The development of local plasticity around void during tensile deformation

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Abstract

Voids can limit the life of engineering components. This motivates us to understand local plasticity around voids in a nickel base superalloy combining experiments and simulations. Single crystal samples were deformed in tension with in-situ high angular resolution electron back scatter diffraction to probe the heterogeneous local stress field under load; the reference stress is informed by crystal plasticity finite element simulations. This information is used to understand the activation of plastic deformation around the void. Our investigation indicates that while the resolved shear stress would indicate slip activity on multiple slip systems, slip is reduced to specific systems due to image forces and forest hardening. This study rationalizes the observed development of plastic deformation around the void, aiding in our understanding of component failure and engineering design.

Graphical Abstract

Key words: deformation; stress analysis; superalloy
1. Introduction

The presence of voids or inclusions that may nucleate voids are common in engineering materials and yet a physical understanding to the process of void growth remain obscure, despite decades of effort studying the ductile fracture of materials relating to the growth and coalescence of voids. In a modern gas turbine engine, nickel base superalloy is widely used to endure the most demanding environment as turbine blade [1,2]. Shrinkage porosity is a common defect seen in cast superalloy [3,4] and it is known to affect the fatigue life by crack nucleation [5,6], creep cavitation [2], and creep crack formation [7].

The mechanical growth of void has been related to stress triaxiality through the Rice and Tracey model [8]. Stress triaxiality is defined as the ratio between hydrostatic stress and von Mises equivalent stress. Gurson and colleagues [9–11] developed a series of continuum constitutive theories linking hydrostatic stress to void growth during ductile fracture. The development of these theories has been facilitated by X-ray computed tomography (XCT) [12,13] which enables a more accurate account of some critical parameters, such as the evolution of void size and volume fraction during deformation. Though these theories have found wide adoption in solving engineering problems, the physical process of void growth was not clear. Progress has been made recently correlating void growth with local microstructures through advanced correlative approach combining XCT with scanning electron microscopy (SEM) and electron back scattered diffraction (EBSD) [14,15].

Molecular and dislocation simulations provide a mechanistic insight into void growth under multiaxial remote loading conditions [16–19]. These, together with some earlier work [20–22] indicate that mass transfer required for the expansion of void is primarily due to dislocation activity around the void surface. The dislocation emission based void growth models, schematically summarised in Fig. 1, generally involve prismatic dislocation loops [23–26], shear loops [23,27], antiparallel dislocations gliding on parallel slip planes [19,23,28], and antiparallel dislocations gliding on angled slip planes [29,30].
The shear loop mechanism has been critically reviewed by Bulatov et al. [24]. The volume change of the void associated with a dislocation loop is [31]:

$$\delta V = \int_{Surface} (\mathbf{b} \cdot \mathbf{n}) \, dA \quad (1)$$

Where $\mathbf{b}$ is Burgers vector of the dislocation loop, $\mathbf{n}$ is the unit normal to the surface $dA$ bounded by the dislocation loop. This is to indicate that for shear loops the net mass transfer equals zero as $\mathbf{b}$ and $\mathbf{n}$ are perpendicular; i.e. When a full shear loop is punched out of the void surface the void would return to its original shape [24].

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**Figure 1: Schematic representation of void growth mechanisms found in literature.**

| Void growth by glide of prismatic loop | Void growth by glide of a pair of shear loop | Void growth by glide of antiparallel dislocations on parallel slip planes | Void growth by glide of antiparallel dislocations on angled slip planes |
For void growth by glide of antiparallel dislocations on angled slip planes, shown in Fig. 1, this would create lattice curvature and need to be supported by the accumulation of geometrically necessary dislocations (GND).

These four mechanisms highlight how local plasticity can be involved in void growth and motivate the present study. Prior work has involved simulations and analytical studies and so in the present work we supplement this with direct measurements of the stress field and dislocation structures near voids during in-situ tensile testing. The stress field is evaluated using a combination of high resolution EBSD (HR-EBSD) [32–34] facilitated by crystal plasticity simulation [35]. Analysis of local plasticity is supported through careful consideration of the resolved shear stress of the 12 FCC slip systems and discussed in the context of image stress analysis [36,37].

2. Experiments and methods

2.1. In-situ EBSD

The material used in this study is a second-generation low rhenium single crystal superalloy designated ‘DD6’. A compositional comparison with other superalloy systems are made by Li. et.al [38] together with some general thermal and mechanical properties. The single crystal in this study was manufactured using directional solidification. From this, tensile test pieces were extracted using electro discharged machining. Before mechanical testing, the samples were ground and polished to a mirror finish (ending on 0.05 μm pH balanced colloidal silica) and the final surface damage was removed using a Gatan PECs ion milling system working at 8 keV with 4° gun tilt for 5 minutes. The tensile sample design and the in-situ tensile testing set up is shown in Fig. 2. The waisted sample gauge is used to focus the region of high strain towards the central region and to avoid probabilistic strain localization in an otherwise straight gauge. The sample was fixed onto a 70° pre-titled sample stage with stainless steel pins, to align the sample with the loading direction, and tightened with tensile clamping grips. For enhanced conductivity the sample was connected to the SEM stage using carbon tape.
The setup shown in Fig. 2 allows for a safe SEM/EBSD working distance of 18 mm and EBSD camera distance of 18 mm (for a phosphor which is 25.4 mm wide). The sample was deformed under displacement control at a rate of 0.1 mm/min and the displacement was paused for EBSD scans. A small region (~71 μm x 52 μm) surrounding voids located in the middle of the gauge section was EBSD scanned at 30 keV and 14 nA with 0.4 μm step size and the diffraction patterns saved to 12-bit TIFF images for subsequent stress analysis.

The saved diffraction pattern images were used to analysis the rotations and strains present in the interaction volume that introduce shifts of the diffraction patterns [39]. The technique is known as high resolution EBSD (HR-EBSD) [32,33]. In principle, in the similarly orientated diffraction patterns (e.g. from one grain) one pattern is chosen as a reference pattern which is compared with other diffraction patterns to estimate displacement gradients. The displacement gradient tensors are then used to estimate strains and rotations which can, respectively, be related to stress, through anisotropic Hooks law, and GND density, by linking rotation gradients to Nye’s dislocation tensor [40,41]. It has been shown [34] that large lattice rotations, e.g. in a heavily deformed...
sample, affects the accuracy of strain measurement, in which case it is necessary to first rotate the intensity distribution within the test patterns close to the reference pattern and perform a second pass of pattern correlation to estimate strains.

2.2. Crystal plasticity finite element simulation

As mentioned above the stress estimated from EBSD are relative to a reference point whose stress are unknown. In this study, we use a lengthscale-dependent and physically based crystal plasticity simulation to extract the stress field at the reference point. The formulations of the simulation can be found in [35] and we introduce it briefly here. The plastic velocity gradient $L_p$ is written in terms of slip rate $\dot{\gamma}$, slip plane normal $n$, and slip direction $s$ as:

$$L_p = \sum_{i=1}^{N_s} \dot{\gamma}^i n^i \otimes s^i \quad (1)$$

where $N_s$ is the number of slip systems. The slip rate is given by:

$$\dot{\gamma}^i = \rho b^2 \nu \exp\left(-\frac{\Delta H}{kT}\right) \sinh\left(\frac{(\tau^i - \tau_{CRSS})\Delta V}{kT}\right) \quad (2)$$

where $\rho$ is the mobile dislocation density, $b$ the Burgers vector, $\nu$ is the thermally assisted dislocation jump frequency (for dislocation to overcome local pinning barrier and become free to glide), $k$ the Boltzmann constant, $T$ the temperature, $\tau_i$ the resolved shear stress on the $i^{th}$ slip system, $\tau_{CRSS}$ the resolved shear stress, and $\Delta V$ is the activation volume. Taylor hardening is employed to capture the dislocation based hardening behaviour. The parameters in equation 2 are chosen to minimize strain rate sensitivity [42] and are listed in Table 1.

| $\rho$     | $b$        | $\nu$     | $\Delta H$ | $T$     | $\tau_{CRSS}$ | $\Delta V$ |
|------------|------------|-----------|------------|---------|---------------|------------|
| 0.01 $\mu$m$^2$ | 2.61 x 10$^{-4}$ $\mu$m | 1.0 x 10$^{11}$ s$^{-1}$ | 3.456 x 10$^{-20}$ J | 293 K   | 450 MPa       | 11.6b$^3$ |

*Table 1: The properties used for slip rule.*

The simulation set up is shown in Fig. 3. The sample is deformed through the application of displacement to the right-hand surface, the $x$, $y$, $z$ plane indicated on the left hand side of the sample shoulder in Fig. 2a is restricted movement along the corresponding directions, and the same is applied to the $y$ and $z$ plane of the right
shoulder to avoid buckling at the sample gauge. To address machine compliance, the model in Fig. 3 includes an isotropic elastic medium attached to the fixed end of the sample grip.

The mechanical behaviour of the sample is simulated using the crystal plasticity constitutive equations implemented through a user-material (UMAT) subroutine in ABAQUS. A local region, of the same size of the EBSD scans, were assigned smaller element size as this is the primary region of interest, as shown in Fig. 3b.

![Figure 3: Model set up for simulating the tensile property and the stress state at the HR-EBSD reference point.](image)

3. Results

The macroscopic load displacement curve and electron micrographs of the sample during a loading-unloading cycle are shown in Fig. 4. The loading is close to <001> direction and the deformation featuring mainly the development of four slip systems, revealed due to the presence of slip bands on the surface. The interactions between the bands likely results in the observed hardening.

EBSD scans were conducted at selected voids on the sample surface at different levels of deformations indicated on the load-displacement curve. The focussed imaging and EBSD has resulted in a local build-up of carbon contamination (the black boxes). The timing of these observations include a pre-deformation scan (#1), a scan at yield point (#2), a scan at intermediate deformation (#3), a scan before fracture (#4,
the displacement to fracture was determined by a separate experiment), and a final scan after unloading (#5).

At later stage of deformation, a tendency for sub-grain boundary formation can be seen towards the centre of the sample (Fig. 4 image #4). Between Fig. 4 image 4 and 5 there are no visible changes of the surface deformation features after unloading.

The higher magnification inserts in Fig. 4, show the surface around one void, indicated with a box in the low magnification images. These inserts reveal the development of local plasticity. Here localised dislocation plasticity has resulted in the formation of two shear bands from a location on the top of the void.

Figure 4: The evolution of the surface deformation feature during a loading-unloading cycle. The inset shows the development of the local plasticity around a void (location indicated by the box).
Figure 5: The accumulated slip from simulation shows slip trace consistent with the experimental slip trace (a), and the comparison between experimental and simulated load-displacement curve shows good agreement(b).

Results from the crystal plasticity model are shown in Fig. 5a, including the accumulated slip, i.e. $\sum_{i=1}^{N_y} \dot{\gamma}^i \Delta t$ (equ. 2). The bands of high strain gradient are spatially co-located with the experimentally observed slip trace (indicated in Fig. 4-2). The statics and kinematics of the simulation show good agreement at the macroscopic scale, as there is agreement of the simulated and experimental load-displacement curve, including the transition from (macroscopic) elastic to plastic deformation as well as macroscopic hardening. This indicates that the simulations capture both the macroscopic mechanical response and the plastic localization that reflects the crystallographic nature of deformation.

Figure 6: The development of the local in-plane stress field at a void during a loading-unloading cycle.

Voids are shown in Fig. 4 and they are stress concentrators during deformation, changing the local deformation conditions. The local stress field at one of the voids (location indicated in Fig.4-2) is demonstrated in Fig. 6, measured underload. The star in Fig.6-2 indicate the location of the reference point and is consistent in other
deformed maps. In Fig. 6 each map presents the stress field around the void with respect to this reference point, i.e. these maps show variations in the Type III stress distributions.

The initial state of the sample reveals a heterogenous residual stress field, and the heterogenous nature of the stress field is homogenised around the yield point as indicated by Fig.6-2.

Comparison of the stress fields under load vs unloading indicates that the variations in these stress fields, from the yield point, persist until after unloading. For example, the $\sigma_{xx}$ stress is compressive to the left and right side of the void and tensile at the top and bottom. After unloading, there is no significant redistribution of the stress field, but the amplitude is reduced.

To address the reference stress issue, the crystal plasticity simulation is used to shift the stress tensor across the map according to the principle of superposition. This follows the approach used previously (e.g. [43]) and this avoids use of a second experimental technique (e.g. synchrotron diffraction [44]). As a note, alternatively the simulation field could be ‘re-referenced’ to reveal only the type III stress variations, but as we will see later the absolute value of the stress field is useful for mechanistic analysis.

To confirm validity of this approach, the ‘re-referenced’ experimental fields are compared with the simulation fields in Fig. 7 and good agreement for the spatial variations are shown. In these figures, the stress tensor at the reference point has a dominant $\sigma_{xx}$ component while the other components are negligible. This reference stress state contribution is likely because the void is located close to the central axis of the specimen and the reference point was chosen far away from the void and therefore was less affected by the local stress field developed around the void. The reference shifted HR-EBSD stress field is shown in Fig. 7c and the magnitude of the reference stress state is sufficient to make the average stress state of the map to have a dominant $\sigma_{xx}$ tensile stress field, with strong variations close to the void.
Figure 7: Full stress field around the void, as measured by HR-EBSD (a) and simulated using crystal plasticity method (b) where the stress tensor at the reference point is demonstrated. The reference shifted HR-EBSD stress field is shown in (c). Two sets of reference frame are defined in (d) together with the method to rotate from sample reference frame to slip reference frame.

The development of local plasticity around the void can be seen from Fig. 4. The shear band is seen at intermediate deformation level (scan 3) without further development of other slip systems between scan 3 and close to fracture (scan 4).

To further understand the local plasticity the resolved shear stress along each of the 12 FCC slip systems are investigated. Here, we use the reference shifted HR-EBSD stress field at scan 3 and scan 4 and assume that these would shed light on the lack of further development of local plasticity around the void between the two deformation levels. As the resolved shear stress is the major factor driving dislocation motion, it is of interest to investigate this for each of the 12 FCC slip systems. This is done by rotating the reference shifted HR-EBSD stress field from the sample reference frame to a reference frame on slip systems defined as X axis parallel to the Burgers vector direction and Z axis perpendicular to the slip plane. The method is schematically...
shown in Fig. 7d. The $\sigma_{XZ}^{\text{slip}}$ component of the rotated stress tensor represent the resolved shear stress of the corresponding slip system. The results are shown in Fig. 8. (Note in Fig. 7 and 8 we only show the stress fields at scan 3 for simplicity, those at scan 4 are shown in Supplementary Figures).

**Figure 8.** The resolved shear stress maps of the 12 FCC slip systems. The clock diagrams indicate the slip trace, represented by the line going through the diameter of the circle, and the in-plane Burgers vector direction, represented by the arrow and the length of this arrow indicates how ‘in-plane’ the Burgers vector is.

In Fig. 8 the slip system is labelled on the corresponding resolved shear stress map together with a clock chart indicating the corresponding slip trace (diametric line) and the in-plane Burgers vector direction (arrow). The length of the arrow represents the magnitude of the out-of-plane Burgers vector component, short arrow indicates
Burgers vector pointing in or out of plane. The location of possible slip band formation can be taken as the location where there is high magnitude of positive (along Burgers vector direction) or negative resolved shear stress field (against Burgers vector direction). The direction of resolved shear stress need to be directed away from the void surface so that dislocations can glide away to facilitate void growth. Based on this principle, the locations around the void where slip band formation is possible are labelled with solid lines in Fig. 8. All the slip systems can potentially be initiated from multiple locations at the void surface, except slip system 2 (where the sign of the stress field does not align with the Burgers vector to result in slip away from the void).

Despite these apparently favourable stress conditions, Fig. 4 shows that only 2 slip systems are observed. These align with slip on the (111) and (111) planes. In the following section, we will explore why only these two systems were activated.

4. Discussion

Measurement of the stress state and observation of dislocation plasticity around the void motives us to test mechanism driven hypotheses for slip activity near the void.

Usually, slip is driven by the resolved shear stress on the slip system exceeding a critical value. The fields in Fig. 8 would indicate that significant slip should occur, either through source nucleation or movement of existing dislocations, near the void.

However these stresses can be shielded and reduced, especially when dislocations are near a free surface due to the additional stress field called image stress [36,37]. The origin of image stress is schematically represented in Fig. 9, where the screw dislocation, represented by a Volterra cut, close to an interface would exert a shear stress, $\sigma_{xz}$ and $\sigma_{yz}$, on the interface under the designated reference frame [36]. The general equilibrium and compatibility requirement at the interface [36] (note the schematic only represent an equilibrium condition) requires an equal but opposite stress at the interface amounting to placing an ‘image’ dislocation of opposite sign at the other side of the interface. Dislocations may be attracted to or repelled from the interface depending on the difference of shear modulus across the interface [36]. In the case of a void where the interface is a free surface, an attraction force is exerted on the dislocations.
For a straight edge dislocation close to a void surface, we adopt here an analytical image stress model proposed by Lubarda [29]:

$$\sigma_{image} = \begin{pmatrix}
\frac{Gb_y}{2\pi(1-\nu)} \frac{x}{x^2-a^2} \left[ a^2 \left( 1 - \frac{a^2}{x^2} \right)^2 \right] & -\frac{Gb_x}{2\pi(1-\nu)} \frac{x}{x^2-a^2} & 0 \\
-\frac{Gb_y}{2\pi(1-\nu)} \frac{x}{x^2-a^2} \left( 1 + \frac{a^2}{x^2} - \frac{a^4}{x^4} \right) & -\frac{Gb_z}{2\pi} \frac{x}{x^2-a^2} & 0
\end{pmatrix}$$

(3)

Where $G$ is shear modulus, $\nu$ is Poisson’s ratio, $a$ is void radius, $x$ is the distance to the centre of the void, and $b = (b_x, b_y, b_z)$ is the Burgers vector. The anisotropic shear modulus is used for the calculation and in the case of $\{111\}$ slip in FCC crystals this is given by (from [45]):

$$G^{(111)} = \frac{3}{1+2A} C_{44}$$

(4)

Where $A = \frac{2C_{44}}{C_{11}-C_{12}}$ is the Zener anisotropy ratio, $C_{11}$, $C_{12}$ and $C_{44}$ are the stiffness constants (the corresponding values of 252 GPa, 161 GPa, and 131 GPa are used in this study). The resolved image stress on the 12 FCC slip systems are obtained by the routine explained in Fig. 7d and the results are presented in Fig. 10.
Figure 10: Image stress of the 12 FCC slip systems. Labelled stress values are those at one full Burgers vector distance from the void surface.

Notice the image stress vs distance plot are made from one Burgers vector away from the interface, and the analytical solution of (3) no longer holds due to distorted dislocation core when the distance to void surface is close to the dislocation core cut-off distance [29]. The 12 FCC slip systems together with the corresponding image stress at one Burgers vector distance are labelled in Fig. 10 and the stress magnitude represent a lower-bound resistance to dislocations gliding away from the void surface.

Now the resolved shear stresses on the 12 slip systems in Fig. 8 are revisited and considered in terms of the image stress field that will need to be overcome to initiate dislocation slip.

Comparing the stress fields in Fig 8. and the dislocation slip based void growth modes shown in Fig. 1, it can be seen that slip systems 5, 8, 9 have local stress fields that facilitate antiparallel dislocations glide on parallel slip planes. For instance, in the shear stress map for slip system 8 in Fig. 8, the negative stress field at the top right corner and the positive stress field at the bottom right corner of the void both tend to direct...
dislocations with opposite sign away from the void surface along the indicated slip traces. For slip system 7, 9, 10, 11, and 12, the corresponding shear field facilitates the formation of single or multiple local shear bands.

To provide quantitative assessment of slip activity, line scans are drawn along the indicated slip traces for the above two categories of slip systems, and the magnitude of the shear stresses are compared to the image stress of the corresponding slip system in Fig. 11. The comparison brings to our attention that the resolved shear stress level for most of the slip systems at scan 3 are on the order of 1 GPa lower than the image stress level. The gap is larger in some cases, notably slip systems 8 and 11. Though the resolved shear stresses show various level of increase due to continued macroscopic loading up to scan 4, the resolved shear stress for most of the slip systems are still below the image stress. The slip system 9-1 has resolved shear stress level that exceeds the image stress close to the free surface at scan 3 and at scan 4 the resolved shear stress for slip system 5-2 also rises to a value above the image stress level, however the corresponding slip traces (indicating that these systems activated) were not observed. A further exploration of the absence of these slip traces follows.

Figure 11: Resolved shear stress maps facilitating the formation of antiparallel dislocations gliding on parallel slip planes (a) and single or multiple slip bands (b). The
magnitude of the resolved shear stresses along the potential slip lines are compared with the image stresses.

The shear stress field for slip systems 1, 3 and 4, 6 facilitates the formation of shear band corroborating the observed slip trace (Fig. 4), a comparison with the image stresses are shown in Fig. 12a and b. These slip systems in general have higher resolved shear stresses compared to those in Fig. 11, however, much higher image stresses for slip systems 1 and 4 suggest that those two slip traces are less likely to be the activated slip trace. The shear stress for slip system 3 is higher than the corresponding image stress, it is likely that this variant consists of one of the observed slip trace (Fig. 4) while slip trace and Schmid factor analysis wouldn’t make this distinction between slip 1 and slip 3 (both give rise to the same slip trace and both have high Schmid factor: 0.43 and 0.40 respectively). The resolved shear stress for slip system 6 is lower than the image stress but note that this observation was made at an instance after the formation of this slip band, it is expected that a certain proportion of the resolved shear stress to be relaxed. Therefore, compared to the more significant stress gap for slip 4, slip 6 is likely the observed slip variant. Notice that slip 3 and slip 6 forms an antiparallel dislocation pair gliding along slip planes at an angle. This configuration may grow the void according to the fourth mechanism in Fig. 1 and the observed GND densities in between those two slip planes (Fig. 12) seems to agree with this growth mechanism. The Accumulation of the GNDs around the void provides forest hardening which further reduce the probability for dislocations gliding away from the void surface. A Taylor hardening model [46], $\tau = 0.9bG\sqrt{\rho_{GND}}$, is used to obtain the stress due to forest hardening and the result is shown in Fig. 12c. There is raised GND density and therefore Taylor stress field at the nucleation site of slip 3 and 6, possibly due to the lattice continuity requirement to assist void growth. The indicated location, where the slip 9-1 (Fig. 11) shows resolved shear stress exceeding the image stress at scan 3 and that for 5-2 at scan 4, has an additional slip resistance on the order of 1 GPa and 0.3 GPa, respectively, imposed by forest hardening. This would bring the net driving force below the image stress. Notice that the forest hardening stress is generally below 1GPa at both deformation scan 3 and scan 4 (GND field at scan 4 is shown in supplementary), except for the location where the observed slip trace emitted from the void surface. For a significant part along the void perimeter the
forest hardening stress is lower than the magnitude of the resolved shear stress for many slip systems shown in Fig. 11, therefore forest hardening itself is not sufficient to rationalize the lack of local plasticity around the void.

![Figure 12](image)

**Figure 12:** Resolved shear stress maps facilitating the formation of slip bands corroborating the observed slip bands (a) and (b). The GND density distribution and the corresponding Taylor stress map are shown in (c).

The void had possibly grown under mechanism 4 in Fig. 1. The equivalent diameter increased from ~16.6 μm (undeformed) to ~18.73 μm (before fracture), giving a growth ratio of ~1.13 at macroscopic strain of 0.085 (based on simulated stress strain curve, supplementary). At similar strain level, copper has shown a growth ratio of 1.29 [47] and aluminium alloy of 1.4 [48]. Additionally, significant and relatively uniform local plasticity can be seen around the void surface in these two materials [47,48].

The image stress can be considered as an energy barrier from moving dislocations away from the void surface. This stress is compared across various material systems in Fig. 14. For simplicity, Burgers vectors are assumed to be parallel to the horizontal
axis and aligned with the centre of a 10 µm void. Therefore, the image stress is simply

\[ \sigma_{xy}^{\text{image}} = -\frac{Gb}{2\pi(1-\nu)} \frac{x}{x^2-a^2}. \]

Figure 13: Image stress vs distance from void surface plot across different material systems. The inset demonstrates image stress vs void size for DD6 superalloy.

It can be seen that the magnitude of the image stresses for Al and Cu do not pose an as significant barrier for void growth based upon dislocation glide, as compared to Ni superalloy (DD6), and this may be a factor influencing the well-developed local plasticity and void growth seen in these two materials [47,48]. Another material given as a comparison is a bainite steel (SA508) where the void imposes a higher image stress compared to DD6 yet demonstrate significant void growth [15,49]. Note that on the one hand, the SA508 steel has BCC crystal structure where slip on the \{110\}, \{112\}, and \{123\} planes [50] offer 36 slip systems compared to 12 in FCC Ni. On the other hand, the significant void growth in SA508 is typically either found in front of a crack [15] or after the UTS and thus during necking [49]. In these cases the local stress state may raise above the image stress for more slip variants and this can be combined with
a consideration of the multiaxial stress state at the crack front and the necked region that may also facilitate dislocation emission by modifying the distribution of the local stress state [19] and this may result in more easy activation of a range of slip systems at the void surface, thus homogenising plasticity assisted void growth. This point can be directly seen from Fig. 6 of reference [17] where the dislocation distribution at void surface against various loading conditions were compared.

For DD6 significant void growth has been observed, but typically at high temperature, for example, in Fig. 6 of reference [51] where transition has been seen from a mixed ductile/cleavage fracture at 850°C to a full ductile fracture featuring void growth and coalescence at 1020°C. Note that at these high temperatures thermally assisted dislocation jump may help overcome the image stress as well as cross slip that facilitates prismatic loop formation. These are likely the factors contributing to the significant void growth seen at high temperature [51] while diffusion may not be a dominant mechanism given the time scale of the experiment in [51].

In the present work, Fig. 10 shows that that the image stress decrease quickly away from the free surface. All slip systems have image stress dropped below 500 MPa at 3 nm distance (i.e. ~90 atomic layers) away from the surface, and at this distance some slip systems have image stresses below 100 MPa. This is important to consider with regards to the measurement of GND densities with EBSD (though a different stress function is needed to properly analysis image stress of a flat surface). For an EBSD experiments, GND density is calculated using spatial gradients of the average lattice orientation of the interaction volume which extends ~20 nm from the surface [52] and therefore dislocations affected by image stress field consists of only a small portion.

The image stress analysis presented in the current work shed light on dislocation mobility after nucleation. The nucleation of dislocations from void surface involves dislocation energy variations across slip systems, elastic anisotropy, dislocation core structure [53], as well as surface energy due to ledge formation [17,23]. Some detailed nucleation process can be found in [27,53]. The surface energy required to create a ledge from a smooth void surface may pose a barrier for dislocation nucleation hence restrict the local plasticity around the void, however, the native ledges or steps on the void surface are found to facilitate dislocation nucleation [17,54]. It has been
suggested that the stress required to nucleate dislocations from void surface could be higher than the strength of the material [55], indeed, the critical stress to nucleate a dislocation from a 4 nm void in copper is on the order of 6 GPa [27] which is far higher than the strength of copper. This can only be practically achieved by high rate or low temperature deformation [55]. However, these molecular dynamics studies concern voids in nano-metre scale where the void size is on the order of the mean dislocation spacing. Under this condition, the dislocation nucleation consists a significant part of void growth due to the starvation of dislocation sources. For a micron-sized void, it is arguable that sufficient dislocations are already present along the void perimeter. Therefore, the growth of micron-voids may primarily involve dislocation gliding rather than nucleation.

5. Conclusion

In this study the stress field around a void in a nickel base superalloy was measured in-situ using HR-EBSD while the specimen was under load. The stress state at the HR-EBSD reference point was estimated by crystal plasticity simulation, enabling adjustment of the full field map to enable assessment of the development of local plasticity at the void surface.

Notable observations are summarized below:

1. Assessment of stress while the sample is still under load is necessary in cases where the magnitude of the stress is important, since there are load drops after unloading (or an unloading model must be used). In this example, the distribution of stress field seems to be less affected by unloading.

2. For the material in this study, there was limited local plasticity development at the void surface throughout the deformation. This is evidenced by the observation that only two slip bands emitted from the top surface of the void.

3. Resolved shear stress analysis found multiple favourable locations around the void surface where slip band formation is possible. However, at these locations the resolved shear stress fell below the image stress and thus it is assumed that dislocations would not be able to move easily and grow the void.
4. GND accumulation at the void surface provide forest hardening and further resist dislocations gliding away from the void surface, but forest hardening alone cannot rationalize the lack of plasticity at void surface.

5. The void growth in this study is due to antiparallel dislocations glide on non-parallel slip plane. This mechanism, however, accumulates GNDs in between the slip bands, which may limit the growth rate by local forest hardening.

6. Image stress can be used to study/predict the dislocation activity close to void surface. It may be used to pin down the specific slip variant that can be activated where the traditional Schmid factor and slip trace analysis fails.

7. The image stress decreases quickly away from the void surface and is a function of shear modulus. The rate of decay depends on Burgers vector orientation. In general, the image stress in Ni superalloy drop below 500 MPa at 3 nm distance away from the void surface.
Author contributions
YG conducted the experiment and simulation, performed the data analysis, and drafted the paper. CZ manufactured the material. TBB supervised the overall project.

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Supplementary Figure 1: Simulated stress-strain curve.

Supplementary Figure 2: Reference shifted HR-EBSD stress maps at scan 4 (a) based on CPFE simulation (b). The star indicates the location of the reference point.
Supplementary Figure 3: Resolve shear stress maps of the 12 FCC slip systems at scan 4.

Supplementary Figure 4: GND density distribution and the corresponding Taylor hardening stress at scan 4.