

Residual stress and localized lattice rotation under fatigue loading

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ABSTRACT

A significant change in residual stress around a fatigue crack has been observed in Al alloys. For Al 7075, the residual stress increases by 200% closer to the crack. However, in commercially pure Al (Al 1100), the residual stress decreases by 80% closer to the crack. The residual stresses are mostly the stored elastic energy as a result of deformation around the fatigue crack. To understand these results, we estimate the dislocation density as a function of distance from the crack using the full width at half maxima of 111 reflection for both alloys. The dislocation density increases by 25–30 % for Al 7075. However, it actually decreases by 10% for Al 1100. Such change in dislocation density cannot explain the observed change in residual stress closer to the crack. Additionally, the deformation leads to a significant lattice rotation, as observed by the x-ray diffraction and the electron backscatter diffraction measurements. We argue that dislocations are needed to produce the lattice rotation and the estimated lattice rotation is $\approx 11^\circ$, and conclude that the deformation associated with the lattice rotation would be a major contributing factor to the residual stress. This provides a new insight on the role of lattice rotation on the residual stress under fatigue loading.

1. Introduction

The characteristics of plastic zone are crucial for understanding the underlying physics of the deformation process and the generation of residual stress around a crack tip under fatigue loading. The residual stresses are introduced in the material as a result of nonlinear deformation or phase transformation during processing of materials or during cooling. Usually, the deformation in metal forming processes, such as rolling, extrusion, welding and metal cutting introduces residual stresses. Here by residual stresses, we mean those stresses existing within the specimen in the absence of any external load. Residual stresses in our context are, by definition, stresses that remain in the solid material after unloading, i.e. when the original cause of the stresses has already been removed. As is well known, these residual stresses can be either compressive or tensile in nature. The only requirement from mechanics is that in unloaded material on the free surfaces these stresses must be zero. This applies to free surfaces, and also on cracks if the cracks are open. The effect of residual stress is quite significant in determining the lifetime of a material. For example, the introduction of compressive residual stress, enhances the fatigue lifetime [1], however, the tensile residual stress is highly damaging under fatigue loading [2].

Considerable work has been done to study the effects of residual stresses on the fatigue crack propagation. The presence of tensile

residual stresses increases the crack growth rate by increasing the effective load ratio. The residual stress can affect the crack propagation by influencing significantly the mean stress during cyclic loading and the fatigue life [3–6]. In particular, the stress intensity factor can be affected by the residual stress [7–12], which is an important parameter for analyzing crack propagation. For the crack increment, the local residual stress may be more significant than the overall large scale distribution of the residual stresses. Our results indicate that in addition to localized internal stress there is a local lattice rotation that may affect the path and mode of crack propagation. Therefore, there is a need to a new experimental and theoretical approach when assessing the influence of residual stresses in the study of fracture and fatigue. Theoretical approaches of mapping the residual stresses may not be sufficient for a realistic approach to these processes. In addition to crack increment, it has been reported that the local residual stresses can affect the initiation fracture toughness values of aluminum alloys [13–15].

It would be of interest to understand how the microstructural change during the fatigue crack propagation affects the residual stress around the crack in the material, which will provide a new insight of how to optimize the microstructure to achieve greater fatigue lifetime. Recently, we have reported a significant lattice rotation and the change of texture in the plastic zone around the crack grown by fatigue loading in Al alloys by employing X-ray diffraction (XRD) and transmission electron microscopy (TEM) techniques [16,17]. The intensity of 111

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peak gradually increases as compared to the intensity of 200 peak as we approach the crack, suggesting lattice rotation. Such lattice rotation during deformation is related to the redistribution of dislocations and slip processes during crack growth, and can increase the residual elastic stresses. Recently, high resolution electron back scattered diffraction (HREBSD) technique has been used to measure the elastic residual stress and associated the lattice rotation close to the twin tips, within twins and with immediate neighboring grains [18]. The XRD-based techniques have also been used by several authors to obtain residual stresses [19,20].

For elastic strain and lattice rotation measurements, the other diffraction-based technique has been employed using the shift in the Kikuchi diffraction patterns from the deformed and undeformed samples [21–23]. There were some prior works on the crack-tip plastic strain fields measured during crack growth (in-situ) by stereo-imaging technique, dislocation etch pitting, X-ray microbeam, backscattered-imaging scanning electron microscopy and selected-area electron-channeling-pattern analysis [23–27]. Here we measure the elastic strain ex-situ, owing to the change in lattice spacing, with the high resolution XRD technique around the crack, and obtain the residual stresses, and correlate with the microstructural change occurred around the crack as a result of crack propagation. The purpose of the present work is to investigate whether the lattice rotation has any effect on the residual stress around the crack in two Al alloys, Al-7075 and Al-1100. We show that the residual stress increases gradually around the fatigue crack by more than 200% for Al-7075. However, it decreases by 80% for Al-1100 as we approach the fatigue crack.

2. Experimental methods

Fatigue crack growth tests were performed along the short transverse (ST) direction in vacuum ($< 6 \times 10^{-6}$ Pa) background pressure at a cyclic load, frequency of 10 Hz using a sine waveform and a load ratio of 0.10 on compact tension (CT) specimen, 80 mm in length, 60 mm in width and 10 mm in thickness. To study the plastic zone, the crack growth was arrested at a different crack length corresponding to a certain difference in stress intensity factor (ΔK) close to the Paris region, $\Delta K \approx 13 \text{ MPa m}^{1/2}$ and $9 \text{ MPa m}^{1/2}$ for Al 7075 and Al 1100, respectively. The forward plastic zone (r_f) is estimated by the expression, $r_f = (1/2\pi) * (K_{\max}/\sigma_{ys})^2$, where σ_{ys} is the yield strength, $K_{\max}$ is the maximum stress intensity factor. As the lattice rotation is a mode of plastic deformation, the distance on the either side of the crack through which the rotation is observed has been estimated as plastic zone. The zone radius calculated with R-ratio = 0.1, $\sigma_{ys} = 105 \text{ MPa}$ and $K_{\max} = 10 \text{ MPa m}^{1/2}$ turns out to be 1.5 mm. On the other hand, the calculated zone radius for Al 7075 for an R-ratio = 0.1, $\sigma_{ys} \approx 350 \text{ MPa}$ and $K_{\max} = 14.44 \text{ MPa m}^{1/2}$ is 0.271 mm. The experimentally estimated plastic zone (radius), however, is more than one order of larger for Al 7075, assuming that the lattice rotation is due to the plastic deformation. The $\sin^2\psi$ method was used for residual stress measurements [28,29] using 113 reflection as it reduces oscillation in the $\sin^2\psi$ plot because of high multiplicity factors of 24. In the x-ray diffraction, we used $\phi = 0.0$ and $\chi = 0.0$. The angle $\psi = \omega - \theta$ is the angle through which the sample is rotated in this method, where 2θ is the Bragg angle and the ω is the angle between the incident X-ray and specimen surface. The stress was measured close to the crack (0.0 mm), and at 5, 10, 15 and 25 mm from the crack on one side of the crack, and also close to the tip of the crack (see Fig. 1). The x-ray spot size of $1 \times 1 \text{ mm}^2$ was used to collect the diffraction data at different locations. The spot size that we have used for measuring the elastic strain covers around 40 grains (grain width 25 μm). However, all the grains will not be oriented to contribute to the 113 reflections. In addition, the stresses very close to the crack (within the plastic zone) were characterized by reducing the x-ray probe size to $100 \mu\text{m} \times 2 \text{ mm}$. The data were collected in the step scan mode around the 113 peak with a step width of 0.02° and the counting time in the range of 15–30 s for ω in the range $40\text{--}60^\circ$. The

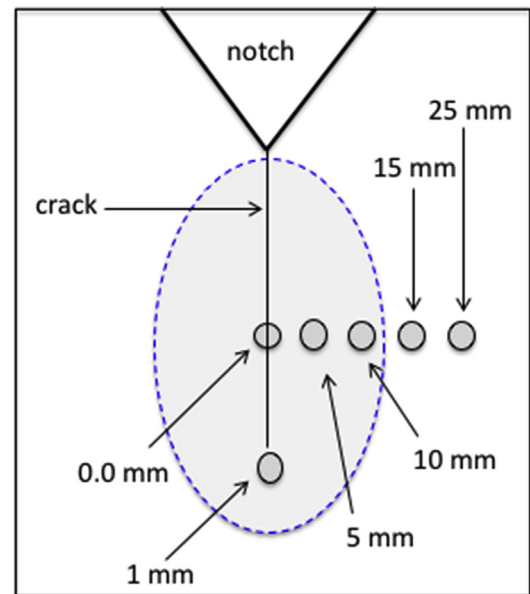


Fig. 1. (a) A schematic showing the notch, crack and relative positions of experimental measurements of stress.

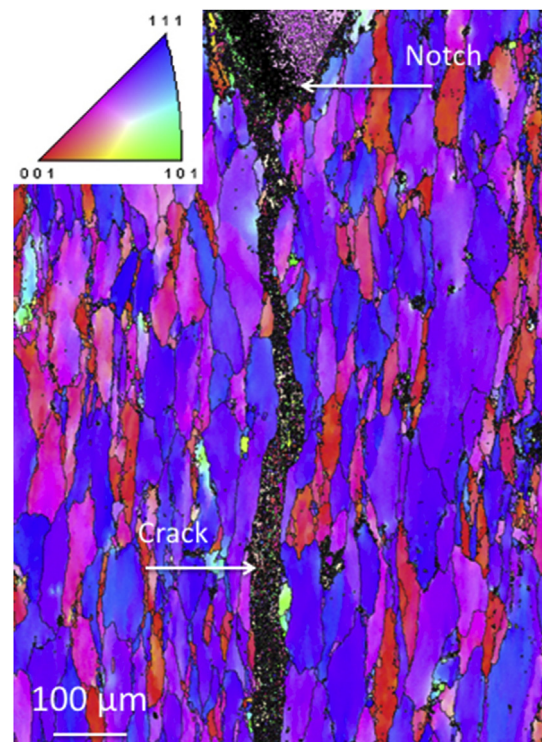


Fig. 2. The EBSD image showing the grain orientation on either side of the crack. Inset shows the color coding of different orientations within a stereographic triangle containing 001, 110 and 111 poles. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

size and orientation using electron back scattered diffraction (EBSD) imaging of different grains for Al-7075 close to the crack were shown in Fig. 2. For EBSD, a JEOL JSM-7001F SEM operating at 30 kV and equipped with a TSL/EDX Hikari EBSD detector was used to acquire the data with a step size of $1.0 \mu\text{m}$. Analysis of the EBSD data and grain size was performed with the TSL/EDX OIM Analysis-5.31 software package. The purpose of the EBSD image (see Fig. 2) is to show the grain

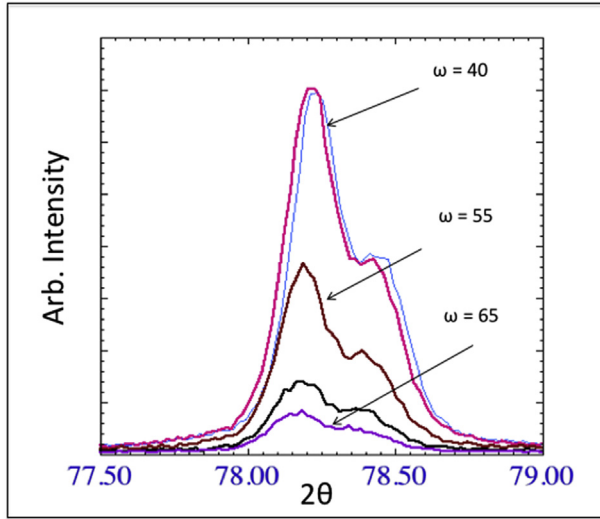


Fig. 3. The 113_{Al} peaks ($K\alpha_1$ and $K\alpha_2$) at a given location collected at different tilt angles, $\omega = 40, 45, 50, 55, 60$ and 65 . The peak shifts towards the left with the increase in tilt angle. The $K\alpha_1$ peak was used for the measurements.

morphology close to the crack. This is a high resolution inverse pole figure (IPF) image. These IPF images were used to generate the orientation distribution maps, close and away from the crack, for both samples. The details of the EBSD images obtained at different distances from the crack are discussed in our earlier paper [17]. The focus of this paper is on the residual stress and the lattice rotation using x-ray diffraction based technique.

3. Results and discussions

The residual stresses in the present case are caused by localized yielding of materials as a result of crack propagation under fatigue loading. We measure here only elastic strains using XRD for the estimation of residual stresses, as the elastic strains change the mean lattice spacing. The measurements are done on a sample (ex-situ) after the fatigue deformation. The mean inter-planar lattice spacing, d , are measured here at different ω tilts. Fig. 3 shows the 113 peaks with two components, $K\alpha_1$ and $K\alpha_2$, at different ω tilts ranging from 40 to 65° from a region. The peak has been shifted towards the lower angles, as the ω tilt increases, suggesting that the lattice spacing increases with the increase in ω tilt. We also observe a considerable decrease in peak intensity as the ω tilt increases. We analyzed the $K\alpha_1$ peak using a Peakfit program and a Gaussian distribution for the measurements. If the stress acting in some direction on the surface is σ_ψ , the strain along an inclined line obtained from the elasticity theory for an isotropic solid can be given by Refs. [28,29].

$$\varepsilon_\psi = \frac{[(1 + \nu)(\sigma_1 \cos^2 \phi + \sigma_2 \sin^2 \phi) \sin^2 2\psi - \nu(\sigma_1 + \sigma_2)]}{E} \quad (1)$$

Considering the strain in terms of inter-planar spacing, d , the stress can be expressed as [28,29]:

$$\sigma_\psi = \frac{E(d_\psi - d_0)}{(1 + \nu) \sin^2 \psi d_0} \quad (2)$$

where E is the Young modulus, ν is the Poisson's ratio, and σ_1 and σ_2 are principal stresses. The residual stress calculated from Eq. (2) knowing the slope and the intercept of the d_ψ vs. $\sin^2 \psi$ lines (see Fig. 4 and Fig. 5) as a function of distance from the crack. The slope is mostly positive at a number of locations for Al 7075, away and close to the crack and in front of the crack, indicating that the stress is tensile in most places.

Fig. 6 shows that the residual stress increases from 60 MPa at

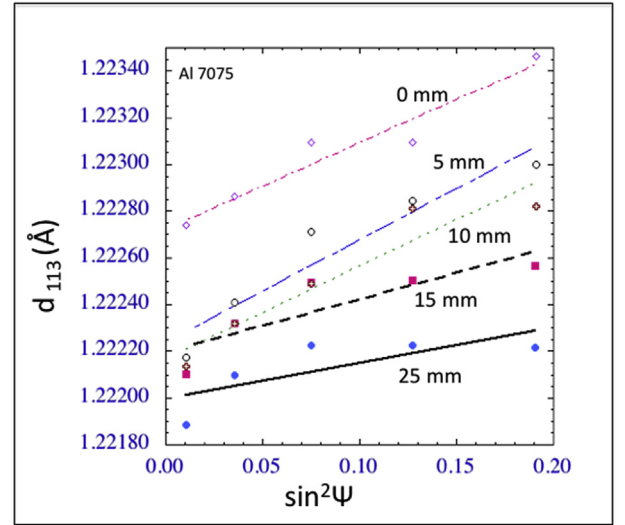


Fig. 4. Plot of d_{113} as a function of $\sin^2 \psi$ at different locations from the crack for Al 7075. The slope of the best fit line is positive, suggesting that the stress is tensile in nature.

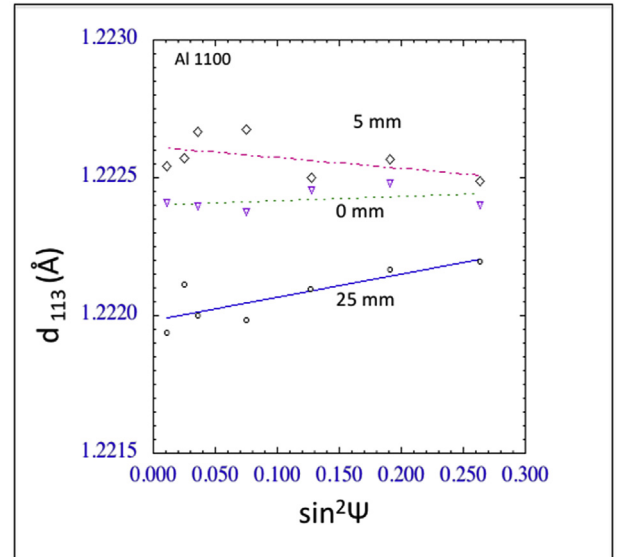


Fig. 5. Plot of d_{113} as a function of $\sin^2 \psi$ for Al 1100. Note that the slope of the best-fit line is negative at 5 mm away from the crack, suggesting compressive residual stress at that location.

25 mm away from the crack to 195 MPa at 5 mm from the crack and then decreases to 165 MPa close to the crack for Al 7075. The stress measured in front of the crack tip is ≈ 220 MPa. For Al-1100, the stress, however, falls down from ≈ 38 MPa tensile at 25 mm from the crack to 17 MPa compressive at 5 mm from the crack to below 5 MPa tensile close to the crack. The results clearly show that the tensile residual stress increases gradually by more than $\approx 200\%$ for Al-7075. However, for Al-1100, it decreases by $\approx 80\%$, as we approach the crack. Note that the stress here has been applied perpendicular to the crack plane, producing a biaxial state of stress in the CT specimen. In order to accurately evaluate the stresses closer to the crack, we further reduce the x-ray probe size to $100 \mu\text{m}$ in width, as the estimated plastic zone size (radius) from mechanics is 0.27 mm for 7075. This ensures that the x-ray probe size is sufficiently lower than the plastic zone size. If there are compressive residual stresses in the Al 7075 near the fatigue crack (within hundred microns), they can be measured. Fig. 7 shows that the d_{113} spacing of Al obtained using the reduced x-ray probe size within

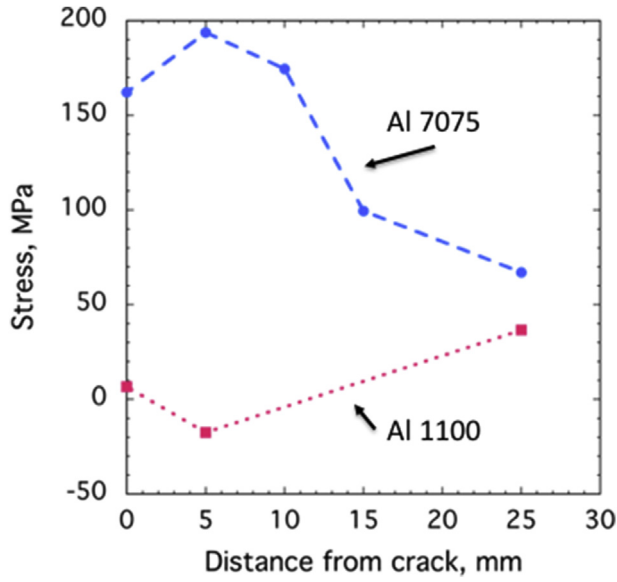


Fig. 6. The residual stress as a function of distance from crack for Al 7075 and Al 1100.

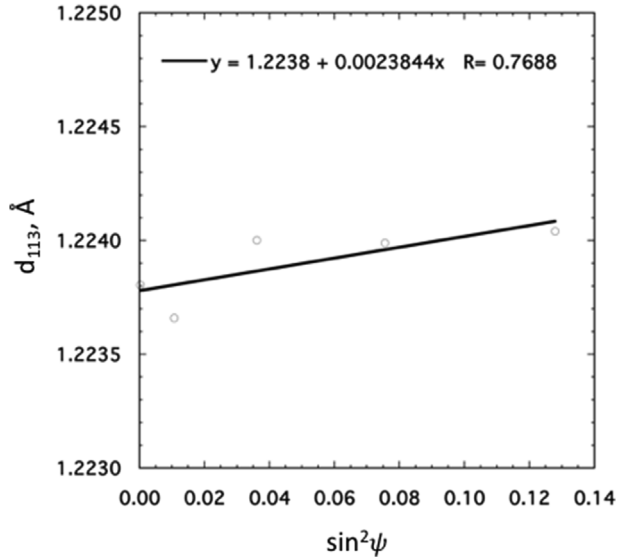


Fig. 7. Plot of d_{113} as a function of $\sin^2\psi$ for Al 7075 with fine x-ray probe size of $100 \mu\text{m} \times 2 \text{ mm}$ very close to the crack (within $100 \mu\text{m}$). Note that the slope of the best-fit line is positive, suggesting tensile residual stress at that location.

$100 \mu\text{m}$ (the crack is within the $100 \mu\text{m}$ slit) of the crack increases with $\sin^2\psi$, suggesting the stress is tensile in nature. The estimated stress from the best-fit straight line of d_{113} vs. $\sin^2\psi$ is 100 MPa . It demonstrates that as the x-ray probe approaches the crack, the nature of the stresses does not change for Al 7075, but they are tending to approach zero as expected (see Figs. 6 and 7).

The residual stresses are mostly the stored elastic energy as a result of deformation around the fatigue crack. Additionally, the increase in dislocation density contributes to the stored elastic energy, and any increase in dislocation density will enhance the elastic stress. The relationship between the residual stress due to dislocation and dislocation density can be given by the following expressions [30]:

$$\sigma_d = A \mu b \sqrt{\rho} + \sigma_0 \quad (3)$$

$$\Delta\sigma_d = A \mu b \sqrt{\rho} \quad (4)$$

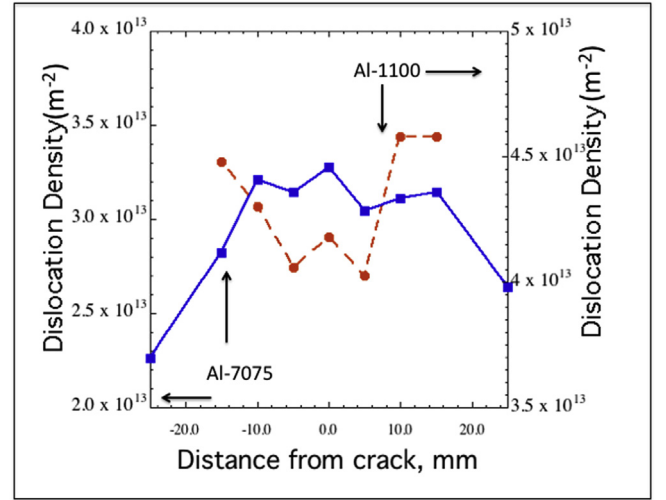


Fig. 8. (a) The dislocation density as a function of distance from crack.

$$\frac{\Delta\sigma_1}{\Delta\sigma_2} = \frac{\sqrt{\rho_1}}{\sqrt{\rho_2}} \quad (5)$$

where A is constant, b is the Burgers vector, μ is the shear modulus and σ_0 is the original stress. Here we measure the dislocation density using the full width at half maxima (FWHM) of the 111 reflection at different locations on either side of the crack. Assuming that the dominant contribution to the width of 111 peak is due to dislocations, as the average grain size (width 25 mm and length 100 mm closer to the crack) is relatively large and the bending is small, the density of dislocations, ρ , can then be calculated using the following expression [31]:

$$\rho = \frac{\beta_D^2}{4.35b^2} \quad (6)$$

where β_D is the FWHM and b is the Burgers vector. Fig. 8 shows the density of dislocation as a.

Function of distance on either side of the crack. While the density was observed to increase by $\approx 25\text{--}35\%$ closer to the crack for Al 7075, it decreases by $\approx 10\text{--}13\%$ closer to the crack for Al 1100, as compared to dislocation away from the crack. In the present case, the increase in stress due to increase in dislocation density from Eq. (5) turns out to be $\approx 20\%$. Thus, the increase in dislocation density alone cannot explain the 200% increase in residual stress due to fatigue crack growth in Al-7075. Similarly, the 80% decrease in residual stress cannot be explained by the 10% decrease in dislocation density in Al-1100. Note that the processing histories of the alloys are different, the dislocation density of 7075 and 1100 alloys would be different before fatigue, which is evident in Fig. 8. Also note that dislocations can be classified into two broad categories, geometrically necessary dislocations (GNDs) and statistically stored dislocations (SSDs). With XRD, we measure both the density of SSDs and GNDs. The results show that the increase or decrease in dislocation density is associated with the change in residual stress. However, if the 200% increase in residual stress is entirely due to the increase in dislocation density of 25% , then one could explain that the 10% reduction in dislocation density would result in an 80% decrease in residual stress. However, we showed (see Eqs. (3)–(5)) that the 25% increase in dislocation density can enhance the residual stress by 10% . This suggests there would be a scaling factor. We emphasized that the change in dislocation density is important for the lattice rotation, and thus for the change in the residual stress. Before the fatigue deformation, the residual stresses were 60 and 30 MPa for 7075 and 1100 Al alloys, respectively.

Furthermore, the residual stresses can result from the lattice rotation owing to the orientation dependence of deformation of individual grains [32]. It is known that deformation can lead to texture formation,

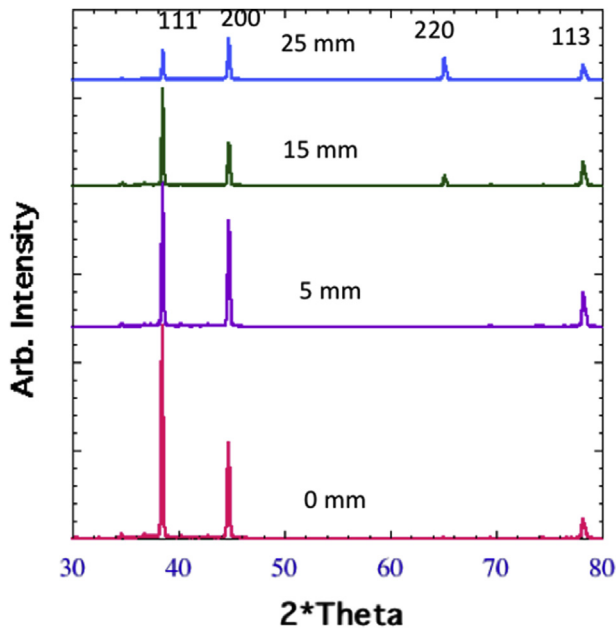


Fig. 9. XRD obtained at different locations from crack for Al 7075.

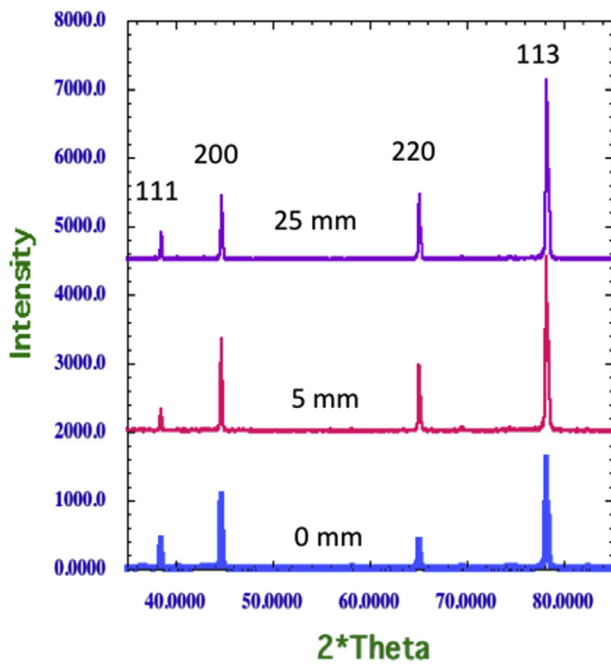


Fig. 10. XRD obtained at different locations from crack for Al 1100.

which involves lattice rotations, and the elastic residual stresses can develop as a result of the change in local texture due to lattice rotation. To investigate the lattice rotation, we study the orientation of grains as a function of distance from the crack using XRD and EBSD techniques. Fig. 9 shows a series of x-ray diffraction (XRD) patterns obtained from Al 7075 at different locations, showing the relative variations of 111, 200, 220 and 113 Al peaks. We could see the intensity ratio of 111/200 peak increases by 175% as we move closer to crack, suggesting more 111 planes are aligned due to lattice rotation as compared to 200 as we move close to the crack. Similarly, Fig. 10 shows a series of XRD patterns obtained from Al 1100 at different locations, close to the crack, 5 mm and 25 mm from the crack. In this case, the intensity ratio of 113–220 drops significantly as we move closer to the crack, suggesting that the reduction of Cu texture, $\{112\} < 111 >$. For texture analysis,

the orientation distribution function (ODF) maps were generated by sectioning through the Euler space parallel to ϕ_2 from 0 to 90° with 5° increment from the inverse pole figures (IPF). Fig. 11 (a and b) are the ODF maps at 25 mm away and close to the crack, respectively, for Al 7075. We observe a considerable increase in texture components, such as shear-2 ($\{111\} < 110 >$) with $\phi_1, \phi, \phi_2 = 0, 55, 45$, near Cu ($\{112\} < 111 >$) with $\phi_1, \phi, \phi_2 = 90, 35, 45$ and S texture ($\{123\} < 634 >$) with $\phi_1, \phi, \phi_2 = 59, 37, 63$ close to the crack. In case of Al 1100, however, the near Cu texture component decreases considerably close to the crack (see Fig. 11(c and d)).

The magnitude of rotation can be estimated by multiplying the number of dislocations, n , gliding through the slip planes with the Burgers vector, b . Thus, the amount of displacement due to slip is nb . The average rotation angle will be given as [16];

$$\theta = \tan^{-1}(nb/l_w), \quad (7)$$

where l_w is the average width of the grain. The number of dislocation can be estimated from the increase in dislocation density due to crack growth. The density of dislocation obtained from the FWHM of 111 peak away (25 mm) from the crack is $\approx 2.3 \times 10^{13} \text{ m}^{-2}$, while the density was observed to increase by $\approx 30\%$ closer to the crack (within 10 mm of the crack) as a result of crack growth. The number of dislocations turns out to be 1.7×10^4 for $100 \mu\text{m} \times 25 \mu\text{m}$ pancake size grain. As the grain width, l_w , is $25 \mu\text{m}$, the estimated average angle from Eq. (7) is approximately 11° , which is sufficient to change the texture close to the crack tip. Another contributory factor in the grain rotation could be due to a change in the content of grain boundary (GB) dislocations. Such a change of GB dislocations must change the GB angle between neighboring grains, due to well-known Frank–Bilby equation [33,34]. All these suggest that dislocations are needed to produce the lattice rotation, and at the same time the remaining dislocation will increase the residual stresses, although the increase in dislocation density in our case is not sufficient to explain the 200% increase in residual stresses.

We observe that the lattice rotation leads to alignment in $\{111\}$ planes for Al-7075. However, for Al-1100, the $\{111\}$ planes are not aligned across the grains. To investigate why such alignment occurs in one case, we study the dislocation configurations using TEM. Fig. 12 (a and b) are the bright-field TEM images, showing the configuration of dislocations close to the crack. Most dislocations are pinned by the precipitates in the matrix at either side of the grain boundary (see Fig. 12(a)). On the other hand, in Al 1100, the extensive dislocation network formation due to dislocation knitting resulting in 3-fold and 4-fold type networks (see Fig. 12(b)) has been observed, which leaves behind regions with depleted dislocations. These dislocation networks mostly block the incoming glide dislocations. A model of dislocation knitting has been put forward by Miekko-Oja and Lindroos [35,36], which requires interaction of forest dislocations with glide dislocations that creates a junction dislocation by the climb. Thus, the alignment of $\{111\}$ lattice planes in Al 7075 occurs as a result of dislocations gliding across the grain boundaries, which facilitates the lattice rotation. However, the glide is restricted for Al 1100 as a result of the extensive dislocation network formation.

4. Summary

In summary, we demonstrated that the residual stress increases gradually around the fatigue crack by $\approx 200\%$ for Al 7075, and decreases by $\approx 80\%$ for Al 1100, as we approach the crack. Such an increase or decrease in residual stress cannot be explained by the change in dislocation density alone in both the alloys due to crack growth. The results indicate that the deformation associated with the lattice rotation is a major factor controlling the stress. The lattice rotation that aligns $\{111\}$ planes across grains, which is consistent with an increase in shear-2 and near Cu texture, increases the stress. However, the decrease

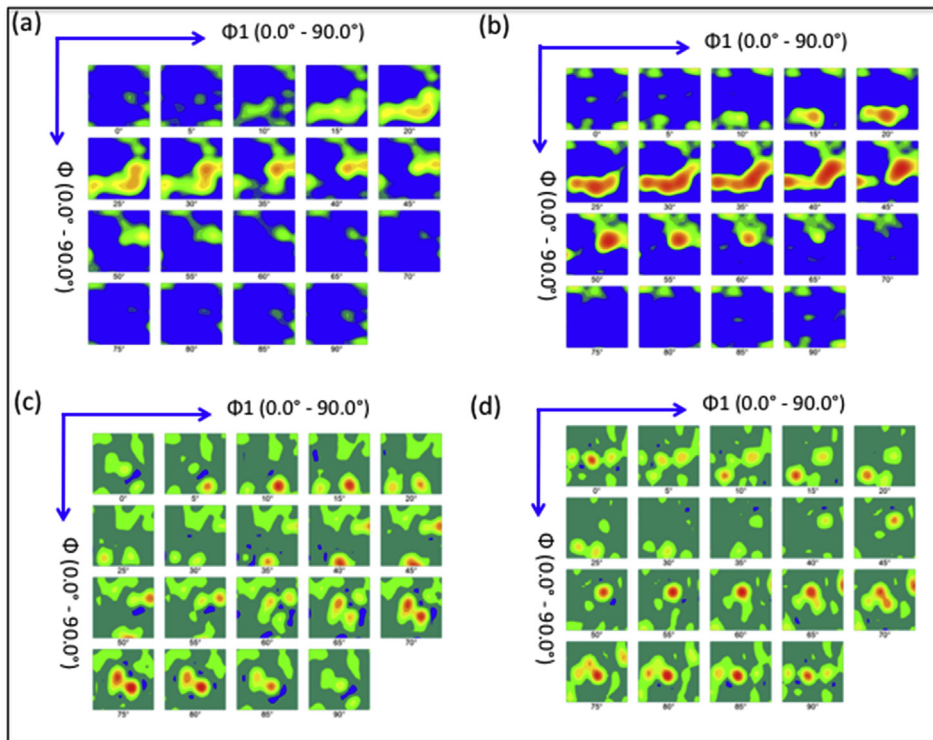


Fig. 11. (a) The ODF sections 25 mm away from crack parallel to ϕ_2 from 0 to 90° with an increment of 5° showing texture for Al 7075 (b) The ODF sections close to the crack for Al 7075. (c) The ODF sections 25 mm away from crack parallel to ϕ_2 from 0 to 90° with an increment of 5° showing texture for Al 1100 (b) The ODF sections close to the crack for Al 1100.

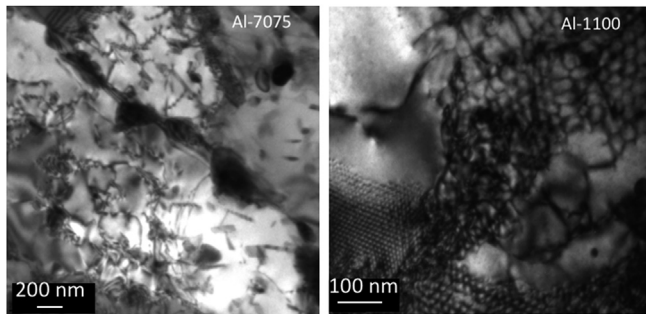


Fig. 12. (a) A bright-field TEM image showing the dislocation configuration close to the crack in Al 7075. (b) A bright-field TEM image showing the dislocation network configuration and the relative dislocation free area close to the crack in Al 1100.

in residual stress in Al 1100 is associated with the lattice rotation that decreases the Cu texture. These findings provide new insights on the role of lattice rotation and dislocation configurations around the fatigue crack in determining the residual stress.

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