrystallization of $\langle 111 \rangle$ -oriented aluminum single crystals⁽⁶⁾. From the orientation relationship between the deformation matrix and recrystallized grains, they proposed a dislocation network model for the nucleus of recrystallized grains⁽⁷⁾.

Previous studies on deformation and recrystallization using single crystal samples have focused on qualitative aspects of these phenomena such as microstructural observations. However, quantitative research related to the residual strain and stress in deformed single crystals has yet to be conducted. It is natural to assume that an increase in the dislocation density due to plastic deformation leads to the evolution of residual strain and stress and that the deformation-induced crystallographic rotation strongly influences the recrystallization behavior. Consequently, residual strain/stress analyses should aid in understanding the true nature of deformation and recrystallization.

The easiest non-destructive method for strain/stress measurements applicable to crystalline materials is X-ray diffraction⁽⁸⁾. Recently, this technique has been applied to synchrotron radiation, which has many advantageous characteristics (e.g., high brilliance, high spatial resolution, excellent directionality, and energy (or wavelength) variability). In a 3rd generation facility such as SPring-8, the internal strain/stress in materials can be evaluated using high energy X-rays exceeding 30 keV (wavelengths shorter than 0.041 nm)⁽⁹⁾.

In the present study, crystallographic orientations and local strain/stress in aluminum single crystals tensile-deformed along the $\langle 111 \rangle$ direction were measured via the strain scanning method⁽¹⁰⁾ using high energy synchrotron radiation.

2. Principles of a single crystal stress measurement by X-ray diffraction⁽¹¹⁾

Figure 1 defines three types of Cartesian coordinate systems; the sample coordinate P_i , the laboratory coordinate L_i , and the crystal coordinate C_i . P_3 is coincident with the tensile direction of the sample, while L_3 is defined as normal to the diffraction plane. ε_{33}^L is the normal strain measured along the L_3 direction by X-ray diffraction.

Figure 1 Relation among the three coordinate systems

Figure 2 Relation of the transformation matrices

Figure 2 shows the relationships of these three types of the coordinate systems, where π , γ , and ω are the directional cosines between coordinate axes.

The normal strain ε_{33}^L in the L_3 direction is written using ε_{ij}^C in the crystal coordinate C_i by the following equation

$$\begin{aligned}\varepsilon_{33}^L &= \gamma_{3i}\gamma_{3j}\varepsilon_{ij}^C \\ &= \gamma_{31}^2\varepsilon_{11}^C + \gamma_{32}^2\varepsilon_{22}^C + \gamma_{33}^2\varepsilon_{33}^C + 2(\gamma_{32}\gamma_{33}\varepsilon_{23}^C + \gamma_{33}\gamma_{31}\varepsilon_{31}^C + \gamma_{31}\gamma_{32}\varepsilon_{12}^C)\end{aligned}\quad (1)$$

In the case of a FCC system, the stress-strain relation is described using the elastic compliances s_{ij} as

$$\left. \begin{aligned}\varepsilon_{11}^C &= (s_{11} - s_{12})\sigma_{11}^C + s_{12}(\sigma_{11}^C + \sigma_{22}^C + \sigma_{33}^C) \\ \varepsilon_{22}^C &= (s_{11} - s_{12})\sigma_{22}^C + s_{12}(\sigma_{11}^C + \sigma_{22}^C + \sigma_{33}^C) \\ \varepsilon_{33}^C &= (s_{11} - s_{12})\sigma_{33}^C + s_{12}(\sigma_{11}^C + \sigma_{22}^C + \sigma_{33}^C) \\ 2\varepsilon_{23}^C &= s_{44}\sigma_{23}^C \\ 2\varepsilon_{31}^C &= s_{44}\sigma_{31}^C \\ 2\varepsilon_{12}^C &= s_{44}\sigma_{12}^C\end{aligned}\right\} \quad (2)$$

Since the main purpose is to correlate the normal strain ε_{33}^L in terms of the stresses σ_{ij}^P in the sample coordinate system, the following relationship between σ_{ij}^C and σ_{ij}^P is used with the help of transformation matrix π in Fig. 2⁽¹²⁾.

$$\sigma_{ij}^C = \pi_{ik}\pi_{jl}\sigma_{kl}^P \quad (3)$$

Substituting eqs. (2) and (3) into eq. (1), ε_{33}^L can be expressed as

$$\varepsilon_{33}^L(n) = A_n\sigma_{11} + B_n\sigma_{12} + C_n\sigma_{22} + D_n\sigma_{23} + E_n\sigma_{31} + F_n\sigma_{33} \quad (4)$$

where

$$\begin{aligned}A_n &= s_0(\gamma_{31}^2\pi_{11}^2 + \gamma_{32}^2\pi_{12}^2 + \gamma_{33}^2\pi_{13}^2) + s_{12} + \frac{1}{2}s_{44}\cos^2\varphi\sin^2\psi \\ B_n &= s_0(\gamma_{31}^2\pi_{21}^2 + \gamma_{32}^2\pi_{22}^2 + \gamma_{33}^2\pi_{23}^2) + s_{12} + \frac{1}{2}s_{44}\sin^2\varphi\sin^2\psi \\ C_n &= s_0(\gamma_{31}^2\pi_{31}^2 + \gamma_{32}^2\pi_{32}^2 + \gamma_{33}^2\pi_{33}^2) + s_{12} + \frac{1}{2}s_{44} - \frac{1}{2}s_{44}\sin^2\psi \\ D_n &= 2s_0(\gamma_{31}^2\pi_{11}\pi_{21} + \gamma_{32}^2\pi_{12}\pi_{22} + \gamma_{33}^2\pi_{13}\pi_{23}) + \frac{1}{2}s_{44}\sin 2\varphi\sin^2\psi \\ E_n &= 2s_0(\gamma_{31}^2\pi_{21}\pi_{31} + \gamma_{32}^2\pi_{22}\pi_{32} + \gamma_{33}^2\pi_{23}\pi_{33}) + \frac{1}{2}s_{44}\sin\varphi\sin 2\psi \\ F_n &= 2s_0(\gamma_{31}^2\pi_{31}\pi_{11} + \gamma_{32}^2\pi_{32}\pi_{12} + \gamma_{33}^2\pi_{33}\pi_{13}) + \frac{1}{2}s_{44}\cos\varphi\sin 2\psi \\ s_0 &= s_{11} - s_{12} - \frac{s_{44}}{2}\end{aligned}\quad (5)$$

Measuring the six ε_{33}^L values in independent directions, the stress components σ_{11}^P , σ_{12}^P , σ_{22}^P , σ_{23}^P , σ_{31}^P , and σ_{33}^P can be calculated from eq. (4).

The direction cosines of the coordinate transformation between L_3 and C_i are defined by γ_{3k} . When the diffraction plane is given by hkl , γ_{3k} is described by the following equations

$$\gamma_{31} = \frac{h}{\sqrt{h^2 + k^2 + l^2}}, \quad \gamma_{32} = \frac{k}{\sqrt{h^2 + k^2 + l^2}}, \quad \gamma_{33} = \frac{l}{\sqrt{h^2 + k^2 + l^2}} \quad (6)$$

Furthermore, the direction cosines of the coordinate transformation between P_i and C_i are defined by π_{ij} . If P_3 and P_1 are designated as the tensile direction $[n_1 \ n_2 \ n_3]$ and the vertical direction $[w_1 \ w_2 \ w_3]$ to the tensile axis, respectively, then the transformation matrix π is described as

$$\pi = \begin{pmatrix} \pi_{11} & \pi_{12} & \pi_{13} \\ \pi_{21} & \pi_{22} & \pi_{23} \\ \pi_{31} & \pi_{32} & \pi_{33} \end{pmatrix}$$

$$\pi_{11} = \frac{w_1}{\sqrt{w_1^2 + w_2^2 + w_3^2}}, \quad \pi_{12} = \frac{w_2}{\sqrt{w_1^2 + w_2^2 + w_3^2}}, \quad \pi_{13} = \frac{w_3}{\sqrt{w_1^2 + w_2^2 + w_3^2}} \quad (7)$$

$$\pi_{31} = \frac{n_1}{\sqrt{n_1^2 + n_2^2 + n_3^2}}, \quad \pi_{32} = \frac{n_2}{\sqrt{n_1^2 + n_2^2 + n_3^2}}, \quad \pi_{33} = \frac{n_3}{\sqrt{n_1^2 + n_2^2 + n_3^2}},$$

$$\pi_{21} = \pi_{13}\pi_{32} - \pi_{12}\pi_{33}, \quad \pi_{22} = \pi_{11}\pi_{33} - \pi_{13}\pi_{31}, \quad \pi_{23} = \pi_{12}\pi_{31} - \pi_{11}\pi_{32}$$

3. Sample preparation

Aluminum single crystals were grown via the Bridgman method. The purity of the material was 99.999 mass%. Figure 3 shows the size of the single crystal sample. The sample with a cylindrical gauge portion was directly formed in the mold. A seed technique controlled the orientation of the crystal, which is shown in Fig. 4. The samples were annealed at 573 K for 3.6 ks before tensile testing.

The $\langle 111 \rangle$ samples were deformed to a tensile strain of 8% at a strain rate of $3.0 \times 10^{-4} \text{ s}^{-1}$ at room temperature. The samples showed very large work-hardening as reported by Tagami et al⁽⁴⁾. The flow stress reached about 60 MPa at a strain of 8%. As will be reported elsewhere, we have found that the minimum value of tensile strain is 8% for $\langle 111 \rangle$ orientation aluminum single crystals to recrystallize after post-deformation annealing at 0.8 T_M (T_M [K]: melting temperature).

Figure 3 Schematic illustration of a single crystal sample

Figure 4 Schematic illustration of the crystal orientation

In the present study, two samples, A and B, were tested. Sample A was measured as-annealed (0% strain) and after tensile testing (8% strain). Sample B was measured only

after deformation (8% strain). Below the measurements are referred to by a combination of the sample name (A or B) and the strain (0% or 8%). For example, A-0 corresponds to Sample A at the as-annealed stage. Table 1 provides the details.

Table 1 Sample details

Sample	1 st experiment	2 nd experiment
A	Anneal (A-0)	Tensile-deformed to 8% strain (A-8)
B	Tensile-deformed to 8% strain (B-8)	-

4. Experimental procedure

Beamline BL22XU of JAEA (Japan Atomic Energy Agency) at SPring-8 was employed to conduct two experiments. The X-ray energies, 30.038 keV and 30.008 keV, were calibrated from a number of diffraction lines of CeO₂ powder sample by regression analysis.

A horizontal type four-circle goniometer was used, and the sample was set on the xyz-stage attached to the goniometer. Figure 5 depicts the relationship between the axes of the goniometer, the xyz-stage and the sample.

Figure 5 Relationship between the axes of the goniometer, the xyz-stage and the sample

The X-ray stress measurement was based on the evaluation of the lattice spacing by Bragg's law, which is expressed as

$$\lambda = 2d \sin \theta \quad (8)$$

where λ is the X-ray wavelength, d is the lattice spacing, and 2θ is the diffraction angle. The relationship between X-ray wavelength and X-ray energy (E) is given by

$$E(keV) = \frac{hc}{\lambda} = \frac{1.2398}{\lambda(nm)} \quad (9)$$

where h is Planck's constant and c is the speed of light.

The lattice spacing was evaluated by measuring diffraction angles, and the lattice strain was then calculated by the following equation

$$\varepsilon = \frac{d - d_0}{d_0} \quad (10)$$

where d_0 is the stress-free lattice spacing. The stress components were calculated by eq. (4). If six strains correspond to six independent poles, then the stress components σ_{11}^P , σ_{12}^P , σ_{22}^P , σ_{23}^P , σ_{31}^P , and σ_{33}^P can be calculated by a simultaneous equation.

Table 2 shows the X-ray conditions for the strain scanning method, and Fig. 6 schematically illustrates the measurement positions in the cross-section of the center of the gauge portion.

Table 2 X-ray conditions for the strain scanning method

Energy	30.038 keV 30.008 keV
Monochromatic crystal	Si (111)
Size of divergent slits	$0.2^H \times 0.05^W \text{ mm}^2$
Size of receiving slits 1	$3.0^H \times 0.05^W \text{ mm}^2$
Size of receiving slits 2	$3.0^H \times 0.05^W \text{ mm}^2$
Distance between sample and receiving slit 1	275 mm
Distance between receiving slits 1 and receiving slit 2	1210 mm
Detector	NaI
Diffraction plane	Al (440)
Diffraction angle 2θ	33.51 deg. @ 30.038 keV 33.54 deg. @ 30.008 keV

Figure 6 Schematic illustration of the measured positions for (a) 0% strain and (b) 8% strain

To evaluate the residual stress in three kinds of aluminum single crystals, peak refinements and profile scans were carried out at each pole of the measured positions. Tables 3–5 show the average values of 2θ , ω , χ , and ϕ at each pole.

Table 3 Average values of 2θ , ω , χ , and ϕ at each pole in the annealed sample (A-0)

Pole No.	2θ	ω	χ	ϕ
1	33.512	16.756	-3.7991	-37.018
2	33.512	16.756	2.4284	23.059
3	33.512	16.756	-1.5196	82.934
4	33.512	16.756	-51.734	-63.686
5	33.512	16.756	-53.987	47.568
6	33.512	16.756	-58.504	175.34

Table 4 Average values of 2θ , ω , χ , and φ at each pole in the 8% tensile-deformed sample (B-8)

Pole No.	2θ	ω	χ	φ
1	33.510	16.755	-1.2811	-81.991
2	33.510	16.755	1.6505	-21.985
3	33.511	16.755	2.8307	38.060
4	33.513	16.757	-57.493	-173.63
5	33.512	16.756	-54.335	-47.956
6	33.512	16.756	-52.469	65.724

Table 5 Average values of 2θ , ω , χ , and φ at each pole in the 8% tensile-deformed sample (A-8)

Pole No.	2θ	ω	χ	φ
1	33.542	16.771	1.5843	-31.259
2	33.542	16.771	2.4586	28.788
3	33.542	16.771	0.77627	88.802
4	33.543	16.772	-54.279	-57.492
5	33.544	16.772	-52.831	56.460
6	33.544	16.772	-57.501	177.44

5. Results and discussion

5.1 Rocking curves of the crystallographic orientation

Figure 7 (a) shows rocking curves about the φ axis scan obtained from pole 1 in A-0. In general, the shape of the rocking curve obtained from a fine single crystal (e.g., silicon or ceramic) is narrow, symmetric, and independent of the measured position. The rocking curves in Fig. 7 (a) are very narrow because the A-0 sample is as-annealed, but the peak position and shape of the rocking curves measured at each sty position differ. Figure 7 (b) shows rocking curves obtained from pole 1 in A-8. Compared to the rocking curves of A-0, which are denoted by square symbols in Fig. 7 (b), the rocking curves of A-8 are much broader than those of A-0, but both exhibit nearly symmetric shapes.

Figure 7 Rocking curves about the φ scan obtained from pole 1 in (a) A-0 and (b) A-8

5.2 Two-dimensional (2D) map of crystallographic orientation

A Gaussian fitting was used to determine the peak position and full width at half maximum (FWHM) of each rocking curve. Figure 8 (a) represents the contour map of the peak positions of pole 1 for a $1.2 \text{ mm} \times 1.2 \text{ mm}$ area in A-0. Although the sample is a single crystal, the measured area has a misorientation of about 0.5° , continuously varying along the stx axis. This continuity is unlike a polycrystal with a grain boundary. Figure 8 (b) is a 2D map of the peak positions of pole 1 in A-8, and has the same area and counter scale as Fig. 8 (a). The range of misorientation in A-8 is almost equal to that in A-0, but the orientation distribution in A-8 is irregular compared to that in A-0. Comparing the as-annealed [Fig. 8 (a)] and deformed [Fig. 8 (b)] states confirms that the crystallographic rotation induced by deformation is not uniform.

Figure 8 Two-dimensional map of the crystallographic orientation at pole 1 in (a) A-0 and (b) A-8

To clarify that the discontinuous distribution [Fig. 8 (b)] is independent of the experimental conditions, sample B was deformed via tension to the same strain (8%) as sample A. Figures 9 (a) and (b) present 2D maps for A-8 and B-8, respectively, where peak positions are determined for pole 1 in A-8 and pole 2 in B-8. For comparison, Fig. 9 (c) maps A-0. In all three maps, variations in the angles use the same color scale. For both samples, the orientation variation in the center is smaller than that in the periphery. Although the crystallographic rotations are non-uniform across the cross-section, the peripheral region has a relatively large rotation.

The above results support the hypothesis that the mosaic structures⁽¹³⁾, which are schematically shown in Fig. 10, exist in the as-annealed single crystal sample without strain. The mosaic blocks are estimated to be 1,000 to 10,000 times the lattice constant⁽¹³⁾. Hence, for aluminum, mosaic blocks typically have diameters between $0.4 \mu\text{m}$ and $4.0 \mu\text{m}$. Because the irradiated region of the present experiment is $173 \mu\text{m} \times 52.2 \mu\text{m}$, a large number of mosaic blocks are present. As shown in Fig. 7 (a), the shape of rocking curves depends on the measurement positions. This can be explained by the difference in the number and orientation of mosaic blocks within the irradiated regions. Presumably, dislocations introduced by plastic deformation break the mosaic blocks into smaller blocks. Hence, the deformed sample contains a large number of blocks. The rocking curves in Fig. 7 (b) are a summation of the rocking curves from such fine blocks, which results in the pseudo-normal distribution shape of the curves. On the other hand, the inhomogeneous crystallographic rotation observed in the deformed sample may be due to a macroscopic influence. In short, we think that this phenomenon is attributed to the misorientation of approximately 3° between the tension load direction and the $\langle 111 \rangle$ direction in the sample. Another factor is the incomplete fixation at the both ends of the sample during tensile testing.

SEM/ECP or SEM/electron backscatter diffraction (EBSD) cannot measure the crystallographic orientation on such an ultrafine scale. The present measurements using synchrotron radiation are the first to unveil the existence of fine mosaic structures and their crystallographic orientation in a metal single crystal.

Figure 9 Two-dimensional map of crystallographic orientation for (a) pole 1 in A-8, (b) pole 2 in B-8, and (c) pole 1 in A-0

Figure 10 Schematic illustration of the mosaic structure

5.3 Peak profile

Because synchrotron radiation measurements of stress or strain must be optically aligned prior to each experiment, the conditions between experiments changed slightly. Therefore, it is appropriate to compare the results measured in the same optical system, especially when discussing the small peak shift between the as-annealed and deformed

states. In the following, the peak profiles obtained from the as-annealed (A-0) and deformed (B-8) samples are compared.

Figure 11 shows examples of peak profiles for both samples. The peak profile of B-8 is three to four times broader than that of A-0, but the shift in the peak position is very small, approximately 0.002° . The lattice strain calculated from this value using eqs. (8) and (10) is approximately $50 \mu\text{St}$. On the other hand, the crystallite size is estimated using Scherrer's equation⁽¹⁴⁾ [eq. (11)]

$$B_s = \frac{0.94\lambda}{D \cos \theta} \quad (11)$$

In eq. (11), B_s is FWHM in radians, λ is the X-ray wavelength, and D is the crystallite diameter. Table 6 shows the calculated values of D for the six poles in both A-0 and B-8. In the preceding section, we estimated that the mosaic blocks are between $0.4 \mu\text{m}$ and $4.0 \mu\text{m}$. The average crystallite size of A-0 ($0.416 \mu\text{m}$) agrees with estimated value of the mosaic blocks. Hence, it is presumed that the mosaic blocks in the as-annealed sample correspond to the crystallite estimated from the rocking curves. For B-8, the average crystallite size is $0.119 \mu\text{m}$, which is approximately one-four of that of A-0, supporting the hypothesis that plastic deformation induces the division of the mosaic blocks into finer ones. Because each finer block has a different crystallographic orientation and is independent, the strain in each block must differ. Hence, the peak profile of B-8 in Fig. 11 is the summation of numerous strained crystallites, i.e., mosaic blocks. The broadening of the peak profile is probably due to increased number of blocks in the measured region.

Note that the calculated values of D are not 'absolute' value because we don't consider the error of the optical system and the angle divergence of incident X-rays. If we consider two factors, the calculated values of D become bigger. Although the modulus evaluation of the crystallite size is difficult, it is clear that the crystallite sizes of the plastically deformation sample become smaller than the as-annealed sample.

Figure 11 Diffraction peak profile

Table 6 Average crystallite size [μm]

Sample	A-0	B-8
1	0.402	0.114
2	0.404	0.121
3	0.425	0.117
4	0.412	0.122
5	0.424	0.115
6	0.431	0.124
Average	0.416	0.119

5.4 Strain/stress analysis

The lattice strain of B-8 was calculated using eq. (10). Then the d_0 value was determined from the average of all lattice spacings obtained from all positions at the six poles in A-0. Figure 12 (a) is the 2D strain map at pole 1 in B-8. Note that the absolute value of the strain is extremely small (less than $100 \mu\text{St}$), even though the strain variation is enhanced in the color map. Hence, the strain distribution is relatively uniform throughout

the measured area. The other poles produced similar results. Therefore, even if the irradiated region is asymmetrical, the value of strain is not affected. If the value of lattice strain is 8%, which is the same as macrostrain in A-8 and B-8, then the calculated value of peak shift $\Delta 2\theta$ between the diffraction angle of the as-annealed sample and that of the deformed sample is about 2.5° . However, the peak shift in the present study is approximately 0.0004° . This discrepancy may be ascribed to two factors. The first is the release of the applied stress associated with the removal of a deformed single crystal from the tensile testing system. The second is the introduction of dislocations during plastic deformation limits the change in the lattice spacing, inducing the division of the mosaic blocks into finer sizes. In short, because the residual strain is small, it may not affect the variation of the lattice spacing among the divided mosaic blocks.

Figures 12 (b), (c), and (d) show stress maps of σ_{11} , σ_{22} , and σ_{33} , respectively. Due to the very small strain, the absolute values of the stresses are less than 10 MPa. Similar to the preceding discussion of strain variation, the internal stress distribution is relatively uniform in a tensile-deformed aluminum single crystal along the $\langle 111 \rangle$ direction.

Figure 12 Two-dimensional maps of (a) strain of pole 1 of B-8, (b) stress of σ_{11} , (c) stress of σ_{22} , and (d) stress of σ_{33}

It has been reported that the layered dense dislocation walls between cells exist as typical microstructures of tensile-deformed aluminum single crystals along the $\langle 111 \rangle$ direction⁽⁴⁾. In addition, the estimated mosaic block size of $0.119 \mu\text{m}$ in B-8 is smaller than

the averaged cell size reported in ref. 7. Therefore, the results of this study suggest that mosaic blocks may be able to construct a cell.

Finally, the present experiments confirmed that it is possible to estimate the internal distribution of the residual stress in a single crystal sample using high-energy synchrotron radiation. However, the crystallographic quality of a metal single crystal is inferior to a perfect crystal. The mosaic block structure increases the complexity of strain/stress analysis. Therefore, using micro or nanometer order beam focused normal synchrotron radiation allow more precise measurements of the microscopic residual stress in dislocated microstructures, which will provide more information of individual mosaic blocks and enable the deformation to recrystallization mechanism to be discussed.

6. Summary

Synchrotron radiation was used to measure the internal distribution of the residual stress in aluminum single crystals deformed along the $\langle 111 \rangle$ direction. From the experimental findings, we conclude:

1. The difference in the rocking curves obtained from each position in the as-annealed sample strongly suggests that a single crystal possesses a mosaic structure with an estimated size of $0.416 \mu\text{m}$. Annealing likely induces a somewhat regular arrangement of blocks. The misorientation within the measured area is approximately 0.5° .
2. In an 8% tensile-deformed sample, the mosaic blocks are divided into smaller size, which is estimated to be $0.119 \mu\text{m}$. The crystallographic rotation is not uniform throughout the measured area.
3. The increased number of mosaic blocks in the deformed sample may lead to broader rocking curves and increased FWHM profiles.
4. The relatively uniform distribution of internal stress with a small absolute value is presumably because plastic deformation negligibly impacts lattice spacing.

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