Plastic heterogeneity in indentation is fundamental for understanding mechanics of hardness testing and impression-based deformation processing methods. The heterogeneous deformation underlying plane-strain indentation was investigated in plastic loading of copper by a flat punch. Deformation parameters were measured, in situ, by tracking the motion of asperities in high-speed optical imaging. These measurements were coupled with multi-scale analyses of strength, microstructure and crystallographic texture in the vicinity of the indentation. Self-consistency is demonstrated in description of the deformation field using the in situ mechanics-based measurements and post-mortem materials characterization. Salient features of the punch indentation process elucidated include, among others, the presence of a dead-metal zone underneath the indenter, regions of intense strain rate (e.g. slip lines) and extent of the plastic flow field. Perhaps more intriguing are the transitions between shear-type and compression-type deformation modes over the indentation region that were quantified by the high-resolution crystallographic texture measurements. The evolution of the field concomitant to the progress of indentation is discussed and primary differences between the mechanics of indentation for a rigid perfectly plastic material and a strain-hardening material are described.
1. Introduction

Characterization of the deformation field around an indentation is especially important for understanding the mechanics of hardness testing and many industrial deformation processing methods, including forging, extrusion, machining and shot peening. More recently, the deformation field in indentation has been the subject of renewed attention with the development of strain gradient plasticity models, analytical solutions and microstructural studies for which accurate measurements of such fields are critical. Modelling approaches have been used to study evolving stress and strain fields in indentation; the predicted strain typically is indirectly validated through post-mortem characterization of the effects of plastic deformation on microstructure and/or mechanical properties. More recently, in situ imaging-based methods have been applied to facilitate direct assessment of deformation in the indentation zone [1]. These methods have extended experimental capabilities to quantify temporal characteristics of the deformation (e.g. strain rate) as well as direct measurement of strain at higher spatial resolution than otherwise possible using classical assessment methods. In prior work, these in situ methods were used to measure strain and strain-rate in flat punch indentation of a perfectly plastic, non-strain-hardening material (e.g. lead) [1]. The studies revealed the location of primary and secondary shear bands in the plastic zone and were consistent with Prandtl’s framework for describing the deformation field [2], wherein material in the region immediately underneath the indenter moves as a rigid body during indentation. In those experiments, lead was chosen as a model system owing to its low flow stress and low strain hardenability. Of significantly greater interest to industrial deformation processing is the indentation response of strain-hardening materials. This study explores measurement of the deformation field in punch indentation in a strain-hardening material—oxygen-free, high-conductivity (OFHC) copper. Copper is an attractive model system as a wealth of information is available regarding constitutive modelling, microstructure response and mechanical properties. Further, microstructure and mechanical properties of copper are far more sensitive to plastic straining when compared with lead. In this regard, image-based assessments of the deformation field can also be coupled and directly compared with post-mortem assessments of the field made through local measurements of microstructure and strength.

2. Background

Prior experimental assessments of strain levels and plastic zone extent in indentation have generally relied on either tracking distortion of internal/external features or assessing strain-induced changes in microstructure and mechanical properties. A common method that relies on tracking deformation of external features is that of grid distortion. In this method, the indentation is performed at the interface of clamped specimen halves whose mating surfaces are imprinted or engraved with fine grids using sharp cutting tools. To assess strain, the specimen is disassembled and the grid distortion measured. The deformation patterns resulting from such impressions have, for example, been used to propose a slip line field theory to predict loading response during wedge indentation [3]. This approach has also been used to study the deformation field in other wedge-type and spherical indentation configurations [4,5]. While the split-specimen technique provides insight into material flow in deformation, a clear drawback is the effect that the mating interface may have in modifying the stress and strain distribution during indentation, as the specimen halves are prone to separate under increasing load.

Monitoring inherent microstructure features in multi-phase materials during indentation avoids the problem of non-uniform deformation that is associated with split specimens. In this case, relative differences in the distortions of the primary and secondary phases reveal the extent of the deformation. For example, visualization of deformed interfaces in layered composites has been used to determine displacement fields during deep spherical indentation [6]. This helped to identify a dead-metal zone in the absence of lubrication, which appeared as a zone of non-deforming material anchored to the indenter face. Alternately, distortion of second phase...
morphology in alloys can be used to visualize deformation patterns. In this regard, the relative displacement of pearlite during indentation in steels has been used to map the deformation zone with regard to strain levels [7]. Although these internal marker methods do not suffer the same issues as split-specimen configurations, the use of intrinsic microstructure markers is generally infeasible for the wider range of engineering materials.

Strain-dependent microstructure transformations and strengthening effects can be exploited to study deformation zone characteristics in a wider range of metallic systems that is compatible with the internal marker method. The deformation occurring during indentation locally generates defects (e.g. dislocations, twins, stacking faults) in quantities directly related to the magnitude of the local strain level. In this regard, the heterogeneous deformation in indentation gives rise to dislocation density gradients in the plastic zone. These gradients can be indirectly visualized through either localized measurement of strength (via hardness) [8–10] or the rate dependence of surface topography in metallographic etching [11,12] and microstructure in static/dynamic recrystallization [13]. Hardness measurement has been used to estimate strain contours through empirically derived relationships between strain and hardness in controlled compression testing [9]. A potential drawback of this technique is that saturation of hardness can occur at rather low strains in systems with high stacking fault energy (e.g. Al- and Mg-alloys) owing to efficient dynamic recovery of the microstructure. Static recrystallization during annealing also is well known to be strongly influenced by initial defect density [14]; this has been used in spherical indentation studies of steels to provide insight into the heterogeneity of the deformation zone [13]. Hardness mapping has also been applied to determine the constitutive properties of plastically graded materials. In this regard, Branch et al. [15] developed an interesting approach that uses an expanding cavity model coupled with finite-element analyses and micro-hardness maps to extract strain-hardening exponent as a function of depth for several plastically graded materials. A subtle drawback of this analysis is the reliance on simulation-based estimates of strain; these likely may differ from the strain levels actually achieved in the experiment, contributing to error in the overall model. Clearly, better methods are needed for assessing deformation in the plastic zone in indentation.

The above methods are limited to post-mortem characterizations of the deformation zone. While these have been useful for providing experimental validation of analytical indentation constructs, opportunities do not exist for studying the evolving field in situ, making temporal-based assessments of deformation (e.g. strain rate) difficult to obtain. This has been addressed with the application of image-based deformation measurements, for example particle image (or tracking) velocimetry (PIV/PTV), to map strain and strain-rate fields in indentation problems [1,16]. In these methods, surface asperities are tracked during deformation to yield displacement and velocity fields from which strain rate and strain maps are derived using continuum mechanics formulations [2]. In addition to providing a direct method for assessing strain rate, the PIV/PTV methods also provide significantly higher spatial resolution for strain measurement when compared with grid-based deformation assessments, as microscopic asperities as small as several micrometres in size can be used to track deformation.

While the PIV/PTV approach for assessing deformation parameters and extent of the plastic zone offer clear advantages over classical experimental methods, no study has provided simultaneous comparisons of these various techniques yet. This was not possible in previous PIV studies [1,16], as these studies focused on understanding the deformation field in indentation of pure lead, which does not show appreciable microstructure/strength changes with plastic straining. In the ensuing, the evolution of the deformation field in punch indentation of OFHC copper is measured using both PIV and conventional post-mortem characterization of microstructure and hardness. The estimates of the deformation measurements made through microstructure and hardness provide a suitable framework for understanding deformation in a conventional continuum construct through evolving strain, strain-rates and velocities as well as with material-related evolution in terms of mechanical properties, microstructure and crystallographic texture. Further, these measurements provide a detailed description of the heterogeneity of the deformation zone in flat punch indentation of copper, including unique
insights into the evolution of the deformation field during indentation. Such an exercise is crucial for establishing the theoretical underpinnings of the indentation problem, which has direct relevance to topics ranging from hardness testing to the development of advanced strain gradient plasticity models.

3. Experimental set-up

A series of punch indentations were made in OFHC copper blocks with a steel flat punch indenter using a tension/compression load frame (MTS Qtest/50LP), a schematic of the set-up is shown in figure 1a. The dimensions of the indenter were 2.54 × 25.4 × 5.4 mm and that of the copper blocks were 127 × 25.4 × 25.4 mm. These components were both machined to a size tolerance of ±0.005 mm and in-plane alignment of these surfaces was maintained during the experiment using tempered glass plates. The indenter was precision ground to a smooth finish. The copper had an initial grain size of 65 ± 37 µm and Vickers micro-hardness of 75 ± 6 kg mm$^{-2}$. The stress–strain response of the copper is reported in figure 1b. The indentation speed was 1 mm s$^{-1}$ and the maximum depth of indentation in these experiments was about 3.5 mm. The indentation region was imaged in situ using an optical microscope connected to a charge coupled-device-based, high-speed imaging system (PCO 1600), resulting in an effective resolution of approximately 7.1 µm/pixel. A sample image taken with the camera is provided in figure 1c and an overlay of multiple such images is provided in figure 1d. Deformation field parameters were obtained by tracking the motion of the asperities created on the surface of the copper using a PIV algorithm. While out-of-plane deformation in the vicinity of these sample edges can affect deformation parameters measured on the surface, transverse material flow in the direction of specimen width was observed to be minimal at the indentation depths investigated herein.

The PIV algorithm, essentially, assesses a degree of match between two images that are collected in a specific time interval apart. Usually, the algorithm is used to identify and track a group of particles (or markers) in a given image pair. In order to track the asperities present on
the copper workpieces, images at sufficiently high temporal and spatial resolution were collected in these experiments. Each image was then divided into a grid, ensuring that two or more of the asperities lie in a given grid element. A cross-correlation algorithm was employed to assess this match, and the peak in the cross-correlation indicates the position of the markers—i.e. the displacement in that region of the workpiece. An entire displacement field is obtained from an image pair through this analysis. Further details of PIV have been reviewed elsewhere [17]. Other deformation parameters related to velocity field, strain rate field and accumulated strains are derived [2] from the displacement using equations (3.1)–(3.4),

\[
\dot{\epsilon}_{xx} = \frac{\partial u}{\partial x},
\]

\[
\dot{\epsilon}_{yy} = \frac{\partial v}{\partial y},
\]

\[
\gamma_{xy} = 2\dot{\epsilon}_{xy} = \frac{\partial v}{\partial x} + \frac{\partial u}{\partial y},
\]

and

\[
\dot{\varepsilon}_{\text{eff}} = \sqrt{\frac{4}{9} \left( \frac{1}{2} \left( \dot{\epsilon}_{xx} - \dot{\epsilon}_{yy} \right)^2 + \dot{\epsilon}_{xx}^2 + \dot{\epsilon}_{yy}^2 + \frac{3}{4} \gamma_{xy}^2 \right)},
\]

where the velocity and strain rate in the \(x\) direction is denoted as \(u\) and \(\dot{\epsilon}_{xx}\), that in the \(y\)-direction as \(v\) and \(\dot{\epsilon}_{yy}\), respectively, and the shear strain rate is denoted as \(\gamma_{xy}\). The spatial resolution of the displacement, strain rate and strain measurements is determined by the length scale over which the cross-correlation is evaluated [17]. In this study, the spatial resolution of the displacement, strain and strain-rate fields resulting from this discretization is 10 pixels or approximately 70 \(\mu\)m. The measurement uncertainty of the numerical PIV code was determined by quantifying errors in measuring simulated displacements of imaged marker fields. For the experimental conditions used in this study, the mean velocity error was 0.0042 mm s\(^{-1}\) and the mean strain rate error was 0.013 s\(^{-1}\). The mean strain error was determined to be 0.040, which is approximately 5% of the maximum strain in these experiments.

Changes in grain structure and material strength were used to characterize the extent of the deformation zone and also to evaluate the magnitude of the deformation in the region of the indentation. To reveal the deformed microstructure, the copper specimen was mechanically polished and subsequently etched using a mixture of hydrogen peroxide (5 ml), ammonium hydroxide (25 ml) and distilled water (25 ml). Electron backscatter diffraction (EBSD) and transmission electron microscopy (TEM) were used for the observation of moderately and highly deformed regions. EBSD measurements were performed on a high-resolution scanning electron microscope (SEM) FEGSEM LEO ULTRA 55 at a working distance of 15 mm and an accelerating voltage of 20 kV. EBSD software from HKL Technology/Oxford Instruments was used to acquire and process the EBSD data. TEM samples were prepared by a focused ion beam lift-out technique using an FEI Nova 200 NanoLab DualBeam SEM. After lift-out, the sample was further processed in the DualBeam SEM using a low-voltage, low-current ion condition so as to remove any damage incurred owing to previous ion milling of the material. The electron transparent sample was then imaged using an FEI Tecnai TEM operating at 200 kV.

Local material strength was characterized through measurements of Vickers hardness on the midplane of the copper sample exposed after sectioning. Indentation load (200 g) was set to yield an approximately 50 \(\mu\)m indent diagonal length and approximately 500 indentations were made with a mean indent spacing of 300 \(\mu\)m. For the purpose of relating hardness to effective strain, an empirical relationship between hardness and strain was established through a series of controlled uniaxial compression tests. Copper billets were deformed to strains from 0.1 to 1.2 and the resulting Vickers hardness measured. To avoid barrelling of the samples, the platen was well lubricated and machined flat between successive passes. The relationship between hardness \((H)\) and strain \((\varepsilon)\) was determined and is given by equation (3.5),

\[
H(\varepsilon) = 140.1e^{0.2117},
\]
Figure 2. Measured velocity fields in the copper specimen deformed at an indentation rate of $V_0 = 1 \text{ mm s}^{-1}$ up to indentation depths of: (a) $d = 0.035$, (b) 0.105, (c) 0.245, (d) 0.490, (e) 0.735 and (f) 0.980 mm. (Online version in colour.)

where $H(\varepsilon)$ has units of kg mm$^{-2}$ and the fit yielded an $R^2$ value of 0.97. Equation (3.5) is similar to comparable measurements made by Chaudhri [9], who found $H(\varepsilon) = 132 \varepsilon^{0.1968}$ for copper.

4. Results

(a) Particle image velocimetry measurements

Figure 2 shows the evolution of the velocity field at the start of indentation up to a depth of approximately 0.980 mm as a series of contour maps with colour indicating the magnitude of the local material velocity. Quiver plots are superposed on these maps, with the size and direction of the vectors indicating magnitude and orientation of the velocity. In the experiments, the copper workpiece was moving while the indenter was held stationary. The flow fields presented here, however, are shown such that the indenter is moving into the copper workpiece. This transformation is made through subtraction of the nominal indentation velocity and is akin to representations typical of indentation research. The progression of the flow field development in the frames shows that the distribution of the velocity field is initially diffuse and sharpens with increasing indentation depth up to 0.245 mm (figure 2a–c). Beyond this depth and up to an indentation depth of 0.980 mm, as in figure 2d–f, the velocity distribution in the deformed copper does not appear to change significantly, indicating that the velocity field is in a quasi-steady state. This is further evident in figure 3, which shows the evolution of the velocity profile along the centre line of the indenter with increasing indentation depth. The changes in the velocity field profile at the start of indentation are clearly greater than those developed at greater indentation depth.

In the velocity fields corresponding to the greater indentation depths, the magnitude of the bulk velocity is negligible in the regions near the extent of the field of view and approaches that of the crosshead ($V_0 = 1 \text{ mm s}^{-1}$) in the vicinity of the indenter. In this regard, the maximum magnitude of the velocity (approx. 0.8 mm s$^{-1}$) occurs in a semicircular area immediately underneath the indenter. Angular deviation of the velocity direction from the indentation axis is the highest in the region that immediately borders this semicircular area (approx. 17°). This deviation of material flow is clearly observed in the composite image in figure 1d. Maximum
velocity in the region underneath the indenter and deviation of material flow around this area are consistent with the notion of material stagnation underneath a flat punch—the so-called dead-metal zone. The dead-metal zone is a region of non-deforming material anchored to the indenter face oft attributed to the effects of the no-slip condition present on the indenter surface [18]. From figure 2, the semicircular dead-metal zone extends the maximum of approximately 0.5 mm into the workpiece subsurface and along the indentation axis. Further into the deformed region, the velocity gradually decreases and the majority of the velocity change ($0.8V_0$) occurred over a region approximately 6 mm into the specimen from the indenter face and along the indentation axis. Gradation of the velocity is the steepest in directions transverse to the indentation axis.

Strain rate is an incremental measure of strain accumulated over unit time and thus is useful in revealing regions wherein strain is locally accommodated during indentation. Figure 4a–f shows the evolution of the strain rate field during indentation, as obtained by spatial differentiation of each velocity field. In the deformed copper, regions of intense strain rate analogous to slip lines, but of finite width, are evident and emanate from the edges of the indenter face at approximately right angles. With increasing distance from the indenter, the slip lines curve towards and meet at the indentation axis. From the progression, evolution of the strain rate field closely follows that observed for the velocity field in figure 2, as the slip line patterns become more sharply defined beyond an indentation depth of 0.245 mm. At the greater indentation depths, these slip lines extend over a region of 0.5–2.5 mm from the indenter face and along the indentation axis. The average strain rate in these regions is approximately $0.4 \text{s}^{-1}$, but there are distinct locations within the bands, especially at the corners of the indentation where the strain rate is as high as approximately $0.6 \text{s}^{-1}$. Further, immediately underneath the indenter is a semicircular region wherein the strain rate is negligible. The absence of appreciable strain rate and/or material flow in this region indicates that strain is not accumulated within this region, consistent with the formation of a dead-metal zone underneath the flat punch.

The overall strain accumulated during indentation, obtained by integration of the strain rate fields, is shown in figure 5. Several characteristics of the field are important. Firstly, a semicircular region of low strain ($\varepsilon \sim 0–0.08$) prevails beneath the indenter over a region extending from the indenter face to a distance of approximately 1 mm along the indentation axis. This low strain confirms material stagnation underneath the indenter. The dead-metal zone is encased by a band approximately 1 mm wide of relatively high strain (up to $\varepsilon \sim 0.7$). Decay of the strain into the workpiece surface is evident with increasing distance from the indenter. Similar to the velocity distribution, the strain decay is relatively gradual along the indentation axis and is steeper in other directions. The maximum extent over which the strain is appreciable ($\varepsilon > 0.2$) in the entire region is approximately 4.25 mm along the indentation axis and approximately 4.6 mm perpendicular to the indentation axis.

Figure 3. Velocity profile with increasing distance along the centre line indenter in the copper specimen deformed at an indentation rate of $V_0 = 1 \text{mm s}^{-1}$. (Online version in colour.)
Figure 4. Strain rate fields in the copper specimen deformed at an indentation rate of \( V_0 = 1 \text{ mm s}^{-1} \) up to indentation depths of: (a) \( d = 0.035 \), (b) 0.105, (c) 0.245, (d) 0.490, (e) 0.735 and (f) 0.980 mm. (Online version in colour.)

Figure 5. Accumulated strain field in the copper specimen deformed at \( V_0 = 1 \text{ mm s}^{-1} \). (Online version in colour.)

(b) Microstructure and mechanical properties

Post-deformation analysis of the indentation zone was made through micro-hardness measurements of the copper. Figure 6a shows a mapping of the hardness measurements made in the region beneath the indentation; an increase in hardness above the undeformed bulk hardness value (approx. 75 kg mm\(^{-2}\)) is a consequence of plastic deformation and can reveal the extent and heterogeneity of the plastic zone. The extent of the plastic zone is approximately 5.5 mm along the indentation axis. Perpendicular to the indentation axis, the extent of the plastic zone is also approximately 5.5 mm. Regions of high, moderate and low hardness are clear in the figure and can be used to further characterize the deformation zone. First, a region (I) roughly semicircular in shape and of low hardness (approx. 90 kg mm\(^{-2}\)) extends approximately 1 mm from the surface of the indenter and along the indentation axis. This low hardness value indicates that the copper in this region was subjected to relatively low levels of deformation during indentation. Second, the hardness increases to 135–140 kg mm\(^{-2}\) with increasing distance from the indenter and along the indentation axis, indicative of substantially higher straining in these regions (II). The broadest extent of this region was approximately 1 mm and occurred along the indentation axis. The hardening was the highest (up to 145 kg mm\(^{-2}\)) near the corners of the indenter and this region (III) itself was relatively sharper, indicative of the presence of a zone of intense straining. Last, the hardness decays gradually from 125 kg mm\(^{-2}\) to the bulk value with increasing distance from the
indenter along the indentation axis. This occurs over a region (IV) that extends from 2 to 5.5 mm from the indenter face. From the figure, the progression of hardness from high to low values with increasing distance from the indenter occurred more gradually along the indentation axis than in other radial directions.

The microstructure evolved underneath the indenter is shown in figures 6b and 7. As in the hardness map, several regions (I–IV) exhibit unique microstructure characteristics that can be used to identify the deformation zone. First, a semicircular area (I) that extends approximately 1 mm away from the indenter and along the indentation axis comprised relatively undeformed grains with size (approx. 65 μm) similar to that of the undeformed (bulk) material (figure 7b). Immediately below this region and along the indentation axis is a zone (II) approximately 1 mm in breadth wherein the grain structure is distorted but still identifiable, as in figure 7c. The grains in this region are elongated in directions tangent to the boundary of the semicircular undistorted area. The grain size, here defined as the shortest dimension of the deformed grain, was 30 ± 21 μm. Flanking this zone at the edges of the indenter face are narrow bands (III) wherein the microstructure is severely deformed, as in figure 7d. Striations or flow lines are visible; these are a characteristic feature of severe plastic deformation [2]. The grain structure in these regions could not be resolved by optical microscopy, suggesting a high degree of refinement of the grain structure. Further away from the indenter and outside of this highly strained region, is an area (IV) where the effects of plastic deformation are difficult to observe (figure 7e). The grain size in this region (50 ± 28 μm) is smaller than that of the undeformed bulk and slight changes in the aspect ratio of the grains are seen. These changes in the microstructure and the overall extent of the deformation zone are similar as that seen in the PIV and hardness maps.

The strain imposed in the deformation field can be estimated based on measurements of the angular orientation of striations in figure 7, as the axis of these striations correspond directly to the elongation axis of the individual grains. The effective strain is given by

\[
\varepsilon = \frac{\tan(\theta)}{\sqrt{3}},
\]

where \(\theta\) is the local orientation of the striations (flow lines) with respect to the normal to the indentation axis [19]. This method for estimating strains is useful when considering...
Figure 7. (a) Enlarged view of the microstructure developed underneath the indenter. Four regions (I–IV) with distinctive microstructure characteristics are identified in (b–e). Effective strains estimated based on the local orientation of the striations (equation (4.1)) are inset in the figure. (b) Region I: underneath the indenter and retaining the original microcrystalline grain structure. (c) Region II: elongated grain structure. (d) Region III: at corners of the indenter and grain structure cannot be resolved owing to the severity of deformation. (e) Region IV: far from indenter where deformation is less pronounced. (Online version in colour.)

Visualization of the deformed microstructure at higher resolution was enabled by EBSD and TEM. Figure 8 shows image quality (IQ) maps with orientation information from the inverse pole figures (IPF) embedded. These IPF maps were taken parallel to the indentation direction in the same regions as figure 7. Figure 8b shows the grain structure in the area (I) immediately underneath the indenter, which is more clearly resolved with the EBSD. The microcrystalline grains appear to be separated by planar, high-angle (more than 10° misorientation) boundaries that do not exhibit significant distortion relative to that of the undeformed bulk material (figure 8a). Although the majority of grains in the micrograph appear completely undistorted, the interiors of select grains show evidence of low-angle boundaries (less than 10° misorientation), which are displayed as thin red lines in the figure. These low-angle boundaries are indicative of the formation of cell walls and sub-grains within larger microcrystalline grains; this is an expected result of dislocation rearrangement and subsequent microstructure evolution at very
Figure 8. IPF and orientation distribution function maps showing grain structure/orientation and texture development in the copper in the (a) bulk and each of the regions I–IV in (b–e), respectively. (Online version in colour.)

low strains [14]. In the region below the undeformed semicircular area (II), shown in figure 8c, the grains show significant distortion, with elongation of the grains perpendicular to the indentation axis. Also clear in the figure is the formation of a greater density of low-angle boundaries within the elongated grains. Distortion of the overall grain morphology and the apparent greater content of low-angle boundaries are both indicative of more severe straining in this region. At higher strains, these low-angle boundaries are expected to evolve into high-angle grain boundaries through strain-dependent continuous recrystallization [14]. Figure 8d shows the microstructure evolved in the regions near the corners of the indenter (III). The grain structure here is clearly more distorted than any other area of the deformation zone; the elongated grains in this region have significantly higher aspect ratios and, in several areas, these grains are further sub-divided into smaller low-aspect ratio grains with different crystallographic orientation. The transformation of elongated grains to equiaxed grains is also an expected result of continuous recrystallization at
very high strains. Lastly, figure 8e shows the microstructure at an area far away from the indenter (IV). The microstructure in this region is similar to that in figure 8a, b—microcrystalline grains with planar boundaries—this also suggesting that the area immediately underneath the indenter has seen negligible straining.

The EBSD information is useful for elucidating the texture evolved throughout the deformation field and in each of the highlighted zones. From the pole figures, random texture prevails in several areas underneath the indenter: the undeformed bulk (figure 8a), in the area (I) directly underneath the indenter (figure 8b) and in the area (IV) far away from the indenter (figure 8e). The texture in the moderately strained region (II) underneath the dead-metal zone (figure 8c) resembles that developed in plane-strain compression of OFHC copper [21, 22], with a predominant $⟨110⟩$ component as well as $⟨100⟩$ and $⟨111⟩$ fibre texture. These components correspond to fibre with $⟨110⟩$ lying parallel to the indentation axis ($y$-direction) and fibre with $⟨110⟩$ tilted by approximately 60° from the indentation axis. These are typically referred to as $α$ and $β$ fibres, respectively, and have been observed following the deformation that occurs during rolling, which can be idealized as plane-strain compression. The deformation texture changes considerably in the highly strained region (III) in figure 8d, where the texture is similar to that observed in deformation of copper by simple shear [23–25]. The variation in textures indicates differences in the modes of deformation in the indentation region, where a state of plane-strain compression is present directly underneath the indenter and a state of simple shear is present at the corners of the indenter. Correlation of these textures to the overall strain tensor obtained by PIV is the subject of ongoing investigations.

Transmission electron microscopy was used to facilitate high-resolution observation of the microstructure in the severely distorted region located at the edges of the contact with the indenter (III). Figure 9 shows a pair of TEM micrographs with selected area diffraction (SAD) patterns (inset) so as to obtain a measure of the deformed microstructure in this region. The bright-field micrographs show evidence of the development of dislocation structures 100–300 nm in extent. Arcs of intensity are evident in the SAD patterns and indicate the development of preferred texture in the deformed microstructure. Grain structure sizes are approximately 225 nm along the major axis and 125 nm along the minor axis. The development of a sub-micrometre grain structure, with increasing levels of grain misorientation, is consistent with large strain deformation of copper [26].

5. Discussion

Knowledge of the deformation field in indentation is potentially useful in providing a basis for the development and validation of admissible velocity and deformation fields that underlie upper bound solutions of plastic flow. Indentation also shares geometric similarities to industrially
relevant deformation processes, including stamping and machining. Thus, a better understanding of the deformation field in indentation can have bearing for the development of insight into other important processes. In this work, a comprehensive study of the deformation field developed in copper during flat punch indentation was undertaken through an in situ imaging analysis coupled with detailed metallographic and microstructure characterization of the deformation zone. This ensemble of measurements provides a systematic characterization of the heterogeneity of this field, revealing how the evolution of strains and strain rates leads to the development of specific microstructures and textures underneath the indenter. A salient characteristic of these measurements is their mutual consistency in indicating deformation zone geometry. From these results, the heterogeneous nature of the deformation field can be described as follows: (i) a region that underwent near-zero straining in the vicinity underneath the indenter (e.g. dead-metal zone), (ii) regions of moderate and low plastic strain at greater distances from the indenter face and (iii) a high-strain region that exists near the edges of the contact (referred to as a ‘slip line’ or ‘shear band’). Figures 2 and 4 show the variations in the velocity and strain rate fields as contour maps underneath the centre line of the indenter; both exhibit temporal variation (transient) at small indentation depths and steady-state behaviour with minimal spatial and temporal variations at greater indentation depths.

(a) Dead-metal zone (I)

The dead-metal zone is a key feature of the mathematical analysis of the flat punch indentation problem by Prandtl [27] and also the simulation results of several other investigators [28–30]. However, experimental evidence of the dead-metal zone has been limited to observations made using PIV-based methods in indentation of pure lead (a non-strain-hardening metal) and granular systems [1,16,31]. These results extend these observations and analyses to a strain-hardening metal (copper) and provide new information in baseline comparisons of these PIV measurements of strain and strain-rate with evolved mechanical properties, microstructure and crystallographic texture in the deformed material.

In the region immediately underneath the indenter (I), the local material velocity closely matched that of the moving crosshead. This suggests stagnation of material flow in this area and is indicative of rigid body motion of the material with either zero or appreciably low levels of deformation. This was confirmed by a PIV strain estimate in the range of \( \varepsilon \sim 0–0.08 \). Low straining in this region also is clear from the hardness and microstructure measurements. Figure 6a shows negligible hardening in this region with values (approx. 80 kg mm\(^{-2}\)), similar to that of the undeformed bulk copper material (approx. 75 kg mm\(^{-2}\)). From equation (3.5), this hardness value corresponds to a strain as low as 0.07, similar to that of the PIV-based measurement.

The etched micrographs and IPF maps also show that the grain structure in this area closely resembles that of the bulk material with regard to grain size and morphology. Another indication of the relatively undisturbed nature of this region comes from crystallographic texture. Orientation information provided in the pole figures showed that the textures developed in this region and that of undeformed bulk material were similar, both exhibiting random texture that is to be expected in the initially annealed and recrystallized copper. While each of these analyses indicated the clear presence of a dead-metal zone, the geometry of this region is more clearly resolved in the PIV strain field and the micro-hardness mapping. In comparison, the transitions between zones are difficult to detect in the etched micrograph owing to the insensitivity of changes in grain morphology at relatively low strain. From the PIV measurement, the dead-metal zone appears to be roughly semicircular and approaching the edges of the indenter contact, extending approximately 1.25 mm (almost equal to the half width of the indenter) into the subsurface from the indenter face. This is somewhat different from our prior investigations in pure lead, wherein the dead-metal zone was found to be significantly larger (almost equal to the width of the indenter) and elliptical in morphology.

Understanding of the dead-metal zone geometry is potentially important as its presence alters the effective indenter geometry and the resulting deformation mechanics. In this study,
for copper, the indentation is more closely related to that of semicircular indentation owing
to the presence of the non-deforming zone of material. The geometry of the dead-metal zone
may be somewhat similar for blunt indenters (i.e. wedge indenters with very large apical
angles) and different material systems. Preliminary investigations of wedge indentation in
copper under different indenter geometries have shown that the deformation zone morphology
is significantly dependent on the indenter geometry. For example, the primary deformation
mechanism transforms from cutting-dominant to compression-dominant modes, as wedge apical
angle increases [32,33]. In this regard, the size and shape of the dead-metal zone likely is a function
of several parameters, including workpiece material, indenter geometry and loading rate,
especially for materials that exhibit rate-dependent loading. Experiments presently underway
will inform a better understanding of the effects of these parameters on the geometry of this zone.

(b) Moderately and severely deformed regions (II–IV)

The strain rate fields in figure 4 show that directly underneath the dead-metal zone are bands of
intense strain rate wherein strain is primarily accumulated, similar to shear bands and velocity
discontinuities in a slip line field. After their initial development, these shear bands do not change
significantly throughout indentation. The hardness values for this region are in the range of
120–130 kg mm\(^{-2}\). The high hardness in these bands is a plausible cause of the ‘bifurcated’
flow field seen in the copper around the indenter (figure 1d), a common characteristic with the
formation of an adiabatic shear band [34]. An understanding of the strain levels, given these
hardness numbers, can come from empirically derived relationships between hardening and the
strain imposed during controlled deformation experiments, e.g. uniaxial tension/compression.
From equation (3.5) and the hardness mapping of figure 6a, \(\epsilon \sim 0.5–0.7\) is expected. This is
comparable with the strain field obtained using PIV, which showed that the strain in the shear
band region (II) is relatively high at \(\epsilon \sim 0.45–0.65\).

The strain rate fields of figure 4 also show that the highest strain rate approximately 0.6 s\(^{-1}\) is
observed near the edges of the indenter, in regions (III) where the shear bands intersect the free
surface of the copper specimen. From the PIV, the strain is \(\epsilon \sim 0.7\), which is the highest found
in the field. Figure 6a shows that the hardness is also the highest at these points approximately
135–140 kg mm\(^{-2}\). This high hardness is similar to that observed in severe plastic straining of
copper using other deformation configurations, e.g. compression [35], extrusion [36] and chip
formation in machining [26]. From equation (3.5), these hardness levels correspond to strains
from \(\epsilon \sim 0.8\) to 1.0. It should be noted that while the empirical relationship in equation (3.5) can
yield a first-pass understanding of the strains in the deformation field, implicit in its application is
the assumption that the strain path is similar to that underlying uniaxial compression throughout
the deformation process. No such assumption is made in the PIV framework.

The high level of straining in this region resulted in severe distortion of the microstructure,
which appeared as ‘flow lines’ in the etched micrographs of figures 6 and 7. The change in
angular orientation of these flow lines indicated strain levels in the range of \(\epsilon \sim 0.7–0.9\), consistent
with the PIV and hardness measurements. The occurrence of the striations as distinct sets of
curved lines in the vicinity of the indenter closely matches the location of the shear bands in
figure 4. While these flow lines are readily seen in the etched micrographs and EBSD IQ maps,
resolution of individual grains was not possible, suggesting that the crystal size was appreciably
small. The TEM micrographs of figure 9 show that the grain/sub-grain size in these regions
was in the range of 200 nm. This sub-micrometre grain structure is expected as the dislocation
generation/accumulation rate has been shown to be inversely proportional with the distance to
the indenter corner in punch indentation [37]. Further, the deformation textures of figure 8 show
evidence that the deformation at the corners of the indenter is consistent with that developed
in intense shear deformation. This can be compared to the texture developed underneath both
the indenter and dead-metal zone, which is more consistent with deformation by compression at
low strains.
Gradual tapering of the strain levels was observed with the PIV and the hardness maps at increasing depth into the deformed material (IV). This gradation of strain was consistent with lower hardness measurements and less severely deformed microstructures further from the indenter. Detection of this gradation is significantly more difficult in the etched microstructures of figures 6b and 7 owing to the insensitivity of changes in grain morphology in response to low strain deformation. In this regard, hardening is a more sensitive post-mortem technique able to reveal the overall extent of the deformed zone at least as well as the \textit{in situ} PIV method.

\textbf{(c) Effect of lower deformation rates and hardening}

A preliminary experiment with indentation at lower speed (0.01 mm s\textsuperscript{-1}) was also conducted; the results are shown in figure 10. The peak load at the lower speed was approximately 50 MPa, as opposed to 60 MPa at the higher deformation rate (figure 10a). These differences are expected owing to strain rate sensitivity of copper [38]. Although the loading response was quite different between the two materials, similarities existed between the hardness maps (figure 10b) and microstructure of the deformation field. The hardness map revealed the same gradation of hardness throughout the deformation zone and also a region of low hardness directly underneath the indenter. One observation made was that the peak hardness values in the shear band are somewhat smaller at 135 kg mm\textsuperscript{-2}. This may be explained by subtle differences in the microstructure developed underneath the indenter owing to the lower strain rate. Measurements of the deformed microstructure in the shear band and PIV measurements of strain and strain-rate should elucidate any such effects in this regard. The microstructure developed underneath the indenter also exhibited similarities to that of the higher deformation rate, in that the same regions of figure 7 are present at the lower deformation rate.

Lastly, the overall character of the deformation field in the case of either deformation rate is significantly different from what has been observed in flat punch indentation of a perfectly plastic material (e.g. no hardening) [1], wherein the deformation is far more localized and occurs in a pattern with three distinct regions: a dead zone, a transition zone and an exit zone. In this case, the hardening causes a semicircular dead zone that is abutted by a deformed zone to form. The deformation field remains predominantly semicircular but extends significantly further into the bulk of the material, causing a bifurcated flow around the indenter over greater distances.
6. Conclusion

This study investigated plastic heterogeneity in plane-strain indentation of copper by a flat punch. Deformation parameters (e.g. strain, strain rate) in the indentation region and their evolution were measured in situ using an image correlation algorithm based on PIV. These measurements were coupled with complimentary post-mortem measurements of strength, microstructure and texture in the indentation region. Self-consistency in the description of the plastic field by the mechanics and materials characterization has elucidated several key insights regarding punch indentation of strain-hardening materials. In this regard, stagnation of material flow underneath the indenter, presence and orientation of shear bands and transitions in plastic modes between shear and compression have been revealed. Additionally, differences between the formation of the dead-metal zones in strain-hardening and rigid-plastic systems were established. Experiments at varying indentation speeds and alternate indentation geometries, as well as more detailed analysis of loading behaviour, will build on these results and highlight differences and similarities between the deformation zones in indentation of strain-hardening and rigid-plastic material systems.

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