Application of aberration-corrected scanning transmission electron microscopy in conjunction with valence electron energy loss spectroscopy for the nanoscale mapping of the elastic properties of Al–Li–Cu alloys

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Abstract

The stress and strain play an important role in strengthening of the precipitation-hardened Aluminum (Al) alloys. Despite the determination of relationship between the mechanical properties and the precipitation existing in the microstructure of these alloys, a quantitative analysis of the local stress and the strain fields at the hardening-precipitates level has been seldom reported. In this paper, the microstructure of a T8 temper AA2195 Al alloy is investigated using aberration corrected scanning transmission electron microscopy (AC-STEM). The strain fields in Al matrix in the vicinity of observed precipitates, namely $T_1$ and $\beta'$, are determined using geometric phase analysis (GPA). Young's modulus ($Y_m$) mapping of the corresponding areas is determined from the valence electron energy loss spectroscopy (VEELS) measured bulk Plasmon energy ($E_p$) of the alloys. The GPA-determined strains were then combined with VEELS-determined $Y_m$ under the linear theory of elasticity to give rise the local stresses in the alloy. The obtained results show that the local stresses in Al matrix having no precipitates were in the range of 138 ± 2 MPa. Whereas, in the vicinity of thin and thick $T_1$ platelet shape precipitates, the stresses were found to be about 202 ± 3 MPa and 195 ±3 MPa, respectively. The stresses measured in the vicinity of $\beta'$ spherical shape precipitates found out to be 140 ± 3 MPa which was near to the local stress value in Al matrix. Our findings suggest that the precipitate hardening in T8 temper AA2195 Al alloy predominantly stems from thin $T_1$ precipitates.

KEYWORDS
Al alloys, bulk plasmons, geometrical phase analysis, local stresses, scanning transmission electron microscopy, valence electron energy loss spectroscopy, Young's modulus

1 | INTRODUCTION

The stresses, elastic moduli, and strain fields around the strengthening precipitates in Al-based alloys play an important role in improving their structural and mechanical properties. The presence of different intermetallic phases in precipitation-hardened Al-alloys usually results from the presence of various micro-alloying elements and adequate subsequent heat treatment applied to respective systems.
If the strain is experimentally quantified, the calculation of the stress can be achieved using the constitutive relation 

$$\sigma = E\varepsilon$$

where $\sigma$ is the stress, $E$ is the Young's modulus, and $\varepsilon$ is the strain. It is worth noting that the strain is defined as 

$$\varepsilon = \frac{Y_m}{Y_m'}$$

in which $Y_m$ is the Young's modulus of the matrix and $Y_m'$ is the reduced Young's modulus of the precipitate material. The Young's modulus is a critical parameter in the determination of the stress, as it directly influences the magnitude of the stress induced in the material.
various industries for light structural materials and in order to improve the fuel efficiency for aircraft and aerospace materials, different structural designs for light materials have been developed (Jeswiet et al., 2008). An example of these light materials is a commonly used Al alloy from 2000 series, which is named as AA2195. This AA2195 alloy material has received considerable attention from both industrial and scientific communities (Pickens, Langan, & Kramer, 1989; Sanders, 1996). It is an attractive choice for direct and engineered substitution in all weight-critical applications. It has been shown that AA2195 alloy conducting T8 tempering (stretched and artificial ageing conditions) has a combination of excellent fracture toughness (60 MPa√m) and improve strength (~550 MPa) (Wang & Shenoy, 1998) with the Ym value of (~69 GPa) (Hertzberg, 1989). However, an exact and experimental determination of the strain field, the Ym, and hence local stress around observed nanoparticles in this alloy system has yet to be achieved.

In the context of above-mentioned applications, the objective of this study is (a) to evaluate the strain fields via the atomic displacements around each type of hardening precipitate in the microstructure of AA2195. It has been accomplished by the acquisition of AC-STEM images followed by the GPA application to the acquired images. (b) Mapping of the Ym around each type of hardening precipitate within the microstructure of AA2195 alloy. This task has been completed by extracting the Ep from acquired STEM-VEELS spectrum-image datasets. (c) In the end, a direct evaluation of local stresses of multi-phases AA2195 alloy having spatial resolution in the nanometers and stresses resolutions on the order of a few MPa was obtained. Quantification of stress and strain fields at nanoscale truly facilitates the design and hence the improvement of such structural materials.

2 MATERIALS AND METHODS

AC-STEM is a method that has been used in this study to characterize, measure, map, and estimate the strain field and Ym in the vicinity of precipitates and Al-matrix of AA2915 alloy. As mentioned earlier, a commercially available alloy AA2195 in T8 temper has been used as a case study. The nominal chemical composition of this alloy is as follows: 4 wt.% Cu, 1 wt.% Li, 0.36 wt.% Mg, 0.28 wt.% Ag, 0.14 wt.% Zr, and 94.22 wt.% Al. Samples on the T8 condition were solutions of 20 vol.% nitric acid in methanol at the bath temperature of ~20 °C. AC-STEM experiments were carried out in a probe-corrected Titan 80-300 ST instrument. During the experiments, the microscope was set to the accelerating voltage of 300 kV. Moreover, high angle annular dark-field (HAADF) detector was utilized to generate conventional DF-STEM and aberration corrected high-resolution scanning transmission electron microscopy (HR-STEM) images of the microstructure of the alloys. Both DF-STEM as well as HAADF-STEM images were using an analog annular dark-field detector of model 3,000 ADF from E. A. Fischione, Instruments by setting the ratio between the radii of direct-beam and annular-aperture smaller and larger than 3, respectively. This condition for the ratio was met at the camera lengths of 220 mm and 115 mm for acquiring DF-STEM and HAADF-STEM images with 3,000 ADF detector, respectively. The experiments were carried out in the [110]Al orientation. This has been achieved by tilting a particular alloy grain of the specimens to [110]Al zone axis. For the sake of completeness, the [110]Al patterns were generated using selected area electron diffraction (SAED) technique of TEM mode and were recorded using a charge-coupled devices (CCD) slow scan camera of model US1000 from Gatan, Inc.

In order to realize strain maps, GPA was applied to several high resolution HAADF-STEM (HR-STEM) images to determine the strain fields around the various types of precipitates present in Al-matrix of AA2195 alloy. The calculation procedures for the strain field around the observed hardening precipitates require (a) obtaining HR-STEM images for the existing precipitates in the Al matrix. (b) Calculating the fast Fourier transformation (FFT) of the obtained images. (c) Selecting different diffraction spots along different lattice directions. (d) Obtaining the inverse FFT (IFFT) of the obtaining images. (e) Calculating the geometric phase image by using the relationship between the phase of local Fourier component (Pg, r) and the component of the displacement field (μr). This relationship can be expressed by (Pg, r) = 2πgu, where g is a reciprocal lattice vector (Hytch et al., 1998). (f) Calculating the displacement field using the relationship for two independent phase images Pg1(r) and Pg2(r). In this case, the two-dimensional displacement field can be calculated by Hytch et al. (1998):

$$\begin{pmatrix} u_x \\ u_y \end{pmatrix} = -\frac{1}{2\pi} \begin{pmatrix} g_{1x}g_{2y} \\ g_{2x}g_{1y} \end{pmatrix}^{-1} \begin{pmatrix} P_{g1} \\ P_{g2} \end{pmatrix}$$

where gx, gy are two parameters of g in the reciprocal space, and ux, uy are atomic displacement fields in normal coordinate. (g) obtaining the strain field by differentiating the displacement field as the following (Hytch et al., 1998):

$$\varepsilon_{xx} = \frac{\partial u_x}{\partial x}, \varepsilon_{yy} = \frac{\partial u_y}{\partial y}, \varepsilon_{xy} = \frac{1}{2} \left( \frac{\partial u_x}{\partial x} + \frac{\partial u_y}{\partial y} \right)$$

These steps for applying GPA to obtain the maps of strain fields were achieved using a GPA Software Package (or plug-in) from HREM Research, Inc. particularly developed for Gatan’s digital micrograph environment of GMS3.2 version.

The value of Ym in AA2195 alloy was determined from its Ep by using the following relationship between Ep and Ym (Howe & Oleshko, 2004):

$$Y_m = 0.08E_p^{2.5}$$

It is to be noted that the generation of Ym maps requires the application of Equation (3) in pixel-by-pixel manner in the acquired
The estimated absolute thickness values have been reconfirmed using Nakayama, & Furuya, 2008; Ohshima, Kaneko, Fujita, & Horita, 2004). The actual form of Equation (5) is tensor form in which both $\sigma$ and $\varepsilon$ vectors and $Y_m$ is a matrix. Using the GPA technique, it is possible to generate $\varepsilon$ fields in vector form, that is, in perpendicular and parallel to the specimen plane. However, the STEM-VEELS technique yields in the isotropic (or scalar) form is insensitive to orientation because of the plasmons are isotropic. Consequently, the application of Equation (5) and STEM-VEELS datasets will give rise the stresses (and then strains) in vector form only and this is what is performed in this study.

$$Y_m = 0.08 \left( E_p \pm 0.1 \right)^{2.5} = Y^{\text{EELS}}_m \pm 1.15 \text{GPa}$$

in the following way.

$$\sigma = Y_m \varepsilon$$

The stresses ($\sigma$) fields or values now can be calculated theoretically multiplying the Equation (2) with Equation (3) under the linear theory of elasticity and is written in the following equation.

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3 | RESULTS AND DISCUSSION

The microstructure of the specimen after conducting T8 temper was found to be enriched with platelets precipitates and spherical precipitates, as shown in Figure 1. It contains AC-STEM images depicting the respective microstructure of those precipitates. According to the annular bright-field STEM (ABF-STEM) image in Figure 1a and HAADF-STEM image in Figure 1b, the distribution of precipitates was present throughout the Al matrix. Although, rod-like precipitates were particularly found to be accumulating more in the grain boundary regions. Furthermore, the presented images have triple grain-boundary junction formed by grains labeled as “1,” “2,” and “3.” Precipitates were characterized by tilting a particular grain of foil along $[110]_{\text{Al}}$ zone axis. The condition of tilting along $[110]_{\text{Al}}$ zone axis was met for the grain “1” and a region encircled by dotted circle is clearly marked in the Figure 1b to indicate the part of the grain from where the corresponding selected area diffraction (SAED) pattern in the $[110]_{\text{Al}}$ direction was acquired. The acquired SAED pattern is shown in Figure 1c, and it contains the expected diffraction spots of Al matrix. It also contains the streaks along <111> directions, and spots adjacent to the <002> positions. These streaks and spots are due to the platelet precipitate, which is usually known as the $T_1$ phase with chemical composition of $\text{Al}_3\text{LiCu}$ (Kumar, Brown, & Pickens, 1996). Additionally, different superlattice reflections can be observed in the pattern in Figure 1c. The superlattice reflection is attributed to the $\beta'$ phase with a chemical composition of $\text{Al}_3\text{Zr}$, and it has spherical morphology and a $L_1_2$ ordered structure (Khushaim et al., 2015). The orientation relationship (OR) of the precipitates, particularly $T_1$, with Al matrix is of interest as they influence the precipitation strength-hardening of the alloy significantly. The lattice structure for this phase was identified by Hardy and Silcock (1956) as a hexagonal phase with lattice constants $a = 0.496 \text{ nm}$, $c = 0.935 \text{ nm}$, and the group symmetry is $P6/mmm$. The $T_1$ phase has a plate-shaped morphology on the [111] planes of the Al matrix. When embedded in the Al matrix, it is usually extremely thin with a thickness of ~1.3 nm, which is equivalent to the stacking of only five [111] Al planes. On the other hand, the platelets can have a large extension, up to ~100 nm, without losing their coherency with [111] planes of the matrix. Further, the orientation relationship (OR) of the platelets with the Al matrix is found to be (0001)$_\parallel$ // (111)$_{\text{Al}}$ (Decres et al., 2013). The small thickness and large length of the observed $T_1$ platelet can be explained in terms of the high coherency of the edge of the $T_1$ precipitates with the [111] Al atomic planes. The final layer of the structure proposed for the $T_1$ platelet precipitates is found to have an extremely low lattice mismatch with the [111] Al atomic planes ( Howe, Lee, & Vasudevan, 1998). The crystal structure of the $\beta'$ phase is the $L_1_2$ structure. The $\beta'$ phase usually appears to be fully coherent with the Al matrix because its lattice parameter $a = 0.400 \text{ nm}$ is close to that of Al matrix ($a = 0.405 \text{ nm}$) (Starke Jr & Staley, 1996).

Several HR-STEM images of both platelet and spherical shape precipitates in Al-matrix were acquired to visualize their structure at atomic scales. The imaging conditions were selected in such a way that the acquired images exhibit optimized image and diffraction contrasts from precipitates as shown by the images presented in Figure 2. The acquired HR-STEM image in Figure 2a contains both $T_1$ platelet and spherical $\beta'$ precipitates and the atomic structure of the precipitates is clearly resolved in that image. In the same vein, the HR-STEM
images of Figure 2b,c convey the same information but only on the atomic structure of a thin and a thick T₁ precipitates, respectively. The presented images possess minimum or no contributions from lens aberrations, and the interference of diffracting beam in samples. Therefore, AC-STEM images enabled a direct interpretation and observation of atom contrast of the T₁ platelet precipitate in the Al-matrix. Moreover, the presence of the two T₁ platelet precipitates with two different thicknesses can be observed in Figure 2b. These have been classified as follows: thick T₁ platelet (thickness ~ 5 nm) and thin T₁ platelet (thickness ~ 2 nm). Regarding the interpretation of the observed T₁ platelet precipitate at atomic scales, different models have been proposed such as Huang and Ardell (1987), Howe et al. (1988), and Van Smaalen, Meetsma, De Boer, and Bronsveld (1990). After performing the analysis, the Van Smaalen et al. model among these three was found to be matching with the structure of T₁ precipitates presented in Figure 2. It should be noted that HR-STEM image in Figure 2c does not reveal the presence of light atoms of Li in the dark-field due to their lower intensity as compared to heavier Al and Cu atoms. Thus, the matching of Cu atoms in Figure 2c with Van Smaalen et al. model indirectly implies the presence of a mix Al-Li layers at the edge of the T₁ platelet and while a mix of Al-Cu layers within the inner layers. This observation demonstrates the ability of AC-STEM technique to be capable of resolving every atom column in the acquired images of the samples used in this study. A detailed study of the nature and the atomic composition of the interface between the thin T₁ platelet and Al matrix can be found in different studies (Donnadieu et al., 2011; Khushaim et al., 2015).

As mentioned earlier, GPA is an efficient method to measure the strain fields in materials from their HRTEM and/or HR-STEM images. The recent studies show that the application of GPA to HR-STEM
over HR-TEM images has several advantages including the lattice patterns in the acquired HR-STEM images are less sensitive to varying thickness and compositions of the analyzed samples. Moreover, using HR-STEM images allows the determination of strain and compositions simultaneously (Rouvière et al., 2011) which is not possible using HR-TEM images. Therefore, the next step is to apply the GPA technique on the obtained HR-STEM images such as presented in Figure 2. GPA was applied to images where Al-matrix contained T₁ and β₀ precipitates. However, before carrying out the analysis, the procedure to apply this technique is demonstrated on a T₁ platelet and the results are shown in Figure 3. The HR-STEM image of the microstructure that contains a T₁ platelet precipitate and Al-matrix in which the FFT was calculated is shown in Figure 3a. Furthermore, the calculated FFT of the image is shown in Figure 3b. The FFT pattern confirmed the incident beam is parallel to [110]Al orientation of the Al-matrix. Additionally, an observation of the sharp and diffuse streaks along [111]Al direction was also recorded which were due to the lying of the T₁ platelet on the [111]Al planes of the Al-matrix. Based on this observation, it is assumed that the x-axis is parallel to the (111)Al plane and the y-axis normal to the (111)Al plane for the sake of calculations. Therefore, two selected sets of [002]Al and [220]Al lattice fringes are marked as indicated by numbers 1 and 2 in Figure 3b, respectively. The geometric phase images were then calculated with respect to a reference selected in Al-matrix as shown by a box in Figure 3c. The resulting phase images from spot 1 and 2 are shown in Figure 3c, d, respectively. After obtaining the geometric phase images, the displacement fields $u_x$, $u_y$ can be calculated according to Equation (1). Finally, the strain fields can accordingly be numerically calculated in terms of the derivative of the displacement by using Equation (2). By applying this procedure, the $\varepsilon_{xx}$ or along [00-1]Al direction and $\varepsilon_{yy}$ or along [−110]Al direction components of the strain field $\varepsilon$ were determined in the regions of the matrix close to thin T₁ platelets, the thick T₁ platelets, and the spherical β₀ precipitates. The experimental results of the strain fields for these respective precipitates are shown in Figure 4. The strain field components ($\varepsilon_{xx}$ and $\varepsilon_{yy}$) for Al matrix in the vicinity of a thin T₁ platelet are presented in Figures 4a–c, of thick T₁ platelet are shown in Figures 4d–f, and finally of β₀ precipitate are shown in Figures 4g–i. The color scale of strain field indicates strain changes of −0.5% to 0.5% for all phases. In this way, colors for positive values represent tensile strains, while colors for negative values represent the compressive strains. Using this scheme, Figure 4 reveals that the strain in Al matrix both near the thin and thick T₁ platelet precipitates is tensile in [00-1]Al direction and while compressive in [−110]Al direction. In general, the presence of strain-dipoles in the platelets indicates relaxation of strain locally via the creation of edge dislocations, and hence imply that the strain fields stem from the presence of the dislocation at the boundary interface between the T₁ platelet precipitate and the Al-matrix. It is also a known fact that the dislocations usually result from the interference of slip planes with the particles precipitating on crystallographic planes (Balluffi et al., 2005).
However, the Figure 4b,e reveals that the thin and thick $T_1$ platelet precipitates predominantly dislocation free and still strained with tensile strain along [00-1]$_{Al}$ direction. The tensile strain along the $T_1$ precipitate can be interpreted due to the aggregation of solute atoms in the crystal lattice, which enhances the hydrostatic stress in the adjacent region containing these solute atoms. For the case of strain along $\gamma$-axis in Figure 4c,f, it has variable values and thus can be averaging to zero. Regarding the strain in Al matrix around the $\beta'$ phase in both [00-1]$_{Al}$ and [-110]$_{Al}$ directions in Figure 4h,i. It is of uniform nature and therefore implies the presence of no dislocations inside the phase. Such absence of dislocations is expected for spherical precipitates due to a complete or coherent matching between the atomic layers of both matrix and precipitate. The $\beta'$ phase with $L1_2$ structure usually appears to be fully coherent with the Al-matrix in perpendicular direction because its lattice parameter ($a = 0.400$ nm) is close to that of Al-matrix ($a = 0.405$ nm) and therefore leads to formation of coherent interfaces (Strake, 1983). Even though, the $\beta'$ phase usually acts to control the grain size and the resistance to recrystallization (Strake, 1983), it also has a clear role in the strengthening mechanism of our respective alloy. That's mainly because of the role of the $\beta'$ phase on providing heterogeneous nucleation sites for the $T_1$ phase (Lee, Lee, & Hiraga, 1998). As presented earlier that the strain

**FIGURE 4** Strain analysis of representative precipitates obtained by applying GPA to their corresponding HR-STEM images. (a–c) Strain analysis of the thin $T_1$ platelet precipitate. (d–f) Strain analysis of the thick $T_1$ platelet precipitate. (g–i) Strain analysis of the $\beta'$ spherical precipitate [Color figure can be viewed at wileyonlinelibrary.com]
fields in Al matrix in the vicinity of \( T_1 \) precipitate are, in some cases, due to presence of dislocations-dipoles in Figure 3a. A clearer illustration on the presence of dislocations there is further elucidated by applying the Fourier filtering technique to the image in Figure 3a and the results are shown in Figure 5. The image in Figure 5a is obtained by filtering [111] planes from a HR-STEM image. In addition, a couple of regions marked by continuous and dashed squares are in Figure 5b and Figure 5c, respectively, in which the edge dislocations in the core of \( T_1 \) platelet precipitate are marked by circles therein.

The next step involved the mapping of \( Y_m \) for thin and thick \( T_1 \) platelet precipitates as well as for \( \beta \) spherical precipitates in order to complete the task of evaluating the local stresses around these observed nano-precipitates below the yield strength of Al-matrix. The determination of specimen thickness around nano-features is required for determining a more accurate value of local stresses. However, the absolute thickness is impossible to determine using conventional optical methods. Therefore, we utilized low-loss STEM-EELS spectrum imaging to determine the absolute foil thickness (Iakoubovskii et al., 2008). After the determination of absolute foil thickness, \( E_p \) can be extracted from the acquired STEM-EELS datasets as described in the experimental section. This can be utilized to deduce the values of \( Y_m \) of Al matrix around any respective precipitates, for example, \( T_1 \), \( \beta \), and even Al matrix alone. These precipitates are metastable in the T8 treated alloy and cannot be fabricated in bulk form due to their nanometer size. Hence, it is reasonable to extract \( E_p \) values of Al matrix in their vicinities. The DF-STEM image of the \( T_1 \) platelets and \( \beta \) phase in which an EELS was obtained is shown in Figure 6a. An example of the represented EELS spectrum is given in Figure 6b. It also contains an insert that is showing overlaid first plasmons peaks of Al from the matrix alone, the matrix containing a \( T_2 \), and the matrix containing \( \beta \) precipitates. A blue shift of about 0.20 eV and 0.27 in the first Al plasmons peaks were observed for the matrix regions containing \( T_1 \) and \( \beta \) precipitates as compared to the matrix region with no precipitates. The acquired VEELS spectra were also utilized to determine absolute thickness map of Al matrix containing \( T_1 \) platelets and \( \beta \) precipitates and the obtained results are shown in Figure 6c. It can be observed from the results presented in Figure 6c that even though the foil-thickness was in the range of 70 to 100 nm across the entire field-of-view, the thickness around \( T_1 \) and \( \beta \) precipitates was also found out to be close to Al-matrix. By performing analysis of the EELS spectra for specific positions of Al-matrix, it was found that the absolute thickness map of Al matrix contains \( 65.1 \pm 1.1 \) GPa, \( 70.41 \pm 1.1 \) GPa for thin \( T_1 \) platelets, respectively. These determined values of \( Y_m \) agreed with the typical literature values (Oleshko et al., 2002). In the same way, the variation in the values of \( Y_m \) for thin and thick \( T_1 \) platelets, as well as for \( \beta \) precipitates, are slightly lower for both cases of thin and thick \( T_1 \) platelets. In fact, these values found out to be \( E_p = 14.8 \pm 0.1 \) eV and \( E_p = 14.6 \pm 0.1 \) eV for thin and thick \( T_1 \) platelet, respectively. While around the regions of \( \beta \) precipitates, the average value of \( E_p \) for the Al matrix was 15.06 ± 0.1 eV. As mentioned earlier, an \( E_p \) map was generated from acquired low-loss STEM-EELS datasets by applying NLLS routine. Afterwards, a procedure of generating the \( Y_m \) map as per Equation (3) was applied to realize the corresponding \( E_p \) map and the results are shown in Figure 6d. The color scale in the image indicates the variation in the \( Y_m \) values from 65 to 71 GPa. A plasmon peak at 14.95 ± 0.1 eV for Al matrix corresponds to an estimated value of 69.13 ± 1.1 GPa for \( Y_m \). It is important to remember that this value was calculated using Equation (3). In this way, using the Plasmon energies of 14.8 ± 0.1 eV, 14.6 ± 0.1 eV, and 15.06 ± 0.1 eV, the corresponding values of \( Y_m \) turned out to be 67.41±1.1 GPa, 65.1±1.1 GPa, and 70.41±1.1 GPa for thin \( T_1 \) platelet, respectively. These determined values of \( Y_m \) agreed with the \( Y_m \) color map presented in Figure 6d. In short, the presented results confirm the value of ~69 GPa represents \( Y_m \) of Al matrix alone, and while the value of ~70.5 GPa represents \( Y_m \) of Al matrix in the vicinity of \( \beta \).
precipitate. Whereas in the vicinity of $T_1$ precipitates, the $Y_m$ color map of Al matrix in Figure 6d gives a range of $Y_m$ values from ~65 GPa and ~67 GPa. A careful analysis leads to the conclusion that these $Y_m$ values of Al matrix stem from thick and thin $T_1$ precipitates, respectively. This finding is contrary to expected values and therefore can be interpreted as a higher $Y_m$ value occurs in Al matrix when $T_1$ platelets interfere with $\beta'$ precipitates. These $\beta'$ precipitates usually act as preferred nucleation sites for the $T_1$ precipitates (Itoh, Cui, & Kanno, 1996). Thus, it can be concluded from the result presented in Figure 6 that every precipitate does not necessarily results in the enhancement of $Y_m$ of Al matrix. By following the methodology described in Figure 6, we generated $Y_m$ map from a larger area of the alloy that contained both thin $T_1$, thick $T_1$, and $\beta'$ precipitates. The obtained results are shown in Figure 7 that gives an estimate on the statistical significance of determined $Y_m$ maps. The $Y_m$ results presented in Table 1 are compiled by taking the average of about 5 precipitates of each type.

In the end, to examine the strain distributions in Al-matrix quantitatively within vicinity of the observed precipitates and to assess the reliability of GPA, we performed the analysis of strain maps shown in Figure 4. The analysis included figuring out average strain in Al matrix along the [00-1]Al direction at the regions of interest, that is, the thin and thick $T_1$ platelets (Figure 4b,e). The outcomes turn out to be that the strains in Al matrix near those precipitates was in the range of 0.3%. In the same manner, the average strain in Al matrix around the $\beta'$ precipitates along [00-1]$_{\text{Al}}$ and [−1 1 0]$_{\text{Al}}$ directions was estimated to be 0.2% (Figure 4h,i). For the Al matrix alone, the strain along the [00-1]$_{\text{Al}}$ is also in the range of 0.2% which was the same of Al matrix near the $\beta'$ precipitates. Our experimental measurements of the strain fields around the nano-precipitates in the microstructure of Al based alloys are in a good agreement with literature measured values for the similar nano-features (Bai et al., 2012; Gammer et al., 2016). In this context, one should note that the strain in Al matrix both near the thin and thick $T_1$ platelet precipitates is tensile in [00−1]$_{\text{Al}}$ direction (positive values for $\varepsilon_{xx}$ components) and while compressive in [−110]$_{\text{Al}}$ direction (negative values for $\varepsilon_{yy}$ components). In the similar case of obtaining positive values for $\varepsilon_{xx}$ components and obtaining negative values for $\varepsilon_{yy}$ components, the off-diagonal elements ($\varepsilon_{xy}$, $\varepsilon_{yx}$), usually
represents small distortions that correspond to a shear, and are on average zero (Hyttch et al., 1998). Therefore, local stresses in Al matrix caused by each nano-precipitate feature can be estimated using the linear theory of elasticity ($\sigma = Y_m \varepsilon$) in a similar way done by Gammer et al. (2016). This was possible to do now since the $Y_m$ is known experimentally from Figure 6. In this way, the determined values of local stress in Al matrix around thin, thick $T_1$ platelets, Al matrix alone, and $\beta'$ precipitates were calculated, and the obtained results are shown in Table 1. It is important to note herein that the $Y_m$ determined in Figure 6 is a scalar quantity since the Plasmon energy is independent of crystal orientation. Therefore, the determined local strain and $Y_m$ values of Al-matrix can be directly compared with their bulk counterparts. Our finding of the local stress at the thin $T_1$ precipitates (Table 1) demonstrate the ability of these nano-features to impede the dislocation motions because the local stress of values higher than 200 MPa are needed in Al alloys to resist the motion of dislocations. Therefore, strengthening of Al alloy presented in this study is expected. In general, the strengthening mechanism of such alloying system is based on the interactions between the moving dislocation and the existing nano precipitates. Indeed, introducing nano-size precipitates in the Al matrix with a specific distribution can lead to a huge improvement in the mechanical properties of such alloys. For instance, it has been shown that microstructure of AA2195 alloys can be tailored in such a way that the alloys has a very specific distribution of precipitates that is, as small thickness of precipitates as 1–2 nm with intermediate spacing of ~9 nm. This approach leads to an optimal situation that corresponds to a high hardness value of 209.3 HV and hence a high level of strengthening is expected as well (Khushaim, 2019). Moreover, a recent study of AA2195 at T8 temper demonstrates a high tensile stress value of 574 MPa; and a high yield strength value of 540 MPa for these alloys (Kim et al., 2016). In our study, the contribution of every observed precipitate to the local stress of Al matrix was estimated as shown in Table 1. Thus, the presented study demonstrates the ability of AC-STEM to measure the local stresses in Al matrix near each respective precipitate. This capability of AC-STEM to allow measuring local stress experimentally at the nanometer scales in Al alloys can be extended to other metallic alloys and hence will also allow determining their mechanical properties (Figure 7).

### Table 1

| Name of the phase   | $\varepsilon_{xx}$ (%) | $\varepsilon_{yy}$ (%) | Young's modulus ($Y_m$/GPa) | Local stress/MPa $\sigma_{xx}$ | Local stress/MPa $\sigma_{yy}$ |
|---------------------|------------------------|------------------------|----------------------------|-------------------------------|-------------------------------|
| Thin $T_1$ platelet | 0.16                   | -0.16                  | 67.41 ± 1.1               | 107 ± 2                       | -107 ± 2                     |
| Thick $T_1$ platelet| 0.13                   | -0.1                   | 65.15 ± 1.1               | 84.69 ± 2                    | -65.15 ± 1                   |
| The $\beta'$ phase  | 0.2                    | 0.05                   | 70.41 ± 1.1               | 140.82 ± 3                   | 35.2 ± 0.5                   |
| Al Matrix           | 0.02                   | 0.01                   | 69.13 ± 1.1               | 13.8 ± 0.2                   | 6.9 ± 0.1                    |

**Figure 7** The mapping of $Y_m$ in the microstructure area showing thin $T_1$, thick $T_1$, and $\beta'$ precipitates. (a) HAADF-STEM image, (b) $Y_m$ map [Color figure can be viewed at wileyonlinelibrary.com]

### 4 | Conclusions

Using AC-STEM, the strain fields and $Y_m$ of the hardening precipitates in Al alloy AA2195 at T8 temper were experimentally mapped at nanometer scale spatial resolutions. GPA is a powerful way of mapping strain fields in Al matrix in the vicinity of the $T_1$ and $\beta'$ precipitates. It was found that the average strain around these precipitates along the [00-1]$_{Al}$ and [−1 1 0]$_{Al}$ directions was estimated in the range of 0.2–0.3%. The mapping of $Y_m$ in Al matrix around each type of hardening precipitates within the alloy microstructure was done by utilizing its correlation $E_p$. The values of $Y_m$ were calculated as 67.41 ± 1.1 GPa, 65.1 ± 1.1 GPa, 70.41±1.1 GPa and 69.13 ±1.1 GPa for Al matrix near thin $T_1$ platelet precipitate, thick $T_1$ platelet precipitate, $\beta'$ spherical precipitate, and Al matrix, respectively. The determined values or maps of $E_p$ and $Y_m$ yield in estimating the local stress in Al matrix around the observed precipitates with resolution on order of a few MPa. The presented results confirm a higher stress value of 202 ± 3 MPa in Al matrix near the regions of the thin $T_1$ platelet precipitate. While the stress were in the range of 195 ± 3 MPa the vicinity of the thick $T_1$ platelets.
The local stresses in Al matrix near to the $\beta'$ spherical precipitate and in Al matrix with no observed precipitates were about $140 \pm 3$ MPa and $138 \pm 3$ MPa, respectively. In this way, it can be concluded that the thin $\beta_1$ precipitates result in alloy hardening more than other precipitate features. Overall, the presented results demonstrate the capability of AC-STEM to enable determining the experimental measurements of the local stress associated with each precipitate features in metallic alloys.

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DATA AVAILABILITY STATEMENT

The data that support the findings of this study are available from the corresponding author upon reasonable request.

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