

To the Graduate Council:

I am submitting herewith a thesis written by Yinan Sun entitled “The Study of Overload Effects on Fatigue-Crack-Growth Behavior by Neutron Diffraction”. I have examined the final electronic copy of this thesis for form and content and recommend that it be accepted in partial fulfillment of the requirements for the degree of Master of Science, with a major in Materials Science and Engineering.

Peter K. Liaw
Major Professor

We have read this thesis
and recommend its acceptance:

Hahn Choo

YanFei Gao

Accepted for the Council:

Anne Mayhew
Vice Chancellor and Dean of
Graduate Studies

(Original signatures are on file with official student records.)

The Study of Overload Effects on Fatigue-Crack-Growth Behavior by Neutron Diffraction

A Thesis
Presented for the
Master of Science Degree
The University of Tennessee, Knoxville

Yinan Sun
May 2006

Acknowledgments

First and foremost, I am deeply indebted to my advisors, Dr. Peter K. Liaw and Dr. Hahn Choo. Their great efforts in guiding and supporting my research have been crucial for me to extend my potentials and develop my academic capabilities. I thank them for providing abundant opportunities and resources. Also, I would like to thank Dr. Yanfei Gao for many helpful discussions on the modeling part of my research. I would further like to thank Dr. Cam Hubbard at the Oak Ridge National Laboratory (ORNL) and Dr. Don Brown at the Los Alamos National Science Center for generously giving me precious neutron-diffraction beam times, which made my thesis work possible.

I owe many thanks to those who did neutron experiments for me. Those are Dr. P. K. Liaw, Dr. H. Choo, Dr. C. Fan, Dr. D Brown, Dr. K. An, Dr. C. Hubbard, and Dr. F. Tang. I also would like to thank Dr. Barabash, who did the X-ray microbeam experiments for me at advanced photon source (APS), and helped me with the dislocation density calculation.

I must express my appreciations to my family and many friends who have helped to relieve stresses of the graduate school, in particular, my husband, Yulin Lu, and

my son, James Lu, for putting up with me as a wife and a mother. I thank God everyday for giving them to me.

Finally, I would like to acknowledge the financial supports from the National Science Foundation (NSF), the Combined Research-Curriculum Development (CRCD) Programs (EEC-9527527 and EEC-0203415), the Integrative Graduate Education and Research Training (IGERT) Program (DGE-9987548), and the International Materials Institutes (IMI) Program (DMR-0231320), and the Major Research Instrumentalities (MRI) Program (DMR-0421219) to the University of Tennessee (UT), Knoxville with Ms. M. Poats, Drs. C. J. Van Hartesveldt, D. Dutta, W. Jennings, L. Goldberger, L. Clesceri, C. Huber, and C. E. Bouldin as the program directors, respectively.

Abstract

During the load-controlled high-cycle fatigue test, when the overload was applied, it is shown that from the crack-growth rate (da/dN) versus stress-intensity-factor range (ΔK) curve, the crack-growth rate decreased, following the overload, which indicates the crack-closure phenomenon. The crack-growth-retardation period was observed after the overload. The goal of this study is to investigate the deformation evolution during tensile loading and unloading cycles using neutron diffraction.

Neutron diffraction is used to investigate the crack-closure phenomenon by measuring the changes in the elastic-lattice-strain profiles around the fatigue-crack tip in a compact-tension (CT) specimen during tensile loading and unloading cycles. Spatially-resolved-strain measurements were performed to determine the in-plane and through-thickness lattice-strain profiles ahead of the crack tip under a constant tensile load. The strain scanning was repeated under various applied loads ranging from 667 to 6,667 N. Subsequently, an overload at 8,889N was applied. The strain scans repeated. After the overload, large compressive strain fields were observed close to the crack tip, indicative of the crack-closure phenomena.

The plastic deformation ahead of a fatigue-crack tip was measured from the diffraction-peak-width changes. A large plastic deformation was observed after the overload. Furthermore, the plastic-zone size in front of the crack tip was estimated from the diffraction-peak broadening, which showed a good agreement with the calculated result. The plasticity-induced crack-closure phenomenon after an overload was observed.

The deformation in the vicinity of the crack tip was not only studied with the neutron diffraction, but also the x-ray microbeam diffraction. The results help understand the overload effect, which induced a large plastic deformation causing dislocations inhomogeneously arranged around the crack tip. From neutron-diffraction measurements, the anisotropic line broadening was observed in front of the crack tip. Furthermore, Laue patterns, obtained from the microbeam diffraction at different locations near the crack, provide a better spatial resolution and show alternating regions with high and low dislocation densities. Overall, the dislocation density was found to decrease with the distance from the crack tip.

Neutron diffraction was used to investigate the retardation phenomenon by mapping the changes in the lattice-strain profiles around the fatigue-crack tip in a series of compact-tension (CT) specimens, which were fatigued to various stages through the retardation period after the overload. Following the overload, compressive-strain fields were observed along the loading direction close to the crack tip. As the crack grows out of the retardation period, the residual compressive strains decreased. The results provide a microscopic understanding of the overload effect during cyclic loading.

Contents

1	Literature Review	1
1.1	Material	1
1.2	Developments of Fracture Mechanics	5
1.3	Basic Concepts	6
1.4	Fatigue Mechanisms	10
1.4.1	Fatigue-Failure Mechanisms	15
1.4.2	Fatigue-Crack-Initiation Mechanisms	16
1.4.3	Fatigue-Crack Growth Mechanisms	18
1.5	Overloading Effects	20
2	Neutron-Diffraction Studies on Lattice-Strain Evolution around a Crack Tip during Tensile Loading and Unloading Cycles	27
2.1	Introduction	27
2.2	Experimental Details	29
2.3	Results and Discussions	32
2.4	Summary	36

3	Changes in Lattice-Strain Profiles around a Fatigue Crack through the Retardation Period after the Overload	37
3.1	Introduction	37
3.2	Experimental Details	38
3.3	Results and Discussions	41
3.4	Summary	44
4	The Study of Crack-Closure Phenomena after the Overload	45
4.1	Introduction	45
4.2	Experimental Details	47
4.2.1	Fatigue Experiments	47
4.2.2	Fatigue-Striation Measurements	48
4.2.3	Neutron-Diffraction Experiments	50
4.3	Results and Discussions	51
4.4	Summary	55
5	Neutron and X-ray Microbeam Diffraction Studies of Elastic Strain and Dislocation Distributions near a Fatigue Crack	56
5.1	Introduction	56
5.2	Experimental Details	57
5.3	Results and Discussions	59
5.4	Summary	65
6	Conclusions	66

7 Future Work	70
Bibliography	72
Vita	81

List of Tables

1.1	Chemical composition of Type 316 low-carbon, nitrogen-added (LN) stainless steel (SS) (wt.%, weight percent)	1
1.2	Mechanical properties of Type 316 stainless steels at 27 and -196 °C .	4

List of Figures

1.1	The three modes of loading a cracked body.	6
1.2	Crack tip coordinate system and the stresses acting on an element ahead of the crack tip.	7
1.3	Elastic and elastic-plastic crack-tip-stress distributions in front of the crack tip and the plastic-zone sizes, r_y and r_p . [1]	9
1.4	Typical fatigue-crack-growth curves under a constant load range. . . .	11
1.5	Typical fatigue-crack-growth diagram showing three different regimes.	14
1.6	Schematic of fatigue-crack growth across a specimen section indicating three stages of fatigue failures.	15
1.7	Schematic shows the roughening of surfaces due to the formation of persistent slip bands (PSBs).	17
1.8	One of the various possible mechanisms of fatigue-crack growth. . . .	19
1.9	Schematic of the ductile rupture showing four stages. Bottom: the fracture surface illustrates dimples formed by large and small particles.	21
1.10	Illustration of the overload effect.	22

2.1	Compact-tension specimen of Type 316 LN stainless steel with a thickness of 6.35 mm, a notch length of 10.16 mm, and a crack length of 22.86 mm. Diffraction patterns were measured along the crack length with a scattering volume of $2 \times 1 \times 2 \text{ mm}^3$ with the 1 mm along the crack length.	29
2.2	Applied tensile loading-unloading sequence. Neutron strain scanning was performed at each load point (LP) from LP1 through LP7.	31
2.3	In-plane lattice-strain profiles in front of the crack tip at applied loads of 667 N, 3,333 N, and 6,667 N, which correspond to LP1, LP2, and LP3, respectively, in Figure 2.2.	32
2.4	In-plane lattice-strain profiles in front of the crack tip at applied loads of 8,889 N, 6,667 N, and 667 N, which correspond to LP5, LP6, and LP7, respectively, in Figure 2.2.	33
2.5	Comparison between the in-plane lattice strain profile and the plastic-zone size (about 7 ~ 9 mm) estimated from the changes in the diffraction-peak width at 667 N (LP7 in Figure 2.2). Note that the plastic-zone size calculated using Equation 2.3 is about 7.9 mm.	35
3.1	The effect of overload on fatigue-crack propagation. During the fatigue tests, seven specimens were prepared.	39
3.2	The lattice-strain profiles of SP1, SP2, SP3, and SP6 along the transverse direction.	42
3.3	The schematic illustration of the perturbed plastic zone. a_0 is the crack length, and r_c is the plastic-zone size.	43

4.1	The load controlled high-cycle fatigue load sketch and crack growth rate curve.	46
4.2	Fracture surface image at different ΔK levels. (a) $34.84 \text{ MPa}\sqrt{m}$, (b) $35.82 \text{ MPa}\sqrt{m}$, (c) $36.34 \text{ MPa}\sqrt{m}$, (d) $37.55 \text{ MPa}\sqrt{m}$, and (e) $39.29 \text{ MPa}\sqrt{m}$	49
4.3	The comparison between the fatigue-striation spacing and the crack-growth rate at a function of the crack length.	52
4.4	The comparison between the in-plane lattice-strain profiles in front of the crack tip at applied loads of 667 N before overload and 667 N after overload, which correspond to (1) and (7) in Figure 2.2.	53
4.5	The comparison between the through-thickness lattice-strain profiles in front of the crack tip at applied loads of 667 N before overload and 667 N after overload, which correspond to (1) and (7) in Figure 2.2.	54
5.1	In-plane and through-thickness lattice-strain profiles in front of the crack tip at applied loads of 667 N after the overload. Insert shows distinct orientation of diffraction vectors G_{TT} and for G_{IP} measurements.	60
5.2	Demonstrate the dependence of FWHM on $\sqrt{\tan \theta}$ for four types of $\langle hkl \rangle$ along IP direction. Inserts the dislocation density as a function of distance from crack tip along IP direction.	62
5.3	Demonstrate the dependence of FWHM on $\sqrt{\tan \theta}$ for four types of $\langle hkl \rangle$ along the TT direction. Inserts the dislocation density as a function of distance from the crack tip along the TT direction.	63

Chapter 1

Literature Review

1.1 Material

The basic composition of the 316 low-carbon nitrogen-added (LN) stainless steel (SS) was listed in Table 1.1. For the selection of the 316 LN stainless steel as a candidate material, the understanding of the main effects of alloying elements in 316 LN SS is needed.

- Carbon (C)

1. A strong austenite former.

Table 1.1: Chemical composition of Type 316 low-carbon, nitrogen-added (LN) stainless steel (SS) (wt.%, weight percent)

Element	C	Mn	P	S	Si	Ni
wt.%	0.009	1.75	0.029	0.002	0.39	10.2
Element	Cr	Mo	Co	Cu	N	Fe
wt.%	16.31	2.07	0.16	0.23	0.11	Bal.

2. Added to some high-strength alloys for hardening and strengthening effects.
 3. Adversely affecting the weld-metal corrosion resistance and toughness at low temperatures.
- Chromium (Cr)
 1. A ferrite and carbide former.
 2. A primary contributor to scaling and corrosion resistances.
 3. In the stainless steels, this element having little or no influence on the high-temperature strength and creep strength.
 - Molybdenum (Mo)
 1. A ferrite and carbide former.
 2. Used to improve the high-temperature strength and creep resistance.
 3. Employed to improve the general corrosion resistance of steels in non-oxidizing media, and the resistance to pitting corrosion in all media.
 - Nitrogen (N)
 1. A strong austenite former.
 2. Used to minimize grains grown in high-chromium steels at high temperatures.
 3. Adversely affecting the weld metal toughness at cryogenic temperatures.

4. Raising strength.
- Nickel (Ni)
 1. An austenite former.
 2. Used to improve the general corrosion resistance against non-oxidizing liquids.
 3. Sometimes added in small amounts to improve the mechanical properties.
 4. Generally improving the weld-metal toughness.

The Type 316 LN has a similar chemical composition to the 316 L stainless steel. The usual 316L austenitic stainless steel is fully austenitic after the solution heat-treatment but has poor mechanical properties. It has already been shown that nitrogen alloying clearly improves the monotonic and cyclic behavior of 316 L stainless steel at room and high temperatures [2]. According to the phase diagram, nitrogen is an austenite stabilizer. The influence of nitrogen on stabilizing the austenite may change in the presence of strains at low temperatures, because the mechanisms of the martensitic transformation may change with the deformation. The work reported by Vogt *et al.* examines the effect of nitrogen and temperature on the fatigue behavior of a 316 L austenitic stainless steel [3]. A correlation between the macroscopic behavior and the microstructure is developed for the low-cycle fatigue (LCF) and fatigue-crack propagation behavior. Two austenitic stainless steels (316 L and 316 LN) were studied, whose chemical compositions differ significantly in their nitrogen contents. Table 1.2 summarizes the different tensile characteristics (the yield strength, σ_y , ultimate tensile strength, σ_u , total elongation, A, and reduction in

Table 1.2: Mechanical properties of Type 316 stainless steels at 27 and -196 °C [3]

Materials	Test Temperature (°C)	σ_y (MPa)	σ_u (MPa)
316 L	27	262	574
	-196	402	1156
316 LN	27	328	697
	-196	902	1415

area, S) deduced from tests performed at 27 and -196°C [3]. It shows that nitrogen increases the fatigue strength at 27 and -196°C. Nitrogen favors the planar slip by reducing the tendency for the cross slip and avoiding a three-dimensional cellular structure. In the 316 LN alloy, the stacking-fault energy (SFE) that is responsible for the planar slip is reduced by low temperatures and by the effect of nitrogen at low temperatures. Note that a lower SFE favors the planar slip instead of the cross slip. The result of the fatigue-crack-growth rate (FCGR) may be interpreted as a consequence of the cyclic hardening due to a combination of the interstitial-solute strengthening, strain-induced phase transformation, and a decrease in the temperature. Lowering the temperature from 27 to -196°C for both 316 L and 316 LN materials decreases SFE, and hardens the materials at the crack tip locally, which produces a decrease of the crack-opening displacement (COD), resulting in a decrease of fatigue-crack-growth rates. Since the decrease in the SFE of 316 LN is even greater with the presence of interstitial solutes of nitrogen, the fatigue resistance of 316 LN is greater than 316 L SS [3].

1.2 Developments of Fracture Mechanics

In 1913, Inglis [4] proposed the concept of stress concentrations at geometrical discontinuities, thus providing an explanation for why fractures emanate from cracks, holes, or defects. In 1920, Griffith [5] proposed his hypothesis about the energy exchanges that take place during the fracture and derived the concept of a critical crack size necessary for a brittle fracture. In 1939, Westergaard [6] published the results of his analysis demonstrating that the stresses at the tips of cracks in elastic bodies varied as a function of $1/\sqrt{r}$, where r is the distance from the crack tip. In 1948, Irwin [7] and E. Orowan [8] proposed modifications to account for the plastic energy. In 1957, Williams [9] used a technique and proposed a more complete description of the crack-tip stress distribution. In mid-1950s, aircraft and ship failures facilitated the U.S. Navy to accept the importance of the fracture mechanics research. More fracture concepts, such fracture toughness, K_{IR} , and crack-tip opening displacement, CTOD, become widely accepted. In 1960, Paris and co-workers [10] first proposed the relationship between the fatigue-crack-growth rate and the cyclic-stress-intensity parameter. As the nuclear industry become prominent in nineteen sixties, ductile steels were used. The elastic-plastic fracture theory had to be extended to include fracture mechanisms. In 1968, three papers of the considerable significance to the development of the elastic-plastic fracture mechanics appeared. Rice [11] idealized the plastic deformation as a nonlinear phenomenon for mathematical purposes and was able to generalize the energy-release rate for such a material. In 1972, the results of the first study to predict the initiation of fracture under elastic-plastic conditions were published by Begley and Landes [12, 13]. In subsequent years, much process

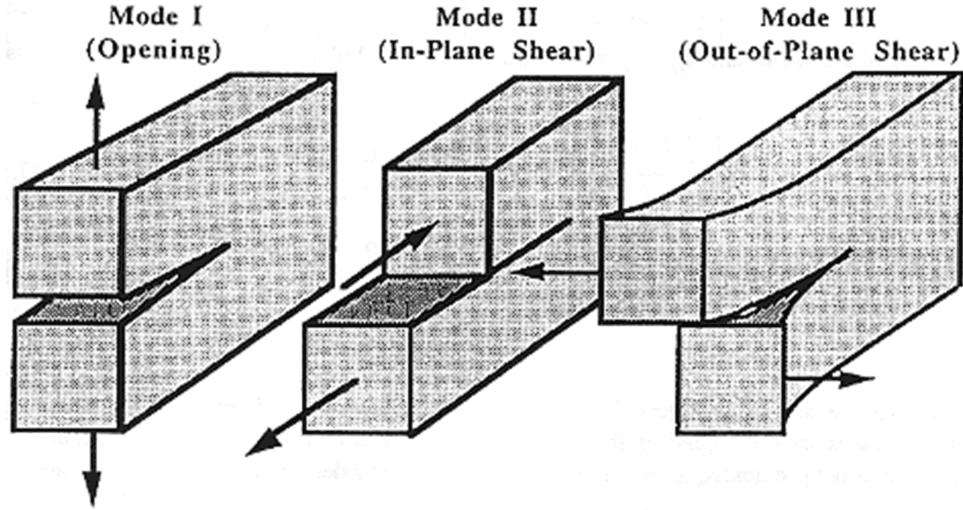


Figure 1.1: The three modes of loading a cracked body.

occurred in the development of test methods, making the elastic-plastic fracture mechanics a viable engineering tool for the structural-integrity analysis. In the early of mid-seventies, creep conditions had already begun to be introduced into the fracture mechanics. Saxena [14] defined the C_t parameter, which can be used to characterize the creep-crack-growth behavior under conditions ranging from the small-scale to extensive creep. Although much progress has occurred in the fracture mechanics over the past 40 years, it still remains an active field of research [15].

1.3 Basic Concepts

There are three distinct modes in which a crack body can be loaded, as shown in Figure 1.1 where the Mode I crack-opening mode is the one we are interested in, because most fracture in metallic components occur under the Mode I condition. In

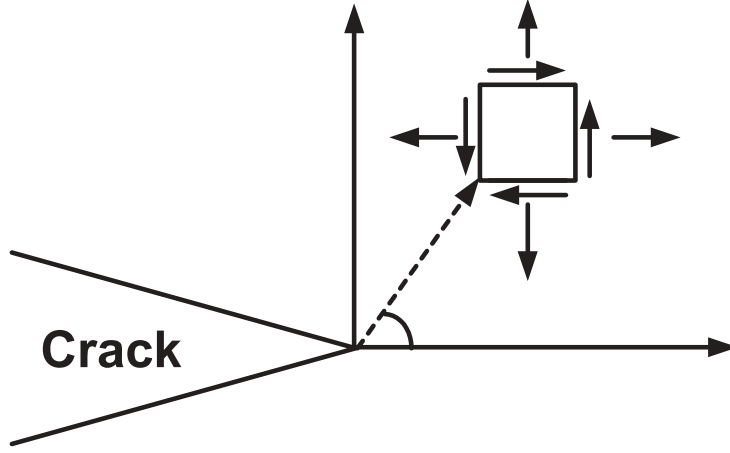


Figure 1.2: Crack tip coordinate system and the stresses acting on an element ahead of the crack tip.

Mode I , the crack separates symmetrically. The stress and displacements in a region near the crack tip are given by Equation 1.1 [1].

$$\begin{aligned}
 \sigma_x &= \frac{K}{\sqrt{2\pi r}} \left[\cos \frac{\theta}{2} \left(1 - \sin \frac{\theta}{2} \sin \frac{3\theta}{2} \right) \right] \\
 \sigma_y &= \frac{K}{\sqrt{2\pi r}} \left[\cos \frac{\theta}{2} \left(1 + \sin \frac{\theta}{2} \sin \frac{3\theta}{2} \right) \right] \\
 \tau_{xy} &= \frac{K}{\sqrt{2\pi r}} \left(\sin \frac{\theta}{2} \cos \frac{\theta}{2} \cos \frac{3\theta}{2} \right)
 \end{aligned} \tag{1.1}$$

see Figure 1.2 where K is the stress-intensity factor, r is the distance from the origin, and θ is the angle to the x axis.

The stress-intensity factor, K , represents the amplitude of the crack tip stress singularity and is dependent on the body geometry, crack size, load level, and loading configuration. For compact-tension specimens, the expressions for estimating K is,

$$K_{\max} = \frac{P_{\max}(2 + \alpha)}{B\sqrt{W}(1 - \alpha)^{3/2}} (0.886 + 4.64\alpha - 13.32\alpha^2 + 14.72\alpha^3 - 5.6\alpha^4) \tag{1.2}$$

where K is the stress-intensity factor, α is the function of total crack length, B is the thickness of the specimen, and W is width of the specimen.

Irwin made a simple estimate of the plastic-zone size along $\theta = 0$ for elastic, perfectly-plastic materials. The simplest estimate can be made by substituting $\theta = 0$ in Equation 1.1 and solving for a distance, r_y , at which $\sigma_y = \sigma_0$, this calculation leads to Equation 1.3.

$$r_p = 2r_y = \frac{1}{\pi} \left(\frac{K}{\sigma_0} \right)^2 \quad (1.3)$$

where r_p is the plastic-zone size, and σ_0 is the yield stress

Figure 1.3 [1] shows the distance, r_y , schematically. This estimate of the plastic zone is incorrect because it is based on the elastic-stress distribution. The elastic-plastic stress distribution is also shown schematically in Figure 1.3. The plastic-zone boundary is located at r_p , according to this distribution. To satisfy the force equilibrium in the y-direction, the areas under the elastic and elastic-plastic stress distributions must be the same. Equation 1.3 has been derived for a nonhardening material. For hardening materials, which follow the Ramberg-Osgood relationship:

$$r_p = \frac{m-1}{m+1} \frac{1}{\pi} \left(\frac{K}{\sigma_0} \right)^2 \quad (1.4)$$

m is the inverse of the strain-hardening exponent, which was a material constant derived from the regression of the stress-strain curve, r is the plastic zone size, and σ_0 is the yield stress.

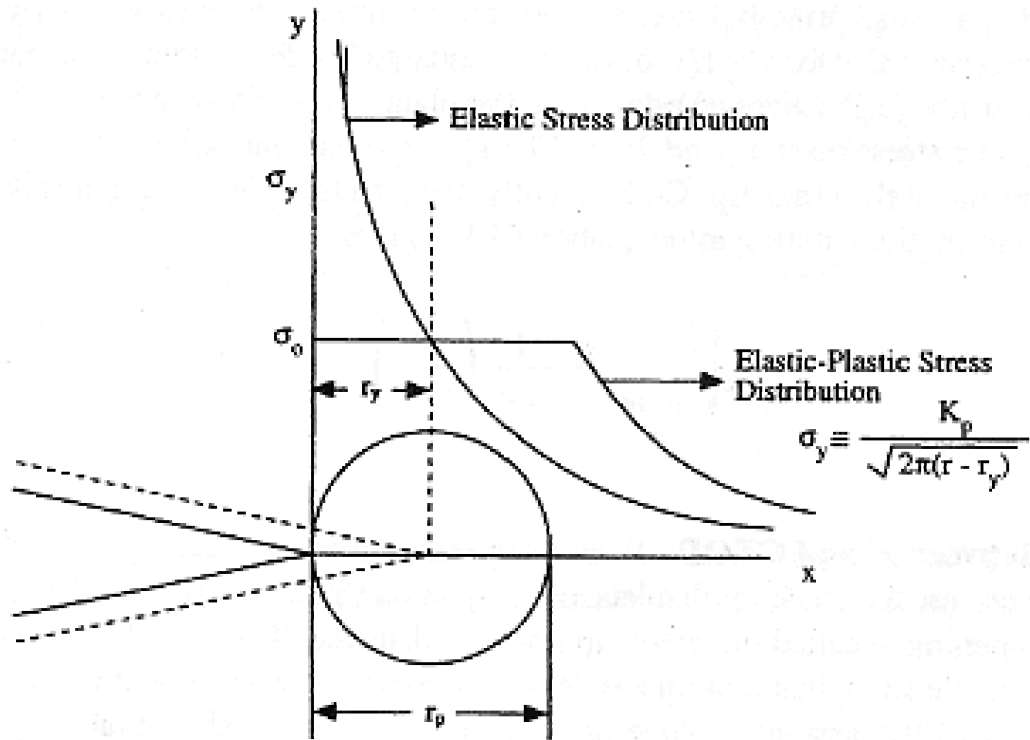


Figure 1.3: Elastic and elastic-plastic crack-tip-stress distributions in front of the crack tip and the plastic-zone sizes, r_y and r_p . [1]

1.4 Fatigue Mechanisms

The fatigue life is composed of both the crack-initiation and crack-growth stages. The useful fatigue life is, then, the time or the number of cycles to grow a dominant flaw of the initial size to a critical dimension, at which the catastrophic failure is expected. To predict the fatigue lives of these components for design and maintenance purposes, it is necessary to know the rates at which these flaws/cracks grow. The measurement of the crack-growth data for different combinations of the applied stress, crack length, and geometrical conditions of the cracked structure, and the mechanisms, which control the crack-growth rates under different conditions of the mean stress, test frequency and environment, are research topics of scientific and practical interest. The crack-growth-rate data are generated through testing in laboratory. The crack-growth rate is expressed in terms of the crack-length increment per cycle or unit time, da/dN or da/dt , respectively. Figure 1.4 illustrates a typical fatigue-crack-growth curve under a constant load range. The crack length increases with increasing the number of loading cycles. The fatigue-crack-growth rate is determined from such curves by graphical procedures or by computations. From these methods, the crack-growth rates resulting from a given cyclic load are $(da/dN)_{a_i}$ and $(da/dN)_{a_j}$ when the crack is of lengths, a_i and a_j , respectively. When the applied load range is held constant, the crack-growth rate most often increases with increasing the number of fatigue cycles. (This is generally the case, though not always.) For the same crack length, the crack-growth rate generally increases with increasing the applied load range, as indicated in Fig. 1.4. With the rapid development of the fracture mechanics, more reliable methods became available in the 1960s. Paris *et al.* [10,16]

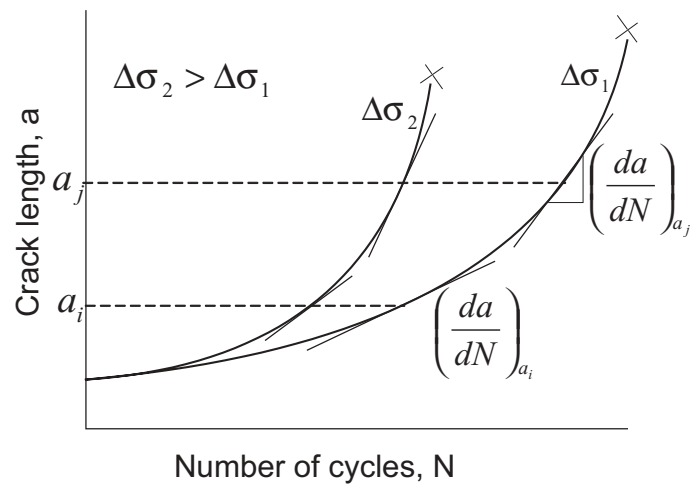


Figure 1.4: Typical fatigue-crack-growth curves under a constant load range.

postulated that the controlling factor in the fatigue-crack-growth process is the stress-intensity-factor range,

$$\Delta K \equiv K_{max.} - K_{min.} \quad (1.5)$$

$K_{max.}$ and $K_{min.}$ are the maximum and minimum stress-intensity factors, respectively, during a fatigue cycle. The stress-intensity-factor range, ΔK , is a function of the load range and crack length. Its particular form depends on the specimen geometry. For an edge-cracked specimen,

$$\begin{aligned} K_{max.} &= Y \sigma_{max.} \sqrt{a} \\ K_{min.} &= Y \sigma_{min.} \sqrt{a} \\ \Delta K &= Y \Delta \sigma \sqrt{a} \\ \Delta \sigma &\equiv \sigma_{max.} - \sigma_{min.} \end{aligned} \quad (1.6)$$

where Y is a geometrical factor related to the ratio of the crack length, a , to the width of the specimen, W , and $\sigma_{max.}$ and $\sigma_{min.}$ are the maximum and minimum stresses in a fatigue cycle, respectively. Paris *et al.* [10, 16] plotted the values of da/dN versus those of ΔK for the same crack length, a , and demonstrated that there exists a relation of the form

$$\frac{da}{dN} = C \Delta K^m \quad (1.7)$$

where C and m are scaling constants that are dependent on material variables, environment, cyclic frequency, temperature, and load ratio, R , which is defined as

$$R = \frac{\sigma_{min.}}{\sigma_{max.}} = \frac{K_{min.}}{K_{max.}} \quad (1.8)$$

This simple relation (Equation 1.7) is commonly called the *Paris-power law*. Although Equation 1.7 is empirical, it has remained one of the most used expressions in the fatigue-crack-growth analysis for a vast spectrum of materials and test conditions.

It is often desirable to estimate the useful component life for the purposes of design and failure analyses. The number of fatigue cycles to failure can be computed, using the Paris-power law, Equation 1.7. Assume that the initial flaw size to be a_0 and the final critical crack size, a_f , and that Y remains constant as the crack grows. Assume also $\Delta K = Y\Delta\sigma\sqrt{a}$. Integrating Equation 1.7 results in

$$CY^m(\Delta\sigma)^m \int_0^{N_f} dN = \int_{a_0}^{a_f} \frac{da}{a^{m/2}} \quad (1.9)$$

The fatigue life is, thus, given by

$$N_f = \frac{2}{(m-2)CY^m(\Delta\sigma)^m} \left[\frac{1}{(a_0)^{(m-2)/2}} - \frac{1}{(a_f)^{(m-2)/2}} \right] \quad (1.10)$$

for $m \neq 2$ and

$$N_f = \frac{1}{CY^2(\Delta\sigma)^2} \ln \frac{a_f}{a_0} \quad (1.11)$$

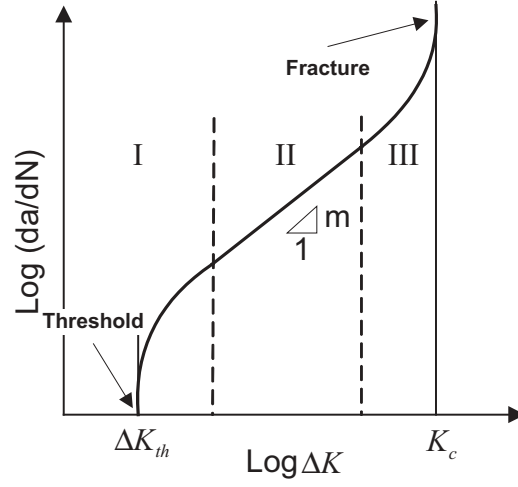


Figure 1.5: Typical fatigue-crack-growth diagram showing three different regimes.

for $m = 2$. The assumption that Y remains constant is generally not true. Therefore, the integration is usually performed numerically. It is easily seen from Equation 1.10 that when $a_0 \ll a_f$ (which is generally true for ductile alloys), the fatigue life, N_f , is not sensitive to the final crack length, a_f , but is strongly dependent on the choice of a_0 .

Many experimental data show a good agreement with the Paris-power-law relationship, Equation 1.7, under a proper choice of the constants, C and m . However, this observation is true only at a particular range of crack-growth rates. At the extreme values of ΔK , both below and above those of the Paris regime, the Paris-power law does not hold. The whole $\log(da/dN)$ versus $\log(\Delta K)$ curve assumes a sigmoidal shape, defining three distinct regimes, as shown in Figure 1.5. In Regime I, there exists a threshold-stress-intensity-factor range, ΔK_{th} , such that at $\Delta K < \Delta K_{th}$, the crack practically does not grow. Above the threshold, there is a steep increase in

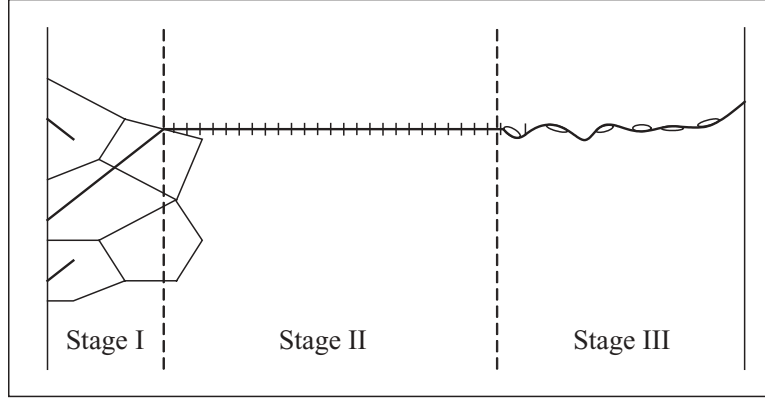


Figure 1.6: Schematic of fatigue-crack growth across a specimen section indicating three stages of fatigue failures. (After Laird [17].)

da/dN with increasing ΔK . The Paris-power law applies in Regime II. In Regime III, the K_{max} value approaches that of the fracture toughness, K_c . The crack-growth rates increase rapidly, causing catastrophic failures.

1.4.1 Fatigue-Failure Mechanisms

The fatigue-failure process begins with the accumulation of damages at localized regions, which leads to the nucleation of cracks and their subsequent propagation. When one of the cracks has grown to such an extent that the remaining net-section is insufficient to carry the applied load, a sudden fracture results. In general, the fatigue cracks nucleate at the free surface, and the crack growth occurs in three stages, as shown schematically in Figure 1.6 [17]. The Stage I crack-propagation region is an extension of the initiation process (the initiated crack length here is of about $10 \mu\text{m}$) and the crack grows in a crack-tip shear plane oriented at approximately 45° to the stress axis. The extent of this stage is usually a few grains. The Stage II growth is

controlled by the continuum response of the materials and is perpendicular to the applied stress. When the crack-tip deformation becomes large enough, the local static fracture (*i.e.*, hole growth) contributes to the crack advance until the final failure, which is in the stage III crack-growth region. Stages I, II, and III are characterized by the featureless, striated, and microvoid-coalescence regimes, respectively. The fractional area of the fracture surface for the three stages depends on the applied strain level. When the strain level is high, the failure life will be small, and the striated regime (stages II) covers almost the entire fracture surface. In contrast, when the strain level is small, the failure life will be long, and the featureless regime (stage I) begins to dominate. Only for very high strain-level tests will stage III occupy a significant area. In the total-life approach to fatigue, both the crack nucleation and the three crack-growth stages are involved. In the fatigue-crack-growth testing of the fracture-mechanics approach, usually only the last two stages of crack propagation will be present.

1.4.2 Fatigue-Crack-Initiation Mechanisms

It is generally agreed that the persistent slip bands (PSBs) are major nucleation sites for cracks in metals and alloys of high purity. The PSBs are thin lamellae where the plastic deformation is mainly concentrated during fatigue cycling. The plastic strain in the PSB lamellae is at least an order of magnitude higher than that in the matrix [18]. The cracks tend to form at the deep narrow intrusion, and the interface between the PSB and matrix, which is a plane of discontinuity and across which there are steep gradients in the density and distribution of dislocations. As schematically

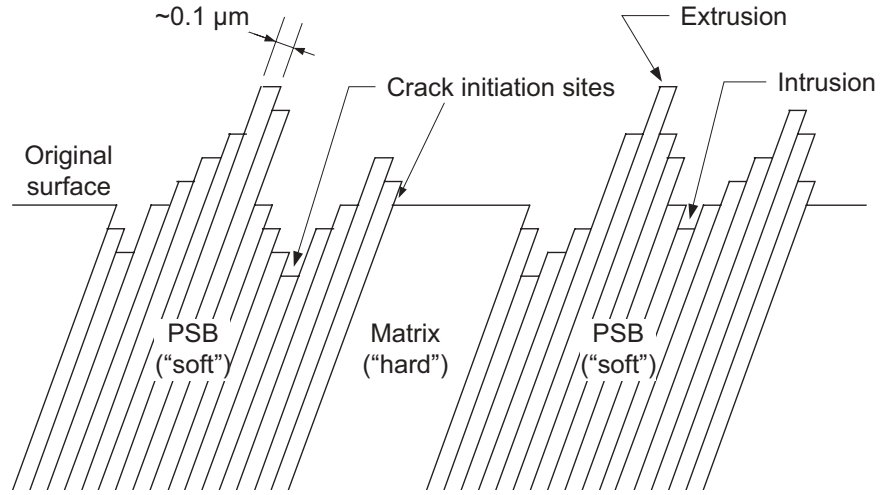


Figure 1.7: Schematic shows the roughening of surfaces due to the formation of persistent slip bands (PSBs). (After Ellyin [19].)

shown in Figure 1.7 [19], the formation of PSBs leads to surface roughening, which is manifested as microscopic *hills* and *valleys*, commonly referred to as *extrusions* and *intrusions*. The intrusions function as micronotches, and the effect of the stress concentration at the root of the intrusions promotes the additional slip and fatigue-crack nucleation. The crack initiation in commercial alloys is very complicated. It could happen at voids, slag or gas entrapments, inclusions, dents, scratches, forging laps and folds, macroscopic stress concentrations, as well as regions of microstructural and chemical non-uniformities [20–22]. In metals and alloys of high purity, cracks usually initiate at the free surface. In commercial alloys, it is common to see fatigue cracks nucleate at both near-surface and interior locations. The exact initiation sites are often specific to the alloy system considered.

1.4.3 Fatigue-Crack Growth Mechanisms

As mentioned earlier, the direction of the stage-I crack growth (see Fig. 1.6) is oriented at about 45° to the stress axis. The direction of the stage II fatigue-crack growth is perpendicular to the applied stress. The stage-II growth is due to some irreversibility in the crack-tip flow within each cycle. A mechanism for the stage-II growth is shown in Figure 1.8 [23]. There are other mechanisms, which are not essentially different. Because of the high stress concentration, the plastic deformation is present at the crack tip even at very low loads. The plastic deformation means slip (due to shear stresses) of atomic planes (Fig. 1.8, stage B). Slip also occurs on the complementary planes, and the crack widens and becomes blunted (Fig. 1.8, stages B-D). Upon unloading (or compressive loading), the slip direction in the end zones is reversed, and the crack faces are crushed together to form a resharpener crack tip (Fig. 1.8, stage E). Theoretically, the whole slip process could be reversed so that the end result after unloading would again be as stage A. However, because of the oxidation of the freshly exposed material along the slip steps, and the general disorder due to the slip, the process is irreversible in practice; and the crack extension, Δa , remains the same after each cycle. The resharpener crack (stage E) is, then, advanced and becomes blunted in the next stress cycle, and so on (stages F-K).

The crack growth per cycle is generally on the order of $10^{-7} - 10^{-3}$ mm ($10^{-8} - 10^{-4}$ inches). If the cycle number is $10^4 - 10^8$, the crack will grow by 25.4 mm (one inch). The repeated blunting and sharpening results in marks on the crack surface, which are visible at high magnifications in an electron microscope. These marks,

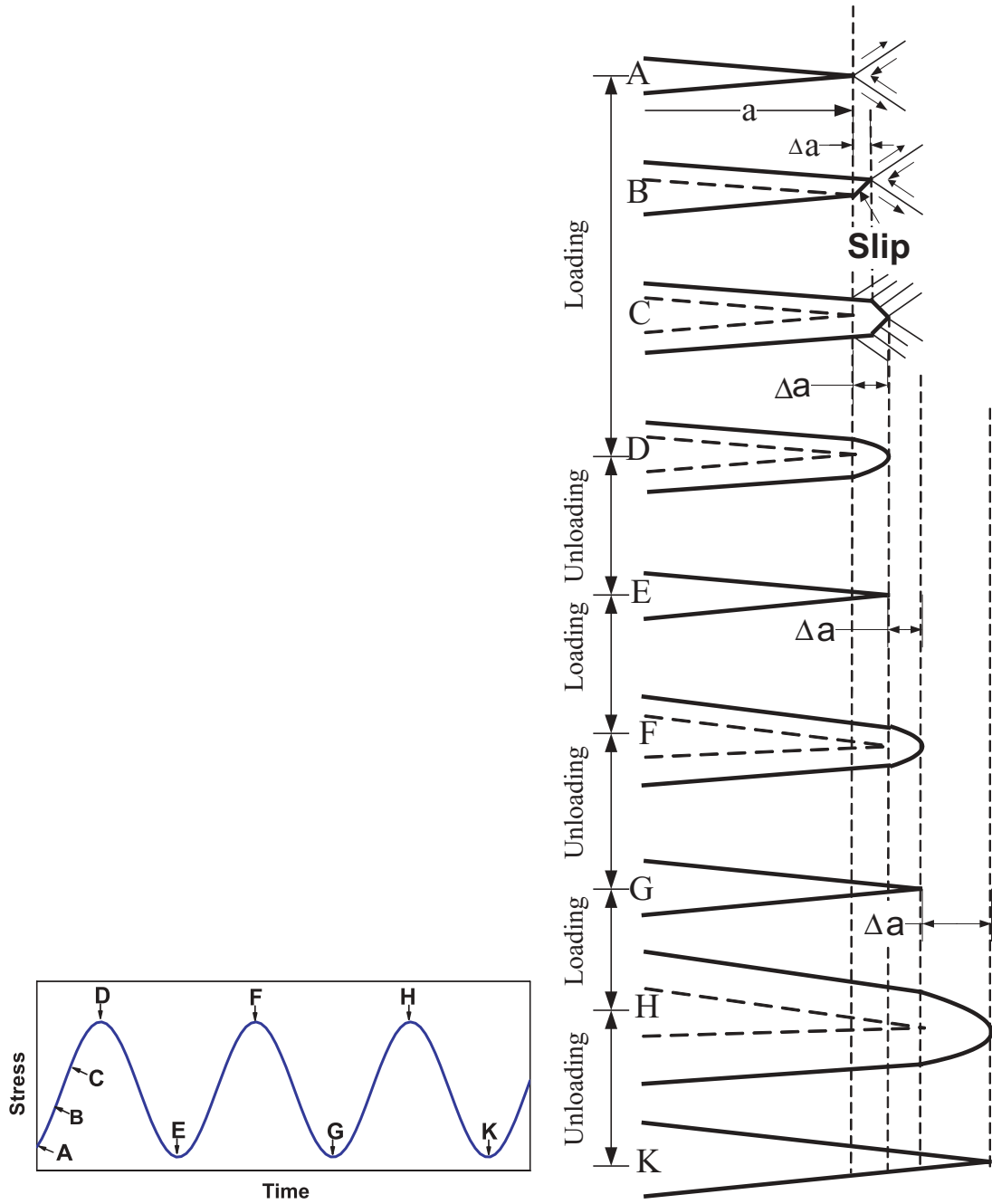


Figure 1.8: One of the various possible mechanisms of fatigue-crack growth. (After Broek [23].)

commonly termed fatigue striations, represent the successive positions of the crack front.

The Stage III fatigue-crack growth is the final fracture. It is unstable and very fast. It occurs by either cleavage or ductile rupture. Cleavage takes place along the low-index atomic planes, leading to a fracture surface of many flat and bright facets. The majority of engineering materials, however, fractures by the ductile rupture. All engineering materials contain particles and inclusions. These particles are usually complex compounds of the alloying elements, which are added to improve the alloy's strength and castability or machinability. The ductile rupture process, which is schematically illustrated in Figure 1.9 [23], is related to these particles and inclusions. Firstly, the large particles are broken or separated from the matrix, forming large holes. Secondly, concentrated slip occurs in channels connecting the cleaved large particles because of the stress concentration and the less load-bearing materials. The slip results in the separation of the small particles from the matrix, forming myriads of small holes. The final and quick separation happens along the small holes inside the channels. Thus, the large holes from the large particles and the small holes from the small particles are present on the fracture surface. These holes are visible with the electron microscopy, and are commonly termed *dimples*.

1.5 Overloading Effects

The fatigue-crack-propagation study is an essential part of structural integrity analysis based on damage-tolerance considerations. Since fatigue-crack growth is driven

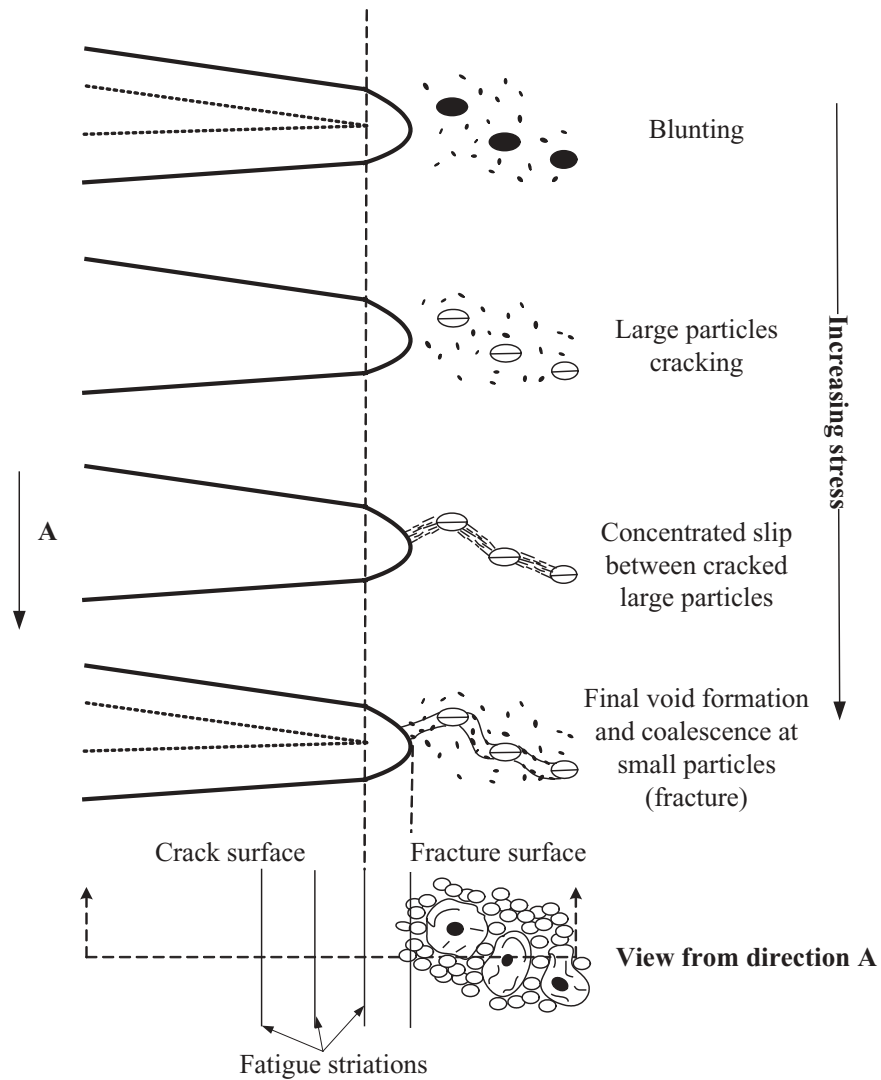


Figure 1.9: Schematic of ductile rupture showing four stages. Bottom: the fracture surface illustrates dimples formed by large and small particles. (After Broek [23].)

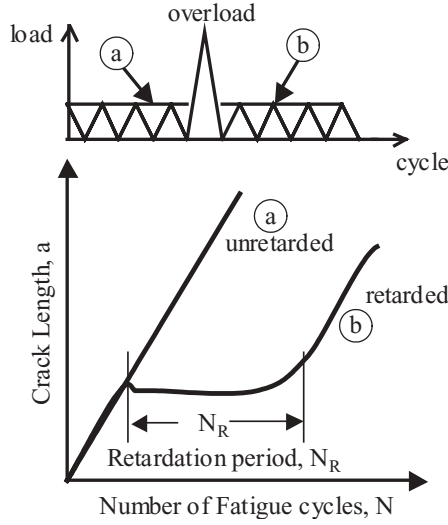


Figure 1.10: Illustration of the overload effect.

predominantly by the crack-tip plasticity, and plastic strains are irreversible, changes in the load patterns invariably result in transient effects, which affect fatigue-crack-growth rates and fatigue lives. Load-interaction phenomena complicate life predictions, and their understanding is crucial to crack-growth predictions. Figure 1.10 is an illustration of the overload effect. In Figure 1.10, constant amplitude loading is interrupted by a single overload. The overload effect was discovered in 1961 [24]. It is known that the retardation cycle period, N_R , in Figure 1.10 increases with the magnitude [25–28] and number of overloads [29–31]. Some general observations can be made, which are summarized below [32]:

1. Overloads produce the retardation while underloads produce acceleration relative to the background growth rates.

2. The retardation generally is measured in terms of the retardation cycle period, N_R , shown in Figure 1.10.
3. The retardation effect depends on the overload ratio, $K_{max.}(overload)/K_{max.}(fatigue)$, ΔK , and the R-ratio.
4. The overload can produce a very short initial acceleration before a significant deceleration happens (Figure 1.10).
5. The maximum deceleration of growth rates occurs a short distance away from the overload point. This effect is a delayed retardation. This delayed retardation depends on overload ratio, $K_{max.}(overload)/K_{max.}(fatigue)$, ΔK , and the R-ratio.
6. For the same overload ratio, $K_{max.}(overload)/K_{max.}(fatigue)$, N_R reaches a minimum as a function of the ΔK .
7. The retardation persists until the crack has propagated out of the perturbed plastic zone, a distance related to both the background plastic zone and the spectrum of the overload. Therefore, N_R depends on both the background plastic zone and the overload-induced plastic zone sizes.
8. The retardation effect depends on the specimen thickness, since plastic zone size, under plane stress and plane strain conditions differ. Retardation effects generally are larger under plane stress conditions.

9. All the factors, including the specimen geometry, temperature, environment, and material properties, which influence the plasticity at the crack tip, will have direct or indirect effects on overload effects.
10. Since interactions between plastic zones are non-linear, sequential effects of overloads, under loads, block loads, and over and under loads are not easily to be quantified.

The current explanations of the crack-closure mechanisms are listed below:

1. Crack-tip blunting
2. Crack deflection, branching, and secondary cracking
3. Crack-tip-strain hardening or residual stresses ahead of the crack tip
4. Plasticity-induced closure
5. Roughness-induced closure
6. Oxide-induced crack closure
7. Transformation-induced crack closure
8. Viscous-fluid-induced crack closure

Crack-tip blunting is that these effects may be important to some extent at high overloads in ductile materials [33–35]. The reason is that the overloads cause crack-tip blunting, affecting crack growth that requires re-sharpening of the crack-tip and the stresses in the immediate vicinity of the crack tip. Crack-tip branching,

deflection, and secondary cracking affect the crack-tip driving forces [34, 36]. The mechanisms [36] are important for materials with a significant planarity of slip, and these mechanisms can be accentuated by certain environments or microstructure. Crack-tip strain-hardening affects the initiation of further slip by the increase of flow stress [37]. This trend retards crack propagation, if the growth is activated by slip. On the other hand, if the crack grows by brittle types of mechanisms, work hardening can have an opposite effect. The effects of strain hardening are of the same as the compressive residual-stress effects, since both are the same manifestations. The extended plastic zone formed during the overload induces a backward force at the crack tip and, hence, future slip from the crack has to overcome this backward force induced by the plasticity ahead. Several investigators have concluded that the crack retardation is due to residual stresses ahead of the crack tip [38–40]. Schijve and others [41] conclude that the residual stresses do not affect directly the crack-tip driving force. Fleck [34] raised several additional criticisms to the residual-stress arguments. First, because the largest residual stresses are present immediately after the overload, the retardation should occur immediately after the overload. Second, retardation persists ever after the crack has propagated out of the overload reverse plastic zone, and beyond the region of the overload-induced compressive residual stresses. Third, the mean stresses relax in the reversed plastic zone as the crack advances. Hence, the full K is experienced by the crack tip regardless of the sign and magnitude of the residual stresses. Finally, K is the only crack-tip-driving force under fatigue, and it is unaffected by residual stresses. The plasticity-induced closure has received a considerable support in the fatigue community because of the above

perceived limitations of the residual stress hypothesis. There are many experimental results in the literature that attribute the retardation effects to the plasticity-induced crack closure [28, 42–44]. Our results about this mechanism will be discussed later. Several authors [33, 43, 44] have attributed the retardation effect to the roughness-induced closure. Roughness arises not from the overload plasticity, but from the slip planarity and crack-path tortuosity. Since overload effects are common across the board, it is unlikely that the roughness-induced closure is the general cause for the overload-retardation effects. Suresh and other co-workers [45] studied the oxide-induced crack closure mechanism in a bainitic and martensite steel. Moist air introduced the oxide-induced crack-closure, and an oxide film formed at the crack tip, induced crack closure and produced the retardation in the fatigue-crack growth. This mechanism requires environments, which is not in common conditions. The phase-transformation-induced crack-closure affects the fatigue-crack-growth behavior too. The deformation-induced martensite transformation around a fatigue crack has been reported by a few researchers [46]. Crack closure is one of the major causes for the retardation effects. It is instructive to examine the magnitude of the crack closure and its relative role in the crack growth process. Detailed and careful analysis of crack-closure measurements by Donald et al. [47] have shown that the ASTM (The American Society for Testing and Materials) criteria for the determination of the K -closure. Sadananda and his co-workers [31] proposed a crack-closure measurement, based on the shape of the load versus displacement curve.

Chapter 2

Neutron-Diffraction Studies on Lattice-Strain Evolution around a Crack Tip during Tensile Loading and Unloading Cycles

2.1 Introduction

Fracture mechanics is a science that relates the global-loading configuration to the stress and strain state around a crack tip. From a material-development perspective, new materials with a better crack resistance could be designed only after a better understanding of the damage mechanism. The effect of overload [30], as shown in Figure 1.10, on the crack-propagation behavior during a cyclic deformation has

drawn much attention since its discovery in 1961 [38]. However, the phenomenon is still not well understood. After the overload, there is a period of crack-growth retardation (Figure 1.10) that is related to the magnitude [24–27] and number of overloads [28–30]. The retardation of the crack-growth behavior can be understood in the context of the plasticity-induced crack-closure mechanism. The plastic deformation of the material in the wake of the crack leads to the premature contact of the crack faces in the unloading sequence, resulting in the reduced cyclic plastic deformation and, therefore, the decreased crack-propagation rate. Measuring the residual strains near the crack tip associated with the prior plastic deformation can help understand the crack-propagation behavior and investigate the overload effect. In general, compressive residual strains/stresses are found to decrease the crack-propagation rates, while tensile residual strains/stresses increase the rates [48]. The size of the plastic-zone at the crack tip created by prior loading can provide a measure of the deformation around the crack tip. Therefore, a method for measuring residual stresses and the plastic zone surrounding a crack is useful for understanding the crack-growth behavior and the crack-tip deformation mechanism [49].

Neutron diffraction, as a nondestructive-evaluation technique, can determine the elastic lattice strains in a polycrystalline material from the shift of the diffraction peaks. At the same time, the plastic deformation can be estimated from the broadening of the diffraction peaks. In the present investigation, lattice strains near a fatigue crack in a compact-tension (CT) specimen were measured during tensile loading and unloading cycles using neutron diffraction to determine the effect of overload.

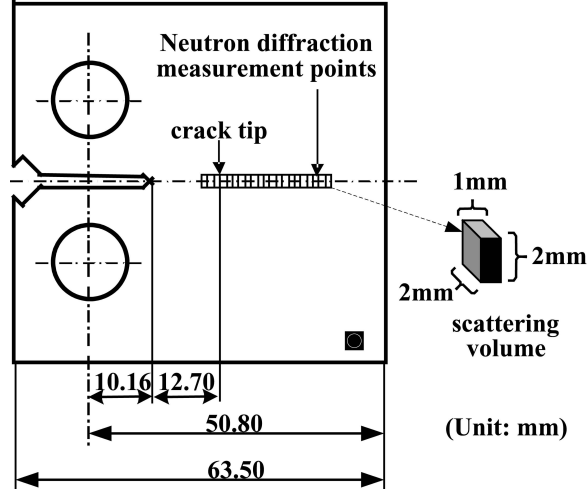


Figure 2.1: Compact-tension specimen of Type 316 LN stainless steel with a thickness of 6.35 mm, a notch length of 10.16 mm, and a crack length of 22.86 mm. Diffraction patterns were measured along the crack length with a scattering volume of $2 \times 1 \times 2 \text{ mm}^3$ with the 1 mm along the crack length.

Diffraction-peak broadening was also determined near the crack tip to estimate the plastic-zone size.

2.2 Experimental Details

The experiment was performed on a CT specimen of a Type 316 LN stainless steel, which is a low-carbon nitrogen-added (LN) austenitic stainless steel. A schematic of the CT specimen, with a width of 50.8 mm, a thickness of 6.35 mm, and a notch length of 10.16 mm, is shown in Figure 2.1. The CT specimen was pre-cracked under a fatigue-loading condition using a Material Test System (MTS) electrohydraulic machine. Fatigue loading was performed in a load-control mode with a frequency of 10 Hz and a constant load ratio, $R = 0.1$, where $R = P_{min.}/P_{max.}$. $P_{min.}$ and

$P_{max.}$ are the applied minimum (667 N) and maximum (6,667 N) loads, respectively. The maximum stress intensity, $K_{max.}$, reached $38.8 \text{ MPa}\sqrt{m}$ during fatigue crack growth. The K value was obtained, using Equation 1.2 [50–52]. The total crack length was extended to 22.86 mm after 169,052 cycles. Crack length was measured by the crack-opening-displacement (COD) gauge using the unloading-compliance technique [53–57]. After the test, the crack length was also confirmed by the dye-penetrant inspection, and the results agree within 0.02 mm.

Neutron-diffraction measurements were performed on the Spectrometer for Materials Research at Temperature and Stress (SMARTS) [58] at the Los Alamos Neutron Science Center (LANSCE). Utilizing the continuous energy spectrum of the incident neutron beam, the entire diffraction pattern (d-spacing ranging from 0.5 to $4\text{\AA}$) was recorded in two stationary detector banks at fixed diffraction angles (2θ) of $\pm 90^\circ$. The samples were aligned such that the tensile load axis was 45° from the incident beam and, thus, the diffraction vectors associated with the two detector banks were parallel to the in-plane and through-thickness directions of the sample. The incident neutron beam was defined by 2 mm and 1 mm (along the crack length) slits, and the diffracted beams were collimated using 2 mm radial collimators, resulting in a 4 mm^3 gauge volume, Figure 2.1. Thirty-two diffraction patterns were recorded along the crack length at a given applied load. The interatomic spacings in the in-plane direction (parallel to the loading direction) were determined by the Rietveld refinement [59] of the diffraction patterns and, subsequently, the lattice strains were calculated from

$$\varepsilon = \frac{a - a_0}{a_0} \quad (2.1)$$

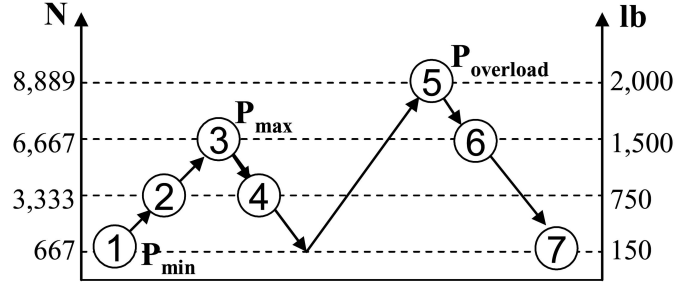


Figure 2.2: Applied tensile loading-unloading sequence. Neutron strain scanning was performed at each load point (LP) from LP1 through LP7.

where a_0 , the stress-free lattice parameter, was measured away from the crack tip (marked $\square$ in Figure 2.1). The analysis of the through-thickness data will not be discussed in this paper, due to the length constraint.

The strain scan was repeated at seven different load values as shown in Figure 2.2. In this paper, the measurement results at load points (LP) 1 (667 N), 2 (3,333 N), 3 (6,667 N), 5 (8,889 N), 6 (6,667 N), and 7 (667 N) are presented. The LP5 is an overload.

In polycrystalline materials, the plastic deformation may generally be related to the broadening of a diffraction peak [60] although other factors, such as small grain sizes, can also contribute to broadening. The extent of the plastic zone in front of the crack tip is a function of the prior loading history of the CT specimen. From the changes in the full-width-half-maximum (FWHM) as a function of scanning positions, the plastic zone near the crack tip was estimated. The peak-width change was determined, using:

$$\Delta SIG = \frac{SIG_f - SIG_0}{SIG_0} \quad (2.2)$$

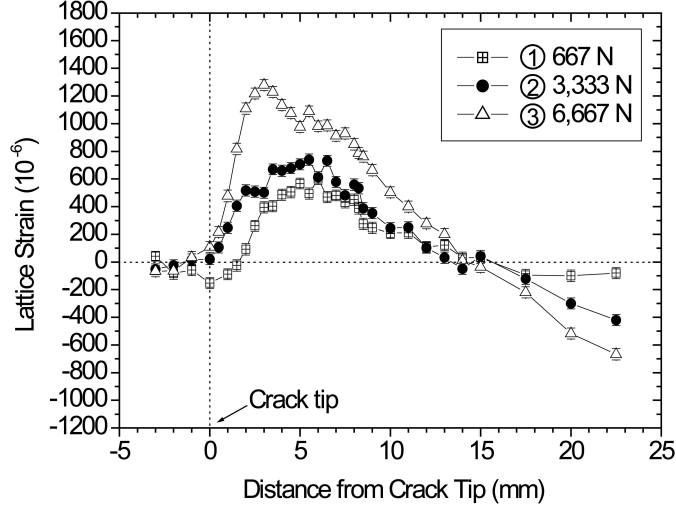


Figure 2.3: In-plane lattice-strain profiles in front of the crack tip at applied loads of 667 N, 3,333 N, and 6,667 N, which correspond to LP1, LP2, and LP3, respectively, in Figure 2.2.

where ΔSIG is the fractional peak-width change, and SIG_f and SIG_0 represent the square of the variance of the diffraction peaks, determined for the strained and unstrained materials, respectively, using the Rietveld refinement [59].

2.3 Results and Discussions

Figure 2.3 shows the in-plane (IP) lattice-strain profiles measured at various applied loads (LP 1, 2, and 3 in Figure 2.2). At 667 N (LP1), in the IP direction, small compressive strains (with a maximum of -194×10^{-6} or $\mu\epsilon$) were observed within ± 3 mm of the crack tip. The strains become tensile in front of the crack tip, reaching a maximum value of 526×10^{-6} (or $\mu\epsilon$) at approximately 5 mm from the crack tip, and then decreasing as the distance increases. At about 15 mm, the strains become compressive. When the load is increased from 667 N to 6,667 N (LP3), the maximum

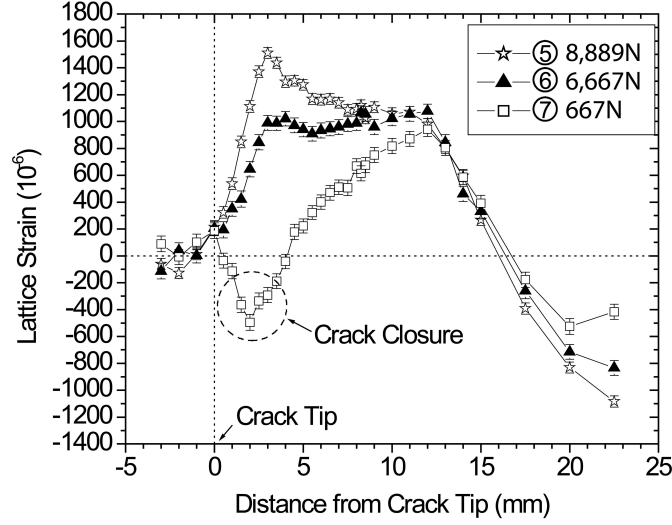


Figure 2.4: In-plane lattice-strain profiles in front of the crack tip at applied loads of 8,889 N, 6,667 N, and 667 N, which correspond to LP5, LP6, and LP7, respectively, in Figure 2.2.

strain gradually increased. Two maxima were observed in the strain profile at 6,667 N. The first maximum (at about 3 mm from the crack tip) is associated with the in-situ applied loading of the sample, while the second maximum (at about 6 mm from the crack tip) is associated with the bending effect due to the long crack length. Note that the bending effect is an unforeseen effect, which caused a tensile to compressive transition at around 15 - 17 mm.

The lattice-strain profiles in the IP direction in front of the crack tip as a function of the applied load after the overload at LP5 is presented in Figure 2.4. The first maximum in the strain profile (at around 2 - 3 mm from the crack tip) diminished as the applied load was decreased from 8,889 N (LP5) to 667 N (LP7). At the tensile load of 667 N following the overload to 8,889 N, compressive lattice strains of up to $-600 \mu\epsilon$ were observed within 4 mm from the crack tip, indicating the

plasticity-induced crack closure. This could, subsequently, retard the crack propagation, decrease the crack-growth rate, and increase the fatigue life. The lattice strain profile associated with the bending continues to show a transition from the tensile to compressive strains at about 15 - 17 mm from the crack tip, showing maximum strains decreasing with a decrease in applied tensile loadings.

The plastic-zone size ($2r$) was calculated to be 7.9 mm at LP7 using [e.g., [1]]:

$$2r = \frac{m-1}{m+1} \times \frac{1}{\pi} \left(\frac{K_{\max}}{\sigma_0} \right)^2 \quad (2.3)$$

where $K_{\max}$ is the maximum stress-intensity factor of $51.8 \text{ MPa}\sqrt{\text{m}}$, which was calculated using Equation 1.2 with $P_{\text{overload}} = 8,889 \text{ N}$, crack length $a = 22.86 \text{ mm}$. Note that the crack length did not grow immediately after the overload (i.e., at LP7). σ_0 is the yield strength of the material (288 MPa) [23-25] and m ($= 7.7$) is the inverse of the strain-hardening exponent, which was a material constant derived from the regression of the stress-strain curve [22, 23].

The elastic lattice-strain profile at LP7 is plotted in Figure 2.5 along with the changes in the diffraction-peak width. The increase in peak width in front of the crack tip indicates the plastic-zone size. Thus, there is a good agreement between the plastic-zone sizes determined from the theoretical prediction (7.9 mm) and peak-width-broadening measurements (about 7 - 9 mm). The plastic deformation during overload caused the compressive lattice strain in front of the crack tip, which could induce the crack-closure phenomenon. Finally, it should be noted that the sharp transition from tensile to compressive strains at around 15 - 17 mm due to the

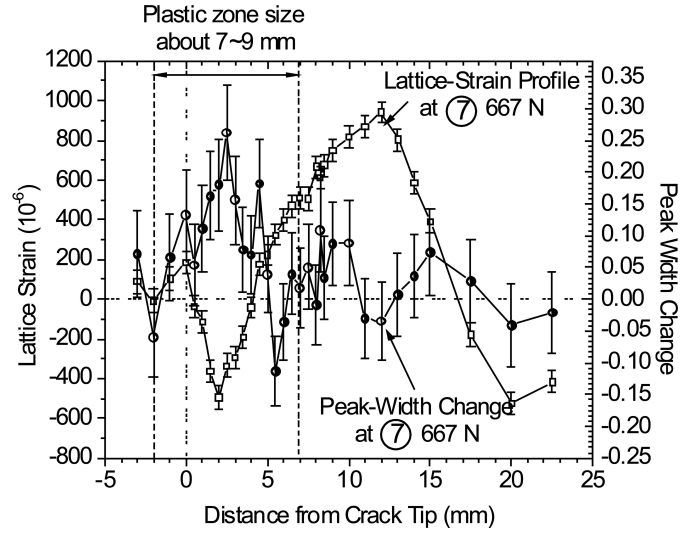


Figure 2.5: Comparison between the in-plane lattice strain profile and the plastic-zone size (about 7 ~ 9 mm) estimated from the changes in the diffraction-peak width at 667 N (LP7 in Figure 2.2). Note that the plastic-zone size calculated using Equation 2.3 is about 7.9 mm.

bending effect did not cause any significant diffraction-peak broadening. The 1-mm spatial resolution used along the crack length may have been small enough to resolve the gradient around 16 mm avoiding significant diffraction peak broadening. More detailed discussion on the correlations among the microstructure, elastic/plastic strain distribution, and the changes in fatigue-crack growth rates will be presented in a future publication.

2.4 Summary

Elastic lattice strains and the associated plastic zone around the crack tip in Type 316 LN steel were investigated using an in-situ neutron-diffraction technique. The elastic strains in the in-plane direction (parallel to the loading direction) were measured as a function of the distance from the crack tip at different applied loads ranging from 667 N to 8,889 N. The results showed the evolution of lattice-strain profiles associated with the in-situ applied loading as well as the residual-strain profile generated by an overload. After the overload, a significant compressive strain of $-600 \mu\epsilon$ (or $\times 10^{-6}$) was observed in front of the crack tip, which is related to the crack-closure phenomenon and retardation of crack growth. The plastic zone size in front of the crack tip was also estimated, using the peak-broadening analysis and it agrees well with the theoretically predicted plastic-zone size of 7.9 mm.

Chapter 3

Changes in Lattice-Strain Profiles around a Fatigue Crack through the Retardation Period after the Overload

3.1 Introduction

Fatigue-crack growth is driven predominantly by the crack-tip plasticity. Sudden changes in the loading patterns during a cyclic deformation could result in complicated plastic zones around the crack tip and influence the crack-growth rate and fatigue life. More specifically, “overloading” beyond the regular stress amplitude during fatigue causes the retardation of crack-growth rates. The retardation persists

until the crack has propagated through the perturbed plastic zone. Therefore, the retardation period depends on both the fatigue-induced and overload-induced plastic zones [31, 61].

Neutron diffraction is a nondestructive method [62] that is useful for the study of the overload effects by mapping the elastic internal strains around the crack tip. Previously, a compressive lattice-strain profile was clearly observed ahead of a fatigue-crack tip after an overload using neutron diffraction [63]. In this paper, the evolution of the residual-strain profiles around a fatigue crack is investigated during the retardation period, and the result is discussed in the context of the perturbed plastic zone. The results provide a microscopic understanding of the overload effect during cyclic loading.

3.2 Experimental Details

Figure 3.1 shows a typical crack-growth rate (da/dN) vs. stress-intensity-factor range (ΔK) curve measured from a Type 316 LN (low-carbon nitrogen-added) austenitic stainless steel. A total of seven compact-tension (CT) specimens were prepared to represent different stages through the da/dN vs. ΔK curve as marked in Figure 3.1. The CT specimens (with a width of 50.8 mm, a thickness of 6.35 mm) were fatigue-deformed using a Material Test System (MTS) electrohydraulic machine with a frequency of 10 Hz and a constant load ratio, $R = 0.1$, where $R = P_{min.}/P_{max.}$, $P_{min.}$ and $P_{max.}$ are the applied minimum (988 N) and maximum (9,880 N) loads, respectively [61, 63]. The crack lengths of the specimens (SP) are: SP1 = 13 mm, SP2

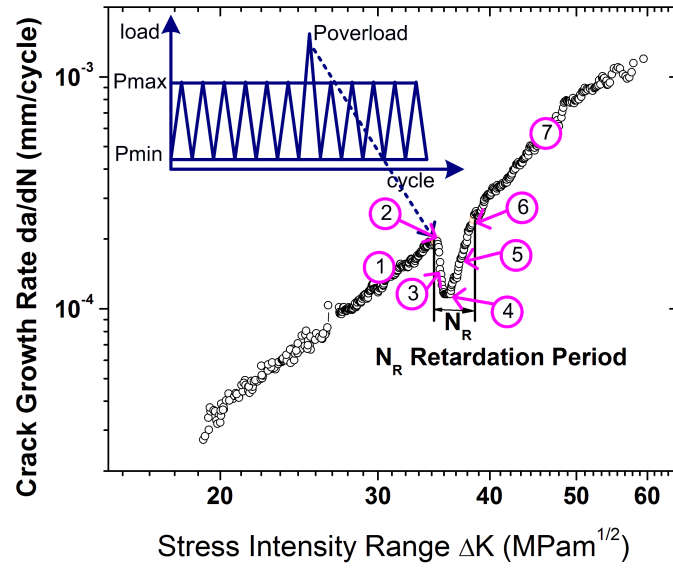


Figure 3.1: The effect of overload on fatigue-crack propagation. During the fatigue tests, seven specimens were prepared.

= 15.3 mm, SP3 = 15.6 mm, SP4 = 16.3 mm, SP5 = 17 mm, SP6 = 17.9 mm, and SP7 = 25.3 mm. Crack lengths were measured by the crack-opening-displacement (COD) gauge using the unloading-compliance technique [52–54]. Specimen 1 (SP1) did not experience the overload and was subjected only to the fatigue deformation. The overload (13,189 N) was applied as the crack length reached 15.3 mm. At the overload, the maximum stress-intensity factor, K_{max} , was $51.8 \text{ MPa}\sqrt{m}$. The specimen 2 (SP2) was stopped right after the overload. Specimens 3 to 7 were fatigue deformed further after the overload through the retardation period as shown in Figure 3.1. Note that the values of ΔK and K_{max} can be calculated using the applied load and crack length for the CT specimens [52–54, 63].

Neutron-diffraction strain measurements were performed using the NRSF2 instrument at the High Flux Isotope Reactor of the Oak Ridge National Laboratory. The wavelength from the Si331 doubly-focusing monochromator was $1.73 \text{ \AA}$. The stainless-steel (311) reflection was used for determining the strain. The incident neutron beam was defined by 2-mm wide and 1-mm tall (along the crack length) incident beam slits, and the diffracted beam was collimated using a 2-mm wide slit, resulting in a 4 mm^3 gauge volume defined at the middle of the specimen thickness. About thirty diffraction patterns were recorded for each specimen along the crack length from approximately 5 mm behind the crack tip to about 20 mm ahead of the crack tip. The d-spacings along transverse (parallel to fatigue loading), longitudinal, and normal directions were determined by the Pseudo-Voigt fitting of the (311)

reflections and, subsequently, the lattice strains were calculated using:

$$\varepsilon = (d - d_0)/d_0 \quad (3.1)$$

where d_0 , is the stress-free lattice parameter, which was measured away from the crack tip at a corner of each CT specimen, using the same gauge volume for all orientations. For brevity, only the strain data of SP1, SP2, SP3, and SP6 along the transverse direction are reported here, and the data for the other specimens and the strain components measured along the longitudinal and normal directions will be discussed in a future publication.

3.3 Results and Discussions

Strain profiles along the transverse direction are shown as a function of the distance from the overload point, Figure 3.2. The error bars are from the diffraction peak fitting. Note that the distance from the overload point is the same as the distance from the crack-tip position of the SP2. The profile of SP1 shows small compressive (negative) strains near its crack-tip position (- 5 - 0 mm) with balancing tensile (positive) residual strains at 0 - 8 mm. SP2 (the overloaded specimen) exhibits large compressive strains (maximum of about - 950×10^{-6} or $\mu\varepsilon$) around 0 - 7 mm in front of its crack tip due to the overload. Tensile strains were observed around 7 - 15 mm. SP3, which was fatigue deformed for 28,414 cycles after the overload, shows a similar profile as SP2. However, SP6 exhibits a significantly different strain profile with a smaller magnitude (about - $600 \mu\varepsilon$) and a wider double-peak shape.

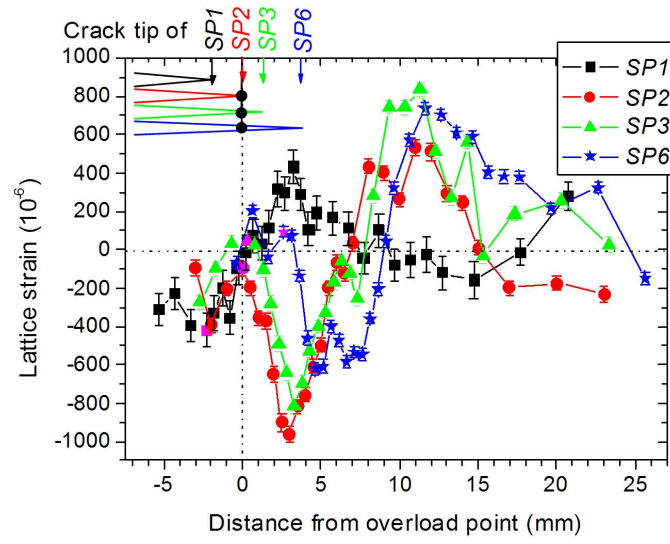


Figure 3.2: The lattice-strain profiles of SP1, SP2, SP3, and SP6 along the transverse direction.

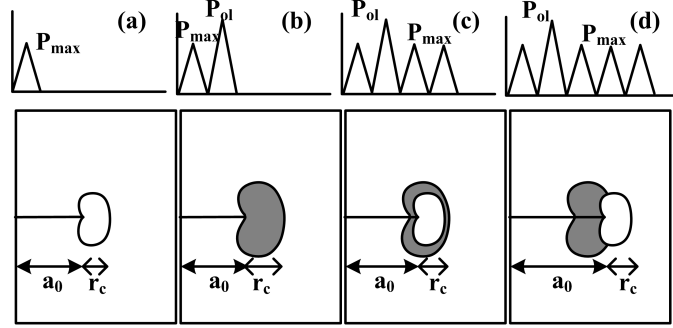


Figure 3.3: The schematic illustration of the perturbed plastic zone. a_0 is the crack length, and r_c is the plastic-zone size.

The evolution of strain profiles described above could explain the retardation phenomenon with a schematic of the perturbed plastic zone represented in Figure 3.3. SP1, without the overload, showed small compressive strains generated by the fatigue deformation, Figure 3.3(a). The large compressive strains observed after the overload, (showed in Figure 3.2) are associated with the overload-induced plastic zone, as illustrated in Figure 3.3(b). The strain profile of SP3 is quite similar to that of SP2 because the increase in the crack length during the subsequent fatigue is too small (0.31 mm) to go through the plastic zone already created by the overload, Figure 3.3(c). SP6, with a sufficiently longer crack length (about 2.3 mm longer than SP3), there are two peaks in the compressive strain profile. The first peak (about 5 mm away from the overload point) is due to the large plastic zone created by the overload, which seems to have relaxed by about $300 \mu\varepsilon$, and the second peak (about 7 mm away from the overload point) is caused by the subsequent fatigue cycles, Figure 3.3(d). This trend indicates that the fatigue-induced plastic zone had grown out of the large plastic zone caused by the overload.

3.4 Summary

In summary, as it was illustrated in Figure 3.1, the overload produced the large plastic deformation (SP2), which, in turn, led to a dramatic decrease in the crack-growth rates (SP3). As the crack grew longer and out of the overload-perturbed region during the subsequent fatigue deformation (SP6), the crack-growth rates recovered to the original slope. These results greatly enhance our understanding of the overload effect during cyclic loading.

Chapter 4

The Study of Crack-Closure Phenomena after the Overload

4.1 Introduction

The material we used is a Type 316 LN SS, which is a low-carbon, nitrogen-added austenitic stainless steel. The classic 316L austenitic stainless steel is fully austenitic after the solution heat treatment, which has poor mechanical properties. It has already been shown that nitrogen alloying clearly improves monotonic and cyclic behavior of 316 L stainless steel at room and high temperature [2]. Type 316 stainless steel (SS) has been widely used in the nuclear industry, because of its excellent ductility, corrosion resistance, and irradiation performance. Its often used over a wide temperature range, and employed in components that are loaded under severe conditions (stress, strain cycling, and thermal cycling) [64]. However, this material hasn't

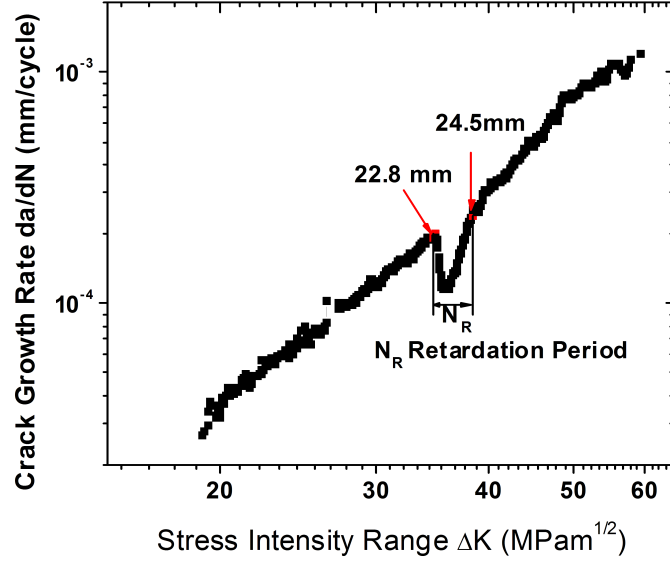


Figure 4.1: The load controlled high-cycle fatigue load sketch and crack growth rate curve.

been systematically investigated regarding its fatigue behavior including the crack-growth resistance. From a material development perspective, new materials with a better crack resistance could be designed only after a better understanding of the damage mechanism. The effect of the overload (as shown in Figure 4.1) on the crack-propagation behavior during a cyclic deformation has drawn much attention since its discovery in 1961 [15, 30]. However, the phenomenon is still not well understood. After the overload, there is a period of the crack-growth retardation (Figure 4.1) that is related to the magnitude [24–27] and number of overloads [29, 30, 65]. The retardation of the crack-growth behavior can be understood in the context of the plasticity-induced crack-closure mechanism. The plastic deformation of the material in the wake of the crack tip leads to the premature contact of the crack faces in the unloading sequence, resulting in the reduced cyclic plastic deformation and,

therefore, decreased the crack-propagation rate. Measuring the residual strains near the crack tip associated with the prior plastic deformation can help understand the crack-propagation behavior and investigate the overload effect. In general, compressive residual strains/stresses are found to decrease the crack-propagation rates, while tensile residual strains/stresses increase the rates [48]. The investigation of the elastic and plastic deformation at the crack tip created by the prior loading can provide a measure of the crack-tip deformation. Therefore, a method for measuring residual stresses and the deformation surrounding a crack is useful for understanding the crack-growth behavior and crack-tip-deformation mechanism [49].

Neutron diffraction, as a nondestructive-evaluation technique, can determine the elastic lattice strains in a polycrystalline material from the shift of the diffraction peaks. At the same time, plastic deformation can be related to the broadening of the diffraction peaks. In the present investigation, lattice strains near a fatigue crack in a compact-tension (CT) specimen were measured during monotonic tensile loading and unloading cycles using neutron diffraction to determine the effect of overload.

4.2 Experimental Details

4.2.1 Fatigue Experiments

The experiment was performed on a CT specimen of a Type 316 LN stainless steel, which is a nitrogen-added, low-carbon austenitic stainless steel. The chemical composition of Type 316 low-carbon, nitrogen-added (LN) stainless steel (SS) (wt.%, weight percent) is shown in Table 1.1. A schematic of the CT specimen, with a

width of 50.8 mm, a thickness of 6.35 mm, and a notch length of 10.16 mm, is shown in Figure 2.1.

The specimens were precracked to approximately 1.27 mm. The final stress-intensity factor range (ΔK) used during precracking was $20 \text{ MPa}\sqrt{m}$. The precracking was performed at a R ratio of 0.1, using a triangular waveform with a frequency of 10 Hz. Fatigue testing was performed at room temperature. The CT specimen was tested under a fatigue loading condition using a Material Test System (MTS) electrohydraulic machine. Fatigue loading was performed in a load-control mode with a frequency of 10 Hz and a constant load ratio. Crack length grew from 12.7 mm to 30.38 mm. The final ΔK used during fatigue test was $62.4 \text{ MPa}\sqrt{m}$. The fatigue crack was extended to 30.38 mm. The crack length was measured by the crack-opening-displacement (COD) gauge using the unloading-compliance technique [57]. An overload of 8,888 N was applied at the crack length corresponding to 22.88 mm, as shown in the Figure 4.1. Righ after the overload, the fatigue test continued with the exactly the same experiment parameter of fatigue testing.

4.2.2 Fatigue-Striation Measurements

Scanning-electron microscopy was used to observe the fracture surface after the fatigue test. Along the crack-growth direction, at different crack length, several micrographs were taken. Some of the micrographs are shown in Figure 4.2. Fatigue striations at different crack lengths were investigated. In general, as the stress intensity increases, the crack growth rate increases. The spacing between striations should increase, as crack-growth rate increases, because the striation development is

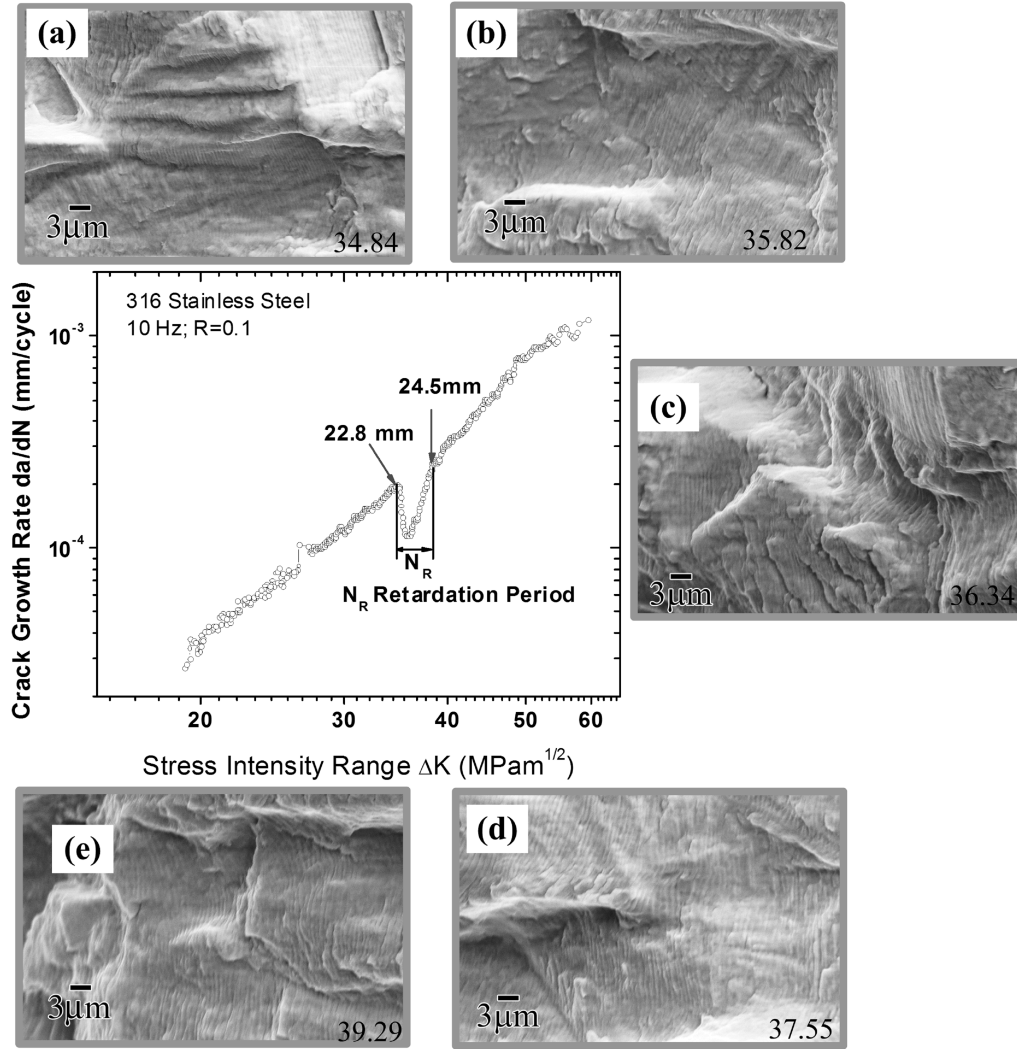


Figure 4.2: Fracture surface image at different ΔK levels. (a) 34.84 MPa $\sqrt{m}$, (b) 35.82 MPa $\sqrt{m}$, (c) 36.34 MPa $\sqrt{m}$, (d) 37.55 MPa $\sqrt{m}$, and (e) 39.29 MPa $\sqrt{m}$.

generally proportional to the crack-growth rate. The measurement of the striation spacing can help investigate the crack growth rate. The average spacing of striation can be calculated by dividing lengths of the lines, which were drawn along the striation propagation direction on the micrographs, over the number of the striation through the lines.

4.2.3 Neutron-Diffraction Experiments

Neutron-diffraction measurements were performed, using the Spectrometer for Materials Research at Temperature and Stress (SMARTS) instrument at the Los Alamos Neutron Science Center (LANSCE). Utilizing the continuous energy spectrum of the incident neutron beam, the entire diffraction pattern (from 0.5 to 4 Å) was recorded in two stationary detector banks at fixed 2θ angles of $\pm 90^\circ$, which are with diffraction vectors in the in-plane and through-thickness directions, respectively. The incident neutron beam was defined by 2 mm horizontal and 1 mm vertical slits, and the diffracted beams were collimated, using 2 mm radial collimators, resulting in a 4 mm³ gauge volume. Thirty-two diffraction patterns were recorded along the crack length at a given applied load. The interatomic spacings in the in-plane direction (parallel to the loading direction) were determined by the Rietveld refinement of the diffraction patterns and, subsequently, the lattice strains were calculated. The strain scan was repeated at seven different load values as shown in Figure 2.2.

In this paper, the measurement results at load points 1 (667 N), 2 (3,333 N), 3 (6,667 N), 5 (8,889 N), 6 (6,667 N), and 7 (667 N) are presented. The load point 5 is an overload.

4.3 Results and Discussions

After the load-controlled high-cycle fatigue test, as shown in Figure 4.1, the da/dN versus ΔK curve was plotted. In general, as ΔK increases da/dN should increase. In Figure 4.1, a retardation period is shown after the overload, which is from the crack length of 22.8 mm to 24.6 mm. The overload was applied at the crack length equal to 22.8 mm. So the overload retarded the crack-growth rate, due to the induced large plastic deformation. After 24.6 mm, the retardation disappeared; the crack-growth rate went back to the normal crack-growth rate. From the calculated average spacing of striations, the striation spacing versus crack length curve was plotting in Figure 4.3. There is an obvious retardation period, which confirmed the retardation period after the overload from the da/dN versus ΔK curve. For neutron-experimental results, this paper focuses on the discussion of the strain variations and peak-width changes after the overload. Figure 2.4 shows the lattice-strain profiles in the IP direction in front of the crack tip as a function of the applied load after the overload. As the applied load was decreased from 8,889 N to 667 N, the maximum value of the lattice-strain profile decreased accordingly. At the tensile load of 667 N following the overload of 8,889 N, compressive lattice strains of up to -500 micro-strain (10^{-6} or $\mu\varepsilon$) were observed within 4 mm from the crack tip, indicating the plasticity-induced crack closure. This trend could, subsequently, retard the crack propagation, decrease the crack-growth rate, and increase the fatigue life.

In Figure 4.4 and 4.5, at LP 1 and LP 7, the same tensile load that are introduced before and after overload, respectively, the strain profiles are different both in IP and TT directions. The comparison between the profiles in LP 1 and LP 7 exhibits the

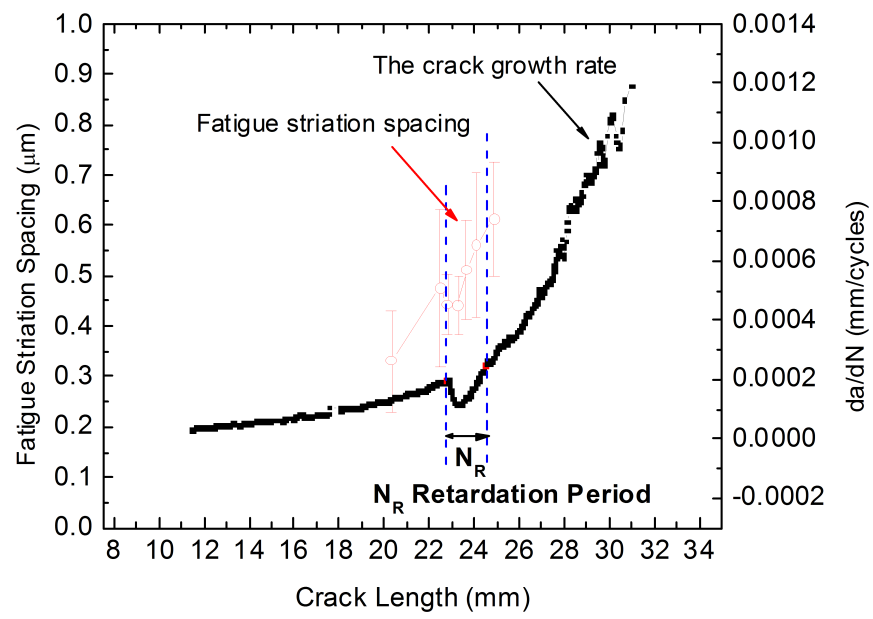


Figure 4.3: The comparison between the fatigue-striation spacing and the crack-growth rate at a function of the crack length.

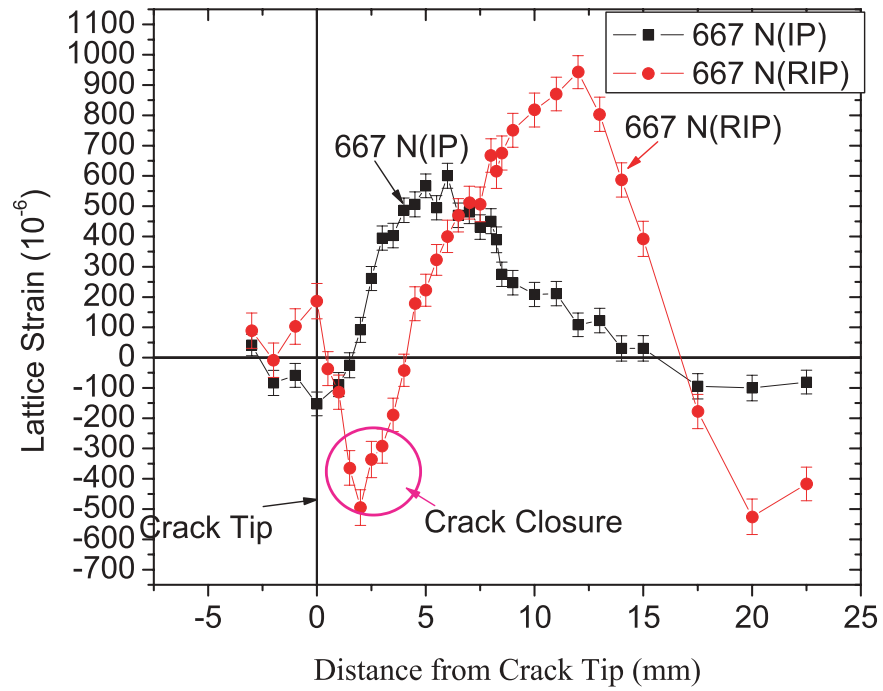


Figure 4.4: The comparison between the in-plane lattice-strain profiles in front of the crack tip at applied loads of 667 N before overload and 667 N after overload, which correspond to (1) and (7) in Figure 2.2.

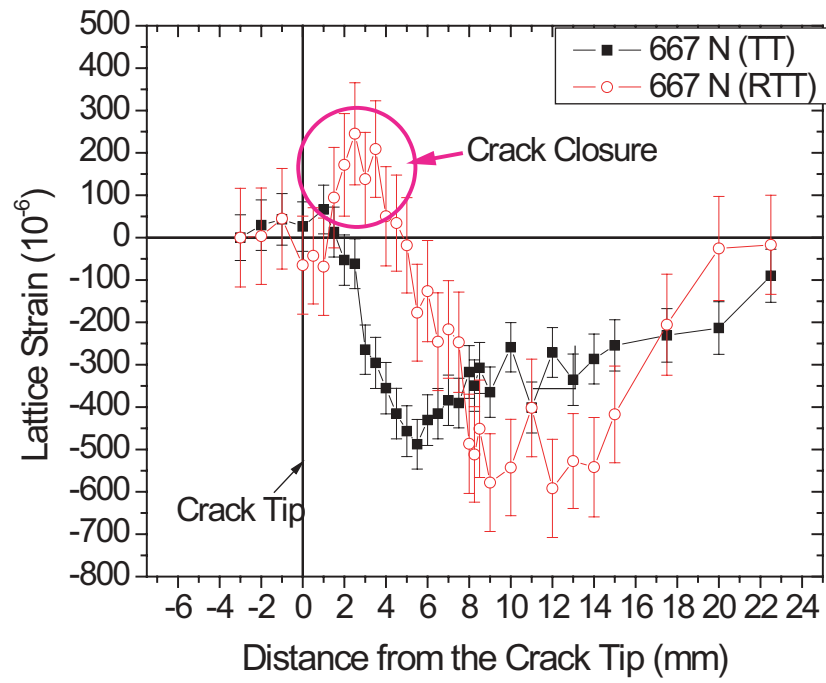


Figure 4.5: The comparison between the through-thickness lattice-strain profiles in front of the crack tip at applied loads of 667 N before overload and 667 N after overload, which correspond to (1) and (7) in Figure 2.2.

overload effect clearly. In Figure 4.4, the strain profiles in the IP direction, there are compressive strains at around 2-3 mm from the crack tip at LP 7. But the strain values at LP 1 are tensile corresponding to the tensile in-situ loading. In Figure 4.5, the strain profiles in the TT direction, there are 250 tensile micro-strain (10^{-6} or $\mu\varepsilon$) at around 2-3 mm from the crack tip at LP 7. But the strain values at LP 1 are compressive due to the Poisson's ratio effect of the tensile in-situ loading. The comparison between the strain profiles at LP 1 and LP 7 includes the overload-induced large plastic deformation around the crack tip. The plastic-deformation-induced crack closure can decrease the crack-growth rate.

4.4 Summary

During fatigue tests, after the overload, the crack-growth retardation period was observed. During the retardation period, the crack growth rate significantly decreased, which is results from the plastic deformation induced by the overload. The retardation period is also confirmed by the measurement of the striation-spacing changes before and after overload. Crack closure at the retardation period due to the plastic deformation was also investigated by neutron diffraction. From Figures 2.4, 4.4, and 4.5, the large compressive strain at LP7 was observed after the overload, which indicates the crack closure, because the compressive stress retards the crack growth. The overload induced a great amount of plastic deformation. Overall, the strain profiles, measured by neutron diffraction, confirm the crack-closure phenomenon.

Chapter 5

Neutron and X-ray Microbeam Diffraction Studies of Elastic Strain and Dislocation Distributions near a Fatigue Crack

5.1 Introduction

In order to improve damage-tolerant designs and residual-life assessments, the predictions of the fatigue-crack growth rate are necessary [30]. The study of the overload effect provides an understanding of the variable-amplitude loading state, which is more realistic in fatigue failures. After the overload, there is a period of the crack-growth retardation, which can be understood in the context of the plasticity-induced

crack-closure mechanism [24–27]. Measurements of the elastic and plastic strain distribution near the crack tip can provide a measure of the severity of the plastic deformation around the crack tip and help understand the crack-propagation behavior. Neutron diffraction (ND), due to its great penetrating power, allows bulk measurements with a typical spatial resolution on the order of a millimeter. ND is a useful technique in the determination of value-averaged elastic lattice strains in a polycrystalline material from the shift of the diffraction peak. At the same time, the plastic deformation can be estimated from the broadening of the diffraction peaks. In addition, polychromatic x-ray microdiffraction (PXM) can provide a submicron spatial resolution, which is an excellent technique for mapping the deformation around the crack tip. As the PXM penetrated a sample, it produces a Laue pattern from each subgrain that it intercepts. The quantitative information on the number and kind of unpaired random dislocation can be recovered from the “streaking” in the Laue image.

In this paper, ND and PXM measurements were used to study the effects of overload during cyclic loading. The elastic lattice strain profile and dislocation arrangement near the crack tip from the shift, and broadening of the diffraction peaks as well as streaking of the Laue spots associated with the large plastic deformation.

5.2 Experimental Details

The compact-tension (CT) specimen, with a width of 50.8 mm, and a thickness of 6.35 mm, were fatigue deformed with a frequency of 10 Hz and a constant load

ratio, $R = 0.1$, where $R = P_{min.}/P_{max.}$, $P_{min.}$ and $P_{max.}$ are the applied minimum (667 N) and maximum loads (6,667 N), respectively. As the crack length reach 22.8 mm, an overload of 8,889 N was applied. Neutron diffraction measurements were performed on the Spectrometer for Materials Research at Temperature and Stress (SMARTS) [58] at the Los Alamos Neutron Science Center (LANSCE). The samples were aligned such that the tensile load axis was 45° from the incident beam and, thus, the diffraction vectors associated with the two detector banks were parallel to the in-plane (IP) (parallel to the loading direction) and through-thickness (TT) directions of the sample. The incident neutron beam was defined by 2 mm and 1 mm (along the crack length) slits, and the diffracted beams were collimated using 2 mm radial collimators, resulting in a 4 mm^3 gauge volume. Thirty-two diffraction patterns were recorded along the crack length at a given applied load. The inter-atomic spacings were determined by the Rietveld refinement [59] of the diffraction patterns. Subsequently, the lattice strains (ϵ) were calculated from

$$\epsilon = \frac{a - a_0}{a_0} \quad (5.1)$$

where a_0 , the stress-free lattice parameter, was measured away from the crack tip. PXM was performed with a focused polychromatic synchrotron beam, which intercepts the sample surface in an area of $0.5 \text{ }\mu\text{m}$ in diameter at a 45° incidence. As the polychromatic beam penetrates the sample, the x-ray produces a Laue pattern in each subgrain that it intercepts. The penetration length depends on the atomic number of the material, and for steel samples, it is about $30 \text{ }\mu\text{m}$. The overlapping

Laue patterns are detected by an x-ray sensitive Charged Coupled Device (CCD) positioned at 90° to the incident beam [66].

5.3 Results and Discussions

First, the lattice-strain and full-width at half-maximum (FWHM) results from neutron measurements are presented. The lattice-strain components measured in both IP and TT directions under in-situ tensile loading are plotted in Figure 5.1. In IP direction, upon unloading to P_{min} of 667 N from the overloading to 8,889 N, compressive lattice strains of up to -600×10^{-6} (or $\mu\varepsilon$) were observed within 4 mm in front of the crack tip. Tensile strains ranging from 5 mm to 17 mm in front of the crack tip were observed, which balances the compressive strains at other locations. Due to the Poisson's effect, the sign of the strain in IP and TT directions are opposite, which is observed experimentally (Figure 5.1). The observed strain distribution: the compressive strain in the IP direction and tensile strains in the TT direction in front of the crack tip strongly supports for the crack closure. The compressive strain along the IP direction could, subsequently, retard the crack propagation, decrease the crack-growth rate, and increase the fatigue life [63].

Crack closure is induced by the plastic deformation caused by the overload. The severity of the plastic deformation can be estimated by observing the diffraction-peak width changes. The dislocation density can be calculated from the peak-width change. The broadening can be expressed by

$$2\delta q_j \propto C_j b \sqrt{n \tan \theta} \quad (5.2)$$

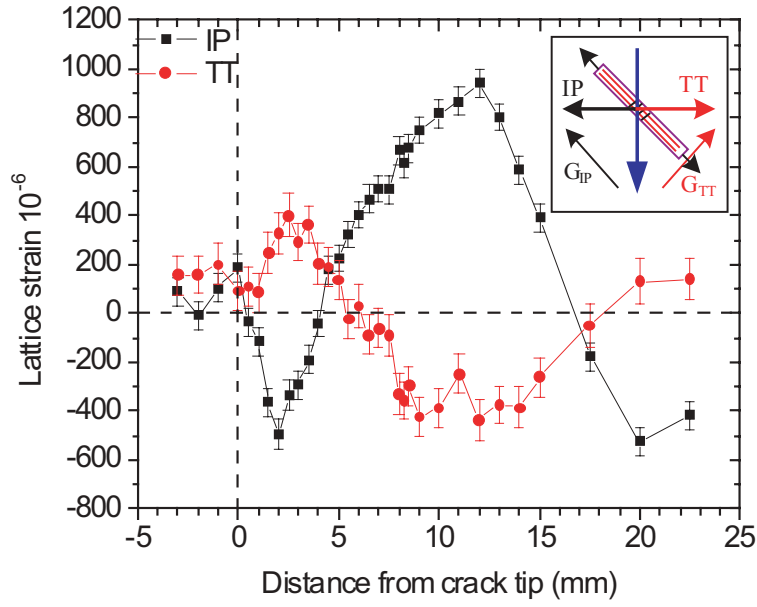


Figure 5.1: In-plane and through-thickness lattice-strain profiles in front of the crack tip at applied loads of 667 N after the overload. Insert shows distinct orientation of diffraction vectors G_{TT} and for G_{IP} measurements.

where θ is the exact Bragg angle of the reflection, b is the Burgers vector, $2\delta q_j$ is the half width of the intensity distribution along the diffraction vector (G_{IP} and G_{TT} for IP and TT directions, respectively), n is the total density of randomly distributed dislocations, and C_j is the orientation factor [67,68] corresponding to the (hkl) reflection. The linear dependence of FWHM on the $\sqrt{\tan \theta}$ predicted by the Equation 5.2 was observed experimentally for the IP and TT directions, as shown in Figures 5.2 and 5.3, respectively. The dislocation density, n , relative to the reference point (which was measured far away from the crack tip) can be calculated from the peak width change.

$$n = \frac{(FWHM - FWHM_0)^2 8\pi(1 - \nu)^2 144}{b^2 g^2 [29 - 76\nu + 72\nu^2 - (1 + 4\nu - 24^2)]} \quad (5.3)$$

FWHM is the full width half maximum of the diffraction peaks. $FWHM_0$ is the full width half maximum at the reference point, ν is the Poisson's ratio, b is the Burger's vector, and g is the diffraction vector. The dislocation density in IP as a function of the distance from crack tip is plotted as the insert in Figure 5.2. The dislocation density is higher in front of the crack tip than far away from the crack tip. The maximum dislocation density reached 3.5×10^{10} . We believe the high dislocation density is induced by the overload effect. A large plastic deformation occurred in front of the crack tip. We also calculated the dislocation density along the TT direction, which is shown as the inserts in Figure 5.3. The average dislocation density along the TT direction is about 3 times larger than that along the IP direction. And the dislocation density as a function of the distance from the crack tip slightly changed. The results from IP and TT directions seem to be distinct. The reason is that not all the slip

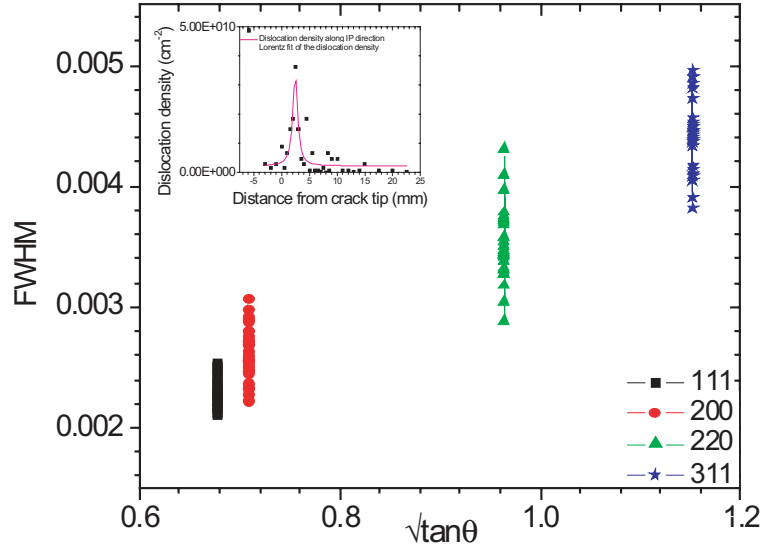


Figure 5.2: Demonstrate the dependence of FWHM on $\sqrt{\tan \theta}$ for four types of $\langle hkl \rangle$ along IP direction. Inserts the dislocation density as a function of distance from crack tip along IP direction.

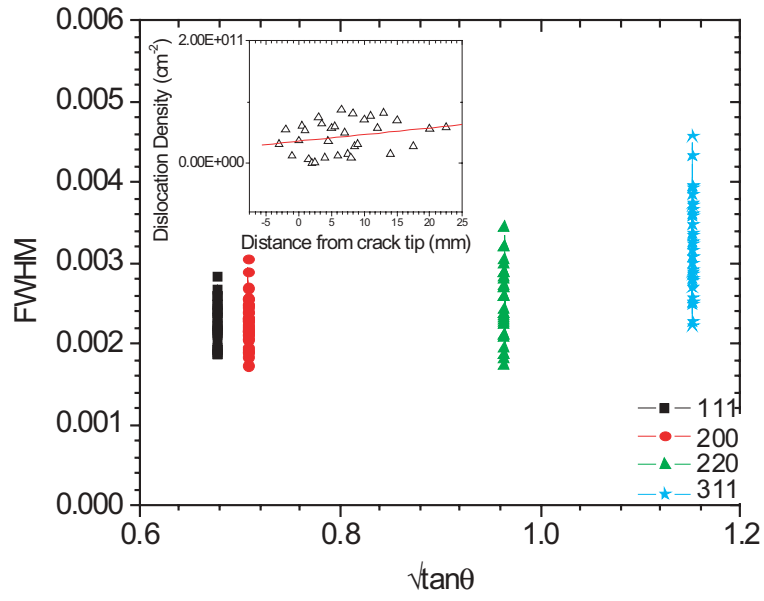


Figure 5.3: Demonstrate the dependence of FWHM on $\sqrt{\tan \theta}$ for four types of $\langle hkl \rangle$ along the TT direction. Inserts the dislocation density as a function of distance from the crack tip along the TT direction.

systems are equally activated. The broadening of the diffraction peak is influenced by the orientation factor, which depends on the type of dislocation ensembles, the mutual orientation between dislocations, and diffraction vectors for the reflection, and the anisotropic elastic constants [67–69]. For example, dislocations parallel to the diffraction vector do not broaden the peak. At the same time, dislocations being perpendicular to the diffraction vector cause the maximal broadening. Around the crack tip, there was the large plastic deformation in all the slip systems. Diffraction vectors corresponding to IP and TT measurements are orthogonal to each other. For this reason, the orientation factor, C_j , in Equation 5.2 is different for IP and TT directions, resulting in the different dislocation-density values and behavior as a function of the distance from the crack tip.

Further study was performed by PXM, which provides the very local structural information within $0.5\ \mu\text{m}$ mapping around the crack tip. The length of the “streak” of the Laue pattern depends mainly of the number of geometrically necessary dislocations (GNDs) within the irradiated volume, causing local lattice rotations [66]. Figure 5.3 shows the Laue patterns obtained at different locations near the crack tip. The analysis of these Laue patterns demonstrates that the plastic-deformation results in the formation of the alternating regions with high and low GNDs densities. Some of the patterns consist of relatively sharp spots corresponding to grains with a relatively low GNDs density. At the same time, some of the patterns demonstrate intensive “streaking” of all Laue spots. Such streaking is indicative of the strong lattice curvature in these grains. The length of streaks characterizes the total lattice

rotation in the probed region of the grain. The polychromatic microdiffraction shows that some grains are deformed severely, and some are not.

5.4 Summary

1. The elastic strain profiles in IP and TT directions clearly show the overload-induced plasticity that is responsible for the crack-closure phenomena and crack-growth-rate retardation.
2. Dislocation densities are high in front of the crack tip.
3. The inhomogeneity of the plastic deformation at the scale of ~ 1 mm was observed by neutron diffraction.
4. The inhomogeneous plastic deformations at meso-scale (~ 1 μm) were observed with the synchrotron PXM. Neighbor grains were deformed differently, depending on their crystallographic orientation, size, and surroundings.

Chapter 6

Conclusions

During fatigue crack growth tests, after the overload, the crack-growth retardation period was observed. During the retardation period, the crack-growth rate significantly decreased, which is because of the plastic deformation induced by the overload. The retardation period is also confirmed by the measurement of the striation-spacing changes before and after the overload. Crack closure at the retardation period due to the plastic deformation was also investigated by neutron diffraction.

Elastic lattice strains and the associated plastic zone around the crack tip in Type 316 LN steel were investigated using in-situ neutron-diffraction technique. The elastic strains in the in-plane direction (parallel to the loading direction) were measured as a function of the distance from the crack tip at different applied loads ranging from 667 N to 8,889 N. The results showed the evolution of lattice-strain profiles associated with the in-situ applied loading as well as the residual-strain profile generated by an overload. Lattice-strain profiles from the through-thickness (TT) and in-plan

(IP)directions have the opposite signs, which agree with the Poisson's effect. As the in-situ load increased from 667 N to 6,667 N, the lattice strain maximum in the IP direction increased. As the load decreased from 8,889 N to 667 N, the lattice strain decreased. After the overload, a significant compressive strain of $-600 \mu\epsilon$ (or $\times 10^{-6}$) was observed in front of the crack tip, which is related to the crack-closure phenomenon and retardation of crack growth. Also from the lattice strain profiles, the large compressive strain at LP7 was observed after the overload, which indicates the crack closure, because the compressive stress retards the crack growth.

The plastic deformation is irreversible. The peak-width change after the overload is due to the plastic deformation. The plastic zone size in front of the crack tip was also estimated using the peak-broadening analysis, which agrees well with the theoretically predicted plastic-zone size of 7.9 mm.

The dislocation density around the crack tip is calculated. The averaged dislocation density is about 10^{10} cm^{-2} . The dislocation density is high in front of the crack tip from the IP direction calculation. The dislocation density changes slightly as a function of the distance from the crack tip in the TT direction calculations. The different neutron diffraction results from IP and TT directions indicates the inhomogeneity of the plastic deformation at the scale of $\sim 1 \text{ mm}$.

The inhomogeneous plastic deformations at meso-scale ($\sim 1 \mu\text{m}$) were observed with the synchrotron PXM, which provides the very local structural information within a $0.5 \mu\text{m}$ mapping around the crack tip. The length of the "streak" of the Laue pattern depends mainly of the number of geometrically-necessary dislocations (GNDs) within the irradiated volume, causing local lattice rotations. The Laue

patterns obtained at different locations near the crack tip. The analysis of these Laue patterns demonstrates that the plastic-deformation results in the formation of the alternating regions with high and low GNDs densities. Some of the patterns consist of relatively sharp spots corresponding to grains with a relatively low GNDs density. At the same time, some of the patterns demonstrate an intensive “streaking” of all Laue spots. Such streaking is indicative of the strong lattice curvature in these grains. Neighbor grains were deformed differently depending on their crystallographic orientation, size, and surroundings.

After the retardation period was observed from fatigue experiments. 7 specimens (SP1, SP2, SP3, SP4, SP5, SP6, and SP7) (in Figure 2.2) through the retardation period were prepared. Neutron-diffraction experiments were designed to investigate what caused the increase of the crack growth rate, and what cause the decrease of the crack-growth rate. Strain profiles along the transverse direction are reported as a function of the distance from the overload point. Note that the distance from the overload point is the same as the distance from the crack-tip position of the SP2. The profile of SP1 shows small compressive (negative) strains near the crack tip. SP2 (the overloaded specimen) exhibits large compressive strains (maximum of about -950×10^{-6} or $\mu\epsilon$) around $0 \sim 7$ mm in front of the crack tip due to the overload. SP3 shows a similar profile as SP2. However, SP6 exhibits a significantly different strain profile with a smaller magnitude (about - 600 $\mu\epsilon$) and a wider double-peak shape.

The evolution of strain profiles described above could explain the retardation phenomenon with a schematic of the perturbed plastic zone represented. SP1, without

the overload, showed small compressive strains generated by the fatigue deformation. The large compressive strains observed after the overload are associated with the overload-induced plastic zone. The strain profile of SP3 is quite similar to that of SP2 because the increase in the crack length during the subsequent fatigue is too small to go through the plastic zone already created by the overload. SP6, with a sufficiently longer crack length, there are two peaks in the compressive strain profile. The first peak is due to the large plastic zone created by the overload, which seems to have relaxed, and the second peak is caused by the subsequent fatigue cycles. This trend indicates that the fatigue-induced plastic zone had grown out of the large plastic zone caused by the overload. In summary, the overload produced the large plastic deformation (SP2), which, in turn, led to a dramatic decrease in the crack-growth rates (SP3). As the crack grew longer and out of the overload-perturbed region during the subsequent fatigue deformation (SP6), the crack-growth rates recovered to the original slope. These results greatly enhance our understanding of the overload effect during cyclic loading.

Chapter 7

Future Work

First, in August 2005 using the SMARTS instrument, the fatigue-crack-tip deformation behavior after the overload was measured during tensile loading and unloading cycles using neutron diffraction. Total 13 in-situ loadings were applied in order to investigate the beginning and the end of the crack-closure phenomenon. The data need to be fully analyzed.

Second, in December 2005 using the HFIR instrument, the experiment studying the strain distribution mapping on three CT specimens, which were fatigued to various stages through the retardation period after the overload, has been done. The strain-mapping contours have been finished. Further carefully studies on the strain contours should be conducted.

Third, finite-element modeling need to be performed. The models for simulating different sets of neutron experiments have to be developed. Meshes should be created

first. The model for explaining the mechanisms of crack growth under cyclic loading, based on an elastic and perfectly plastic model, need to be developed.

Fourth, publish papers on all sets of data.

Bibliography

Bibliography

- [1] A. Saxena. *Nonlinear Fracture Mechanics for Engineers*. CRC PRESS New York, 1998.
- [2] G. Degallaix S. Degallaix and J. Foct. In *ASTM STP 942*, pages 798–811. ASTM, Philadelphia, PA, 1988.
- [3] J. B. Vogt, J. Foct, C. Regnard, G. Robert, and J. Dhers. *Metallurgical Transactions A*, 22A:2385–2392, October 1991.
- [4] C. E. Inglis. *Transactions of the institute of Naval Architects*, 55:219–241, 1913.
- [5] A. A. Griffith. *Philosophical Transactions*, A221:163–198, 1920.
- [6] H. M. Westergaard. *Journal of Applied Mechnics*, 6:49–53, 1939.
- [7] G. R. Irwin. In *Americn Society for Metals*, pages 147–166, Cleveland, 1948.
- [8] E. Orowan. *Reports on Progress in Physics*, X11:185–232, 1948.
- [9] M. L. Williams. *Journal of applied mechanics*. 24:361–364, 1957.

- [10] P. C. Paris, M. P. Gomez, and W. P. Anderson. *The Trend in Engineering*, 13:9–14, 1961.
- [11] J. R. Rice. *Journal of Applied Mechanics*, 35:379–386, 1968.
- [12] J. A. Begley and J. D. Landes. volume 1. the Crack Propagation symposium, 1961.
- [13] J. A. Begley and J. D. Landes. *American Society for Testing and Materials*, pages 1–23, 1972.
- [14] A. Saxena. In *In fracture Mechanics: ASTM STP 905*, volume Seventeenth, pages 185–201. American Society for Testing the Materials, 1986.
- [15] J. Schijve. chapter Advances in Aeronautical Sciences 3. Oxford: Pergamon Press, 1961.
- [16] P. C. Paris and F. Erdogan. *Journal of Basic Engineering*, 85:528–534, 1960.
- [17] C. Laird. *Fatigue Crack Propagation, ASTM STP 415*, pages 131–180, 1967.
- [18] L. Bairstow. *Trans. R. Soc. London*, A210:35–55, 1910.
- [19] F. Ellyin. *Fatigue damage, crack growth and life prediction*. Chapman & Hall, 1997.
- [20] H. Wang Y. H. He R. L. McDaniels L. Jiang B. Yang L. J. Chen, P. K. Liaw and D. L. Klarstrom. *Mechanics of Materials*, 36(1-2):85–98, 2004.

- [21] P. K. Liaw R. L. McDaniels C. R. Brooks R. R. Seeley Y. H. He, L. J. Chen and D. L. Klarstrom. *International Journal of Fatigue*, 24(9):931–942, 2002.
- [22] C. R. Brooks B. Somieski and D. L. Klarstrom L. Jiang, P. K. Liaw. *Materials Science and Engineering A-Structural Materials Properties Microstructure and Processing*, 313(1-2):153–159, 2001.
- [23] D. Broek. *The practical use of fracture mechanics*. Kluwer Academic Publishers, 1989.
- [24] J. Schijve, D. Rijk, and J. Broek. Technical Report NLR-TN M. Amsterdam, National Aero- and Astronautical Research Institute, 1961.
- [25] R. C. Rice and R. I. Stevens. In *ASTM STP*, volume Seventeenth of 536, page 95. 1973.
- [26] R. P. Wei, T. T. Shih, and J. H. Fitzgerald. Technical Report CR-2239, Fitzgerald JH, 1973.
- [27] L. G. Vargas and R. I. Stephens. In *3rd Int. Conference on Fracture*. Munich, Germany, 1973.
- [28] R. Roberts V. W. Trebules Jr and R. W. Hertzberg. In *ASTM STP*, volume 536. 1973.
- [29] W. J. Mills and R. W. Hertzberg. *Eng Fract Mech*, 8:657, 1976.
- [30] M. Lang and G. Marci. *Fatigue Fract Eng Mater Struct*, 22:257, 1999.

- [31] K. Sadananda, A. K. Vasudevan, R. L. Holtz, and E. U. Lee. *International Journal of Fatigue*, page S233, 1999.
- [32] D. Taylor, I. A. N. Staniaszek, and J. F. Knott. *International Journal of Fatigue*, 12:397–402, 1990.
- [33] S. Suresh. *Eng Fract Mech*, 18:577–93, 1983.
- [34] N. A. Fleck. *Eng Fract Mech*, 16:52–65, 1988.
- [35] K. Sadananda. *Advances in fracture research*. New York: Elsevier, 1986.
- [36] R. J. Bucci, A. B. Thakker, T. H. Sanders, R. R. Sawtell, and J. T. Staley. In *ASTM STP*, volume ASTM 714, pages 41–78. 1980.
- [37] J. F. Knott and A. C. Pickard. *Met Sci*, 11:399–404, 1977.
- [38] J. Schijve. *Advances in Aeronautical Sciences*, 3:387–408, 1962.
- [39] D. R. Donald and W. E. Anderson. *Proceedings of the Crack Propagation Symposium Cranfield, England*, 2:399–404, 1961.
- [40] R. D. Brown and J. Weertman. *Eng Fract Mech 1978*, 10:867–78.
- [41] J. Schijve. *Engineering advisory Materials Service Ltd*, pages 1685–721, 1987.
- [42] A. P. Renolds. *Fat Fract Eng Mat Struct*, 15:551–62, 1992.
- [43] E. P. Philips. *NASA Technical Memorandum 101601*. Langley Research Center, Hampton. VA, 1989.

- [44] C. M. Ward-Close and R. O. Ritchie. In *ASTM STP 982*, pages 529–35. ASTM, 1988.
- [45] S. Suresh, G. F. Zamiski, and R. O. Ritchie. *Metallurgical Transactions A*, 12A:1435–1443, 1983.
- [46] Z. Mei and W. Morris. *Metallurgical Transactions A*, 21A:3137–3152, 1990.
- [47] J. K. J. K. Donald and G. M. Connelly. *Fatigue Damage of Structural Materials II*. 1998 September.
- [48] J. D. Almer, J. B. Cohen, and R. A. Winholtz. The effects of residual macrostresses and microstresses on fatigue crack propagation. *Metallurgical and Materials Transactions A- Physical Metallurgy and Materials Science*, 29(8):2127–2136, 1998.
- [49] J. Smith, M. N. Bassim, C. D. Liu, and T. M. Holden. Measurement of crack-tip strains using neutron-diffraction. *Engineering Fracture Mechanics*, 52(5):843–851, 1995.
- [50] H. Tada, P. C. Paris, and G. R. Irwin. *The stress analysis of cracks handbook*. Paris Productions, Inc., St Louis, MO, 2nd edition, 1985.
- [51] R. S. Williams P. K. Liaw, T. R. Leax and M. G. Peck. Influence of oxide-induced crack closure on near-threshold fatigue crack-growth behavior. *Acta, Metallurgica*, 30(12):2071–2078, 1982.

- [52] R. S. Williams P. K. Liaw, T. R. Leax and M. G. Peck. Near-threshold fatigue crack growth-behavior in copper. *Metallurgical Transactions A-Physical Metallurgy and Materials Science*, 13(9):1607–1618, 1982.
- [53] V. P. Swaminathan P. K. Liaw, A. Saxena and T. T. Shin. Effects of load ratio and temperature on the near-threshold fatigue crack-propagation behavior in a crmov steel. *Metallurgical Transactions A-Physical Metallurgy and Materials Science*, 14(8):1631–1640, 1983.
- [54] W. A. Logsdon, P. K. Liaw, and M. A. Burke. Fracture-behavior of 63sn-37pb solder. *Engineering Fracture Mechanics*, 36(2):183–218, 1990.
- [55] W. A. Logsdon and P. K. Liaw. Tensile, fracture-toughness and fatigue crack-growth rate properties of silicon-carbide whisker and particulate reinforced aluminum metal matrix composites. *Engineering Fracture Mechanics*, 24(5):737–751, 1986.
- [56] T. R. Leax P. K. Liaw and W. A. Logsdon. Near-threshold fatigue crack-growth behavior in metals. *Acta, Metallurgica*, 31(10):1581–1587, 1983.
- [57] L. J. Chen, P. K. Liaw, R. L. McDaniels, and D. L. Klarstrom. The low-cycle fatigue and fatigue-crack-growth behavior of haynes (r) hr-120 alloy. *Metallurgical and Materials Transactions A-Physical Metallurgy and Materials Science*, 34A(7):1451–1460, 2003.

- [58] D. C. Dunand M. A. M. Bourke and E. Ustundag. Smartsa spectrometer for strain measurement in engineering materials. *Applied Physics A-Materials Science&Processing*, 74:S1707–S1709, 2002.
- [59] H. M. Rietveld. A profile refinement method for nuclear and magnetic structures. *Journal of Applied Crystallography*, 2:65–71, 1969.
- [60] G. A. Webster H. J. Macgillivray P. J. Webster D. J. Smith, R. H. Leggatt and G. Mill. Neutron-diffraction measurements of residual-stress and plastic-deformation in an aluminum-alloy weld. *Journal of Strain Analysis for Engineering Design*, 23(4):201–211, 1988.
- [61] C. Y. Hou and J. J. Charng. Estimation of plasticity-induced crack closure in a pre-existing plastic zone. *International Journal of Fatigue*, 18(7):463–474, 1996.
- [62] L. Clapham, R. Sabet-Sharghi, T. Holden, and D. L. Atherton. A neutron diffraction study of local stress concentrations surrounding defects in pipeline steel. *Physica B*, 241:1240–1243, 1997.
- [63] Y. N. Sun, H. Choo, P. K. Liaw, Y. L. Lu, B. Yang, D. W. Brown, and M. A. M. Bourke. Neutron diffraction studies on lattice strain evolution around a crack-tip during tensile loading and unloading cycles. *Scripta Materialia*, 53(8):971–975, 2005.
- [64] H. Tian. *Effects of Environment and Frequency on the Fatigue Behavior of the Spallation Neutron Source (SNS) Target Container Material C 316 LN Stainless Steel*. PhD thesis.

- [65] R. Roberts V. W. Trebules Jr. and R. W. Hertzberg. Effect of multiple overloads on fatigue crack propagation in 2024-t3 aluminum alloy. In *ASTM Special Technical Publication 536*, pages 115–116.
- [66] J. D. Budai, W. G. Yang, N. Tamura, J. S. Chung, J. Z. Tischler, B. C. Larson, G. E. Ice, C. Park, and D. P. Norton. X-ray microdiffraction study of growth modes and crystallographic tilts in oxide films on metal substrates. *Nature Materials*, 2(7):487–492, 2003.
- [67] M. A. Krivoglaz. *Theory of X-ray and Thermal Neutron Scattering by Real Crystals*. Plenum Press, New York.
- [68] M. A. Krivoglaz. *X-ray and Neutron Diffraction in Non-ideal Crystals*. Springer, Berlin.
- [69] R. Barabash. X-ray and neutron scattering by different dislocation ensembles. *Materials Science and Engineering A-Structural Materials Properties of Microstructure and Processing*, 309:49–54, 2001.

Vita

Yinan Sun (Julia) was born in Shenyang, Liaoning, China, on March, 1979. She attended Shenyang Univeristy, where she received a Bachelor of Science degree from the Civil Engineering department in 2001. She entered the Material Science and Engineering graduate program in spring 2003 and earned her Master of Science degree in 2006.