Evaluation of the overload effect on fatigue crack growth with the help of synchrotron XRD strain mapping

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Abstract

The overload retardation effect on fatigue crack growth rate (FCGR) in titanium alloy Ti-6Al-4V is studied. Synchrotron X-ray diffraction strain mapping of near-crack tip regions of pre-cracked fatigued samples is used to determine the effective stress intensity factors experienced by the crack tip. The effective stress intensity factor values are computed by finding the best match between the experimental strain maps and linear elastic fracture mechanics (LEFM) predictions. The dependence of the effective stress intensity factor, $K$, on the applied load is plotted, and an interpretation of the overload retardation effect is proposed. The present approach permits to reconcile the traditional LEFM fatigue crack propagation prediction and the experimental measurement of strain fields.

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1. Introduction

Fatigue (which is defined as the rupture of components under cyclic loading) is known to be one of the main phenomena leading to the possible failure of aerospace components. The need to develop and validate predictive modelling capabilities for fatigue crack propagation is therefore evident. Although fatigue has been the subject of thorough studies for many decades now (e.g. see [1–8]), it remains a highly topical research direction for the mechanical engineering community. Prediction of fatigue crack propagation is complicated by the fact that components in-service are not subjected to cyclic loading of constant amplitude and frequency (a better understood phenomenon now), but in fact experience a much more complex loading history involving changing amplitudes and frequencies, significant overloads and underloads, giving rise to crack growth rate variations [9]. It is generally well known that overloads produce a retardation effect on the fatigue crack growth rate, while underloads accelerate the crack propagation. Moreover, even if the major cycles associated with flight are of LCF (Low Cycle Fatigue, high amplitude and low frequency) type, in-service components are also subjected to additional HCF loading (High Cycle Fatigue, relatively high frequency and lower amplitude). The effect of dwell (load hold) at maximum load is also a phenomenon that has to be taken into consideration, even at low temperature.

Linear elastic fracture mechanics (LEFM) [10] has become a powerful tool to deal with fatigue crack propagation. One of the fundamentals of the LEFM is that, under the conditions of Mode I rupture, one variable is sufficient to describe the stress and strain field at the tip of a crack, and also to predict the onset of unstable crack propagation. This variable is the so-called stress intensity factor, $K$, and is the function of the crack length $a$, the component geometry and the applied loading. The
stress intensity factor provides a meaningful measure of crack resistance only if the plastic zone remains relatively small compared to the crack length and the component size. Based on the success of the stress intensity factor concept for static fracture, Paris [5] and Paris et al. [6] postulated that the rate of fatigue crack propagation per cycle should be determined by the stress intensity factor range $\Delta K$ defined as the difference between the value of $K$ at the maximum load, $K_{\text{max}}$, and the value at the minimum load, $K_{\text{min}}$. The crack advance per cycle, $da/dN$ (where $N$ is number of cycles) was postulated to depend on the parameter $\Delta K$ via a power law (see Eq. (1)).

$$\frac{da}{dn} = C\Delta K^m$$

(1)

In Eq. (1), $C$ and $m$ depend on the material, environment, frequency, temperature, stress ratio, etc. At first this proposal was judged to be too empirical and received sceptically by many experts in the field. However, in this form and with modifications (e.g. including explicit power law dependence on another parameter, $K_{\text{max}}$) Paris law has served the engineering design community well for decades. It now continues to be considered as the principal basis for fatigue crack growth rate (FCGR) predictions. The beauty of LEFM and the main reason of its success is that it allows to consider that fracture is only controlled by elastic phenomena occurring around the crack tip. Paris law, however, suffers from two major drawbacks: the lack of explicit dependence on the maximum load or $R$-ratio (that control the extent of plasticity within each cycle), and the lack of dependence on the loading history. In theory, a difference between the “applied” stress intensity factor range, $\Delta K$, and the effective $\Delta K_{\text{eff}}$ is, thus, likely to be observed. Ideally that should be reflected within the Paris law and this has been tried for decades. In the late 60’s, Elber [7] proposed to account for plastic deformation via the concept of crack closure. On the basis of simple closure analysis, he derived expressions for $\Delta K_{\text{eff}}$ (2) and re-formulated Paris law accordingly (3):

$$\Delta K_{\text{eff}} = (\sigma_{\text{op}} - \sigma_{\text{max}})\sqrt{\pi a} = \Delta \sigma_{\text{eff}} \sqrt{\pi a}$$

(2)

$$\frac{da}{dn} = C\Delta K_{\text{eff}}^m$$

(3)

In Eq. (2), $\sigma_{\text{op}}$ and $\sigma_{\text{max}}$ stand respectively for the opening stress and the maximum applied stress. Elber’s pioneering work has then been enhanced by Newman [11,12]. More recently, Pommier and Risbet [13], Lopez-Crespo and Pommier [14] and Korsunsky et al. [15] pursued the idea of correlating the crack tip plastic deformation with FCGR. Inelastic deformation at the crack tip was accounted for via a single crack tip parameter, either the blunting radius $\rho$ or the energy dissipation density.

Despite their demonstrated utility for FCGR prediction, crack closure-based approaches remain the subject of ongoing argument. In particular, it has been reported that different measurement techniques gives different closure magnitudes. For this reason attempts have also been made to reformulate Paris law without making recourse to crack closure concepts. Approaches of this kind can be found in the works of Walker [16], Dinda and Kujawski [17], Vasudevan et al. [18], and Noroozi et al. [19,20]. These studies rely on the empirical law for fatigue crack growth rate proposed by Walker [16] and later by Donald and Paris [21]:

$$\frac{da}{dn} = C(1 - R^\gamma K_{\text{max}})^\gamma = C(\Delta K_{\text{eff}}^\gamma)^\gamma$$

(4)

In Eq. (4), $R$ is the ratio of the minimum, $K_{\text{min}}$, to maximum, $K_{\text{max}}$, stress intensity factor. $p$ and $\gamma$ are fatigue crack growth law parameters.

The central idea in the so-called “two parameter” models is to relate the (macroscopic) remote $K$ (and the stress intensity factor range, for fatigue) to some parameters related to the (local) crack tip stress state (crack tip blunting in [13–15], energy dissipation in the vicinity of the crack tip in [15], crack tip yielding obtained from Ramberg–Osgood stress–strain relationship in [19,20]). Indeed, it is widely accepted by the scientific community that crack advance in metals is mainly determined by the damage state and the failure behavior of a highly localised volume immediately ahead of the crack tip. This small volume, called the process zone, is embedded within the crack tip plastic zone. In its turn, the zone of domination of the elastic crack tip stress field solution (K-field) surrounds the plastic zone. Traditionally, under small scale yielding conditions it was assumed that the elastic parameters of this K-field may control all the inelastic processes at its core. However, it is clear that this assumption fails once history dependence of crack growth is observed: the same applied load would produce different effects depending e.g. on whether the crack has undergone an overload. This discussion highlights the need for local (near crack tip) deformation and damage analysis and measurement.

At present, due to the lack of experimental data, fatigue crack models (whether making use of the crack closure concept or not) seem to contain the following paradox. On the one hand, these models are formulated using local descriptions of the crack tip and the parameters that need to be determined are local. However, the models are calibrated indirectly by trying to find the parameter values for the macroscopic fatigue crack growth criterion in the form of Paris law ($da/dN$ as a function of $\Delta K$). A standard experimental technique used to characterise crack growth consists in observing the crack extension in a compact tension specimen subjected to constant range remote cyclic load. The stress intensity factor $K$ is computed as the crack is propagating, for example, using expressions such as that proposed by Srawley [22]. This technique requires continuous measurement of the crack length and relating it with the number of cycles from initiation. Another relatively widespread alternative is the use of crack closure measurement methods. Here the idea is to measure the $\sigma_{\text{op}}$ of Eq. (2) and to use (3) to calculate $\Delta K_{\text{eff}}$. Various methods have been used by researchers to measure $\sigma_{\text{op}}$ experimentally [23].
An alternative approach to the problem would consist of measuring directly the relevant local physical parameters of crack tip deformation. If this route were to become open experimentally, it could be used to obtain datasets to calibrate the FCGR model versus local physical parameters. Synchrotron X-ray diffraction permits the mapping \cite{24} of elastic lattice strains in components made of (mono- or poly-)crystalline materials using the determination of the difference in lattice spacing between undeformed and deformed structures. This technique has been first used in the context of fatigue crack propagation by Allison \cite{25} and then by Ramos et al. \cite{26}. Recently, Croft et al. \cite{27–29} performed high-resolution mapping at the crack tips of 4140 steel CT specimens with different levels of overload using synchrotron X-ray diffraction. By collecting quantitative results under different \textit{in situ} loading conditions, they were able to emphasise that the nature of the overload retardation effect appears to be peak strain dominated (and therefore, by extension, $K_{\text{max}}$-dominated). In their most recent paper \cite{29}, the authors also confirm experimentally the existence of crack closure under plane strain conditions, by internal mapping of strain within the cross section of an unloaded specimen that has been overloaded previously. They demonstrated the existence of a compressive front within the cracked region ahead of the crack tip. Even more importantly, they performed \textit{in situ} loading measurements on overloaded specimens (both at maximum retardation effect and 50\% recovery of the initial crack growth rate) showing the existence of a critical force $F_c$, under which the strain range differs from the one expected using LEFM solutions due to the existence of crack closure. Above $F_c$, the specimen response is dominated by the tip (the crack is fully open) and the strain distribution is the one given by LEFM. $F_c$ increases with increasing retardation effect (FCGR reduction).

While Croft et al. \cite{27–29} results are expressed in terms of the local strains and strain ranges, most current models for FCGR predictions remain formulated in terms of the stress intensity factor, $K$, and the stress intensity factor range, $\Delta K$. In the present study we attempt to reconcile the local physical basis for FCGR prediction and the global formulaic expression for FCGR laws. We describe the \textit{in situ} loading study using synchrotron X-ray diffraction to perform strain mapping of Ti–6Al–4V CT specimens overloaded to 100\% ($F_{\text{OL}} = 2 \times F_{\text{max}}$) considered either at maximum retardation condition, or after growing the crack through the overload retardation stage. The method used to interpret the experimental data is similar to that presented by Korsunsky et al. \cite{15} for determining the residual stress intensity factor. The LEFM predictions for $K$-fields are matched to the measured strain maps for the specimens under different load levels, so that the stress intensity factors can be determined.

The \textit{in situ} study described in this paper allowed us to plot the dependence of the apparent stress intensity factor, $K$, on the applied load. The analysis reveals similar insight to that obtained by Croft et al. \cite{29} using the maximum strain at crack tip, in that the evidence is obtained of the crack tip remaining propped open even in the absence of external loading. The approach developed here possesses the advantage of furnishing experimental datasets that obviate the need to resolve the details of the local deformation (stress and strain) states at the crack, but nevertheless allow the residual stress effects to be explicitly quantified and taken into account. Finally, the effect of the overload can be expressed directly in terms of the stress intensity factor range and maximum values, i.e. in terms of the parameters used for the prediction and modelling using the extension of LEFM approaches.

2. Experimental investigation of crack growth in Ti–6Al–4V

In this study, further investigations were carried out on the set of specimens used in \cite{15}. As a reminder, CT specimens (Fig. 1a) were machined out of 5 mm-thick rolled plates of alloy Ti–6Al–4V. Fatigue crack propagation was investigated in
specimens machined so that the cracks propagate along the longitudinal (rolling) direction within the plate. The setup used is illustrated in Fig. 1b and is similar to that used by De Matos [30] for the study of crack closure in aluminium plates.

A sinusoidal load waveform with maximum load of 7.5 kN and minimum load of 0.75 kN was applied to the specimens by a servo-hydraulic loading device, equipped with a 15 kN load cell and driven by a programmable controller allowing flexible variation of amplitude and frequency of the load and counting the number of cycles applied. Samples were gripped in the loading device using two pins introduced in the two holes, separated by 33 mm (inter-centre distance), as shown in Fig. 1a. The crack length was measured optically with a digital camera attached to a Questar long range telescope. The camera used was a simple webcam allowing continuous filming at 30 frames per second at the resolution of 640 × 420 pixels. Since the telescope field of view was adjusted to about 0.65 mm, each pixel corresponded to about 1 μm. The actual image resolution, and, in particular, of displacement analysis by digital image correlation (DIC) is very dependent of the quality of the surface preparation, sample lighting conditions, image contrast, etc. The telescope was mounted on a transverse translation table with digital position readout. The digital camera was connected to a laptop so that live imaging of the area around the crack tip was possible. The crack length was measured by aligning a cross-hairs marker on screen with the crack tip, and using the positioning readout to determine the crack length to the accuracy of about 10 μm.

Samples were pre-cracked at the relatively high frequency (~10–12 Hz) and were then subjected to the desired constant amplitude signal with the frequency of 6 Hz. Tests were interrupted regularly to measure the crack length precisely by cross-hairs alignment with the crack tip under maximum load, when the crack is fully open. Telescope re-focusing was required in some cases. Crack length and the number of cycles from initiation were noted down and crack growth curves (Fig. 2a) and associated Paris’s diagrams plotted (Fig. 2b). K and AK on the Paris diagram were determined using Srawley’s equation [22].

Fig. 2a shows the crack growth curves for the sample series CT 1–CT 6. Retardation effect due to the application of a 100% overload is clearly distinguishable for specimens CT 1, CT 3, CT 5 and CT 6. The physical nature of this effect remains a subject of controversy. The insight into this phenomenon obtained in the present study is discussed in more detail below.

Fig. 2b shows the same data set plotted together with archival data for two samples from the AGARD report [31,32], on a bilogarithmic scale as the Paris fatigue crack growth diagram. Satisfactory agreement between the two data sets is observed. It is also found that a power law relationship of the form of Eq. (1) provides a good description of the dependence of the FCGR on the applied stress intensity factor range for non-overloaded specimens, as illustrated for sample CT 4. The retardation effect due to overload is conspicuous. The downward “spike” (sharp reduction of fatigue crack growth rates) is observed on overloaded samples is the result of additional localised plastic deformation experienced by the crack tip, and the associated local residual stresses that arise. It is interesting to note that once the overload-induced retardation effect has disappeared, the FCGR curve largely returns to the trend line observed for a crack of the same length in a sample without overload.

3. Evaluation of the residual stress intensity factor

Samples CT 1 and CT 6 have been chosen for further investigations using white beam synchrotron X-ray diffraction on the ID 15A beamline at the European Synchrotron Radiation Facility (ESRF, Grenoble, France). Both specimens have been subjected to constant amplitude cyclic loading and then to a 100% overload. However, while crack growth has been stopped in CT 6 when reaching maximum retardation of the FCGR, cyclic loading of CT 1 has been carried on for a while and full recovery of the FCGR has been reached (see Fig. 2a and b).

Synchrotron X-ray diffraction occupies a special place in the field of strain measurement since it can provide information relevant at all structural length scales, and is non-destructive [33]. Diffraction utilizes the simple relationship known as Bragg’s law, \( n \lambda = 2d \sin \theta \), where \( n \) is an integer, \( \lambda \) is the wavelength of the radiation, \( d \) is the interplanar distance between planes with Miller indices \((hkl)\), and \( \theta \) is the angle that the incident beam makes with the lattice planes (called the Bragg angle). Bragg’s law provides the relation between the spacing of atomic planes and the angle of incidence at which these planes produce the most intense reflection.

High energy white beam synchrotron X-ray diffraction was used to collect the experimental data. The use of a white beam in an energy-dispersive mode gives an access to a higher energy range than the monochromatic radiation and offers high-spatial resolution with short counting times [34] and reaches penetration depths up to several centimeters in most engineering materials [35]. Two strain components were measured simultaneously in the directions parallel \(\varepsilon_x\) and perpendicular \(\varepsilon_y\) to the crack growth direction in CT samples. Both detectors were mounted at fixed scattering angle of 2θ = 5°, and a beam spot size of 0.25 mm (x direction) × 0.25 mm (y direction) was used. Specimens CT 1 and CT 6 where mounted on the Instron servo-hydraulic loading rig available at beamline ID15A, ESRF. 2D strain mapping around crack tips was performed for both samples loaded in situ under three different load levels: 0 kN, 2 kN and 7 kN. The diffraction patterns collected were interpreted and analysed using the calibration procedures described by Liu et al. [36] and the software package General Structure Analysis System (GSAS) (see [37]). The lattice strains in the direction of the scattering vectors were calculated as:

\[
\varepsilon = \frac{(d_{\text{hkl}} - d_{\text{hkl}}^{\text{ref}})}{d_{\text{hkl}}^{\text{ref}}} \quad (5)
\]

In the absence of reference strain-free samples, the unstrained lattice parameters \(d_{\text{hkl}}^{\text{ref}}\) were obtained from patterns collected by placing the beam at the very corner of the plate, which was assumed to carry no or little residual macroscopic strain. Typical nominal strain error was only 50 × 10⁻⁶, corresponding to stress uncertainty of only about 6 MPa for Ti-6Al-4V.
Strain maps (in the crack opening direction) of areas surrounding crack tips of CT 1 and CT 6 under different load levels can be found on Figs. 3a–c and 4a–c. It is interesting to note that, similarly to what is shown in Croft et al. [28,29], a region of compressive strain is observed behind the crack tip. For both specimens it was evident that the lower the applied load was, the bigger was the compressive strain. When the applied load became sufficiently high, the compressive strains completely disappeared, as seen in Figs. 3c and 4c. Simultaneously, a highly tensioned zone was observed ahead of the crack tip. As expected, the higher the applied load was, the bigger were the maximum and average tension strains.

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In order to interpret these experimental strain maps quantitatively, the obtained crack tip strain distributions were analysed in terms of the corresponding value of the Mode I stress intensity factor, $K$ [38]. The method used to interpret these maps consists of seeking the best match (in terms of the sum-of-squares over all experimental points) between the theoretical $K$-field and the measurements [15]. The value of $K$ for which the LEFM prediction matches best the residual strain map of overloaded samples was declared to be the effective macroscopic stress intensity factor. The LEFM expression for the opening direction strain around a plane strain crack is given by [17]:

$$e(r, \theta) = \frac{(1 + v)}{E} \frac{K_I}{\sqrt{2\pi r}} \cos \left( \frac{\theta}{2} \right) \left[ 1 - 2v + \sin \left( \frac{\theta}{2} \right) \sin \left( \frac{3\theta}{2} \right) \right] + e_0 = K_I \cdot e^h + e_0$$

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Note that in the above expression a constant strain offset, $e_0$, is present. This is introduced in order to account for the possible error in the determination of the strain-free lattice parameter, $d_0$, since, to a good approximation, small error in $d_0$ results in a constant offset of the computed strain values. The only adjustable parameters in Eq. (6) are the stress intensity factor $K_I$ and the strain offset, $e_0$. These parameters were determined by seeking the values of $K_I$ and $e_0$ that minimize the sum-of-squares measure of mismatch between model and experiment, as follows.

Let the experimentally measured strain at position $(x_j, y_j)$ be denoted $e_j$ and its corresponding value calculated from Eq. (6) be denoted $e_j$. The sum-of-squares measure of the misfit between the predicted strain (given by Eq. (6)) and experiment is formed as follows:

$$ J = \sum_{j=1}^{N} (e_j - \tilde{e}_j)^2, \quad (7) $$
where N designates the number of experimental measurements in the entire map. The sought values of $K_I$ and $e_0$ are those that minimize $J$. This condition leads to a straightforward procedure for finding $K_I$ and $e_0$. If Eq. (6) is substituted in Eq. (7), and the derivative of $J$ is equated to zero, then the following system of equations is obtained:

$$\frac{\partial J}{\partial K_I} = 2 \sum_{j=1}^{N} \left( e_j - K_I \cdot e_{ij}^h - e_0 \right) \cdot e_{ij}^h = 0$$

$$\frac{\partial J}{\partial e_0} = 2 \sum_{j=1}^{N} \left( e_j - K_I \cdot e_{ij}^h \right) - N \cdot e_0 = 0$$

(8)

The system of equations presented in (8) is rearranged and written in matrix form, so that:

$$\begin{pmatrix} \sum_{j=1}^{N} e_{ij}^h \\ \sum_{j=1}^{N} e_{ij}^h \\ \sum_{j=1}^{N} e_{ij} \end{pmatrix} \begin{pmatrix} K_I \\ e_0 \end{pmatrix} = \begin{pmatrix} \sum_{j=1}^{N} e_{ij} \cdot e_{ij}^h \\ \sum_{j=1}^{N} e_{ij} \end{pmatrix}$$

(9)

$K_I$ and $e_0$ are obtained by solving the above system. Their substitution in Eq. (6) allows computing contour plots of LEFM best match to experimental strain maps of Figs. 3a–c and 4a–c. The results are presented in Figs. 3d–f and 4d–f, respectively. The $K_I$ values corresponding to Fig. 3d–f and 4d–f can be found in Table 1.

It is worth noticing from Figs. 3 and 4 that the LEFM model predictions show clear similarities with the experimental maps. The maximum values of strain and the strain distributions are captured well by the LEFM fitting. On the contrary, the zones of compressive strain behind the crack tip observed in the experimental maps under low applied loads are impossible to capture in this simple model. The good agreement between the computed LEFM distribution and the experimental maps suggests that the $K_I$ values obtained by the method presented above constitute a reasonable interpretation of the measurements. The physical meaning of the present results is discussed in more detail in the next section.

### 4. Discussion and analysis

The compressive zone behind the crack tip that is observed by Croft et al. [28,29] and our measurements is of principal importance for understanding the process of crack propagation in metallic alloys subjected to cyclic loading. The results appear to confirm the presence of crack closure [7]. A simple schematic diagram of the crack closure mechanism in this particular instance is illustrated in Fig. 5. When the material is loaded cyclically, a plastic zone that develops ahead of the crack tip is of a size that is determined by the stress intensity factor range, $\Delta K$ [10]. This plastic zone is contained within a larger region where the material behaves elastically: the $K$-dominant region. The classical LEFM equations are only valid in this annular $K$-field region (excluding the plastic zone). Outside this area the strain distribution is further dependent on the overall sample geometry, boundary conditions, etc. Within the plastic zone, the process zone is contained where new dislocations are nucleated each cycle (e.g. by activation of Frank-Read sources), while those dislocations already present within the material are propagated by the local shear stresses created by remote loading [39]. Kinematic hardening and the closely related Bauschinger effect are observed e.g. due to dislocation pile-up on stress concentrators such as grain boundaries, inclusions or precipitates. Voids are formed when the stress applied by dislocations pile-up is sufficient to create grain decohesion or inclusions debonding. Crack propagation, then, occurs due to void growth [40] and coalescence. When a crack propagates, it travels across the plastic zone created around the tip in its previous positions. A new crack tip is formed and a new plastic zone develops ahead of it, etc. Fatigue-induced cracks are thus entirely surrounded by a band of plastically deformed material. The application of an overload results in the increase of the plastic strain (stretch) at the current crack tip position. If the crack is grown beyond that position, the elevated profile of the crack faces created by the overload cause premature contact during load shedding. The two phenomena described above coexist, and both can be used to explain the overload-induced retardation effect, as discussed below.

The first phenomenon concerns the retardation effect induced by residual stresses, as pointed out by Drew et al. [41] and Ling and Schijve [42]. This has already been discussed in the introduction, but it is worth summarizing that in a cracked sample that has been overloaded and unloaded, compressive residual stresses develop within the plastically deformed material. The effective stress intensity factor $K$ that acts on the material within the elastic annular region surrounding the crack tip is thus decreased in comparison to the applied $K$.
The second phenomenon that is operative during the overload-induced retardation effect is the crack closure described above [30]. If an overload has been applied and the crack grown beyond its original position, the region behind the new crack tip position retains the “memory” of the plastic stretch, so that on the decreasing part of the load cycle, contact between crack faces at this point occurs sooner than the minimum load is reached. That premature crack closure keeps the crack “wedged open”, and prevents further reduction of the stress intensity at the crack tip. The overall consequence of this is the reduction of the stress intensity factor range, $\Delta K$, and the associated retardation of FCGR. As nicely shown in Croft et al. [29], when the load is at minimum, the crack is closed and the compressive strains are at maximum. When the load is increased, the crack remains closed for a while, while the compressive strains decrease continuously till they reach zero. The crack, then, starts to open. Simultaneously, and only after this point, the tensile strains ahead of the crack tip start to increase again, but only once the crack becomes fully open. The actual range of strains experienced by the region ahead of the crack tip is thus smaller than what would be expected by considering the applied load range only. The crack is then

Fig. 5. Schema representing phenomena that happen at the crack tip of a cyclically loaded ductile material when a crack is propagating.

Fig. 6. $K=f$ (load) plots for CT 1 and CT 6. $K$ values have been obtained in finding the best match between the experimental result and LEFM solutions.
less likely to propagate, since under smaller cyclic strain ranges, fewer dislocations are formed and void nucleation is less likely to occur. It seems self-evident that with greater applied overload this phenomenon becomes stronger. Croft et al. [29] also demonstrate that the bigger the overload, the later the crack opens in a cycle.

The phenomenon described above also becomes evident by using a different method of analysis that we propose here and that is illustrated in Fig. 6. The K values presented in Table 1 are plotted in Fig. 6 as a function of the applied load for both specimens. If no closure happened, both curves should be straight lines with the slope given by the Srawley equation. However, due to the contact between crack faces, closure happens at some intermediate load between the maximum and minimum loads in the cycle. Based on this assumption we draw straight lines passing through the points corresponding to the load of 7 kN (close to the maximum load in the cycle, when the crack is fully open) with the slope given by the Srawley equation. At high loads therefore the plot of SIF vs load can be reconstructed using the Srawley equation. The intersections between the closure behavior and fully open behavior provides an estimate of the load needed to fully open (or close) the crack, $P_{\text{closure}}$.

Although the plots in Fig. 6 provide a comparison between two specimens at different stages of retardation (CT 1 and CT 6), quantitative comparison is hard to establish since the crack lengths for the two specimens is not the same (which explains a different behavior in the post-opening regime). However, it is interesting to make a comparison (for both specimens) between the effective stress intensity factor range, $\Delta K_{\text{eff}}$, deduced from diffraction experiments with the applied SIF range predicted by the Srawley equation. For specimen CT 1 (that was grown beyond the maximum retardation point) the $\Delta K_{\text{eff}}$ range is decreased compared to $\Delta K_{\text{applied}}$ by a factor of about 2, corresponding (according to the power law observed on the Paris diagram for this material) to the FCGR reduction by a factor of 15. In comparison, for specimen CT 6 (where test was interrupted immediately after the overload) the FCGR knock-down factor amounts to nearly 30. These observations explain the profound retardation effect of the overload.

5. Conclusion

In the present study, the retardation effect on the crack growth rate in overloaded CT specimens experiencing cyclic loading was studied using synchrotron X-ray diffraction. Two different stages of crack propagation were studied. In one case (specimen CT 1) the crack was grown under constant amplitude fatigue conditions after that the specimen has been overloaded until the original crack growth rate was almost recovered. In the second case (specimen CT 6) loading was interrupted immediately after the overload. Synchrotron X-ray diffraction 2D strain mapping of the near – region at three different load levels was carried out for both specimens. The effective stress intensity factor K values were computed by finding the best match between LEFM predictions (with strain offset) and experimental data.

In this study, some observations by Croft et al. [27–29] were corroborated. Namely, the existence of a compressive strain region behind the crack tip which disappears progressively when the applied load is increased. On the other hand, the elastic strains ahead of the crack tip increase only from the time when the crack becomes fully open. The existence of a threshold (from which the stresses ahead of the crack tip start to increase and the compressive strain zone disappears) suggests the existence and operation of some sort of crack closure mechanism. This is illustrated in this study by the strong deviation from the Srawley equation predictions observed at low load levels. This approach (as opposed to that used by Croft et al. [29]) provides a less detailed (in terms of spatial resolution) picture of the phenomenon happening in the near-crack tip region. On the other hand, the current interpretation uses a large number of experimental measurement points in the process of matching the LEFM predictions to observations. This approach therefore leads to more reliable and statistically accurate evaluation of the SIF. It also permits to reconcile the traditional LEFM fatigue crack propagation prediction and experimental measurements of strain fields.

References


