Experimental characterization of meso-scale deformation mechanisms and the RVE size in plastically deformed carbon steel

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Abstract
The local deformation response of low carbon steel subjected to uniaxial tensile loading is investigated, and the local strain field at sub-grain scale is obtained using high-spatial-resolution digital image correlation. The implemented digital image correlation method enables the observation and study of inhomogeneous deformation response at microstructural levels. Detailed local deformation mechanisms including mesoscopic slip bands are captured. Furthermore, the local information is used for the determination of representative volume element size in polycrystalline low carbon steel. To obtain the representative volume element size, we proposed and successfully implemented a strain variation method. Further, the influence of global strain on the local deformation mechanisms and representative volume element size is discussed. The challenges associated with the local strain measurement using digital image correlation are also discussed.

KEYWORDS
digital image correlation, low carbon steel, multiscale experiments, representative volume element, subset

1 | INTRODUCTION

It is well established that polycrystalline metals exhibit significant deformation heterogeneity at micro-scale/meso-scale. This local inhomogeneous deformation response can be due to grain–grain interactions, slip bands, twinning, and so forth and may lead to deformation patterning. Such micro-scale/meso-scale local deformation heterogeneity has so far been investigated in a great body of literature, revealing the significant influence of microstructural parameters, for example, grain orientation, grain size, texture, and crystal structure on the development of plastic deformation heterogeneities in crystalline materials.[1–9] Micro-scale/meso-scale experiments conducted on polycrystalline copper,[1] aluminum,[5–7] and titanium[2] have confirmed high degrees of plastic deformation heterogeneities at grain and sub-grain scales, whereas the degree of such deformation heterogeneity has been observed to be more pronounced at triple junctions and grain boundaries.[6,7]

Majority of the studies on the subject have essentially intended to capture the local deformation response, in order to validate micromechanical and crystal plasticity modeling approaches.[10] The micromechanics models basically take advantage of homogenization techniques to determine the bulk material parameters from the local constitutive response of the material. However, the so-called homogenization algorithms are required to be performed within a portion of the material which, according to Hill,[11] is (a) “entirely typical of the whole mixture on average” and (b) “contains a sufficient number of inclusions for the apparent overall moduli to be effectively independent of the surface values of traction and displacement, so long as these values are macroscopically uniform.” Therefore, the micromechanical modeling schemes are always accompanied by the concept of representative volume element (RVE). Accordingly, the determination of the RVE size in a crystalline material is of great significance.

Identification of the length scales of the RVE in elastically deformed composites and polycrystalline metals has been a subject of study for decades, whereas a wide range of values for the RVE size has so far been documented.[2,12–20] There have also been a number of studies...
 attempting to determine the RVE size of polycrystalline structures deformed within nonlinear/plastic regions.\textsuperscript{[16,20]} A list of relevant studies documenting various values for the RVE size of several material types is shown in Table 1. As detailed in Table 1, the RVE size values reported for a variety of material systems are within a very widespread range, reportedly from 8 to more than 600 grains. It is also worth noting that the research on the subject of RVE has mainly been conducted on the basis of numerical approaches. To the authors’ knowledge, there have been very few studies to date attempted to characterize the RVE size experimentally. To mention a few, Liu\textsuperscript{[14]} determined the RVE size of an elastically deforming PBS9501 using an experimental-based approach and obtained an RVE size of 3,375 crystals for the examined material. Another experimental approach documented by Efstathiou et al.\textsuperscript{[2]} indicated that the RVE of plastically deformed titanium encompasses ~27–30 grains. Efstathiou’s method required a large window size to get an accurate RVE size. If only small window size is considered, the line fitting method will give a converging value but could underestimate the RVE size.

The reason behind such scarce experimental-based studies might have been due to the relatively impractical methods of full-field deformation observation at microstructural levels. However, in recent years, the study of local deformation response of materials at micro-scale and meso-scale has been facilitated following the development of small-scale digital image correlation (DIC).\textsuperscript{[21–38]} A combination of scanning electron microscopy (SEM) and DIC has made it possible to study the full-field deformation phenomena at nanoscale and micro-scale.\textsuperscript{[26]} However, there are complex challenges associated with SEM-DIC, which need to be dealt with for a correct implementation of the method. The limitations associated with non-conductive materials, application of high-quality speckle pattern suitable for image correlation purposes, necessity for noise minimization, and the presence of a variety of image distortion sources are challenges introduced in the application of SEM-DIC.\textsuperscript{[26,36,37]} On the other hand, optical DIC is proven to be the preferred and easier-to-implement method particularly for meso-scale deformation study of materials.\textsuperscript{[12,24,28]} However, the speckling of the sample for high-spatial-resolution experiments is still a challenge. Recently, different speckling methods including direct deposition of nanometer-sized particles have emerged as a promising method for high-resolution micro-scale and meso-scale DIC at grain and sub-grain scales.\textsuperscript{[31]}

Although several studies have been dedicated to investigate the full-field deformation response at grain-size scales, there still exists a gap in a thorough experimental-based quantitative analysis of the RVE in metals, an experimental analysis that takes into account the influences of plastic deformation at grain and sub-grain levels, not to mention that the only research relating to the subject has been conducted on pure titanium with a hexagonal close-packed crystalline structure. Accordingly, the present work mainly focuses on the characterization of the RVE in a polycrystalline metallic specimen under plastic deformation, with an emphasis to obtain the length scale at which the divergence between the meso-scale and continuum scale response takes place. In line with the previous investigations, the present study also uses 2D DIC to conduct surface measurements, and thus, the results obtained here basically characterize representative surface element. Although there is still no universal correlation between representative surface element and RVE, the RVE size of the material can be estimated by simply assuming that the thickness direction also includes a similar number of grains in the representative surface element.\textsuperscript{[2]} On the basis of the results obtained from macro-scale and meso-scale experiments, and following statistical approaches discussed in forthcoming sections, the scale at which separation of meso-scale and macro-scale deformation takes place is determined for the case of a metallic specimen with body-centered cubic crystalline structure. The RVE size of the material during plastic deformation, along with the effects of length scale, and the strain magnitude on RVE size have also been presented and discussed.

**Table 1** Summary of RVE size of polycrystalline materials obtained through numerical and/or experimental approaches

<table>
<thead>
<tr>
<th>Type of material</th>
<th>Type of study</th>
<th>RVE size (number of grains = N)</th>
<th>Ratio of linear length scale to grain size (N\textsuperscript{1/3})</th>
<th>Deformation region</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>Polycrystalline (copper)</td>
<td>Numerical</td>
<td>445</td>
<td>7.63</td>
<td>Linear elastic</td>
<td>[20]</td>
</tr>
<tr>
<td>Polycrystalline (copper)</td>
<td>Numerical</td>
<td>550</td>
<td>8.19</td>
<td>Linear elastic</td>
<td>[17]</td>
</tr>
<tr>
<td>Polycrystalline (bi-phase and single phase)</td>
<td>Numerical</td>
<td>8</td>
<td>2</td>
<td>Inelastic</td>
<td>[16]</td>
</tr>
<tr>
<td>Polycrystalline (cubic)</td>
<td>Numerical</td>
<td>400 or less</td>
<td>7.37</td>
<td>Linear elastic</td>
<td>[18]</td>
</tr>
<tr>
<td>Polycrystalline (cubic)</td>
<td>Numerical</td>
<td>200 or less</td>
<td>5.85</td>
<td>Linear elastic</td>
<td>[19]</td>
</tr>
<tr>
<td>Polycrystalline (mild steel)</td>
<td>Numerical</td>
<td>632</td>
<td>8.58</td>
<td>Viscoplastic</td>
<td>[15]</td>
</tr>
<tr>
<td>Polycrystalline (copper)</td>
<td>Numerical</td>
<td>400</td>
<td>7.37</td>
<td>Viscoplastic</td>
<td>[20]</td>
</tr>
<tr>
<td>Polycrystalline (titanium)</td>
<td>Experimental</td>
<td>27</td>
<td>3</td>
<td>Plastic</td>
<td>[2]</td>
</tr>
<tr>
<td>Polymer-bonded explosive</td>
<td>Experimental</td>
<td>3,375</td>
<td>15</td>
<td>Linear elastic</td>
<td>[14]</td>
</tr>
</tbody>
</table>

R = scale at which the average meso-scale strain is equal to the continuum scale strain; δ = average grain size; RVE = representative volume element.

\[N = (R/δ)^{1/3}\].
2 MATERIALS AND METHODS

2.1 Specimen geometry and preparation

Commercially available cold-rolled and partially annealed AISI 1018 steel specimens were examined in the present study. The crystalline structure of the examined steel in this work is body-centered cubic. Fully annealed specimens were not regarded here in order to minimize the effects of Lüders band formation and yield point phenomena. The occurrence of such phenomena can substantially elevate the complexity of the meso-scale measurements.\[39\] Sub-size flat dog-bone specimens were extracted from \~0.8-mm-thick as-received sheets. The dimensions of the specimen in the gage area were 16 × 4 × 0.73 mm\(^3\), whereas the gage length-to-width ratio was maintained at 4 to meet the American Society for Testing and Materials (ASTM) standard and obtain the macro-scale constitutive response under tension. To facilitate the microstructural observations, particularly in order to capture images of the grain structures before and after deformation, each specimen was mechanically ground using grits ranging from 240 to 1,200, whereas standard metallography procedure was followed during the subsequent polishing. Finally, specimens were etched using a 2% Nital etchant solution to reveal the grain structure. Grain structure of the examined material is shown in Figure 1. Using optical microscopy and standard intercept method, the average grain size of the steel specimens was estimated as 20 \(\mu\)m.

To align the strain mapping area with its corresponding grain structure, a small region on the specimen surface was marked with Vickers micro-indenters as reference points (see Figure 1). Accordingly, 8 \(\mu\)m indentation marks were used to inscribe a rectangular area of interest on the center of the specimen. Rectangular areas of interest with three different dimensions were marked and speckled to allow meso-scale DIC analyses at three different magnifications (see Figure 2). As shown in Figure 2, the size of the inscribed rectangular area (field of view) is inversely proportional to the magnification. The largest rectangle marking is used for DIC experiments at 20\(\times\). The rectangular area inscribed at the center of the 20\(\times\) rectangle is used for the analysis at 40\(\times\) magnification, whereas the smallest rectangle marked at the bottom corner is used for measurements at 100\(\times\) magnification. The size of the smallest rectangle corresponding to the highest magnification in this experiment is approximately 124.8 × 100 \(\mu\)m\(^2\), containing approximately 31 grains.

To prepare the specimen for high-resolution DIC, a high-contrast fine speckle pattern must be applied on the specimen surface. Also, to allow for the in-grain strain measurement, the speckle size must be sufficiently small such that a large number of speckles can be incorporated within a single grain. In the present work, directly deposited submicron-sized particles of Rhodamine 6G were used to apply a uniform speckle pattern on the specimen. For this purpose, 1 mg of Rhodamine 6G was first dissolved in 10 ml of methanol to form a solution with a dark pink color. A droplet of the mixture was then placed on the surface of a polished and etched specimen and was given enough time to dry. Upon evaporation of the methanol, a fairly uniform high-contrast discrete speckle pattern was produced by the retained solid Rhodamine 6G particles. The solid particles adhere well to the specimen surface, and the speckle size achieved by this method is in the range of 500–1,000 nm, small enough to facilitate intergranular and intra-granular DIC measurements. Typical images showing the achieved speckle pattern at different magnifications, along with their corresponding gray-scale histograms, are depicted in Figure 3. The gray-scale intensity of the speckle patterns in all magnifications shows a bell-shape distribution, suitable for DIC.\[37\] For macro-scale strain measurement, the specimen is flipped and speckled with macro-scale speckles.

2.2 Tensile experiment

A miniature electric-driven tensile load frame, as shown in Figure 4, with the maximum load capacity of 2,250 N was
used to apply quasi-static tensile loading on the specimen. Uniaxial tensile testing was conducted at room temperature, and at a displacement control mode with constant cross-head speed of $1.7 \times 10^{-3}$ mm/s, corresponding to a mean strain rate of $10^{-4}$ 1/s. Tensile loading was conducted in three cycles, applying an overall global strain of 4.23% after the last loading cycle. In each cycle, the macro-scale strain was measured in situ with the help of a macro-scale DIC on the other face of the specimen. All the data points acquired from the macro-scale DIC is averaged to get the global applied strain.

The loading increments are shown in detail in Figure 5. Nominal residual strains accumulated in the specimen after each loading increment were used to compare the global and local strain distributions within the specimen. After each deformation cycle, the specimen was removed from the tensile frame and imaged at different magnifications from its speckled area. High-magnification imaging was performed using an inverted Olympus microscope equipped with a Grasshopper-3 camera, collecting images at a resolution of 3,376 $\times$ 2,706 pixel$^2$. Low depth of field is a challenge in the case of in situ high-magnification experiments. In this
study, the images of the deformed specimens are captured \textit{ex situ} after each deformation cycle. Therefore, refocusing of the area of interest is possible which eliminate the depth of field problems associated with the high magnification DIC. Care is taken to make sure that the test area remains the same during imaging in every cycle. First, the specimen is rigidly secured at one side in a predefined slot made to fit the width and length of the specimen. Every time the specimen is placed in this slot, the location of the area of interest remains the same. In the case of higher magnifications, we had four indentation marks on the lower and left corner of the specimen bounding the area of interest, based on which a small adjustment is maintained in every cycle with the help of a micrometer-assisted microscope stage. The slot and the rigid support also help to minimize any rigid body translation; therefore, the parasite deformation due to rigid body motion of the specimen is negligibly small. The distortion correction procedure was performed by rigidly moving an undeformed specimen to a known distance with the help of micro-scale stage. The correlation of the images shows very small uncertainty in the strain calculation.

3.2 | Distortion correction

The presence of spatial image distortion in optical lenses can result in inaccurate DIC measurements if left uncorrected. This type of image distortion mainly arises from the spherical geometry of the imaging lenses and causes large errors in the calculation of displacement and the corresponding strain fields. The distortion correction algorithm implemented in Vic-2D is used for correcting the field. It uses nonparametric distortion models to correct the strain field. In the present work, the procedure detailed in Schreier et al. and Sutton et al. was followed as briefly discussed below.

A speckled area on the undeformed specimen is marked first. The specimen is translated horizontally to a known distance using a micrometer stage and imaged. The horizontal translation and imaging are proceeded for four intervals. The same procedure is then performed by translating the specimen vertically, capturing images at each vertical interval, as well (see Figure 6). The total translation along $x$ and $y$ directions must be approximately one fourth of the dimensions of the field of view at each magnification. The vertical and horizontal movements are for correcting the displacement in $x$ and $y$ directions. Image correlation is then performed on the images acquired during the described calibration process, keeping the image acquired at positions

![Image](image.png)

\textbf{FIGURE 5} Nominal stress–strain curves obtained at consecutive loading cycles. The residual accumulated plastic strain after each cycle is marked.
$x = 0$ and $y = 0$ as the reference. A B-spline vector function (warping function) is generated using the correlated images to correct the spatial distortion present in the imaging system. A detailed explanation of the technique can be found in Schreier et al.\cite{35}

3.3 | Representative volume element

In the present work, the length scale of the RVE was estimated using the surface strain fields measured at high magnifications. One may argue that conceptually, the RVE size is a three-dimensional quantity and may not be determined by surface measurement. It must be emphasized here that there are two assumptions made while determining the size of the RVE from surface measurement: (a) The number of grains in the RVE aggregate through the thickness direction is similar to that of the surface and (b) the deformation behavior of the material below the surface is similar to the surface deformation response.\cite{5} On the basis of these assumptions, the number of grains included in the RVE, $N$, can be given as follows:

$$N = \left(\frac{R}{\delta}\right)^3,$$

(1)

where $R$ denotes the scale at which the meso-scale strain value is sufficiently close to the continuum scale strain and $\delta$ is the average grain diameter. Note that Equation 1 is valid as long as the grains are equiaxed; that is, the aspect ratio of each grain is close to unity. The criterion used in the present work to determine the RVE size is average strain method and is discussed below.

The RVE size in average strain method is determined considering different window sizes with small increment as shown schematically in Figure 7. To estimate the RVE length scales in this method, an $R \times R \, \mu m^2$ square window is considered at the center of the field of view. The axial strain value within this window is calculated by averaging the local strain values encompassed inside the window. Denoting the calculated average local strain as $\varepsilon_{local}$, the magnitude of the variation associated with an $R \times R$ window size can be defined as follows:

$$\text{strain variation} = \frac{\varepsilon_{local} - \varepsilon_{macro}}{\varepsilon_{macro}} \times 100,$$

(2)
where $\varepsilon_{\text{macro}}$ is the global axial strain at which the image is acquired. The dimensions of the square window are then increased progressively at a step of 2 pixels per iteration, increasing the number of grains encompassed within the window. The window size is increased in steps of 0.368 $\mu$m per iteration, which results in a total of 625 window sizes to reach the final size of 230 $\mu$m. The data considered for the RVE size calculation are an area at a distance of 24 $\mu$m away from the indent marks. Therefore, the influence of indentation marks on the strain field will be negligible as per ASTM E384. By plotting the strain variation values as a function of $R$, one can expect that the variation would take smaller values at larger $R$'s, indicating the convergence of locally measured strain with the strain applied globally. The $R$ values corresponding to strain variation of 1% are regarded in this work as the length scale at which meso-scale and macro-scale strains are sufficiently close, and the corresponding $R$ value is statically representative of the RVE length scale. The rationale for the 1% convergence criterion has been explained in the following section. Accordingly, the number of grains contained in the RVE can be estimated using Equation 1. Different magnifications are used and implemented at a range of plastic strains. The RVE size presented here is on the basis of 40× magnification images.

To compare the RVE size obtained from the proposed average strain method, a box average method by Efstahiou et al. [2] is also used. In this method, the entire field of view at a given magnification is first divided into a finite number of boxes of the same dimensions, each containing several data points. The average value of all the data points within each box is calculated and referred to as box average, $x_i$. The strain averaged over the entire field of view is also calculated as $\mu$. Having obtained the box averages for all the boxes within the field of view, the standard deviation, $\sigma$, of all the box averages is calculated as follows:

$$\sigma = \sqrt{\frac{1}{n} \sum_{i=1}^{n} (x_i - \mu)^2},$$

(3)

where $n$ is the total number of boxes within the field of view (see Figure 8). It is good to mention that the maximum window size that can be achieved with the field of view is half of the field of view in the area of interest. Therefore, to have a window size of 260 $\mu$m, four images at the magnification of 40× are stitched together in the box-averaging technique.

By uniformly increasing the box size at each iteration, the variation of $\sigma$ with respect to the box size can be obtained. Note that by increasing the length scale of the box size, the standard deviation is expected to decrease, because the strain values are averaged over a larger area consisting of a larger number of data points resulting in smaller $\sigma$. Accordingly, the RVE length scale in this method is identified by the box size at which a dramatic change in the value of $\sigma$ with box size takes place. To find the location at which a dramatic change in standard deviation occurs, a line is fitted to the tail end of the curve similar to the procedure discussed in Efstahiou et al. [2] Efstahiou’s method required a large window size to get an accurate RVE size.

### 3.4 Measurement noise consideration for the appropriate RVE criterion

DIC technique has inherent noise from camera, microscope, experimental conditions, and so forth. While calculating RVE size, careful attention has to be taken to achieve the accurate strain measurement for identification of RVE length scale by taking into account the contribution of the measurement noise. In this work, the strain error due to noise is calculated by capturing 10 images of the unloaded specimen and correlating them with the same subset sizes that are used in the RVE size estimation. It is seen that the maximum strain error due to noise in our experiments is 0.01% for a window size of 20 $\mu$m, whereas this strain error decreases as the window size increases (see Figure 9). Using the maximum strain error calculated at 20 $\mu$m window size, the percentage strain error in our measurements can be evaluated on the basis of the smallest plastic strain applied on the specimen (i.e., $\varepsilon_{\text{macro}}$ = 1.33%) as $0.010 / 1.33 = 0.75\%$. This means that strain variation in the order of 0.75% is expected from our measurement uncertainty at the lowest strain levels considered.
Accordingly, the window size at which the global strain value is within 1% difference with the averaged local strain over the window ($R$) can be considered as the RVE size in the average strain method.

## 4 RESULTS AND DISCUSSION

### 4.1 Meso-scale deformation response

Full-field contour maps showing the accumulated residual strain distribution after each loading cycle are illustrated in Figure 10. The strain maps show the evolution of axial strain, $\varepsilon_x$, at different magnifications. The contours indicate a wide range of locally developed strain magnitudes for the axial strain component, also revealing the formation of regions with highly localized strain values, particularly at larger global strains. The strain field exhibits deformation heterogeneity by the formation of patterns initially inclined in approximately $\pm 50^\circ$ angle relative to the loading direction, evidencing the formation of mesoscopic slip bands.\(^{13}\) Although uniaxial tensile deformation was applied on the specimen, Figure 10 exhibits the presence of local compressive (negative) strains, even at global tensile strains of as high as 4.23%. The presence of residual compressive strains after uniaxial tensile deformation has also been evidenced in.\(^{2,40}\) The reason for this type of local deformation response might be due to the boundary conditions imposed on a single grain with a specific crystallographic orientation by its neighboring grains.\(^{10,41}\) Note that the constraints provided by the neighboring grains can result in a significantly complex state of deformation on a specific grain. Additionally, the original processing route (possibly cold rolling followed by temper rolling and heat treatment) of the as-received material might have also resulted in the development of residual stresses within the material.\(^{42}\) Tensile loading of the specimen may or may not be in line with the direction of the residual stress, especially on the specimen surface, possibly yielding in a local deformation response completely different from what is anticipated.

To further probe into the mesoscopic strain field, a microstructure overlay image was shown in Figure 11. The overlay plot was prepared by aligning the Vickers indent mark at the lower left corner of the microstructure (see Figure 1) on the axial strain field obtained from DIC. Overlay plot shows a highly heterogeneous strain (marked by white arrows) within grains. Some grains undergo negligible deformation even at an applied global strain of 4.23% (note white marks in 4.23% global strain). To confirm the possible slip band formations on the surface of the sample, experiments are conducted on specimens highly polished and ready for
microstructure imaging. Figure 12 indicates the slip marks on the surface of the specimen after the application of 4.4% global strain. As shown in figures, the slip lines are not present in all grains; the presence of such strain-free grains was indicated earlier in Figure 11.

An interesting point in the study of meso-scale strain distribution is the trend observed in the level of deformation heterogeneity at different global stains and length scales. This quantitatively shows in Figure 13, where the histograms of the frequency of local strain magnitudes are plotted at different magnifications and strains. Please note that the frequency distribution is normalized by the maximum frequency for each magnification (Figure 13a). A widening trend of the strain histogram indicates the increase in heterogeneity owing to the high displacement resolution. The strain histogram plotted at 1.33% strain at 100× magnification indicates that the strain heterogeneity is significantly increased at higher magnifications (see Figure 13a). On the other hand, for the same magnification, the strain histogram tends to narrow in the case of 2.89% global strain compared with at 1.33% global strain but not substantially as shown in Figure 13. This points out that the strain heterogeneity slightly decreases as the plastic strain increases but the difference is not large enough to change RVE size as shown in Figure 16.

To better understand the effect of magnification on the measurement resolution, full-field strain maps are presented in Figure 14, depicting the local strain distribution at the same location but in different magnifications. For that...
purpose, the strain field is shown only for the area that corresponds to the size of 100× magnification. In this case, the rest of the area for case 20× and 40× are cropped and resize to match the size of 100×. It is clearly shown that at low magnifications, the local details tend to smear out, deteriorating the resolution of sub-grain-level strain measurement. Such smearing effect was similarly observed in Efstahiu et al.\[2\] and is believed to affect the characteristics of the strain distribution, particularly at mesoscopic scales as well as at lower global strains, where the local strain magnitudes are small and can be overlooked at low measurement resolutions. In our study, we have seen no substantial difference by going 100× compared with 40×, except reducing the size of the field of view. Hence, the RVE size is determined on the basis of imaging at 40× magnification.

### 4.2 Representative volume element

Figure 15 shows the locally averaged strain as a function of window size at three different global strains for three independent experiments. As discussed in Section 3.3, the locally averaged strain is obtained by numerically averaging the full-field strain at the specific selected window size. It is clear from the figures that the locally averaged strain approaches the applied global strain as the window size increases. At lower window size, the locally averaged strains are far from the applied global strains. The trends of strains, especially at smaller window sizes, are different for the three experiments. This is expected, as the location of the area of interest is random for each experiment and the strain averaging is performed over a small number of grains, which may not be sufficient to represent the size of an RVE. The area of interest for different experiments, particularly at smaller window size, could contain only a single grain, a boundary between two grains, or even a triple point. Correspondingly, the local strain averaged at smaller window size would be different for different experiments. It can be expected that if the number of grains increases (window size increases) over which the strain averaging is performed, the discrepancies between the global strain and the locally averaged strain reduces as the window size approaches the RVE size of the material. When the size of the RVE is achieved, the locally averaged strain and global strain should be the same for all the experiments performed. It is clear from Figure 15 that the locally averaged strain converges to globally applied strain for both experiments at larger window size. It should be mentioned here that, the strain converges almost at similar window size, starting around 120 μm, for all strain applied and experiments considered, indicating the full-field method can be used to calculate the RVE size of the material.

Further, the length scale of RVE was determined using the methods described in Section 3.3. Figure 16 depicts the strain variation parameter defined earlier in Equation 2), as a function of the window size at different global strain values.
Three distinct regions are identified in this figure, labeled as I, II, and III. In region I ($R < 120 \, \mu m$), the strain variation parameter is oscillating for experiment 1 (Figure 16a) and experiment 2 (Figure 16b) but monotonically decreasing for experiment 3 (Figure 16c) as the window size ($R$) increases. Also, it is clear that for $R$ value lower than 120 $\mu m$, the difference between the global and local strains is substantial; thus, the window size is too small to incorporate a sufficient number of grains and to represent the RVE of the examined material.

In region II ($120 \, \mu m < R < 170 \, \mu m$), the strain variation parameter decreases monotonically for all the experiments and remains below 5%. In the case of a loose convergence criterion, the window dimensions in this region may be regarded as the length scales of RVE for the examined material. However, the RVE length scales determined in this region cannot be considered as an absolute RVE size for two main reasons: (a) The strain variation parameter is still decreasing, indicating there is an optimal value at larger window sizes and (b) as shown in Figure 16, at $R$ value between 120 and 170 $\mu m$, the strain variation parameter is shown to be a function of the applied strain magnitude, clearly visible in experiment 1 (Figure 16a) and experiment 2 (Figure 16b) in Figure 16. This contradicts the fundamental definition of RVE, the independency of the boundary condition for an optimal RVE size. Finally, in region III, that is, $R > 170 \, \mu m$, the strain variation parameter takes very small magnitudes lower than 1%, whereas it is obvious that this parameter remains below 1% and further converges to 0 at larger $R$ values. In addition, Figure 16d shows that the strain variation parameter is substantially different at smaller window sizes for two independent experiments at the same global strain. However, the strain variation becomes smaller and equal at a larger window size indicating the convergence of the window size to the RVE size of the material. It is evident that in Figure 16d, at a larger window size ($R = 177 \, \mu m$), the locally averaged strain is equal to global strain of 1.83% for two independent experiments considered. Accordingly, the optimum linear length scale of the RVE for the material considered in this work was obtained as 177 $\mu m$.

Finally, the estimated RVE size using the strain averaging method was compared with the well-known box-averaging method described in Section 3.3. The standard deviation for different box sizes is plotted in Figure 17. It is seen that the standard deviation decreases monotonically with increasing the box size. To obtain the point at which the drastic variations in standard deviation occurs, a line is fitted to the tail end of the curve as shown in Figure 17 similar to the procedure discussed in. It is apparent that the line departs from the curve at a box size of 165 $\mu m$. This indicates that the RVE size obtained from both methods is close and any of these methods can be used to characterize the RVE length scale of the polycrystalline material.
Using 177 μm as an optimal RVE length scale and knowing the average grain size of the materials (δ = 20 μm), the ratio of length scale to grain size was found to be 8.85. The number of grains (N) within the RVE was calculated using Equation 1 as $N = 8.85^3 \approx 694$ grains. This value is slightly higher than those computed in most numerical-based algorithms but agrees well with the values obtained by Nakamachi et al. for low carbon steel ($N = 8.58$), under viscoplastic loading condition. The discrepancies in the RVE size obtained numerically by researchers could be due to the influence of texture that can significantly alter the results of a numerical approach, whereas its effects are automatically incorporated in experimental-based analyses such as the one presented in this work.

5 | CONCLUSIONS

The local deformation mechanisms in low-carbon steel subjected to uniaxial tension are investigated. On the basis of the full-field measurement, detail local inhomogeneous deformation mechanisms including mesoscopic slip bands are captured and discussed. Furthermore, the RVE in a polycrystalline metal was characterized experimentally using meso-scale 2D DIC. Full-field distribution of axial strain at mesoscopic scales was captured and studied in a systematic approach, and the effects of globally applied strain and magnification were investigated. On the basis of the results obtained from macro-scale and meso-scale experiments and following statistical approaches, the scale at which separation of meso-scale and macro-scale deformation takes place was determined for the case of a metallic specimen with body-centered cubic crystalline structure. The RVE size obtained in our work for a plastically deformed steel specimen was found to be very close to numerically computed value of RVE for the plastically deforming polycrystalline materials. The methodology presented in this work is a general approach and can be implemented to provide reliable evidence on the deformation inhomogeneity at mesoscopic length scales, and to determine the RVE dimensions on any material system.

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