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Microstructure-sensitive prediction and experimental validation of fretting fatigue
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Abstract: This paper is concerned with the development of a micro-mechanical methodology for prediction of fretting fatigue crack nucleation life and short crack propagation. The methodology employs critical accumulated plastic slip as a fatigue indicator parameter for microstructure-sensitive crack nucleation. Crystal plasticity unit-cell models are employed for calibration of constitutive and crack nucleation parameters and a crystal plasticity frictional contact model of the fretting test rig is developed to predict the micro-scale surface fretting damage, crack nucleation and estimated micro-crack propagation. The methodology is validated against fretting fatigue tests using a fretting bridge arrangement for 316L stainless steel.

1 Introduction
Fretting is a surface damage phenomenon that occurs in a wide range of industries such as offshore oil and gas, biomedical and aerospace. It typically arises when two surfaces in contact under a normal (clamping) load, $P$, experience cyclic relative tangential displacement (slip), due to superimposed cyclic loading conditions (displacement or load-controlled). In numerous engineering applications, this damage occurs at size scales competitive with the material microstructure, e.g. 5 to 100 μm. Three main sliding regimes are commonly associated with fretting, namely, partial, gross and mixed slip, primarily dependent on normal load, displacement amplitude and coefficient of friction (COF) [1]. When there is no stick region and all points of the contact region experience sliding at some instant in the load cycle, the regime is described as gross slip. Partial slip is characterised by a mixture of stick and slip regions across the contact region, so that part of the contact region remains stuck throughout the loading cycle. The mixed slip regime has initial gross slip
characteristics but generally stabilises into partial slip. Fretting damage is commonly considered to take either of two forms, namely fretting wear or fretting fatigue (cracking), with the former being broadly, but not exclusively, associated with gross slip and the latter with partial slip, e.g. Vingsbo and Soderberg [2], Ding et al. [3] and Fouvry et al. [4]. Wear can be considered as a microscale fatigue cracking of asperities leading to material removal and debris generation. A fundamental objective of the present work is to address microscale fatigue crack initiation in fretting with the potential to unify both fretting wear and crack nucleation predictions.

Ding et al. [3] observed cracks initiating in Ti-6Al-4V under partial slip conditions. These cracks were about 50 μm in length, of comparable length scale to the material microstructure (grain size). Goh et al. [5] presented similar scanning electron microscope (SEM) images of fretting crack nucleation sites at contact edges, also for Ti-6Al-4V. A crystal plasticity (CP) formulation was implemented for microstructure-sensitive modelling of fretting fatigue and plastic strain ratchetting was used to predict the orientation and location of experimental cracks, but without explicit prediction of life (numbers of cycles) to crack initiation due to the absence of a microstructure-sensitive fatigue parameter. Comparisons with $J_2$ continuum plasticity showed that the microstructure-sensitive model predicted more realistic and accurate results in terms of plastic strain ratchetting [6]. Numerous studies, e.g. Ding et al. [7], Madge et al. [8], have applied fatigue indicator parameters (FIP's) such as the Smith Watson-Topper (SWT) parameter, to predict fretting fatigue cracking, using macromechanical material modelling e.g. $J_2$ plasticity. Madge et al. [8] predicted the effect of slip amplitude on fatigue life by modelling the effects of material removal based on a wear simulation technique that incorporated the Archard wear equation. These life predictions were based on total life estimations, i.e. did not distinguish between crack initiation and propagation lives. For realistic prediction of micro-cracking in fretting
fatigue, it is argued here that a CP formulation is required. Numerous authors have dealt with CP modelling of fretting and fretting fatigue, e.g. Goh, McDowell and colleagues [6], [9], [10], Dick and Cailletaud [11] but again, without explicit prediction of the cycles to initiation, propagation and hence failure.

Recently, however Dunne and co-workers, e.g. [12] and [13], have presented a microstructure-sensitive FIP, based on accumulated crystallographic slip. This approach was successfully used to predict the low cycle fatigue response (numbers of cycles to nucleation) of a nickel base alloy C263, based on identifying a critical value of accumulated crystallographic plastic slip, corresponding to the number of cycles to cracking. The authors have recently described the implementation of this approach to fretting wear of 316L stainless steel (SS) [14], using the constitutive formulation of Huang [15], as implemented previously for modelling the deformation of 316L SS cardiovascular stents [16].

This paper combines experimental plain and fretting fatigue testing of 316L SS with presentation of finite element implementation of a micro-mechanical life prediction methodology, which separately predicts crack nucleation and crack propagation. An experimental fretting bridge type arrangement has been developed to study the effects of fretting fatigue using a cylinder-on-flat configuration. A crystal plasticity frictional contact model with a microstructure-sensitive crack initiation methodology is implemented to study the micro scale surface damage that is typical of fretting situations. Short and long crack growth behaviour is incorporated for both plain and fretting fatigue through the use of El-Haddad and Paris equations via the use of mixed mode cracking analysis and a weight function method. Total life predictions for fretting fatigue are calculated combining the microstructural sensitive crack initiation predictions with the mixed mode cracking analysis and these are compared to experimentally obtained fretting fatigue data.
2 Methodology

2.1 Crystal plasticity theory

Fatigue life, $N_f$, can be broken up into three different regimes:

$$N_f = N_i + N_{scg} + N_p$$  \hspace{1cm} (1)

$N_i$ is the number of cycles to crack initiation of a known crack dimension, typically less than 10 $\mu$m. $N_{scg}$ is the number of cycles within the short crack growth regime (SCG) regime and $N_p$ is the number of cycles of crack propagation until specimen fracture. Crack initiation is difficult to experimentally detect due to the microscopic length scales involved. In fatigue problems dislocation motion along persistent slip bands cause a saw tooth surface roughness profile. This surface damage eventually leads to the nucleation of small cracks as a result of microstructural defects, grain boundaries or weak grain orientations. For accurate modelling of slip system deformation a micromechanical model based on CP theory is used within this work.

316L is a face centred cubic (FCC) material comprising of 12 slip systems per crystal lattice, 4 slip planes and 3 slip directions, see Figure 1. Modelling the individual deformation of metallic crystal grains is done through a physically based, rate dependent crystallographic theory [17]. In this work isotropic elasticity is assumed within the CP user subroutine, with a Young's modulus of 213 GPa and Poisson's ratio, $\nu$, of 0.34. Plastic slip is assumed to obey Schmidt's law [18], where the rate of plastic shear strain, $\dot{\gamma}^\alpha$, for a particular slip system, $\alpha$, is assumed to depend on the resolved shear stress, $\tau^\alpha$, through the following power law:

$$\dot{\gamma}^\alpha = \dot{\alpha} \text{sgn}(\tau^\alpha) \left\{ \left| \frac{\tau^\alpha}{g^\alpha} \right| \right\}^n$$  \hspace{1cm} (2)

where $\dot{\alpha}$ and $n$ are the reference strain rate and rate sensitivity exponent, respectively. Material strain hardening is specified by the slip system strain hardness, $g^\alpha$, which is defined by the integral of the following equation:
\[
g^\alpha = \sum_\beta h_{\alpha\beta} \dot{\gamma}^\beta
\]  
(3)

where \( h_{\alpha\beta} \) are the strain hardness moduli and \( h_{\alpha\alpha} \) and \( h_{\alpha\beta} \) are the self and latent hardening moduli, respectively. In this work Taylor isotropic hardening is assumed and self and latent hardening moduli are considered equal. \( g(\gamma_\alpha) \) is the slip system strain hardness defined by the following hardness function [17]:

\[
g(\gamma_\alpha) = g_0 + (g_\infty - g_0) \tanh \left( \frac{h_0 \gamma_\alpha}{(g_\infty - g_0)} \right)
\]  
(4)

where \( h_0 \) is the initial hardening modulus, \( g_\infty \) is the saturation stress and \( g_0 \) is the critical resolved shear stress. The hardening moduli can be found through differentiation of the above equation, as follows:

\[
h_{\alpha\alpha} = h_{\alpha\beta} = h(\gamma) = h_0 \text{sech}^2 \left( \frac{h_0 \gamma_\alpha}{g_\infty - g_0} \right)
\]  
(5)

The accumulated slip, \( \gamma_\alpha \) is defined as follows:

\[
\gamma_\alpha = \sum_\alpha \int_0^t |\dot{\gamma}^\alpha| \, dt
\]  
(6)

This theory is implemented here in Abaqus 6.10 via a user defined material, (UMAT) user subroutine following the approach of [15].

2.2 Crystal plasticity calibration

An important aspect of modelling the deformation on individual slip systems within a metallic grain is the calibration process used to identify the constitutive constants that relate the resolved shear stress to the shear strain. A unit-cell polycrystalline model of uniform hexagonal morphology is developed here (Figure 2) following the work of Savage et al. [19]. This allows the inclusion of triple points and grain boundaries which are important features of
microstructural modelling. Random crystallographic orientations are assigned to each grain using a material grain size of 19 μm, as seen in Figure 3 which shows an optical image of the microstructure of the 316L SS material tested here. In this work the CP constants are identified via calibration with respect to the macroscopic cyclic stress strain curve of 316L SS, as represented by a non-linear kinematic hardening (NLKH) J₂ material model, using the material data shown in Table 1 and the following equation, from [20]:

\[
\frac{\Delta \sigma}{2} - k = \frac{C}{\gamma} \tanh \left( \frac{\Delta \varepsilon}{2} \right)
\] (7)

where \( k \) is the cyclic yield stress, \( C \) is the hardening modulus and \( \gamma \) is the rate of decay of the hardening modulus. The initial CP hardening modulus, \( h_0 \), saturation stress, \( g \), and critical resolved shear stress, \( g_c \), thus identified by matching the stabilised cyclic response of the unit-cell model of Figure 2, to the macroscopic cyclic CP response, across a range of applied stress-ranges are listed in Table 2. Figure 4 shows the comparison between the cyclic stress strain response of the CP unit-cell model, using the identified constants, and the macroscopic cyclic stress-strain curve (CSSC) of 316L SS, using the NLKH constants of Table 1.

2.3 Microstructure-sensitive crack initiation parameter

Previous work on modelling of fretting and fretting fatigue has employed macroscopic fatigue indicator parameters e.g. [3] [8]. However it is argued here that for accurate prediction of crack nucleation, it is necessary to employ a microstructure-sensitive FIP for scale consistency when using CP modelling, e.g. see also Sweeney et al. [21]. Hence, the FIP used here is the accumulated plastic slip, \( p \), defined by Manonukul and Dunne [13] as follows:

\[
\dot{p} = \left( \frac{2}{3} L^P : L^P \right)^{\frac{1}{2}} \quad ; \quad p = \int_0^\tau \dot{p} \, dt
\] (8)

where the plastic velocity gradient \( L^P \) is defined by:
\[ L^p = \sum_{\alpha=1}^{n} \dot{\gamma} s^\alpha n^{\alpha T} \]  

with \( s^\alpha \) and \( n^\alpha \) as the slip direction and normal vectors, respectively, for a given slip system, \( \alpha \), with \( n \) slip systems. \( \dot{\gamma} \) and \( p \) are coupled in the CP user subroutine. The criterion for crack initiation is then \( p = p_{\text{crit}} \) where \( p_{\text{crit}} \) has been argued to be a fundamental material constant. The approach adopted here, following that of \[13\], is to identify \( p_{\text{crit}} \) from a specific LCF data point for the material. Due to the quick stabilisation of the stress-strain response, it is possible to determine a stabilised maximum value of accumulated crystallographic plastic slip per cycle, \( p_{\text{cyc}} \), in the unit-cell model. The crack initiation criterion is then written as:

\[ p_{\text{crit}} = N_t p_{\text{cyc}} \]  

The critical accumulated slip, \( p_{\text{crit}} \), was shown in \[13\] to predict fatigue crack initiation for both low cycle fatigue (LCF) and high cycle fatigue (HCF) and over a range of temperatures for C263, a FCC nickel alloy.

2.4 Fracture mechanics

\( N_p \) can be calculated by the integration of the Paris equation to calculate the number of cycles for a small crack of a known size to propagate, as follows:

\[ \frac{da}{dN} = C(\Delta K)^m \]  

where \( C \) and \( m \) are material constants, \( a \) is the crack length and the stress intensity factor (SIF) \( \Delta K \) is

\[ \Delta K = \Delta \sigma Y \sqrt{\pi a} \]
Δσ is equal to \((σ_{max} - σ_{min})\) and \(Y\) is a geometrical factor, for the fatigue test specimens of the present work (see below) this is taken to be that of an edge crack under uniaxial loading given by

\[
Y = 1.12 - 0.281 \left(\frac{a}{w}\right) + 10.55 \left(\frac{a}{w}\right)^2 - 21.72 \left(\frac{a}{w}\right)^3 + 30.39 \left(\frac{a}{w}\right)^4
\]  

(13)

where \(w\) is the specimen thickness. Short cracks have been observed to propagate at a faster rate than long cracks [22]. However self-arrest occurs if \(ΔK\) is below a threshold value, \(ΔK_{th}\).

The El-Haddad approach [23] incorporates a threshold crack length, \(a_0\), which represents the transition from SCG to conventional crack growth, as can be seen from the Kitagawa and Takahashi diagram in Figure 5. \(a_0\) can be found from the following equation

\[
a_0 = \frac{1}{π} \left(\frac{ΔK_{th}}{σ_e}\right)^2
\]  

(14)

where \(σ_e\) is the fatigue limit and \(ΔK_{th}\) is the threshold stress intensity factor \(ΔK_{th}\) for \(a > a_0\).

The El-Haddad correction is an empirical approach which allows for the prediction of SCG by substituting \((a + a_0)\) for \(a\) in the SIF equation, when \(a < a_0\) as follows:

\[
ΔK = ΔσY\sqrt{π(a + a_0)}
\]  

(15)

2.5 Fretting fatigue crack growth

Crack propagation under fretting fatigue conditions is a complex issue. Houghton et al. [24] implemented a weight function method, based on the work of Nicholas et al. [25] which analysed the mixed mode cracking of Ti-6Al-4V to successfully predict multiaxial fretting fatigue in a simplified representative fretting fatigue test for spline coupling teeth. Houghton et al. [24] used back calculated fatigue constants with a critical plane SWT approach to estimate crack initiation at a length scale of 10 \(μm\). Once crack location was established, local stress ranges were identified for mode I and mode II weight functions, \(h_I\) and \(h_{II}\) respectively, as seen in Fett and Munz [26]:
where $\rho = x/a$ and $\alpha = a/W$. $a$ is the crack length, $W$ the specimen width and $A_{\nu \mu}$ are the influence coefficients [26] for each weight function. These weight functions are implemented within the following equations for mode I and II stress intensity factors.

\[ h_I = \frac{2}{\pi a \sqrt{1 - \rho}} \left[ 1 + \sum_{\nu, \mu} A_{\nu \mu} \alpha^\mu (1 - \rho)^{\nu + 1} \right] \quad (16) \]

\[ h_{II} = \frac{2}{\pi a \sqrt{1 - \rho (1 - \alpha)^{1/2}}} \left[ (1 - \alpha)^{1/2} + \sum_{\nu, \mu} A_{\nu \mu} (1 - \rho)^{\nu + 1} \alpha^\mu \right] \quad (17) \]

where $\rho = x/a$ and $\alpha = a/W$. $a$ is the crack length, $W$ the specimen width and $A_{\nu \mu}$ are the influence coefficients [26] for each weight function. These weight functions are implemented within the following equations for mode I and II stress intensity factors.

\[ \Delta K_I = \int_0^a \Delta \sigma_{xx}(x) h_I(x, \alpha) \, dx \quad (18) \]

\[ \Delta K_{II} = \int_0^a \Delta \sigma_{xy}(x) h_{II}(x, \alpha) \, dx \quad (19) \]

where $x$ is the independent variable representing direction along the crack and $\Delta \sigma_{xx}$ and $\Delta \sigma_{xy}$ are local FE-predicted normal and shear stresses during crack growth in model I and mode II respectively. This approach allows incorporation of the effect of the contact-induced stress gradients on crack growth. An effective stress intensity factor is then defined as:

\[ \Delta K_{eff} = \sqrt{\Delta K_{I,eff}^2 + \Delta K_{II}^2} \quad (20) \]

where

\[ \Delta K_{I,eff} = \Delta K_I (1 - R)^{(1-n)} \quad (21) \]

where $R > 0$ and $n$ is a material constant where $0 < n < 1$. In this work $n$ is chosen to have a value of 0.5. If $R < 0$, $\Delta K_I = \Delta K_{max}$. This is as a result of the dependency of crack
propagation rate on stress ratio $R$ expressed by Walker [27] based on empirical results. $R$ is defined as:

$$
R = \frac{\sigma_{\text{min}}}{\sigma_{\text{max}}}
$$

(22)

where $\sigma_{\text{min}}$ is the minimum applied stress and $\sigma_{\text{max}}$ is the maximum applied stress.

The increment of crack growth $\Delta a$ is defined by:

$$
\Delta a = C(\Delta K_{\text{eff}} - \Delta K_{\text{th}})^m \Delta N
$$

(23)

where $\Delta N$ is a cycle jumping factor. The process is repeated until the crack length, $a$, propagates through the specimen width.

2.6 Experimental Methodology

2.6.1 Material

The fatigue test coupons were manufactured from cold rolled austenitic 316L SS plate in accordance with ASTM standards for force controlled axial fatigue tests [28]. Figure 6 (a) shows the dog bone specimen design and dimensions. The surface roughness of 0.2 $\mu$m was verified using an atomic force microscope (AFM). Tests on mechanical properties and material composition were carried out and the results were found to be consistent with material data sheets provided by the supplier, as given in Table 3 and Table 4. These coupons were designed to be used in both plain and fretting fatigue tests. Chemical etching was performed on material samples cut from the specimens for microstructural characterisation. Samples underwent a grinding and polishing process using a range of diamond abrasive suspensions from 30 to 0.25 $\mu$m. A solution of 20% HCL and 80% distilled H$_2$O with an addition of 0.5-1.0 g of K$_2$S$_2$O$_5$ per 100 ml is used to colour etch the samples between 30-120 s [29]. Multiple micrographs of the material microstructure were examined using this technique and sample grain areas were taken from each to obtain an average grain area. This
grain area is then transformed into square grain dimensions as seen in Figure 3, to give an average square grain dimension of 19 μm.

2.6.2 Plain fatigue testing
The samples were cyclically tested in a servo-hydraulic Instron testing machine at a range of stress amplitude, $\sigma_{amp}$, values, under sinusoidal loading at a frequency of 7 Hz. Table 5 shows the range of loading conditions used for the plain fatigue testing. All tests were conducted at a stress ratio, $R$, of 0.1. Two plain fatigue tests were conducted at each stress level ($\sigma_{amp}$) to establish experimental scatter.

2.6.3 Fretting fatigue
A bridge type fretting arrangement as seen in Goh [5] and Pape [30] was developed here to validate the FE-predicted crack initiation lives. The rig consists of a cylindrical on flat fretting configuration. The interchangeable cylindrical fretting pads were manufactured from the same grade of 316L SS as the fatigue coupons. The fretting pads were 6 mm in radius and had an average surface roughness, $R_a$, value of 0.18 μm when measured with a Mitutoyo Surftest-211 profilometer. Ball bearings were used to transfer the load between the loading screws and the fretting pads supports to ensure self alignment and consistent distribution of contact pressure. The two fretting pads were symmetrically clamped onto each side of the gauge length of the fatigue coupon, as shown schematically in Figure 6 (b), and the assembly was cyclically loaded in an Instron testing machine, as shown in Figure 7. The proving ring was designed so that no plastic deformation would occur during the application of the clamping load. The proving ring was calibrated so that a known clamping pressure could be applied to the specimen gauge length via the loading screws. Calibration was performed by loading the proving ring in tension via the loading screws and acquiring a load-strain curve via strain gauges placed on the proving ring. This allowed a measurable clamping pressure to be applied to the surface of the fretting fatigue specimen. Sufficiently fine pitch threads on
the loading screws allowed for small incremental adjustments to the clamping pressure during
initial loading on the fatigue coupon. The loading screws were tightened to give in a contact
pressure of \(0.5 P_y\) at the substrate surface, where \(P_y\) is the load needed to cause yielding,
defined as follows [31]:

\[
P_y = \frac{\pi r}{E^c} (P_o) r^2
\]

(24)

\[
(P_o) Y = 1.8k
\]

(25)

where \(r\) is pad radius, \(E^c\) composite modulus of the two materials and \(k\) is the material yield
stress. The two strain gauges on the proving rig ensured consistent contact pressure
throughout a range of applied substrate loading. The specimen was cyclically loaded at 1 Hz
and the relative tangential movement between the fretting pads and the substrates induces
fretting fatigue. Fretting fatigue tests were conducted using the same applied stress ranges as
tabulated in Table 5 for plain fatigue.

2.7 Fretting fatigue modelling

The finite element model is based on a quarter segment of the experimental bridge type
fretting rig. The experimental arrangement is modelled as a 2D plane strain cylinder-on-flat
fretting model. A 6 mm radius cylinder is held in contact with a 4.5 x 13.5 mm substrate
under a fixed normal (clamping) load. The substrate represents one quarter of the gauge
length with symmetry boundary conditions on the bottom and right edges. The substrate is
subjected to the same cyclic loading conditions as in the experiments. Linear equation
constraints are defined on the left edge of the substrate to ensure uniform nodal displacements
in the horizontal \(X\) direction between the master and slave nodes when applying cyclic
loading. The same method is used on the fretting pad to enforce uniform nodal displacements
in the vertical \(Y\) direction when applying a normal force \(P\). The normal load \(P\) is applied in
the first step and held constant throughout the analysis. In the second and subsequent steps
the cyclic loading is applied to the substrate to simulate the experimental conditions as shown in Figure 7. A coefficient of friction (COF), $\mu$, of 0.8 is used throughout this work based on unlubricated metallic contact as seen in McColl et al. [32]. A Lagrange multiplier contact algorithm is used which enforces a stick condition on nodes that are less than a critical shear stress, $\tau_{\text{crit}}$, where $\tau_{\text{crit}} = \mu p$ and $p$ is the local contact pressure on that node. The ADJUST parameter is used to reposition nodes on the slave surface directly onto the surface of the master surface. Additionally, the HCRIT parameter is used to set the maximum interpenetration of the slave nodes into the master surface before a smaller increment is attempted. This value is set to 0.1 of the slave surface element length. A similar contact model is presented in [14] for further reference. Figure 8 illustrates the hybrid model whereby a CP contact region is embedded within a non-linear kinematic hardening (NLKH) $J_2$ plasticity bulk model. This CP region is 20 grains wide and 10 grains deep. Square grains are used to give better mesh control. A comparison study between square and hexagonal grains was carried out in McCarthy et al. [14] showing almost identical results (in terms of $p$) for this type of fretting analysis. An element size of 2.5 $\mu$m is used in the contact region with a decreasing mesh density further away from the CP region. Previous mesh refinement studies [14] have shown that a 2.5 $\mu$m contact element size is more than sufficient for accurate results while the coarser mesh at the outer edges allows faster overall model run-times. A sensitivity study on orientation effects of grains within the CP contact region was conducted. Five different sets of random orientations were studied across all test stress amplitudes, to study the dependency of crack initiation life on crystallographic orientation.

3 Results

3.1 Plain fatigue

Figure 9 shows the S-N curve summarising the plain fatigue test results. A fatigue limit of 191.25 MPa, $\sigma_{\text{amp}}$, is evident when the results are graphically presented. Repeat tests carried
out at the same applied stresses showed good consistency of the life. It is worth pointing out
also, in particular the significantly larger degree of scatter at lower loads (larger lives), e.g. \(\sim 10^5\) at \(\sigma_{amp} = 202\) MPa. Table 6 shows the experimental failure lives along with the average
data used below in the computational work. All failed specimens fractured within the gauge
length. Figure 10 shows an SEM image of a typical fracture surface seen in the plain fatigue
specimens. The distinction between the crack initiation and propagation region compared to
the fast fracture region is evident. In this case, initiation began in the top left corner and
propagated radially throughout the specimen until failure. An average dimension of the
propagation region for all plain fatigue specimens is used in the following crack growth
analysis. Failure is deemed to have occurred when the propagation region has grown to 1.5
mm in length.

3.2 Fretting fatigue

The fretting fatigue tests were carried out until complete failure of the fatigue specimen. All
of these fretting failures occurred in the contact regions, i.e. under the fretting pads, on the
fatigue specimen. The resulting cracks propagated from the surface at angles of between 70°
and 110°, resulting in a fracture surface comparable to those of the plain fatigue case as seen
in Figure 10. The measured fretting fatigue lives are plotted in Figure 11 along with the plain
fatigue results. Clearly the fretting action leads to a significant reduction in fatigue life. An
average life reduction factor of 3.5 is observed, as tabulated in Table 6. Furthermore there is
no longer a fatigue limit at \(\sigma_{amp} = 191.25\) MPa. SEM imaging carried out on the failed
specimens shows fretting wear scars and crack nucleation sites at the contact edges. Figure 12
shows the typical surface damage and debris measured for \(\sigma_{amp} = 225\) MPa. Quantification
of fretting wear scar depths and areas was carried out via profilometry and SEM. A Taylor-
Hobson Surtronic 3+ was used to obtain wear scar profiles in the contact regions. Two
dimensional longitudinal profiles were measured at regular intervals across the width of the
wear scar; Figure 13 shows one such measured image, highlighting the U-shaped wear scar, which is typical of gross slip fretting conditions. To allow for more accurate microscopy, characterisation and validation of the wear scar, a technique described by Ding et al. [3], for example, was implemented. Specimen gauge lengths were sectioned via a slow diamond cutter and mounted in an epoxy resin. After mounting, the specimens were subjected to plane grinding using a range of SiC paper and polished through a range of abrasive diamond suspensions, finishing with a 3 μm paste. The finely polished surface allowed for quantification of the wear scar via optical microscopy techniques. Figure 14 shows the polished cross sectional area of a fretting wear scar. Good agreement exists between the two wear scar measurements methods with respect to shape, width and depth, as seen in the sample comparative data of Table 7.

A reddish-brown oxide layer was observed on the outer edges of the wear scar. EDX (Energy dispersive X-ray) spectroscopy of three distinct regions was carried out on the specimen surface to assess the levels of oxidation. Figure 15 shows the undamaged, worn and debris region of the gauge length where spectroscopy was applied. The highest levels of oxygen were present in spectrum 2 and 3, where surface damage had occurred in the form of wear and debris build up, respectively. The oxygen levels compared to the undamaged surface had increased from 2% to 36% on average, highlighting the presence of fretting corrosion; see Table 8 for more detail.

3.3 Computational

The numbers of cycles to crack initiation, \( N_p \), for the plain fatigue test data were inferred by implementing the Paris and El-Haddad methodology for short and long crack growth. Using \( \Delta K_{th} = 5.81 \text{ MPa m}^{0.5} \) [33] a threshold crack length, \( a_{th} \), of 59.5 μm is calculated. Incorporating this transitional crack length with the Paris equation constants \( C \) and \( m \), 2.0 × 10^{-10} and 1.9, respectively, from [34] based on short crack growth testing of 316L SS,
an "experimental" number of cycles for propagation $N_p^{\text{exp}}$ is calculated, where $N_p^{\text{exp}}$ is defined as the number of cycles for a 1.2 μm crack to propagate to failure, in this case to a length of 1.5 mm. Hence it is possible to infer an "experimental" number of cycles to initiation, as follows:

$$N_i^{\text{exp}} = N_f - N_p^{\text{exp}}$$  \hspace{1cm} (26)

$N_p^{\text{exp}}$ is relatively small, about 15-30%, compared to $N_i^{\text{exp}}$ as shown in Table 9. Using $N_i^{\text{exp}}$ it is possible to deduce a value of critical accumulated plastic slip parameter, $p_{\text{crit}}$, for one test data point. Figure 16 and Table 9 show the predicted CP crack initiation response versus the inferred $N_i^{\text{exp}}$ data, based on the identified value of $p_{\text{crit}} = 37.73$.

Contact variable distributions are presented in Figure 17 which show the evolutions of contact pressure and shear. The predicted evolutions of the $p_{\text{cyc}}$ damage parameter and contact slip distributions are presented in Figure 18. The position of maximum stabilised $p_{\text{cyc}}$ is the predicted location of crack initiation. In all cases, $p_{\text{cyc}}$ has stabilised after 6 cycles. In Figure 18 (a), (c) and (e) crack initiation is predicted to occur at the edge of contact, which correlates well with experimental observations. Figure 12 provides a typical location for crack initiation at the edge of contact, further corroborating the FE predictions. The distinct localised peaks of the $p_{\text{cyc}}$ parameter, particularly after 6 cycles, is evident in contrast to the more uniform distributions of conventional FIP's such as SWT as seen in [3], for example. Hence, a key benefit of $p_{\text{crit}}$ as an FIP is that it is less ambiguous than non-micro structure-sensitive FIP's. Also, it is worth noting that a number of such localised peaks in $p_{\text{cyc}}$ are predicted. This is consistent with experimental observations of multiple micro-cracking sites in fretting contact regions (see also [3]). The gross slip fretting condition is evident in the FE model from the contact slip distributions presented in Figure 18 (b), (d) and (f). 
The microstructure-sensitive methodology was implemented within the hybrid FE fretting fatigue model to predict the fretting fatigue crack initiation life of 316L SS under the simulated experimental conditions. These microstructure-sensitive crack initiation lives, calculated at a depth of 1.2 μm, were predicted for the different applied stresses via the accumulated plastic slip per cycle, \( p_{\text{cycle}} \), values predicted by the hybrid FE models in equation 10. The result of the crystallographic orientation study carried out using five sets of randomly generated orientations assigned to each grain in the FE model are used to quantify scatter. Error bars signify this scatter of predicted fretting fatigue crack initiation lives. Figure 19 compares these CP microstructure-sensitive predictions against the experimental lives. Table 10 tabulates the data in greater detail for comparative purposes.

Using the hybrid FE crack initiation methodology, described previously, in conjunction with the weight function method for mixed mode crack propagation, to calculate total life for the fretting fatigue cases, comparisons are made against experimental total life fretting fatigue data in Figure 20. The hybrid model initial crack location was assumed to propagate in a direction normal to the substrate surface. The process was implemented automatically within a computer program which incrementally calculated crack length until the crack had propagated through the substrate to a critical failure length. This critical length, based on experimental observations, was taken to be 1.5 mm. The predicted initiation, propagation and total CP microstructure-sensitive predictions for one specific random orientation set are given in Table 11.

An experimental life reduction factor of 3.5 was expressed previously as the ratio between the plain and fretting fatigue total life. A similar ratio can be expressed between the CP fretting fatigue total life results, as seen in Table 11, and a combination of the (i) CP plain fatigue crack initiation life and (ii) the Paris-El Haddad experimental plain fatigue
propagation life presented in Table 9. This results in a similar knockdown factor in life of 2.15.

4 Discussion

Fretting was induced on fatigue specimens through the implementation of a bridge type fretting rig. Experimental observations showing the presence of fretting fatigue is associated with: (i) the significant reduction in life compared to plain fatigue testing, (ii) the presence of a red oxide layer and (iii) wear scars at the contact regions on the specimen-pad contact surfaces. The two methods used to obtain measured wear scars show consistent results for scar depth and width. SEM imaging (Figure 10) shows crack nucleation sites at the trailing edge of contact, consistent with published work e.g. [35]. EDX carried out on sample points on the wear scar highlights the presence of a metallic oxide layer (Table 8). Fretting causes material surface damage which removes the corrosion resistant passive layer and allows for oxidation to occur on the material beneath the surface. This phenomenon typically results in debris which can have a higher hardness than the original material and accelerates the process of further material removal. An experimental anomaly is observed for the test result at 202.5 MPa which deviates from the experimental trend. A much lower fatigue life is observed at this stress amplitude. It is evident from the plain fatigue testing that a significant amount of experimental scatter is present at the mid to lower stress amplitudes (213.75 to 202.5 MPa). It is therefore reasonable to assume that this irregularity can be explained by experimental fretting fatigue scatter.

An FE model was generated to study the effects of the microstructure under experimental fretting fatigue conditions. The initial Hertzian calculated and FE-predicted contact semi-width, are 33 μm and 36 μm, respectively. However, significant widening is predicted in the early fretting cycles combined with non-uniform distributions of both contact pressure and shear, as shown in Figure 17. The significant widening of the contact area is the
result of the micro-plastic deformation of the substrate microstructure during cyclic loading. The applied loads to the bridge type fretting rig resulted in a gross sliding fretting situation which is evident from the wear scars shown in Figure 13 and Figure 14. The FE model predicts the same fretting condition, which is evident from the contact slip distributions of Figure 18 (b), (d) and (f). The CP model results in contact slip distributions that are non-uniform and asymmetric in nature. This again, is caused by material inhomogeneity leading to differential yielding across different grains. During plain fatigue testing crack propagation made up a small proportion of total life (15-30 %) with the majority of cycles spent initiating a surface crack, especially for the lower stress range. In contrast, for the fretting fatigue specimens, a significant proportion of specimen life involved crack propagation. This is attributed, obviously, to surface fretting damage significantly shortening crack initiation life. Highly localised concentrations of surface traction and plastic slip are predicted in Figure 17 and Figure 18 (a), (c) and (e) respectively.

For high load levels, the CP method, with combined crack initiation, \( N_p \) and propagation, \( N_p \), predictions is very accurate. At lower load levels more conservative results are observed. Interestingly for \( \sigma_{app} = 202.5 \) MPa case, the experimental life seems to be out of trend with respect to other results. The plain fatigue scatter for this case is significant and if assumed to apply to the fretting fatigue case for this load, would encompass the predicted CP result. It is possible that this low experimental life is due to complex interaction of wear and crack nucleation. In this case, the fact that wear-induced material removal is not explicitly simulated here (e.g. see Zhang et al. [36]), can explain the fact that the CP predictions over estimate life, for this load level. Furthermore, nucleation life prediction trends will be affected by fatigue damage accumulation effects, i.e. material degradation due to damage (e.g. void growth and micro-cracking). This has not been incorporated into the present work although an approach similar to that of Zhang et al. [37] could be adopted.
within the CP formulation in further studies. Furthermore, the absence of an initial surface roughness from the CP model will undoubtedly affect predicted crack initiation and propagation.

The FE model of Figure 8 only represents one quarter of the fretting rig, based on an assumption of symmetry. In reality, due to the finite (axial) spacing of the bridge feet and the axial variation of (fatigue) strain-induced displacement of the substrate (fatigue specimen), there is likely to be at least a small difference in pad-substrate relative displacement between the two fretting pads. Clearly, it is therefore possible that different slip regimes could exist at the two pads (on either side of the fatigue specimen), particularly if the imposed slip amplitude is close to the gross-partial slip threshold displacement and also due to variations in surface roughness or specimen-pad manufacturing variations, for example. It is also possible that the fretting regime varies with respect to time (fretting cycles) between partial, mixed and gross slip. In such a situation, different pads would accumulate different types of fretting damage (associated with the different slip regimes) for different numbers of fretting cycles and there would no longer be a one-to-one correspondence between number of (imposed) fatigue cycles and number of fretting damage cycles under any one pad. This is an interesting (and complex) modelling problem which would require the development of a micro-plasticity wear-fatigue approach, e.g. combining the present crystal (micro-) plasticity model, for example, with a material removal simulation capability (as described above). However, as shown above, the microscopy and profilometry of the measured wear scars from the tests, indicating significant increase in contact width (~ 300 \( \mu \)m, approximately 8\( a_o \), where \( a_o \) is initial contact width), combined with the FE-predicted slip amplitude (~ 35 \( \mu \)m), corroborate the assumption of negligible difference in relative slip between the two pads. Thus, the symmetry assumption for the fretting rig analysis is considered to be reasonable.
Furthermore, a constant value of COF is assumed throughout this work. It has been observed experimentally [32] that the COF value typically evolves, in unlubricated cases, during the fretting process. Typically, there is an initial (bedding-in) period with the COF value increasing from as low as 0.3 to a steady-state value of about 0.7 or higher, as the passive surface oxide layer is removed and eventually stabilises due to third body formation [38]. Experimental evidence in McColl et al. [32] for a high strength aeroengine steel, observes that the COF stabilises after about 1000 cycles. Future work will examine the effect of a more realistic model for the cyclic evolution of COF. For the moment, the present work adopts a conservative approach with respect to fretting fatigue damage, by assuming an initially high (stabilised) COF value.

5 Conclusions

- A bridge-type fretting fatigue rig was developed and applied to study the effects of fretting fatigue on 316L SS. Comparisons against plain fatigue tests demonstrated a significant life reduction factor, within the range of $\approx 2$ to $\approx 6$ (on average 3.5), due to the gross slip fretting action.
- The realistic microstructure grain size of the 316L material was identified and incorporated into a micromechanical, physically-based, rate-dependent crystal plasticity model.
- A microstructure-sensitive crack nucleation prediction methodology, based on a unit-cell model, was calibrated to identify the critical accumulated plastic slip parameter $p_{\text{crit}}$ for plain fatigue.
- Short and long crack growth calculations based on the El-Haddad and Paris equations allowed for estimations of the number of cycles to crack initiation based on the plain fatigue response of the material and the micromechanical (unit-cell) model.
An FE fretting fatigue cylinder-on-flat arrangement of the experimental bridge-type rig was implemented, incorporating the microstructure-sensitive crack initiation methodology within the contact region. Crack initiation life predictions were made across a range of stress amplitudes.

A mixed-mode crack growth methodology was combined with a microstructural sensitive crack initiation prediction methodology and successfully compared to experimental failure lives. Excellent predictions are achieved at the higher stress amplitudes where less scatter is observed in plain fatigue. At lower stress amplitudes, where greater scatter in plain fatigue is observed, the crystal plasticity model over-predicts propagation, and hence total life for one load level, but under-predicts propagation, and hence total life, for the other load levels. In all cases the predicted crack initiation fretting lives are conservative relative to the experimental fretting fatigue (total) lives.

6 Acknowledgements

SFI/HEA Irish Centre for High-End Computing (ICHEC) for the provision of computational facilities and support. The technical staff of the College of Engineering & Informatics, Mr Seamus Carey for rig design and manufacture assistance and Dr. Barry O'Brien for helpful discussions during the chemical etching stage of this work.
7 References


[38] S. Fouvry, “Shakedown analysis and fretting wear response under gross slip condition,” 
8 Tables

Table 1. Material constants, including non-linear kinematic hardening data, for 316L SS [20].

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>$E$</td>
<td>213 GPa</td>
</tr>
<tr>
<td>$v$</td>
<td>0.32</td>
</tr>
<tr>
<td>$k$</td>
<td>300 MPa</td>
</tr>
<tr>
<td>$C$</td>
<td>30000 MPa</td>
</tr>
<tr>
<td>$\gamma$</td>
<td>60</td>
</tr>
</tbody>
</table>

Table 2. Identified CP constitutive constants for cyclic behaviour of 316L SS.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>$h_0$</td>
<td>10.5 GPa</td>
</tr>
<tr>
<td>$g_\infty$</td>
<td>207.84 MPa</td>
</tr>
<tr>
<td>$g_0$</td>
<td>82.68 MPa</td>
</tr>
<tr>
<td>$\dot{\alpha}$</td>
<td>0.0023 s$^{-1}$</td>
</tr>
<tr>
<td>$n$</td>
<td>30</td>
</tr>
</tbody>
</table>

Table 3. Element composition of 316L SS from EDX.

<table>
<thead>
<tr>
<th>Element</th>
<th>Weight %</th>
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</thead>
<tbody>
<tr>
<td>C</td>
<td>12.30</td>
</tr>
<tr>
<td>O</td>
<td>2.32</td>
</tr>
<tr>
<td>Si</td>
<td>0.51</td>
</tr>
<tr>
<td>Cr</td>
<td>15.84</td>
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<td>Mn</td>
<td>1.74</td>
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<td>Fe</td>
<td>59.11</td>
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<tr>
<td>Ni</td>
<td>6.86</td>
</tr>
<tr>
<td>Mo</td>
<td>1.31</td>
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</table>

Table 4. Experimentally measured material properties for 316L SS.

<table>
<thead>
<tr>
<th>Property</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>$E$ (GPa)</td>
<td>213</td>
</tr>
<tr>
<td>YS (MPa)</td>
<td>253</td>
</tr>
<tr>
<td>UTS (MPa)</td>
<td>636</td>
</tr>
</tbody>
</table>

Table 5. Experimental loading range for both plain and fretting fatigue experiments using $R = 0.1$.

<table>
<thead>
<tr>
<th>$\sigma_{max}$ (MPa)</th>
<th>$\sigma_{min}$ (MPa)</th>
<th>$\sigma_{amp}$ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>500</td>
<td>50</td>
<td>225</td>
</tr>
<tr>
<td>475</td>
<td>47.5</td>
<td>213.75</td>
</tr>
<tr>
<td>450</td>
<td>45</td>
<td>202.5</td>
</tr>
<tr>
<td>425</td>
<td>42.5</td>
<td>191.25</td>
</tr>
</tbody>
</table>
Table 6. Tabulated data of the experimental results for both plain and fretting fatigue over a range of $\sigma_{amp}$ values. On average a fretting fatigue life reduction factor of 3.5 is observed relative to the plain fatigue results.

<table>
<thead>
<tr>
<th>Test</th>
<th>Test type</th>
<th>$\sigma_{amp}$</th>
<th>$N_f$</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>PF</td>
<td>225</td>
<td>112108</td>
</tr>
<tr>
<td>2</td>
<td>PF</td>
<td>225</td>
<td>102284</td>
</tr>
<tr>
<td>3</td>
<td>PF</td>
<td>213.75</td>
<td>177862</td>
</tr>
<tr>
<td>4</td>
<td>PF</td>
<td>213.75</td>
<td>113796</td>
</tr>
<tr>
<td>5</td>
<td>PF</td>
<td>202.5</td>
<td>279580</td>
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<tr>
<td>6</td>
<td>PF</td>
<td>202.5</td>
<td>160816</td>
</tr>
<tr>
<td>7</td>
<td>PF</td>
<td>191.25</td>
<td>Run-off</td>
</tr>
<tr>
<td>8</td>
<td>FF</td>
<td>225</td>
<td>56669</td>
</tr>
<tr>
<td>9</td>
<td>FF</td>
<td>213.75</td>
<td>59780</td>
</tr>
<tr>
<td>10</td>
<td>FF</td>
<td>202.5</td>
<td>35091</td>
</tr>
<tr>
<td>11</td>
<td>FF</td>
<td>191.25</td>
<td>245806</td>
</tr>
</tbody>
</table>

Note: PF = Plain Fatigue; FF = Fretting Fatigue

Table 7. Comparison of wear scar measurement techniques as illustrated in Figure 13 and Figure 14 for $\sigma_{amp}$ of 225 MPa.

<table>
<thead>
<tr>
<th>Wear scar 1 (Profilometer)</th>
<th>Wear scar 2 (Microscopy)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Shape</td>
<td>U shaped</td>
</tr>
<tr>
<td>Width (μm)</td>
<td>349.0</td>
</tr>
<tr>
<td>Depth (μm)</td>
<td>9.46</td>
</tr>
</tbody>
</table>

Table 8. Percentage weight of oxygen levels in three regions, undamaged, worn and debris from fretting fatigue specimen.

<table>
<thead>
<tr>
<th>Spectrum</th>
<th>Description</th>
<th>Oxygen %</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Undamaged</td>
<td>2</td>
</tr>
<tr>
<td>2</td>
<td>Worn</td>
<td>35</td>
</tr>
<tr>
<td>3</td>
<td>Debris</td>
<td>38</td>
</tr>
</tbody>
</table>

Table 9. Calculated experimental plain fatigue data separated into crack initiation and crack propagation lives compared against the predicted CP crack initiation data.

<table>
<thead>
<tr>
<th>$\sigma_{amp}$ (MPa)</th>
<th>$N_i^{exp}$</th>
<th>$N_p^{exp}$</th>
<th>$N_i^{CP}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>225</td>
<td>76340</td>
<td>30856</td>
<td>95182</td>
</tr>
<tr>
<td>213.75</td>
<td>111815</td>
<td>34014</td>
<td>111815</td>
</tr>
<tr>
<td>202.5</td>
<td>182505</td>
<td>37693</td>
<td>122033</td>
</tr>
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</table>
Table 10. Microstructure-sensitive fretting fatigue crack initiation lives for different crystallographic orientation sets.

<table>
<thead>
<tr>
<th>$\sigma_{\text{amp}}$ (MPa)</th>
<th>SET-1</th>
<th>SET-2</th>
<th>SET-3</th>
<th>SET-4</th>
<th>SET-5</th>
</tr>
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<tbody>
<tr>
<td>225</td>
<td>9093</td>
<td>15481</td>
<td>2117</td>
<td>2008</td>
<td>5753</td>
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<tr>
<td>213.75</td>
<td>15350</td>
<td>34451</td>
<td>2368</td>
<td>2572</td>
<td>6927</td>
</tr>
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<td>27044</td>
<td>32762</td>
<td>2911</td>
<td>3602</td>
<td>8027</td>
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<tr>
<td>191.25</td>
<td>58989</td>
<td>35798</td>
<td>3212</td>
<td>4513</td>
<td>8412</td>
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</table>

Table 11. Combined crack initiation and propagation lives for the microstructural sensitive model based on one set of random orientations compared against the experimental fretting fatigue data.

<table>
<thead>
<tr>
<th>$\sigma_{\text{amp}}$ (MPa)</th>
<th>$N_i^{CP}$</th>
<th>$N_p^{CF}$</th>
<th>$N_f^{CF}$</th>
<th>$N_f$ (Experimental)</th>
</tr>
</thead>
<tbody>
<tr>
<td>225</td>
<td>9093</td>
<td>41868</td>
<td>50961</td>
<td>56669</td>
</tr>
<tr>
<td>213.75</td>
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<td>202.5</td>
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<td>191.25</td>
<td>58989</td>
<td>72531</td>
<td>131520</td>
<td>245806</td>
</tr>
</tbody>
</table>
9 Figures

Figure 1. (a) Atomic structure of a FCC crystal lattice and (b) illustration of the slip directions and corresponding slip plane for a FCC material.

Figure 2. Unit-cell model under stressed controlled loading showing a contour plot of accumulated plastic slip, $p$.

Figure 3. Tint etching of 316L SS highlighting an average grain size, $d$, of 19 $\mu$m.
Figure 4. Comparison of the crystal plasticity (CP) predicted response against the $J_2$ macroscopic cyclic stress strain curve (CSSC).

Figure 5. A simple schematic of the Kitagawa and Takahashi diagram.
Figure 6. (a) Schematic of the plain fatigue specimen and (b) the fretting pads clamped to the gauge length of the fatigue coupon.
Figure 7. Simple schematic highlighting the fatigue specimen clamped between the fretting feet and cyclically loaded ($\sigma_{app}$) within the servo hydraulic testing machine along with the load history.

Figure 8. A schematic of the fretting fatigue FE model based on the experimental test arrangement, highlighting the CP embedded region in the contact zone.
Figure 9. Experimental plain fatigue results including repeat tests also highlighting the material fatigue limit in terms of stress amplitude, $\sigma_{amp}$, 191.25 MPa.
Figure 10. SEM image of the plain fatigue specimen subjected to a $\sigma_{amp} = 202.5$ MPa highlighting the dimensions of the crack propagation region.

Figure 11. Fretting fatigue lives compared to corresponding plain fatigue data.
Figure 12. SEM image showing edge view of a gauge length of a fretting fatigue specimen where crack initiation has occurred at the edge of the wear scar. The specimen was subjected to a $\sigma_{amp}$ of 225 MPa and failed after 56,669 cycles.

Figure 13. A typical gross slip wear scar, tested at $0.5P_y$ normal load and $\sigma_{amp}$ of 225 MPa, obtained using a profilometer.
Figure 14. A wear scar profile for specimen 9 tested at $0.5P_y$ normal load and a $\sigma_{amp}$ of 225 MPa obtained via microscopy techniques.

Figure 15. EDX was carried out in three different zones on a contact region of a fretting fatigue specimen highlighting the presence fretting corrosion. Results are tabulated in Table 8.
Figure 16. Comparison between experimental and CP based predictions for number of cycles to crack initiation, $N_i$ for plain fatigue.
Figure 17. CP fretting model predicted evolutions of contact pressure and shear under different stress amplitudes, $\sigma_{\text{amp}}$. The CP fretting models have an initial contact semi width ($a_0$) of 0.036 mm and the contact variables are sampled when the applied stress equals $\sigma_{\text{max}}$. 
Figure 18. CP fretting model predicted evolutions of $p_{cyce}$ and contact slip distributions under different stress amplitudes, $\sigma_{amp}$, $a_0$ equals 0.036 mm.
Figure 19. Comparison of experimental failure lives and microstructural sensitive crack initiation lives, including the effect of random orientations for fretting fatigue. Error bars represent the maximum and minimum microstructure-sensitive initiation lives.

Figure 20. Total experimental failure lives compared against the total predicted failure lives based on the microstructure-sensitive model.
Highlights

- Development of a bridge-type fretting rig to quantify the detrimental effects of fretting fatigue
- Microstructural-sensitive fretting fatigue crack initiation prediction methodology
- Mixed mode crack propagation methodology based on El Haddad and Paris equation
- Excellent correlation between the FE predicted total life and experimental fretting fatigue data