Stress Gradient Induced Strain Localization in Metals: High Resolution Strain Cross Sectioning via Synchrotron X-Ray Radiation

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Abstract

Strain localization is a phenomenon common to many systems described by continuum mechanics in the strong material deformation (or growth) régime. Variations of this complex phenomenon lead to interesting nonlinear effects in fields as diverse as materials science, geology, biological structures, and general relativity. Here, high energy x-ray diffraction on small length scales is used to characterize, in unprecedented detail, strain localization and large plastic deformation in metallic systems, induced by both compressive and tensile applied stresses. The examples of compressive stress induced strain localization are are shot peened materials where ballistic impact is used to produce plastic flow and yield a residual surface compressive layer. This peened surface layer greatly toughens materials for a wide range of aerospace, automotive, and medical applications – from turbine engines to dental picks. Tensile stress induced strain localization example is that of the plastically deformed zone left in the wake of a propagating the crack tip. In both the tensile and compressive nonlinear plastic flow cases the details of the intrinsically anisotropic strain/stress fields are cross-sectioned in small steps through the localization zone.

Since antiquity it has been empirically recognized that surface hardness/durability could be enhanced by impact-cold-working [A] (e.g. hammering) by which hardening and a near surface compressive stress are produced. Of these processes, shot peening has found the most pervasive use. Shot peening impacts the treated surface repeatedly with small, hard shot – cold working the surface, inducing compressive stresses, and greatly inhibiting fatigue cracking. The use of shot peening to systematically inhibit surface initiated cracking dates back to the early 1900’s [A,A1,B,B1?].
Fatigue cracks, of course, occur as the result of localized tensile stresses which develop at crack tips [C], while the shot peening produces localized compressive stresses which reduce or cancel out the crack tip driving stresses. Both phenomena, while in competition, have in common that they are the result of localized plastic deformations; i.e., localized residual strain. Understanding of these phenomena, both crucial in materials lifecycle engineering have suffered, heretofore, from the lack of direct, nondestructive experimental probes capable of characterizing the local strain field gradients on the germane short lengths. In this paper we describe a recently developed high energy synchrotron x-ray scattering method capable of precisely such a detailed profiling. Specifically we address the local, anisotropic strain fields in metallic specimens driven to localized severe plastic deformation under compression-yielding induced by shot peening; and tension-yielding induced by the previous passage of a fatigue crack tip.

It should be recognized that strain-localization continuum mechanics problems have diverse manifestations in fields well beyond materials science, such as in general relativity, biology, and geology. The strongly nonlinear deformation régime is typically of current interest in and all of these areas. For example the interesting biological problem of mechanical buckling, or folding, at the edge of leaves or flowers has been addressed in several of articles [D,E,F,G]. In these works leaf-edge buckling was compared to the mechanical elasto-plastic buckling near a tear-edge of a polyethylene sheet torn a constant rate. In both cases an increasing compressive stress gradient arises approaching the edge, caused in the former by new material growth at the leaf-edge and in the latter by stresses left behind along the tear-edge in the wake of the plastic deformation near the propagating tear/crack-tip. These effects are, like the ones considered in this paper, examples of strain localization with a rapid gradient in the strain/stress approaching a boundary. In these cases however the thinness of the medium dictates a buckling instability to relieve the stress (minimize the elastic energy) with a decreasing sequence of buckling wavelengths approaching the edge reflecting the strong strain/stress localization. Interestingly Audoly et. al. [F,G] also pointed out the similarities of these examples of strain localization with local curvature to the mass induced curvature of space-time in general relativity.

Similarities between the present work the continuum mechanics approach in geology to treat deformations of the tectonic plates is worth highlighting. Two particular geological examples are the detailed continental plate plastic flow and uplift under compressive stress. As for example in Tibetan plateau [H], and the tension induced subsidence (or “necking of the lithosphere”) in rift valley structures [I]. Interestingly these strongly nonlinear compression and tension deformation examples bear a striking resemblance to the work presented here but with scale changes in time from $\sim 10^6$ years to $\sim 10^{-2}$ seconds and in space from $\sim 10^3$ kilometers to $10^{-1}$ millimeters.

It is rare in any of these strongly nonlinear continuum mechanics problems that one has the opportunity of experimentally performing a nondestructive local profiling of the underlying strain fields to decisively evaluate the complex theoretical modeling deployed to attack the problem. As should be clear below the ability to nondestructively cross-section the strain response in the direction of the steep gradient of the strain localization is now highly feasible in certain classes of problems in elasto-plastic deformation of solids. For specificity consider a metal subjected to an increasing uniaxial tensile stress [J]. Under small loads the deformation response is linear elongation along the load direction with a transverse contraction due to volume conservation (the Poisson effect). Larger stresses lead to a homogeneous breakdown (yielding) of the material with permanent plastic flow. At still larger stress, an inhomogeneous
concentration of the plastic deformation (strain localization) onsets with strong longitudinal elongation and transverse contraction (necking) changing the local specimen geometry. Eventually increased loading will lead to fracture but with the necking region still evidenced below the fracture surface. Under compressive loading a similar sequence occurs, however with the plastic deformation in the strain localization régime exhibiting a bulging (rather than necking) effect.

As we show herein, recently developed energy dispersive synchrotron x-ray diffraction techniques now allow routine, nondestructive measurements of highly localized elastic strains and strain gradients deep within the interior of metallic specimens, on feasible experimental time scales. We examine the two cases described above, residual strains due to shot peening, and internal strains associated with fatigue cracks.

Figure 1 3D perspective surface topological maps of surface height for shot peened and fatigue cracked materials.

- a) Surface map of the shot-peened-surface of the 1070 steel specimen. Note the craters from the ballistic impacts.
- b) Surface map of the edge of the 1070 steel peened specimen. Note the budging of the peened layer forma a lip (see Figure 2-inset for a clarifying schematic) at the edge due to the outward plastic flow transverse to the compression induced yielding.
- c) Surface map of the edge of the Ti-6Al-4V peened specimen. Again note the budging of the plastic peened

**Compressive Stress Material Breakdown: Shot Peening**

In shot peening the high impact velocity of the shot causes a transient compression well beyond the material yield limit at the impact site [A,A1,B,B1]. Simply speaking, by the work
energy theorem, the kinetic energy of the shot is partially converted into localized plastic flow near the material’s surface [B,B1]. The plastic flow occurs laterally away from the center of the impact crater, creating a layer of extremely high dislocation density on top of the underlying bulk material.

The surface height mapping results in Figure 1a (measured by white light microscopy methods) [K] for the peened surface of a 1070-steel specimen clearly shows the dense impact cratering typical of shot peening treatment. Figure 1b illustrates the lateral plastic flow (in the $x_1$ and $x_2$ directions as defined in Figure 1a) for the same 1070-steel peened by showing a surface height mapping of the edge of this specimen. Here the peened surface is viewed edge-on and the plasticity induced bulging at the edge, typical of shot peened surface edge [A1], is dramatically apparent. In Figure 1c a similar edge-on surface height map for a Ti-6Al-4V peened specimen is shown. The bulging morphology is the same in the Ti-alloy specimen, albeit with much smaller spatial scale due to much smaller peening intensity (relative to the 1070 specimen). The surface height profile in Figure 1d shows the “necking” or dimpling effect in the vicinity of a fatigue crack which will be discussed below. It is amusing to note that the deformation by bulging in Figures 1b and 1c and by subsidence-necking in Figure 1d bear resemblances to compression and tension induced deformation features in geology [H,I]. (Distinct boundary condition differences should be remembered before pushing this analogy too far however.)

After the impact the biaxial surface expansion leaves the near-surface plastic region in a state of compression with an interface to the underlying bulk material. The key parameters in this surface toughening are the magnitude and depth of the surface compressed plastic zone [A,B,B1]. Over the years, exhaustive empirical optimization of peening intensity for surface toughening in a multitude of components, varying from dental picks to turbine engine components, has been successfully pursued. Although destructive techniques [L,M,N] for profiling the underlying compressive stress/strain magnitude and depth have been used, direct nondestructive techniques to probe these key parameters on the required small spatial scale have been essentially unavailable. In recent years high intensity/energy synchrotron radiation has begun to be used for deeply penetrating X-ray diffraction strain profiling [N,O,P,Q]. Even in these techniques however, the elongation of the diffraction volume along the beam direction often limited the spatial resolution [N]. In the strain profiling studies discussed here the incident x-ray beam is carefully aligned, and the diffraction volume is allowed to elongate, parallel to the shot peened surface (see the insets of Figures 2 and 3). Perpendicular to peened surface, on the other hand, the beam collimation is kept very small (10-30 µm). With appropriate beam collimation slit choices this allows the profiling of both the in-plane and out-of-plane components of the plastic zone strain with extraordinary spatial resolution.

In the shot peening studies the plastic layer is in the $x_1$-$x_2$ plane and perpendicular to $x_3$ (see the insets of Figures 2 and 3). Figure 2 shows the variation of $\varepsilon_{33}$ across the entire depth of an ~5 mm thick Ti-6Al-4V shot peened specimen, along with an inset schematic of the x-ray scattering and sample geometry. Note the symmetric tensile (+) $\varepsilon_{33}$ strain on both peened surfaces along with a balancing compressive (-) strain in the intervening bulk material. It is important to note that the height of the incident x-ray beam (and spatial resolution) in the $x_3$ direction in these measurements was 20 µm.

In Figure 3 the $\varepsilon_{33}$ strain near the peened layer is reproduced along with the variation of the $\varepsilon_{11}$ strain. The very different detailed-shaping of the diffraction (gauge) volumes in these two measurements are indicated schematically in the insets of Figure 3. For the $\varepsilon_{33}$ measurement: the
incident beam collimation $d_i$ was 20 $\mu$m along $x_3$ (as noted above); whereas to increase the diffraction signal, the scattered beam collimation, $d_s$, was increased to $\sim 200\mu$m and the width along $x_2$ (parallel to the surface) was expanded to several hundred microns. For the $\varepsilon_{11}$ measurement: the sample was rotated by 90°; the collimation along $x_3$ was 30 $\mu$m the incident; $d_i$ was allowed to expand to $\sim 100$ $\mu$m (for increased signal) since it was now along the in-plane $x_1$ direction; and $d_s$ was increased to $\sim 300\mu$m also to increase the scattered signal.
Figure 2) Strain profile of $\varepsilon_{33}$ across the entire thickness of a Ti-6Al-4V double-sided shot peened specimen. The inset shows schematic of the x-ray scattering geometry along with the definition of the coordinate directions. Note the schematic representation of the lip which was optically profiled in Figure 1b and c.

Figure 3) The strain profiles of $\varepsilon_{33}$ and $\varepsilon_{11}$ in the vicinity of the peened surface layer and the underlying bulk material of the Ti-6Al-4V specimen. The insets illustrate the x-ray scattering geometries for the $\varepsilon_{33}$ (top) and $\varepsilon_{11}$ (bottom) measurements. Note the stress scale (lower right) uses $E=118$ GPa and $\nu=0.33$ (see text).
The compressive in plane $\varepsilon_{11}$ strain and tensile $\varepsilon_{33}$ strain results in Figures 2 and 3 are in accord with the biaxial in-plane stress expected in the plastically deformed shot peened layer [A,B,B1,N]. For biaxial symmetry $\varepsilon_{11}=\varepsilon_{22}$ and $\sigma_{1}=\sigma_{2}$ where $\sigma_{i}$, is the stress in the i’th direction. In this case

$$\sigma_{i} = \left[ \frac{E}{(1+\nu)(1-2\nu)} \right] \left\{ \varepsilon_{11} + \nu \varepsilon_{33} \right\}$$

and

$$\sigma_{3} = \left[ \frac{E2\nu}{(1+\nu)(1-2\nu)} \right] \frac{1}{2\nu} \left\{ (1-\nu) \varepsilon_{33} + \frac{\varepsilon_{11}}{\varepsilon_{33}} \right\}$$

(1a&b)

where $\nu$ is Poisson’s ratio and E is Young's modulus. For Ti-6Al-4V $\nu=0.33$ and one obtains

$$\sigma_{3} = \left[ \frac{3E}{2} \right] \varepsilon_{33} \left\{ 1 + \frac{\varepsilon_{11}}{\varepsilon_{33}} \right\}$$

(2)

Since the experimental results in figure 3 indicate that, to a very good approximation $\varepsilon_{11} \sim \varepsilon_{33}$ one has $\sigma_{3} \sim 0$. This is not at all unexpected in view of the free surface in the $x_{3}$ direction. Indeed in general for $\sigma_{3}=0$ leads to

$$\frac{\varepsilon_{11}}{\varepsilon_{33}} = -\frac{1-\nu}{2\nu}$$

and

$$\sigma_{1} = \varepsilon_{11} \left[ \frac{E}{(1-\nu)} \right]$$

(3a&b)

which reduces to the observed $\varepsilon_{11} \sim \varepsilon_{33}$ for $\nu=0.33$. Further using $E=118$ GPa for Ti-6Al-4V one obtains the in-plane stress scale $\sigma_{1}=175 \varepsilon_{11}$ [MPa] which is shown in the right and scale of Figure 3. It is important to note here that this stress calculation is as a first approximation since it ignores the differences between the crystallographic-direction-dependent and bulk elastic properties.

We will now consider a second illustration of the application of this x-ray strain mapping technique to a heavily shot peened (on just one surface) 1070 spring steel placket of ~4 mm thickness. Here, as in the case of a bi-metallic strip when heated, the stress differential between the two layers (the peened and bulk) introduces a bending moment $M$ and a curvature of the sample (see Figure 4a for an exaggerated schematic of this curvature.). Indeed the degree of such curvature in standard “Almen strips” is routinely used as an empirical measure to determine the peening intensity in industrial applications [A].

$\varepsilon_{11}$ strain measurements on this sample were performed by our group some time ago [N] in the geometry illustrated in Figure 4b and are shown in Figure 5. The compressive $\varepsilon_{11}$ strain in the plastically deformed peened layer and the linear spatial variation in the strain elastic response in the underlying material [N] are strikingly apparent in Figure 5. The thickness of the peened layer in this specimen was thick enough that the orientation of the diffraction volume, with its elongated direction along $x_{3}$, resulted in only modest rounding of the peened-layer/substrate interface.

Also shown in Figure 5 is a recent measurement of the $\varepsilon_{33}$ strain collected in the geometry shown in the inset of Figure 3. The very analogous behavior, but opposite sign, of the $\varepsilon_{33}$ strain, (relative to the $\varepsilon_{11}$ strain) is clear upon comparing the two curves in Figure 4. Indeed the linear spatial variation in the bending moment elastic response portions of the $\varepsilon_{33}$ and $\varepsilon_{11}$ curves indicate a specimen curvature radii in the 1700-1800 mm range consistent (with experimental uncertainties) with the three–point microscopic measurement of 1650 mm. It should be noted that the spatial resolution of the $\varepsilon_{33}$ strain measurement (along the $x_{3}$ direction) is 30 $\mu$m so that essentially no finite gauge volume rounding is present in the data.
As in the Ti6Al-4V specimen discussed above, theoretical expectations and the qualitative mirror-image-type of $\varepsilon_{11}$ versus $\varepsilon_{33}$ strain behavior for this 1070-steel specimen, warrant the assumption biaxial stress. Accordingly using equation (3b) with a Poisson’s ratio of $\nu=0.3$ and $E=200$ GPa the approximate equivalent in-plane $\sigma_1$ stress scale has been included on the right in Figure 5. The results on this peened steel specimen emphasize that both the compressive strain in the localized plastic surface region and the complex long-range elastic response of the underlying bulk material can be characterized in detail with the synchrotron x-ray technique used here.

**Tensile Stress Material Breakdown: Fatigue Crack Wake Effect**

The singular enhancement in the local tensile stress which occurs near a fatigue crack tip naturally leads to strain localization, strong plastic deformation, and localized fracture in a region around the tip [C]. The propagation of the crack tip then leaves behind it a deformed plastic
wake at the crack face which has been created by tensile fracture. The large tensile stress (along the y-direction, with the crack propagation in the x-direction) induces a y-elongation and a transverse x-z contraction “necking” deformation. To underscore this lateral necking plastic flow, surface height measurements (noted earlier) across a fatigue crack in a 4140-steel specimen are shown in Figure 1d. The plasticity induced necking upon approaching the crack face is clear from this nonlinear surface dimpling. Indeed varying loading will leave varying spatial extents of the concave surface dimple along the crack [O].

For these fatigue crack measurements the gauge volume was began tailored to the sample geometry in which the crack was aligned closely parallel to the x-direction. The \( \varepsilon_{yy} \) measurements had a rectangular x-y cross-section of 200 \( \mu \)m X 60 \( \mu \)m. The \( \varepsilon_{xx} \) measurements had an x-y cross-section of ~100 \( \mu \)m X 60 \( \mu \)m. The z-y gauge volume cross-section was an elongated regular parallelogram the maximal z extent being ~300 \( \mu \)m [N,O].

The schematic inset in Figure 6 illustrates the fatigue crack with the conventional x-axis parallel to the crack, y-axis perpendicular to the crack and coordinate origin at the crack tip. The variation of \( \varepsilon_{yy} \) (\( \varepsilon_{xx} \)) crossing the crack perpendicularly, along the y direction, (at x=-2mm behind the tip) for a fatigue cracked 4140 steel specimen are shown in Figure 6. We wish to draw attention to the region of strongly nonlinear plastic flow in the strain results. Specifically, the sharp negative (positive) peak, labeled 1 in Figure 6, in \( \varepsilon_{yy} \) (\( \varepsilon_{xx} \)) within a region of ± 0.15 mm of the center of the crack (y=0) should be noted. The results in Figure 6 clearly show that in this near-crack plastic wake region the strain anomaly is highly anisotropic with \( \varepsilon_{yy}/\varepsilon_{xx} \approx 0.7 \). It should be note that this phenomenon is morphologically typical of the “at-crack” behavior for all of the fatigued specimens study by our group [O].
Assuming in this case that $\sigma_y=0$ in the proximity the crack surface (at $y=0$) one finds:

$$\epsilon_{zz} = -\epsilon_{xx} - \left(1 - \frac{1}{2}\right)\epsilon_{yy}.$$ 

Using the experimental observation $\epsilon_{yy}/\epsilon_{xx} \sim -0.7$ and $\nu=0.3$ one finds $\epsilon_{zz}/\epsilon_{xx} \sim -0.63$, which is close to (but depressed from) $\epsilon_{xx}/\epsilon_{zz}$ biaxiality. Such a depression is not unexpected in view of the proximity of free surfaces at $z=\pm2$mm from this center-specimen measurement. Moreover, at the crack tip, where this wake-zone was created, there is a real physical difference between the strain along ($\epsilon_{xx}$) and transverse ($\epsilon_{zz}$) to the crack propagation direction. The stresses, within the that $\sigma_y=0$ assumption, are

$$\sigma_x = \frac{E}{(1+\nu)} \left\{ \epsilon_{xx} - \epsilon_{yy} \right\}; \quad \text{and} \quad \sigma_z = -\frac{E}{(1+\nu)} \left\{ \epsilon_{yy} + \nu \epsilon_{xx} \right\}.$$ 

Again inserting $\epsilon_{yy}/\epsilon_{xx} \sim -0.7$ with $\nu=0.3$ in these expressions one obtains $\sigma_x=1.31E \epsilon_{xx}$ and $\sigma_z=1.03E \epsilon_{xx}$. Using $E=200$ GPa the maximal stresses in the wake zone are $\sigma_x=287$ MPa and $\sigma_z=225$ MPa. Further work on this detail is in progress.

Thus these results evidence the tensile strain localization phenomenon in a narrow plastic zone immediately on both sides of a fatigue crack. In this plastic wake region the anisotropic large $\epsilon_{xx}$ contraction and smaller $\epsilon_{yy}$ expansion are indicate an anisotropic residual stress after tensile fracture. Longer-ranged oscillation below this plastic wake zone (labeled 2 and 3) appear to be the elastic response of the bulk due its interface coupling to plastic wake zone. These longer range or elastic effects are not central here and their discussion will be differed to a expanded paper.

**Concluding**

The results presented here present a clear and detailed experimental picture of strain-localization in the strong plastic deformation régime in metallic systems under both compression and tension. The important result is that it is clear that an experimental fine grained, nondestructive sectioning of the nonlinear strain gradients along a strain localization direction in real metallic systems is now routinely available. In particular the precise spatial extent, magnitude and anisotropies of the strains in the plasticity region, the interface to the bulk, and in the underlying bulk material can now be mapped in detail. This is great importance for the quantitatively addressing the myriad of parameters which create such strain localize zones and their modification and evolution under varying processing conditions, thermal annealing, or cyclic loading. It all of these real-life applications general expectations are insufficient for engineering design and real strain profile data is required for model construction and validation.

In the introduction we tried to motivate thinking of continuum mechanical phenomenon in a way which crosses fields but which also clearly identifies the myriad of effects specialized to differing problems. Examples of such specialized effects are the gravitational potential and Mantle-interface effects in planetary crust mechanics and the membrane thickness and buckling response in biological systems. With such a cross field spirit and the differing-problem caveats in mind it is intriguing to consider an analogy of the ballistic impact surface-toughening and upward plate-curvature, discussed here in shot-peened metals, to the heavily impacted cratered regions of the moon. Such heavily cratered lunar surface regions: are on average elevated in height (hence the “lunar highlands” appellation); rest upon thickened regions of the lunar crust;
and have the greatest antiquity of the lunar surface. Pursuit of the details of such a suggestion about the lunar crust is clearly beyond the expertise of the authors, however the mechanisms of high impact-site densities inducing surface uplift, and surface preserving toughening is worthy of bringing to the attention of experts in this field.

Experimental Methods

EDXRD The energy dispersive x-ray diffraction (EDXRD) measurements used for this work were performed at the Brookhaven National Synchrotron Light Source (NSLS) on the superconducting wiggler beam line X17-B1. The experimental setup, described in more detail elsewhere [N,O], involves “white beam” incident radiation with scattering at a fixed angle 20. The energies (E in keV) of the scattered Bragg peaks are given by 

\[ E = \frac{6.199}{d_{hlk} \sin(\theta)} \]

where \( d_{hlk} \) (in Angstroms) is the inter-atomic plane spacing associated with a specific \( \{hlk\} \) inter-atomic plane. In this work the variation from position to position within the sample of the inter-atomic spacing (d) of the a single, well isolated line has been used to determined the strain \( \varepsilon = \frac{(d - d_0)}{d_0} = \frac{(E_0 - E)}{E_0} \). Here E is the fitted energy of the Bragg line and the reference \( d_0 \) (and \( E_0 \)) represent the stress-free lattice spacing (Bragg line energy). In the case of the fatigue crack sample \( d_0 \) (and \( E_0 \)) were determined by the Bragg line position far from the crack [N,O]. In the case of the shot peened samples \( d_0 \) (and \( E_0 \)) were determined by the condition that the net stress across the entire line profile of the specimen be balanced (i.e. zero total force on the specimen). In the 4140 steel and recent (\( \varepsilon_{33} \)) 1070 steel measurements the bcc Fe \{321\} Bragg line was used. In the prior work on \( \varepsilon_{11} \) in 1070 steel, a weighted average of seven to ten bcc-Fe Bragg lines were used [N]. For the Ti-6Al-4V measurements the hcp \( \alpha\)-Ti \{110\} Bragg line was used.

For the strain experiments (\( \varepsilon_{33} \) for specificity) discussed herein, the orientation of the sample x3 direction deviates by the angle \( \theta \) (3°) from the scattering vector direction. Hence, strain measured in this case technically deviates from the true \( \varepsilon_{33} \) by \( \sin(\theta) \), which entails less than 2 % systematic error and is negligible compared to the other sources of errors in the experiment. Rotation of the sample by \( \theta \) would remove this systematic error however this entail some sacrifice of the high spatial resolution and scattered intensity achievable by aligning the incident beam parallel to the peened surface.

The incident and diffracted beams were each collimated by two 10 mm thick Ta slits thereby defining the small of the gauge volume. The gauge volume was positioned in the center of the specimens being studied. For the shot peened materials studies the profile directions were was from surface to surface through the interior of the specimen, and for the fatigue crack studies the profile perpendicularly crosses the crack in the center of the center (2mm from each surface) of the specimen [O].

The optical surface height profiling measurements were performed using a Zygo Inc. New View 5200 optical profiler. Discussion of white light optical surface profiling can be found in reference [K].

Materials The Ti-6Al-4V alloy specimen (where the 6Al and 4V are weight percent constituents) sample was a 23x23x5 mm\(^3\) placket peened on both large surface area sides for symmetry. The SAE 1070 carbon spring steel had (by definition) ~0.7 wt. % C and ~0.7 wt. % Mn and was 25x75x 4 mm\(^3\). The SAE 4140 Mo-steel specimen had nominal compositions of ~0.4 wt. % C, ~0.7 wt. % Mn, ~0.9 wt. % Cr, and ~0.2 wt. % Mo and was 25x75x 3 mm\(^3\). The normalized 4140-steel specimen was cut into a single edge notched geometry, as discussed previously [O] and fatigue with a constant amplitude loading with the maximum-minimum excursion in the tip stress intensity factor being \( K_{max} = 49.8 \) MPa m\(^{1/2}\) to \( K_{max} = 5 \) MPa m\(^{1/2}\).

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References


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