Quantification of slip band distribution in polycrystals: an automated fast Fourier transform decomposition approach

Fernando D. León-Cázares, Bradley Rowsland, and Enrique I. Galindo-Nava

Department of Materials Science and Metallurgy, University of Cambridge, 27 Charles Babbage Rd, Cambridge, CB3 0FS, UK, Sandia National Laboratories, 7011 East Avenue, Livermore, CA, 94550, USA and Department of Mechanical Engineering, University College London, Torrington Place, London, WC1E 7JE, UK

Abstract

Plastic deformation is accumulated in slip bands in a wide variety of engineering alloys. Multiple material and loading conditions impact their distribution and degree of slip localisation, but these effects are rarely quantified. To tackle this, the current work introduces a fast Fourier transform (FFT) decomposition method and applies it to a tensile loaded polycrystalline nickel-based superalloy imaged via high resolution digital image correlation and electron backscatter diffraction. This approach identifies active slip planes over the FFT images of individual grains and performs inverse transforms such that slip band traces with shared orientations are isolated. This technique enabled the largest quantification of slip band spacings and in-plane strains to date, with a total of 6557 slip bands detected. The results show that the slip band spacings increase with grain size, with no evident dependence on grain orientation and Schmid factor. Slip bands are found to develop similar spacings along different octahedral planes and continue to spread over larger regions of the grain as the resolved shear stress of the active slip system increases. The FFT decomposition technique, which could be employed with multiple microscopy techniques, will allow for much needed large scale quantitative studies of slip localisation.

Key words: plastic deformation, slip band, slip localisation; slip planarity, fast Fourier transform, electron microscopy, digital image correlation, electron backscatter diffraction

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Introduction

Several structural metals and alloys accommodate plastic deformation in the form of slip bands. For example, these appear in face-centred cubic (fcc) materials with low stacking-fault energies, such as nickel-based superalloys (León-Cázares et al., 2020b) and some austenitic stainless steels (Kim et al., 2019; Li et al., 2021), and in hexagonal close packed metals like titanium (Guo et al., 2014) and zirconium (Andani et al., 2020). Slip bands are complex planar dislocation structures that develop along neighbouring slip planes, and are characterised by the presence of dipoles, multipoles, dislocation debris and other defects (Lukáš & Kunz, 2004; Risbet & Feaugas, 2008; León-Cázares et al., 2020b). Critically, slip is highly localised in these regions, which makes them preferential sites for crack nucleation (Sangid, 2013; Polášek & M不止, 2014; Stinville et al., 2015). Recent research (Stinville et al., 2022) has shown that the distribution of slip developed during the first cycle is indicative of the lifetime of polycrystalline materials under cyclic loading. Thus, there is a need to accurately characterise the slip band morphologies and distributions.

A plethora of factors affect the slip localisation behaviour. Slip bands form during the early stages of plasticity (Petrenec et al., 2014; Ho et al., 2015), and as deformation progresses, the slip activity within them intensifies, their dislocation densities increase and additional bands form (Di Gioacchino & Quinta da Fonseca, 2015; Lavenstein et al., 2020). During cyclic loading regimes they often saturate after a number of cycles (Lukáš & Kunz, 2004; León-Cázares et al., 2020a, and the slip irreversibilities may continue to damage the microstructure (Mughabi, 2009, 2013; León-Cázares et al., 2020a). Loading conditions such as temperature (Lerch et al., 1984), strain rate (Valsan et al., 1994) and strain amplitude (Kundu et al., 2019) influence the resulting slip band distribution and dislocation densities. Material properties that increase the slip planarity are low stacking fault energies (Risbet & Feaugas, 2008), and the presence of second-phases (Lunt et al., 2017) (particularly of smaller sizes (León-Cázares et al., 2020a)) and short range order (He et al., 2021; Kim et al., 2023; Abu-Odeh & Asta, 2022). Moreover, physical phenomena such as dynamic strain ageing (Aboulfadl et al., 2015) and hydrogen trapping leading to hydrogen embrittlement (Nabur et al., 2006; Menard et al., 2008; Yu et al., 2019) have also been linked to more localised plastic deformation behaviours. However, while the literature concerning these matters has widely discussed the multiple roles of slip localisation, this is rarely done quantitatively and with a high statistical significance.

Slip bands can easily be observed with a variety of imaging techniques, from optical (Lerch et al., 1984; Lerch & Gerold, 1987; Valsan et al., 1994) to electron (e.g. with electron channeling contrast imaging (ECCI) (León-Cázares et al., 2020a),...
Accurately detecting the orientation of the slip band traces in each grain, as any reasonable measurement of their distribution must be taken perpendicular to them.

(ii) Quantifying the distribution of the slip bands with a particular orientation while ignoring the contrast of other features, such as second phases or non-coplanar slip bands.

(iii) Extracting real distances from the ones measured on the slip band projections at the sample surface.

Different techniques can be leveraged to overcome some of the above-mentioned challenges. High resolution digital image correlation (HR-DIC) and electron backscatter diffraction (EBSD) used in tandem in a scanning electron microscope (SEM), often employed to characterise slip band behaviour (Lunt et al., 2017; Jiang et al., 2017; Harte et al., 2020; Sangid et al., 2020; Stinville et al., 2020), provide all the geometrical information needed to assess their distribution. In spite of this, the quantification cannot be performed reliably mainly due to challenge (ii), as measurements for a given slip plane orientation become impossible due to the presence of other sources of contrast (the strain of slip bands with a different orientation in this case). Sperry et al. (Sperry et al., 2020) applied a Radon transfer to the HR-DIC dataset to automate the detection of slip bands, but manual verification still had to be done. Bourdin et al. (Bourdin et al., 2018) developed a Heaviside-DIC method that detects the discontinuities in the kinematic fields measured, and Charpaigne et al. (Charpaigne et al., 2020) further refined this technique by complementing it with Hough transforms and merging of contiguous and closely oriented steps, effectively capturing individual slip bands. However, this approach loses some spatial information (e.g. width of slip bands). Moreover, these approaches apply exclusively to DIC datasets. Alternatively, León-Cázares et al. (León-Cázares et al., 2020a) addressed the detection of slip bands by performing a fast Fourier transform (FFT) on ECCI micrographs, which effectively decomposes them into images each with the slip bands that share a single orientation. This enabled a more accurate quantification of slip band volume fractions in fatigue loaded nickel-based superalloys with different ageing conditions. However, the image processing was done manually, on just a few grains, on regions smaller than the grains and without the crystallographic information needed to overcome challenge (iii). Yet, this study showed the potential of FFT image decomposition, successfully employed in the past to deconvolute Kikuchi bands in Kikuchi diffraction patterns (Ram et al., 2014) and in other fields to measure varied feature shapes and orientations (e.g. Ayres et al., 2008; Davis et al., 2012; Malek et al., 2013), as a means to better characterise slip localisation.

The current work introduces a new automated technique to quantify the slip band distribution. The capabilities of this approach are demonstrated on a tensile loaded polycrystalline nickel-based superalloy. Challenges (i) and (ii) are tackled with a FFT that decomposes the strain data in a HR-DIC dataset into its components along different orientations, and challenge (iii) is addressed by incorporating the corresponding crystallographic data from EBSD. In return, this methodology outputs an extensive characterisation of slip band distribution across a sample. The results obtained provide a quantitative understanding of the slip localisation behaviour and new insights on the slip band formation and evolution in polycrystals.

Materials and methods

Material

This study was performed on the coarse-grained nickel-based superalloy RR1000, supplied by Rolls-Royce plc. Following powder processing to shape, the material was subject to a super-solvus solution treatment at 1170 °C/1 h, fan air cooled close to -1 °C/s, and then aged at 760 °C/16 h and static air cooled. The heat treatment implemented produces a uniform grain orientation distribution with mean planar grain diameter of 48.4 ± 18.2 µm including twins (obtained via EBSD). The microstructure is also known to correspond to a solid solution fcc γ matrix phase and two distinct families of the ordered Li2 γ′ phase, namely secondary and tertiary precipitates approximately 300 nm and 20 nm in size, respectively.

A dogbone tensile specimen was produced by electro-discharge machining, with a gauge length of 8 mm, width of 2 mm and thickness of 1 mm. Both surfaces of the sample were finished by grinding with 4000 grit paper. The surface for EBSD and HR-DIC image collection was further polished to 0.04 µm colloidal silica for 20 minutes.

Mechanical testing

An interrupted tensile test was performed on a 100 kN servo-hydraulic Instron machine at room temperature. A crosshead displacement rate of 8 × 10⁻³ mm/s was employed, which corresponds to a strain rate of 0.1 %/s, and the test was interrupted at a total strain of 2.1 %.

Full-field DIC

Full-field DIC to monitor the in-plane strains across the sample was performed on the reverse face to that used for EBSD and HR-DIC. A continuous white coating followed by a black speckle pattern, both of automotive flameproof paint, were applied via airbrush in multiple passes. Features were tracked in situ at 10 frames per second with an Imager E-Lite 1600 × 1200 px camera with a 100 mm fixed focal length lens, at a stand-off distance of 10 cm. This was enough to image the whole gauge length with a pixel size of 10 µm.

EBSD

An area of interest for EBSD and HR-DIC imaging of the coarse-grained nickel-based superalloy RR1000 was defined at the centre of the gauge width. EBSD mapping was performed at a working distance of 8.5 mm on a Zeiss GeminiSEM 300, equipped with an Oxford Instruments EBSD detector and Aztec software version 3, to determine the crystallographic orientation of each grain. Image acquisition and stage movement were automated across 20 frames of 700 × 525 px², with 10 %
overlap per frame, needed to cover the area of interest plus a further \( \sim 10\% \) padded area to ensure the HR-DIC dataset was fully contained. Acquisition was performed at 25 kV with a pixel size of \( p_{\text{EBSD}} = 0.5 \mu\text{m} \) and a dwell time of 4 ms/px.

**HR-DIC**

A surface pattern for HR-DIC imaging was applied through the reconstruction of a thin gold film following a method similar to that reported by Edwards et al. [Edwards et al., 2017] in TiAl and Di Gioacchino et al. [Di Gioacchino & Quinta da Fonseca, 2013] in an austenitic stainless steel. Sputtering was performed with an Emitech K550 sputter coater with a gold target, backfilled with argon to 0.1 atm and depositing at 20 mA for 25 s, resulting in a film approximately 2.5 nm thick. Subsequent remodelling was performed by heat treating at 300 °C/8 h in a water vapour atmosphere followed by a stabilisation heat treatment at 600 °C/24 h in fused silica ampoules under a vacuum of \( < 5 \times 10^{-5} \) mbar. This pattern was masked when applying the airbrush speckles for full-field DIC.

HR-DIC image collection was performed ex-situ before and after loading in the Zeiss GeminiSEM 300. This was operated at 15 kV with a four quadrant Si-drift backsaturated electron detector, found to give high atomic number contrast between the gold features and the superalloy substrate. To traverse the area of interest, the automated collection of approximately \( 400 \times 12 \) frames was performed with the SmartStitch (Zeiss) software with 10 % overlap by moving the stage in a rectangular grid, defining a focal plane non coincident with the imaged plane. Individual frames had a field of view of 25.8 \( \times \) 19.4 \( \mu\text{m}^2 \) with a pixel size of 12.6 nm, compared to the feature size of \( \sim 30-80 \) nm. Images were recorded using a 60 \( \mu\text{m} \) aperture, with a dwell time of 13 ms/px and no noise reduction to 8 bit depth, attempting to match the brightness, contrast and working distance in pre- and post-test images. Fig. 1(a) shows an example of the speckle pattern recorded, where the feature sizes range between 3-8 pixels. A histogram of the pixel intensities across the area of interest is plotted in Fig. 1(b), indicative of a good contrast for image acquisition.

Individual HR-DIC images were stitched together with the FIJI stitching plugin, computing the optimal overlap, and employing linear blending in the overlap region. Stitching lines were not visible in this process. Owing to image correlation software memory limitations, the stitched images were divided into 8000 \( \times \) 18000 px matched pre- and post-test frames, forming a grid of \( \sim 100 \times 1 \) frames with 10 % overlap covering the region of interest. Correlation was performed via least squares subset matching with the DaVis 8.4.0 (LaVision) software using a 21 \( \times \) 21 px\(^2\) subset size, 8 px step size between subsets and fitting second order deformation. Such image correlation routines output the \( u_i \) \( (i = 1, 2) \) in-plane components of the displacement tensor across the defined area. The in-plane strains can then be calculated as

\[
\varepsilon_{i i} = \frac{d u_i}{d x_i}
\]  

and

\[
\varepsilon_{12} = \frac{1}{2} \left( \frac{d u_1}{d x_2} - \frac{d u_2}{d x_1} \right). \tag{2}
\]

Results were visualised by plotting the maximum in-plane shear strain component, defined as

\[
\gamma_{\text{max}} = \sqrt{\left( \frac{\varepsilon_{11} - \varepsilon_{22}}{2} \right)^2 + \varepsilon_{12}^2}. \tag{3}
\]

The pixel size used to match the distance between correlated displacement vectors, i.e. 8 raw image pixels, corresponds to \( p_{\text{DIC}} = 100.7 \mu\text{m} \). The resulting strain fields were then merged again using the stitching plugin of FIJI to obtain a strain map of the full area.

To assess the non-random stitching artifacts, a small region without deformation was imaged and correlated by HR-DIC using the same steps. The largest strain variations introduced in this control test had a magnitude of \( \sim 0.01 \). This is much lower than the strains detected along the slip bands (see the Results section), which proves the efficacy of the method at resolving these deformation features.

**FFT decomposition method**

The technique introduced in this section, denoted Fast Fourier Transform Decomposition (FFTD) (open source code available at [source code](https://github.com/ferleoncazares/Slip-bands-FFT-decomposition)), makes use of the FFT to isolate the slip band information contained in the HR-DIC dataset. This allows for the decomposition of the contrast arising from differently oriented sets of slip bands. Table I summarises the steps followed. The corresponding scripts were coded into Matlab R2021a, making use of the image processing toolbox (IPT), the signal processing toolbox (SPT) and

![Image](https://via.placeholder.com/150)

Fig. 1.: (a) Speckle pattern at imaging conditions showing the characteristic feature size. The box shows the subset size for image matching. (b) Histogram of pixel intensities across the area of interest, which contains few saturated pixels.

### Table 1: Steps followed for FFTD

<table>
<thead>
<tr>
<th>Step</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.</td>
<td>Acquire HR-DIC images</td>
</tr>
<tr>
<td>2.</td>
<td>Stitch images using FIJI</td>
</tr>
<tr>
<td>3.</td>
<td>Correlate images using DaVis</td>
</tr>
<tr>
<td>4.</td>
<td>Calculate in-plane strains</td>
</tr>
<tr>
<td>5.</td>
<td>Apply FFT decomposition</td>
</tr>
</tbody>
</table>

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*Source code for the FFTD technique: [https://github.com/ferleoncazares/Slip-bands-FFT-decomposition](https://github.com/ferleoncazares/Slip-bands-FFT-decomposition)*
Table 1. Summary of the preprocessing and FFT decomposition steps.

<table>
<thead>
<tr>
<th>Preprocessing</th>
<th>Inputs: HR-DIC and EBSD datasets</th>
<th>Outputs: ( I^B ) and ( I^G ) images of each grain</th>
</tr>
</thead>
<tbody>
<tr>
<td>1. Build binary and greyscale versions of DIC dataset</td>
<td>1. Perform FFT ( \rightarrow ) ( I^{FFT_B} ) and ( I^{FFT_G} )</td>
<td></td>
</tr>
<tr>
<td>2. Merge EBSD and DIC datasets</td>
<td>2. Adjust contrast on ( I^{FFT_B} ) ( \rightarrow ) ( I^{FFT_d} )</td>
<td></td>
</tr>
<tr>
<td>3. Identify individual grains from EBSD dataset</td>
<td>3. Detection of active slip plane trace orientations:</td>
<td></td>
</tr>
<tr>
<td>4. Smooth and denoise grain boundaries</td>
<td>4. Build masks of each grain with point in polygon algorithm</td>
<td></td>
</tr>
<tr>
<td>5. Topological analysis:</td>
<td>5. Topological analysis is performed to determine which grains</td>
<td></td>
</tr>
<tr>
<td>• Sort grains by size</td>
<td>are fully contained within another grain as seen on the surface</td>
<td></td>
</tr>
<tr>
<td>• Compute rectangles enclosing each grain</td>
<td>imaged. This will be helpful later, considering the large amount of</td>
<td></td>
</tr>
<tr>
<td>• Identify candidate grains that may be within a larger grain</td>
<td>twins in the material, to determine which pixels within the bound-</td>
<td></td>
</tr>
<tr>
<td>• Point in polygon algorithm in candidate grains to determine topology</td>
<td>ary of a grain must be ignored when performing the slip band analysis. The</td>
<td></td>
</tr>
<tr>
<td>6. Removal of DIC pixels near grain boundaries:</td>
<td>grains are firstly sorted by size, and the rectangles</td>
<td></td>
</tr>
<tr>
<td>• Boundary detection algorithm to remove outer pixels, repeat ( w^- ) times</td>
<td>that contain them are computed from the minimum and maximum</td>
<td></td>
</tr>
<tr>
<td>• Extend boundaries of inner grain boundaries and remove pixels, repeat ( w^+ ) times</td>
<td>( x ) and ( y ) coordinates of their pixels in the EBSD dataset. The</td>
<td></td>
</tr>
<tr>
<td>7. Crop DIC datasets to individual grains</td>
<td>grains with boxes that fully lie within that of a larger grain are</td>
<td></td>
</tr>
<tr>
<td>8. Build masks of each grain with point in polygon algorithm</td>
<td>considered candidates. A point in polygon algorithm (Hormann &amp; Agathos</td>
<td></td>
</tr>
<tr>
<td>9. Apply masks to cropped binary images and skeletonise</td>
<td>2001) (inpolygon function in Matlab) is then used for an arbitrary</td>
<td></td>
</tr>
<tr>
<td>10. Apply masks to cropped greyscale images</td>
<td>pixel of each of the candidate grains to determine whether</td>
<td></td>
</tr>
</tbody>
</table>

FFT decomposition

<table>
<thead>
<tr>
<th>Inputs: ( I^B ) and ( I^G ) images of each grain</th>
<th>Outputs: ( I^{FFT(b)} ), slip band trace locations and angles</th>
</tr>
</thead>
<tbody>
<tr>
<td>1. Perform FFT ( \rightarrow ) ( I^{FFT_B} ) and ( I^{FFT_G} )</td>
<td>1. ( I^{FFT(b)} ) ( \rightarrow ) ( I^{FFT_d} )</td>
</tr>
<tr>
<td>2. Adjust contrast on ( I^{FFT_B} ) ( \rightarrow ) ( I^{FFT_d} )</td>
<td>2. Detection of active slip plane trace orientations:</td>
</tr>
<tr>
<td>3. Detection of active slip plane trace orientations:</td>
<td>3. Peak detection algorithm to determine the angles ( \theta_{s(b)} ) of b</td>
</tr>
<tr>
<td>• Build spectrum of angular intensity in ( I^{FFT_d} )</td>
<td>bright bands</td>
</tr>
<tr>
<td>• Peak detection algorithm to determine the angles ( \theta_{s(b)} ) of b</td>
<td></td>
</tr>
<tr>
<td>4. Crop ( I^{FFT_d} ) spectra for each bright band ( \rightarrow ) ( I^{(b)} )</td>
<td>4. Crop ( I^{FFT_d} ) for each bright band ( \rightarrow ) ( I^{FFT(b)} )</td>
</tr>
<tr>
<td>5. Perform IFFT on ( F^{(b)} ) and apply mask ( \rightarrow ) ( F^{FFT(b)} )</td>
<td>5. Perform IFFT on ( F^{(b)} ) and apply mask ( \rightarrow ) ( I^{FFT(b)} )</td>
</tr>
<tr>
<td>6. Slip band detection:</td>
<td>6. Interpolate intensities in ( I^{FFT(b)} ) along search line at angle</td>
</tr>
<tr>
<td>• Interpolate intensities in ( I^{FFT(b)} ) along search line at angle</td>
<td>( \theta_{q(b)} ) perpendicular to the slip bands</td>
</tr>
<tr>
<td>• Peak detection algorithm to detect slip bands</td>
<td>( \theta_{q(b)} )</td>
</tr>
</tbody>
</table>

The MTex 5.1.0 crystallographic toolbox (Hielscher & Schaeben 2008; Bachmann et al. 2011). Relevant built-in functions used are referenced throughout the text.

**Image preprocessing**

A series of steps are needed to combine the EBSD and HR-DIC datasets in a meaningful way. The objective of the preprocessing stage is to create images of individual grains with HR-DIC strain-related data, each linked to their corresponding EBSD crystallographic information. Fig 2 shows the different preprocessing steps.

Firstly, the EBSD and HR-DIC datasets need to be merged. Note that the sizes of the regions imaged by each technique are different, and so are their pixel sizes (denoted \( p_{EBSD} \) and \( p_{DIC} \), respectively). For each EBSD frame, the coordinates of approximately 30 triple junctions are manually collected in both coordinate systems. These are used to calculate the affine transformation and shift of the EBSD dataset giving the least squares differences to match those in the HR-DIC image (Atkinson et al. 2020; Harte et al. 2020). A maximum distance error of 6.8 EBSD pixels was obtained in the current study, and stitching lines were invisible without the need of smoothing filters. The grains, calculated using a threshold of 5° in neighbouring pixels, are then smoothed and denoised (smooth function in MTex) to better delimit their boundaries. Fig 2(a) shows a region of the DIC image with superimposed grain boundaries calculated from the EBSD dataset.

A topological analysis is performed to determine which grains are fully contained within another grain as seen on the surface imaged. This will be helpful later, considering the large amount of twins in the material, to determine which pixels within the boundary of a grain must be ignored when performing the slip band analysis. The grains are firstly sorted by size, and the rectangles that contain them are computed from the minimum and maximum \( x \) and \( y \) coordinates of their pixels in the EBSD dataset. The grains with boxes that fully lie within that of a larger grain are considered candidates. A point in polygon algorithm (Hormann & Agathos 2001) (inpolygon function in Matlab) is then used for an arbitrary pixel of each of the candidate grains to determine whether they are inside of the larger grain. This is illustrated in Fig 2(b).

The DIC data near the boundary of each grain is removed from the analysis because of two reasons. Firstly, to avoid including pixels of neighbouring grains that may have been misplaced due to errors during the merging of the EBSD and DIC datasets. Secondly, because strain localisation behaviour and lattice rotation near grain boundaries vary substantially from those in the bulk of grains, and we are mainly concerned with the latter. To achieve this, a boundary detection algorithm (boundary function in Matlab) is performed on the EBSD data of the grain of interest, and the corresponding pixels are excluded. This step is repeated multiple times, for a combined width of \( w^- \) DIC pixels removed. Similarly, pixels within a layer of width \( w^+ \) extended outwards from the grain boundaries of grains in the interior (polybuffer function in Matlab) are excluded. The removal of these layers of pixels in a grain is exemplified in Fig 2(c).

Individual pictures can be built for each grain for the subsequent image analysis. A HR-DIC micrograph is cropped into the smallest rectangle of \( X \times Y \) pixels that encloses the remaining EBSD data points of the grain of interest, as shown in Fig 2(d), setting the new \( x \) and \( y \) coordinates of the bottom left pixel as \((1,1)\). Masks \( I^{mask} \in \{0,1\}^{X\times Y} \) are constructed such that its elements \( x = 1, \ldots, X \) and \( y = 1, \ldots, Y \) are

\[
I^{mask}_{x,y} = \begin{cases} 
1 & \text{if pixel (x,y) belongs to the grain of interest} \\
0 & \text{otherwise}
\end{cases}
\]

These pixels are identified by first detecting on the EBSD data the boundaries of that grain and all grains inside of it, and then performing the point in polygon algorithm for every pixel of the cropped DIC image and each of these boundaries. The resulting mask can be applied to any image, i.e. multiplying the intensity of each of its pixels by the homologous one in \( I^{mask} \). This can be done for any version of the HR-DIC figure, such as for different strain components or colouring formats (binary, greyscale or coloured).
Two versions of the DIC dataset are used in the following analysis. Firstly, a binary (pixel intensities of 0 or 1) image is created by setting a strain threshold \( \gamma_{th} \) (maximum in-plane total shear strain in this work) such that many slip bands are evident and most regions in-between show no deformation. Afterwards, this image is skeletonised (bwskel function in IPT), a standard operation that turns shapes, slip bands in the present case, into thin lines equidistant to their boundaries. Secondly, a greyscale (pixel intensities from 0 to 1) image is built such that pixels where the strains are above threshold \( \gamma_{th} \) have an intensity of 1, and these linearly decrease until an intensity of 0 for no strain. The masks \( I^m \) for each grain are then applied to these two versions of the DIC dataset to exclude the pixels that do not belong to them. This results in skeletonised and greyscale images of individual grains, such as those in Fig. 2(e) and (f), respectively.

**FFT decomposition**

The FFTD technique enables the detection of slip bands from the images generated in the preprocessing stage. Fig. 3 shows the steps followed to achieve this. Two-dimensional FFTs \( \mathcal{F} \) (fft2 function in Matlab) are applied to both the binary skeletonised \( I^B \in [0,1]^{X \times Y} \) (Fig. 3(a)) and the greyscale \( I^G \in [0,1]^{X \times Y} \) (Fig. 3(b)) images of individual grains; the former will later be employed to discern slip band orientations and the latter to detect slip bands. Such transforms \( \mathcal{F} \{ I^B \} = I^{FFT_B} \) and \( \mathcal{F} \{ I^G \} = I^{FFT_G} \) of images in the space domain result in images \( I^{FFT_B} \in \mathbb{C}^{X \times Y} \) and \( I^{FFT_G} \in \mathbb{C}^{X \times Y} \) that contain the frequency components instead. Note that the intensity of pixel \((x,y)\) for a given transform \( \mathcal{F} \{ I \} = I^{FFT} \) is computed as

\[
I^{FFT}_{x+1,y+1} = \sum_{j=0}^{X-1} \sum_{k=0}^{Y-1} \omega_X^j \omega_Y^k I_{j+1,k+1},
\]

where \( \omega_X \) and \( \omega_Y \) are complex roots of unity

\[
\omega_X = e^{-2\pi i / X} \quad \text{(6a)}
\]
\[
\omega_Y = e^{-2\pi i / Y}, \quad \text{(6b)}
\]

and the zero-frequency component is then shifted to the centre of the spectrum (fftsift function in Matlab) at point \((x_0, y_0)\) by swapping the first and third quadrants of \( I^{FFT} \) with the second and fourth.

The search for active slip planes is developed on an additional version \( I^{FFTd} \in \mathbb{R}^{X \times Y} \) of the frequency spectrum. The intensity of each of its pixels is computed as

\[
I^{FFTd}_{x,y} = \log_2 \left( I^{FFT_B}_{x,y} \right). \quad \text{(7)}
\]

The absolute value turns the pixel intensities to real numbers, as opposed to complex in \( I^{FFT_B} \), and the logarithm serves to make the range of pixel intensities narrower, which helps for image readability. An example of this FFT is shown in Fig. 3(c). Each of the bright bands in the frequency domain captures a set of parallel slip band projections in the space domain.

A line search algorithm is implemented to spot the orientation of the brightest bands in \( I^{FFTd} \). Pixel intensities \( z_s(r_s, \theta_s) \) at radii \( r_s \) and angles \( \theta_s \) are linearly interpolated from \( I^{FFTd} \) within a circle with radius of

\[
R_s = \text{floor} \left( \frac{\min(X, Y)}{2} - 1 \right) \quad \text{(8)}
\]

pixels centred at \((x_0, y_0)\), where the floor function rounds the value towards the nearest smaller integer and \( \min \) is the minimum value function. The interpolation points are spaced one DIC pixel and \( \Delta \theta_s \) degrees apart, for angles between \( 0^\circ \leq \theta_s < 180^\circ \).
Fig. 3.: Steps of the FFTD shown for the same grain used in Fig. 2 (a) Binary skeletonised and (b) greyscale images of a grain. (c) Two-dimensional FFT of the binary image and (d) angular variation of the mean intensity with two peaks marked that correspond to the bright bands in the FFT, shown in (e). (f,g) IFFT performed on the two cropped versions of the FFT of the greyscale image (shown in the inserts), one for each active slip plane orientation. The search lines (cyan) and slip bands (red) detected from the (h,i) intensity profiles along them are displayed. The contrast of all greyscale images was adjusted for better visualisation.

The cartesian coordinates of the interpolation points are thus 
\[ (r_s \cos \theta_s + x_0, r_s \sin \theta_s + y_0). \]

The mean intensity \( z_s(\theta_s) \) of interpolation points along one of these lines is computed as

\[ z_s(\theta_s) = (2R_s + 1)^{-1} \sum_{r_s = -R_s}^{R_s} z_s(r_s, \theta_s). \]  

An example of the angular variation of the mean intensity is shown in Fig. 3(d). A peak detection algorithm (findpeaks function in the SPT) with set minimum peak height \( z_{s_{\text{min}}} \) and minimum peak prominence \( z_{s_{\text{rel}}} \) is used on this spectrum to identify the angles \( \theta_{sb} \) of the \( B \) brightest bands. This is shown back in \( I_{\text{FFTd}} \) in Fig. 3(e). It was found that adding to \( I_{\text{FFTd}} \) a two-dimensional Gaussian smoothing operation (imgaussfilt function in the IPT) with standard deviation \( \sigma_S \) before performing the pixel intensity interpolations reduces the noise of the \( z_s(\theta_s) \) spectrum, making the peak detection more consistent.

Excluding the pixels far away from a line at angle \( \theta_{sb} \) crossing through \((x_0, y_0)\) in a FFT image effectively isolates the frequency components along that orientation. The distance \( d^{(b)}(x, y) \) to that line can be calculated for a pixel situated at \((x, y)\) as

\[ d^{(b)}(x, y) = \frac{|m_s^{(b)}(x - x_0) - (y - y_0)|}{\sqrt{(m_s^{(b)})^2 + 1}} \]  

where \( m_s^{(b)} = \tan \theta_s^{(b)} \) is the slope of the line. For each bright band \( b \) in \( I_{\text{FFT,G}} \), a cropped image \( I^{(c)(b)} \in \mathbb{C}^{X \times Y} \) is constructed such that

\[ I_{x,y}^{(c)(b)} = \begin{cases} I_{\text{FFT,G}}^{(b)} & \text{if } d^{(b)}(x, y) \leq d^*/2, \\ 0 & \text{otherwise}. \end{cases} \]

Performing an inverse fast Fourier transform (IFFT) \( \mathcal{F}^{-1} \) (ifft2 function in Matlab) to these cropped spectra and applying the grain binary mask generates greyscale images \( I_{\text{FFT}}^{(b)} \in \{0, 1\}^{X \times Y} \) with slip bands of a single orientation that resemble those in the original image, i.e. \( I_{\text{FFT}}^{(b)} = \mathcal{F}^{-1}(I^{(c)(b)}) \circ I_{\text{mask}} \) where \( \circ \) denotes an element-by-element matrix multiplication. The intensities of the individual pixels are calculated as

\[ I_{x,y}^{(c)(b)} = \left| \frac{1}{X} \sum_{j=1}^{X} \frac{1}{Y} \sum_{k=1}^{Y} \omega_X^{(j-1)(x-1)} \omega_Y^{(k-1)(y-1)} I_{x,y}^{(c)(b)} \right| I_{x,y}^{(c)(b)} \]

\[ I_{x,y}^{(c)(b)} \cdot \]  

(12)
Fig. 3(f,g) show the IFFT images computed from the two cropped spectra (displayed in the inserts), each for a different bright band in FFTd. Note that slip bands in the IFFT images may project beyond the grain boundaries if the binary mask is not reapplied (further details in the [Discussion] section). While this would not affect the detection of slip bands, the mask is reapplied here to facilitate the measurement of slip band spacings.

**Measurement of slip band spacings**

Slip bands can readily be identified from the IFFT images. Slip band volume fractions can be measured by selecting an intensity threshold, as done in reference [Leon-Cázares et al. 2020]. Instead, we focus our current efforts in quantifying the slip band spacings. For each bright band in a FFT spectrum there is an associated line in the space domain perpendicular to a set of slip band projections that share a particular orientation. Thus, slip bands can be detected from the corresponding IFFT(b). This is performed along the search line at an angle

\[ \theta_q^{(b)} = \arctan \left( \frac{X}{Y \tan \theta_s^{(b)}} \right) , \]

perpendicular to the slip bands and crossing through \((x_0, y_0)\), although any other reference point could be chosen. The length in pixels \(2R_q^{(b)} + 1\) of that search line is chosen to ensure that it goes through the whole image. Intensities \(I_q^{(b)}(r_q^{(b)})\) at positions \(r_q^{(b)} = -R_q^{(b)}, -R_q^{(b)} + 1, \ldots, R_q^{(b)}\) spaced one DIC pixel apart along that line are linearly interpolated from \(I^{FFT(b)}\), producing profiles like those in Fig. 3(h,i). Another peak detection algorithm, with set minimum peak half-width \(w_{\text{min}}\), minimum peak prominence \(w_{\text{rel}}\) and minimum peak separation \(r_{\text{min}}\), is then used to locate the positions \(q_s^{(b)}(s = 1, \ldots, S^{(b)})\) of the \(S^{(b)}\) high intensity peaks detected along line \(b\), which correspond to individual slip bands. The actual coordinates \((x_q^{(b)}, y_q^{(b)})\) of the points found are

\[ x_q^{(b,s)} = r_q^{(b,s)} \cos \left( \theta_q^{(b)} \right) + x_0 \quad \text{(14a)} \]
\[ y_q^{(b,s)} = r_q^{(b,s)} \sin \left( \theta_q^{(b)} \right) + y_0 \quad \text{(14b)} \]

with pixel intensities \(I_q^{(b,s)}\). Note that these intensities are directly related to the HR-DIC strains used to build the greyscale image \(I^G\), in-plane shear strains in this case. Such strains can be recovered as

\[ \gamma_{SB}^{(b,s)} = z_q^{(b,s)} \gamma_{\text{th,G}} , \]

for the physical meaning of which is further explained in the Discussion section.

With this data, the slip band spacings for each slip plane orientation spotted can be quantified. The peak-to-peak spacing between two adjacent slip bands are calculated as

\[ \lambda^{(b,s)} = (x_q^{(b,s+1)} - x_q^{(b,s)}) \sin \alpha^{(b)} \text{DIC} \quad \text{(16)} \]

where \(\alpha^{(b)}\) is the angle between the sample surface (neglecting local height variations due to deformation) and slip plane \(b\), as long as there are no pixels in-between that do not belong to the grain analysed (in which case this spacing is ignored). Similarly, the mean linear slip band spacing (inverse to the linear slip band density) is

\[ \lambda^{(b)} = \frac{I^{(b)} \sin \alpha^{(b)} \text{DIC}}{S^{(b)}} , \]

where \(I^{(b)}\) is the number of pixels that fall within the grain analysed along the search line \(b\). The value of \(\lambda^{(b)}\) is slightly different and more accurate than that obtained by computing the mean value of all \(\lambda^{(b,s)}\) in equation (15), which would not fully consider the slip bands nearest to the grain boundaries.

Finally, the orientation information can be recovered from the EBSD data. The angles of the slip band traces

\[ \theta_{SB}^{(b)} = \theta_q^{(b)} + 90^\circ \]

found via the FFT method are matched to those of octahedral planes predicted with the EBSD dataset [Gibson et al. 2021] by selecting the permutation that adds the smallest angle difference when considering all \(B\) active slip planes detected. Note that these differences must consider the cyclic nature of angles, e.g., a line at \(-90^\circ\) is equivalent to one at \(90^\circ\), and so forth. Once assigned, the \(\alpha^{(b)}\) angles are related to the \(z\)-component \(z_q^{(b)}\) (in-and-out of the sample surface) of the corresponding slip plane unit normals as \(\sin \alpha^{(b)} = \sqrt{1 - z_q^{(b)}^2}\). This is all performed using the mean grain orientations.

**Parameter definition**

This section outlines the parameters of the image preprocessing and FFTD method.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Width of outer boundary excluded</td>
<td>(w^-) 5 (\text{PEBSD})</td>
</tr>
<tr>
<td>Width of inner boundary excluded</td>
<td>(w^+) 3 (\text{PEBSD})</td>
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<td>Threshold for greyscale image</td>
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<tr>
<td>FFT spectrum minimum peak prominence</td>
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<td>Width of cropped FFT</td>
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<td>IFFT profile minimum peak height</td>
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<tr>
<td>IFFT profile minimum peak prominence</td>
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<tr>
<td>IFFT profile minimum peak separation</td>
<td>(r_{q_{\text{min}}}) 3 (\text{PEDIC})</td>
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</tbody>
</table>

This table displays such parameters. Grains with less than two EBSD pixels were excluded from the beginning of the preprocessing stage because their effects on the slip band analysis were deemed negligible. A minimum of 1000 EBSD pixels, which corresponds to an area of \(~250 \mu m^2\), was imposed after performing the grain topology analysis to determine which grains would be analysed. This threshold, set because the detection of bright bands in the FFT image becomes less consistent for smaller grains, was imposed both before and after removing the grain boundary layers \((w^-\) and \(w^+)\).

For the slip band distribution analysis, the Gaussian smoothing of \(I^{FFTd}\) drastically minimises the false positive detections of grain boundary layers (\(~250 \mu m^2\)).
bright bands, although nearby peaks become more likely to merge. Thus, \( \sigma_S \) was kept to a small value of 1 \( \rho_{PC} \). The FFT peak detection parameters were first chosen so that bright bands apparent to the naked eye were always detected, and the minimum peak prominence was set to avoid false positives from noise in the FFT. The width of the cropped FFT spectra was selected to produce IFFTs that resembled the original binarised images; a deeper analysis on the effects of this parameter are given in the [Discussion] section. The peak detection parameters to identify slip bands were chosen accounting for the background noise produced in the IFFT images. A value of 0.0457 was chosen for the minimum peak height, which corresponds to a strain of 0.016 in the original greyscale image, and the minimum peak prominence was adjusted manually to minimise the number of false positives encountered while including slip bands observed.

### Results

The stress-strain curve of the alloy tested is plotted in Fig. 4. This shows that a plastic strain of about 1.64% was achieved at the highest extension, but the strain is distributed heterogeneously across the sample. Fig. 5(a) shows that the axial strain, obtained via full-field DIC, is distributed evenly across the gauge length of the sample, but has a degree of heterogeneity at the sub-millimetre length scale. However, part of the strain fluctuations observed must be attributed to noise associated to DIC, e.g. from lens aberrations. The EBSD of the area of interest, enclosed in a rectangle in Fig. 5(a), is shown at a higher magnification in Fig. 5(b). Note that this area is considerably larger than that used in most HR-DIC studies. A subregion of this area is displayed at a larger magnification in Fig. 5(c), where individual grains and twins can be discerned. The HR-DIC image of this subregion, Fig. 5(d), shows that most of the plastic strain in the material is accumulated in slip bands, with a clearly heterogeneous distribution across grains. It is this slip band distribution which our new FFTD technique quantifies.

The above analysis returned five instances of active slip planes detected with the FFTD method resembling those of octahedral planes calculated with the EBSD data \( \theta_{EBSD} \), but there are some deviations. The angle differences \( \Delta \theta = \theta_{SB} - \theta_{EBSD} \) show a mean absolute error of 2.1\(^\circ\) and some individual deviations above 9\(^\circ\). The highest angle differences per grain, plotted in Fig. 6(b) as a function of the crystallographic direction in-and-out of the sample surface in an inverse pole figure (IPF), show systematic deviations for particular grain orientations. For example, large positive errors consistently appear for orientations between \( \langle 001 \rangle \) and \( \langle 011 \rangle \). Additionally, upon visual inspection, slip plane traces drawn with the angles obtained via FFTD closely follow the slip bands in the micrographs. These hints indicate that there was a misalignment when obtaining the EBSD data. Thus, we performed an optimisation algorithm to determine the rotations by the Euler angles along the \( z \), \( x \) and \( z \)-axes that the crystals as detected by EBSD would have to undergo to minimise the angle differences. A small rotation of 4\(^\circ\) resulted in the angle differences shown in Fig. 7(b), with a mean absolute error of 0.55\(^\circ\) and 99% of the individual angle differences below 2.5\(^\circ\). Fig. 7(c) shows the histograms of the angle differences for the original and optimised orientations, indicating a considerable improvement. Moreover, Fig. 7(d) plots the mean angle differences and standard deviations for the primary, secondary and tertiary slip planes. The errors are clearly lower on the former, indicating that the FFT band detection algorithm improves when more slip bands are visible.

The above analysis returned five instances of active slip planes with traces that did not match any of those predicted by EBSD. Further inspection revealed that one of those false positives was due to the detection of slip bands from a neighbouring grain that the preprocessing grain boundary removal operation \( \setminus w \) did not eliminate. Two more were detections of slip bands in a small twin within the grains of interest, with clear signs of slip band transmission through the twin boundaries, that was not indexed by the EBSD. Another case occurred due to high strains near a grain boundary above the threshold \( \gamma_{th,b} \), which introduced an artifact in the skeletonised version of the grain image. Lastly, a false positive was obtained in the grain shown in Fig. 8. The bands indicated correspond to what is often referred to as wavy slip, a zig-zag pattern that arises from the sequential activation of dislocations on two slip planes. This may affect the slip band detection algorithm only in Fig. 6(a), with the majority displaying two distinct ones. Fig. 6(b) shows the distribution of slip bands based on the type of slip plane, where that associated with the brightest band in each FFT image is denominated primary slip plane, the second brightest secondary, and so on. This demonstrates that brighter bands in FFT images are the result of higher numbers of slip bands with a shared orientation.

<table>
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<th>Step</th>
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<th>Time [s]</th>
</tr>
</thead>
<tbody>
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<td>Grain calculations and denoising</td>
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<td>445</td>
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<tr>
<td>Grain topology</td>
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<td>49</td>
</tr>
<tr>
<td>Removing grain boundary pixels</td>
<td>527</td>
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<td>Building grain images</td>
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<td>1240</td>
</tr>
<tr>
<td>FFT decomposition</td>
<td>365</td>
<td>240</td>
</tr>
</tbody>
</table>

Table 3: Computing times and numbers of grains of the pre-processing and FFTD steps.
Fig. 5: (a) Full-field DIC displaying the axial strain at the highest extension of the sample. (b) Pre-test EBSD of the area of interest (colour coded according to the crystallographic direction normal to the sample surface), and (c) EBSD and (d) post-test HR-DIC of a subregion shown at larger magnification. The full-field DIC was performed on the opposite face to that used for EBSD and HR-DIC.

Fig. 6: Numbers of (a) grains with 0, 1, 2, 3 and 4 active slip plane orientations as detected from the bright bands in the FFT images, and (b) slip bands detected in primary, secondary, tertiary and quaternary slip planes. The individual counts are shown above each bar.

when the period of the zig-zag pattern is close to the pixel size; otherwise, the FFTD method is capable of discerning the active slip planes (see Fig. 3). These false positives were removed from all the quantitative analyses in the current work.

The FFTD algorithm distinguishes closely oriented slip band traces from the FFT images. The minimum angular separation detected was 7.1°. However, it is not clear whether slip band traces closer than that can be accurately differentiated. Fig. 9 shows the minimum angular difference between the traces of all octahedral planes in an fcc crystal calculated as a function of the surface normal direction. Two non-coplanar slip planes develop similarly oriented traces when the surface normal falls near the \( \langle 001 \rangle \) or \( \langle 111 \rangle \) edges of the IPF. However, from the 10 grains with the closest slip band traces found, also plotted in Fig. 9, only two had a surface normal further than 10° away from a \( \langle 001 \rangle \) orientation, for which there are two pairs of octahedral planes with equal traces. This occurs because the two slip planes that share a trace orientation possess active slip systems with similar Schmid factors, which happens more consistently for orientations in this region of the IPF. Fortunately, in grains with a \( \langle 001 \rangle \) surface normal, those two slip planes are also positioned at the same angle \( \alpha \) from the sample surface. Thus, any FFT band detection error would affect the true slip band spacings measured (equations (16) and (17)) to a small extent. Measurements will only be affected largely if both planes develop slip bands such that their traces cannot be distinguished.

A mean slip band spacing of 1.85 µm was measured across all primary slip planes of the grains analysed, but individual values varied considerably. Fig. 10(a) shows the distribution of mean spacings across grains calculated with equation (17), although it must be taken into account that some grains have considerably more slip bands than others. More defined profiles that resemble log-normal distributions appear in Fig. 10(b) when considering...
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Fig. 7: FFTD and EBSD trace angle differences shown as a function of the surface normal for the (a) original and (b) optimised orientations. (c) Histograms of the angle differences, and (d) their means and standard deviations (half error bar) along primary, secondary and tertiary planes for the original and optimised orientations.

Fig. 8: Grain where a false positive due to wavy slip was identified. The arrows point at two deformation bands with the characteristic zig-zag pattern of wavy slip. The contrast of the image was adjusted for better visualisation.

the 5756 individual spacings measured with equation (16) between contiguous slip bands.

The large number of slip bands detected allows us to analyse the dependence of the spacings on other variables. For instance, a large scatter in slip band spacings is immediately observed for smaller grain sizes in Fig. 10(c). The dashed line, which represents a linear regression taking into account all the spacings measured (not only the mean values per grain), has a positive slope with a high statistical significance (low p-value). This indicates that, on average and despite the scatter, shorter spacings are achieved in smaller grains down to the size threshold used in this study.

Fig. 9: Minimum trace angle difference between octahedral planes of an fcc crystal as a function of the surface normal. The 10 grains with the more closely aligned slip band orientations detected are plotted.
Fig. 10. Histograms of the (a) mean slip band spacings per grain and (b) individual spacings along different slip planes. Mean slip band spacings on primary planes as a function of (c) grain area, (d) Schmid factor, (e) Taylor factor and (f) loading direction. The dotted lines in (c), (d) and (e) correspond to linear regressions taking into account all the spacings measured, with the equations and p-values shown. The strain tensor assumed to calculate the Taylor factors is indicated in (e), with $\nu = 1/3$. Larger marker sizes in (f) correspond to bigger grains.

Fig. 10(d) explores the effect of the uniaxial loading Schmid factor (using the highest value for the corresponding slip plane). As expected, low values result in more widely spaced slip bands because dislocations experience lower shear stresses. However, unlike in the previous case, a linear regression shows a very poor statistical significance. A comparably low correlation is shown in Fig. 10(e) with respect to the Taylor factor. This means that there is no clear dependence of the slip band spacing on any of these geometric parameters. A similar lack of trend is obtained if plotted against other stress components (e.g. Escaig or normal stresses (León-Cázares & Raï, 2020)). It is also hard to distinguish any pattern when plotted against the grain orientation, as seen in Fig. 10(f).

A similar analysis can be performed on the slip band strains measured with the FFTD approach. The distribution of mean slip band strains $\gamma_{SB}$ per grain and plane orientation as detected from the IFFTs is shown in Fig. 11(a). These resemble Gaussian distributions where, as expected, strains are larger on primary slip planes. Note that no values under 0.016 appear due to the minimum peak height chosen for the slip band detection algorithm. If such low strain slip bands were considered, these distributions would shift towards lower values. Fig. 11(b) shows the distribution of slip band strains measured across the whole sample. All the distributions have a large number of low strain slip bands, but those on secondary and tertiary planes decrease more steeply. We refrain from showing additional dependences of the recorded strains by reminding the reader that these are only in-plane strains. True three-dimensional values could only be recovered if the main Burgers vector of the dislocations that form each slip band was known.

Discussion

The results of the methodology introduced offer a new picture of how the plastic deformation, localised in the form of slip bands, is distributed across the grains of a polycrystal. However, the results obtained are not definite properties of the material, which are impossible to obtain because of two main reasons. Firstly, the detection of slip bands is affected by the resolution of the HR-DIC, and the FFTD and peak detection parameters. Edwards et al. (Edwards et al., 2017) showed by performing DIC of an area with different subset sizes that what appears to be a single slip band may actually be a collection of thinner bands. Secondly, even if a perfect resolution and slip decomposition could be achieved, the very concept of slip band is often ill-defined. For instance, it is not clear when some form of planar slip can be named a slip band; it could be a single dislocation pile-up, two of them in close proximity to each other (which still requires arbitrarily selecting a maximum separation distance) or more. Similarly, it is not always clear where a slip band ends and another one begins (León-Cázares et al., 2020), so the same dislocation structure could be classified as two thin slip bands or as a single thicker one. Despite these caveats,
we argue that the FFTD does capture important aspects of the slip localisation behaviour in the material. Moreover, it will detect the differences in slip band spacings and strain profiles if performed on different samples using the same microscopy setups and FFTD parameters, allowing for direct quantitative comparisons.

Another point that needs to be discussed is the physical meaning of the slip band strains recorded after performing the FFTD. For this, a comparison between the slip band strain profiles in an original grain image ($I^G$) and its corresponding IFFT ($I^{IFFTG}$) decompositions is shown in Fig. 12(a-e). The pixel intensities for an original grain image ($I^G$) and its corresponding IFFT on a secondary slip plane using different values of $d^*$, Fig. 12(b-d). The corresponding strain profiles of the slip band pointed at by the arrows are compared in Fig. 13(e). The profile of this slip band, which starts at a grain boundary and disappears as it approaches the centre of the grain, is displayed as a decreasing slope from the left followed by intensity peaks that correspond to its intersections with non-coplanar slip bands. However, when a value of $d^* = 1$ is used, Fig. 13(b), this slip band extends through the entirety of the grain with an almost constant contrast. By preserving a single line of pixels from the FFT image, the resulting IFFT can be thought of as a one-dimensional transform. For larger values of $d^*$, Fig. 13(c,d), the shorter slip band is recovered thanks to the contributions of frequency components slightly misaligned with the FFT bright band. Yet, these components also introduce artifacts in the form of intensity oscillations further along the slip band profile. Generally, such artifacts become smaller for more homogeneous slip bands, which develop more prominently on primary slip planes.

The results obtained offer insights into the heterogeneous distribution of plastic deformation across the sample. Slip bands are on average further spaced apart in larger grains, as shown in Fig. 13(c). A similar finding was reported in 316L stainless steel over a smaller dataset (Menard et al., 2008). This phenomenon is likely an effect of the grain boundaries, which often develop larger stresses in their vicinity compared to those in the grain interiors as a consequence of strain partitioning (Fallahi & Ataee, 2010). It is the larger von Misses stresses developed in alloys with smaller grain sizes which ultimately give rise to the Hall-Petch relationship (Ospina-Correa et al., 2021). Thus, smaller grains in the polycrystal may develop larger stresses on average, resulting in the shorter mean slip band spacings observed. Despite this rationale, the way in which our approach performs the quantification must also be taken into account. Slip bands are detected along lines that cross the grains near their centres, which in bigger grains are overall further away from grain boundaries. The trends reported are those towards the grain interiors, which in bigger grains have stresses that deviate more from those near the grain boundaries.

Moreover, the stress orientation does not show a clear effect on the slip band statistics. Low Schmid factors promote larger slip band spacings, but no significant variation is observed overall. In addition to the linear regression shown in Fig. 13(d), we performed a Spearman correlation test, i.e. a nonparametric measure of rank correlation that does not assume any particular functional dependence between the variables. The Spearman’s rank coefficient of 0.0566 indicates only a very weak correlation, consistent with the previous findings. The same conclusion is reached when
Fig. 12: (a) Greyscale image of the planar strain in a grain and (b,c) its corresponding IFFT images for each active slip plane orientation. The search lines (cyan) and slip bands detected (red) are plotted. (d,e) Pixel intensity profiles along each of the slip bands detected in both images. The slip band highlighted in (a) and its profiles enclosed in (c) are shown in more detail in (f). Arrows pointing at the peaks in intensity correspond to the slip band intersections indicated in (a), sources of contrast external to the slip band of interest that do not appear in the IFFT image. The contrast of the greyscale images was adjusted for better visualisation.
analysing the Taylor factor. A similar behaviour, although considering mean grain strains instead, was reported in a previous study in nickel-based superalloys (Harte et al., 2020). This lack of a clear correlation can be explained by acknowledging that the deformation distribution is a function not only of grain orientation, but also of the configuration of neighbouring grains. The local arrangements of differently oriented grains promote strain partitioning across the material. Neighbourhoods of grains can show collective behaviours based on their intrinsic properties and topological connectivities (Mangal & Holm, 2018). An example of this is seen in Fig. 14, which shows the grains imaged coloured according to the mean slip band spacings on their primary planes. Two regions, pointed at by the arrows, contain multiple contiguous grains with statistically large slip band spacings. Factors that influence this collective behaviour are the differences in stiffness and Schmid factor of neighbouring grains. Birosca (Birosca, 2015) showed in superalloy RR1000 that this results in the formation of soft and hard grains, with very different deformation behaviours as shown by the distributions of their geometrically necessary dislocations.

Moreover, slip transfer across individual grain boundaries also plays a big role in the redistribution of slip. Multiple cases of slip band transmission, which effectively provide slip bands in highly stressed grains with long-range influence in their vicinities, were observed in the region imaged. All these phenomena introduce heterogeneous stresses across the material making the mesoscale deformation not crystallographic in nature, thus reducing the relevance of crystal orientation and Schmid factor as predictors for the slip band distribution in a given grain.

While only a time frame was captured, the large dataset captured allows us to better understand the evolution of plastic deformation. The distributions of individual slip band spacings on secondary and tertiary slip planes have similar modes to that on primary ones, but a larger one when considering the mean spacings per grain (Fig. 10(a,b)). Note the difference on how these distances are measured, i.e. equations (16) and (17). Moreover, the modes and domains of the strain distributions are similar across all active slip planes (Fig. 11(b)), but the proportion of slip bands with a higher strain is considerably larger on primary ones. These trends indicate that, during the initial stages of plasticity, slip bands develop in localised regions with characteristic spacing distributions. As the resolved shear stress increases, these slip bands spread into larger regions of the grain, likely via a double cross-slip mechanism (Koehler, 1952; León-Cázares et al., 2020b). As this happens, the amount of plastic deformation per slip band increases unevenly. Once slip bands occupy the entirety of the grain, more of these become increasingly active, shifting the mean strains measured towards higher values. This occurs because the most active slip bands approach a level of strain saturation, forcing additional bands to accumulate more plastic deformation. These steps later develop on secondary and tertiary slip planes. Statistically, these develop similar slip band spacings over smaller regions and with a lower fraction of high strain slip bands due to the lower stresses attained.

Important aspects of the slip localisation behaviour are also captured in this dataset, despite not fully quantifying the amount of slip per slip band. Such a measurement would need to include three-dimensional strains, not available from DIC, and the quantification of slip band widths, which we consider unviable with the resolution employed and qualitatively appears to vary to a lower degree. Fig. 15 shows the recorded mean slip band spacings and in-plane strains per grain, as well as the overall means and standard deviations from individual measurements. The larger resolved shear stresses on the primary planes promote shorter slip band spacings and larger strain levels to accommodate the additional deformation. Mapped in a plot like this, grains, alloys and loading conditions with more planar slip would manifest further to the top right for equal amounts of plastic deformation. This approach could facilitate much needed large scale quantitative studies of slip planarity, with potential applications in the study...
Fig. 14. Grains coloured by the mean slip band spacings on their primary planes. White regions correspond to grains smaller than the size threshold, not analysed with the FFTD technique. The two arrows point at regions where multiple contiguous grains have large mean slip band spacings.

Fig. 15. Map of the recorded mean slip band in-plane shear strains and spacings per grain on primary, secondary and tertiary slip planes. The overall means from the individual measurements are plotted as triangles, together with their standard deviations (half error bars).

Conclusions
A novel methodology was developed to automate the quantification of the slip band distribution in polycrystalline alloys. The slip band projections are identified from HR-DIC micrographs and the crystal orientations detected via EBSD. Performing FFTs of individual grain images, followed by IFFTs along independent directions, effectively decomposes the original strains observed into images each with the slip bands that share a single orientation. From these, slip band spacings and in-plane shear strains can be measured.

The application of this technique on a tensile loaded polycrystalline nickel-based superalloy returned the largest slip band quantitative analysis to date. The distributions measured indicate that slip bands are on average further spaced apart in bigger grains, with no clear dependence on grain orientation and Schmid factor. Slip bands on secondary and tertiary slip planes develop similar spacings to those on primary planes, although the latter extend over larger regions of the grains. Once they occupy the entirety of a grain, the plastic strains within the slip bands keep increasing to accommodate the macroscopic deformation imposed.

The FFTD approach, with potential applications using other microscopy techniques, could enable much needed large scale quantitative studies of slip localisation for a variety of materials and loading conditions.

Competing interests
The authors declare none.

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