

Making Stress Visible:

A method to quantify local stress distribution in metals using digital image correlation method

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Abstract

Stress (force) induces the deformation of materials. Stress distribution in materials is generally heterogeneous depending on their shapes and microstructures, which governs their global deformation and failure, and sometimes stress concentration induces serious incidents in society. However, it has been challenging to experimentally quantify stress distributions in materials in macro- and micro-scales by now. Here, we newly propose and establish a simple but promising method to quantify local stress distributions in materials by applying the digital image correlation (DIC) technique, which is applicable to all length scales. We expect that the newly developed method would become the first step to give an impact on fundamentally understanding the deformation and fracture of materials in various scales as well as on practically reducing risks in structural components.

Key words: Local stress distributions; Elastic strain; Digital image correlation (DIC); Structural materials; Stress visualization.

1 Materials deform when a force (stress) is applied externally. The way of deformation, i.e.,
2 the response against an applied force, is different depending on the nature of materials.
3 Mechanical properties (deformation responses) of materials are evaluated by various kinds of
4 testing methods. The uniaxial tensile test is the most popular method, and the obtained results
5 are expressed as stress-strain curves. Ideally, the tensile tests are designed to have uniform
6 distributions of tensile stress and strain within the gage part, but the occurrence of macroscopic
7 necking introduces non-uniform deformation. Moreover, at micro-scales, distributions of
8 stress and strain are inherently non-uniform even before necking occurs, attributed to the
9 presence of microstructures. In metallic materials, for example, most bulky metals are
10 polycrystalline materials composed of a number of variously shaped grains with different
11 crystallographic orientations and dislocation-slip directions. Such complicated features in
12 microstructures contribute to heterogeneous deformation. Furthermore, recently developed
13 advanced high-strength materials tend to exhibit more and more complex microstructures,
14 comprising more than two phases with different mechanical properties. Such complexity leads
15 to more heterogeneous deformation, thereby emphasizing the necessity for quantitatively
16 characterizing distributions of stress and strain.
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18 As a strain-visualizing tool, the digital image correlation (DIC) technique [1-6] that can
19 provide local strain distributions has been widely used in the study of local deformation
20 behavior. On the other hand, it is still challenging to experimentally quantify local stress
21 distribution, which has still been assessed by computer simulations such as the finite element
22 method (FEM) [7, 8] or crystal plasticity FEM (CPFEM) [9-13]. Although the FEM
23 simulations offer relatively reliable stress outputs at a macroscopic scale for bulky materials,
24 microscopic analysis performed by CPFEM still has some difficulties due to the following
25 obstacles: (i) Uncertainties related to fitting parameters lacking clear physical interpretations
26 and (ii) complexity of the calibration process requiring precise materials parameters, including
27 crystal orientation, grain size and grain boundary properties. For this reason, the exploration
28 through experimental approaches has also been conducted. At macroscopic scales, Xu *et al.*
29 [14-16] developed a technique to visualize stress distribution in materials using
30 mechanoluminescent ceramic powders applied on the surface of bulky materials, although the
31 method has limitations, particularly in quantitatively evaluating values of each stress
32 component. At microstructural scales, the EBSD-Wilkinson method has been spotlighted as
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an advanced stress-visualizing technique, which can measure elastic strain/stress through changes in the Kikuchi pattern [17-20]. However, the accuracy of stress measurements strongly depends on the quality of the Kikuchi pattern, which poses challenges for its application in heavily deformed materials (e.g., cold-rolled materials) having a degraded quality of Kikuchi patterns. An analytical problem where the stress value sensitively varies with the position of a reference point is another critical issue. In this study, we introduce a new and straightforward method for visualizing local stress distribution at both macro- and micro-scales in metallic materials. In this approach, two-dimensional local stress components are mapped from the measurement of local elastic strain during loading-unloading processes, utilizing the digital image correlation (DIC) method. Limitations to be improved in the method are also addressed in the later part of this paper.

Figure 1 (a) shows a schematic illustration of the global engineering stress-strain (S-S) curve and the corresponding true S-S curve obtained from a uniaxial tensile test. Until uniform elongation, the gage part of the tensile specimen exhibits macroscopically homogeneous deformation, during which the true stress (σ) is estimated as $s(1+e)$, where s and e denote engineering stress and strain, respectively. After the occurrence of necking, deformation becomes heterogeneous macroscopically, which makes it difficult to estimate localized stress value in the necked region.

For estimating heterogeneous stress distributions, the current study focuses on the elastic strain (ϵ^E) because the stress is a function of elastic strain according to elastic theories like Hooke's law ($\sigma = C\epsilon^E$, where C represents elastic modulus) [21]. **Figure 1** (b) shows schematical illustrations of expected S-S curves in the non-necked region (Region A) and necked region (Region B) of the tensile specimen after macroscopic necking. Although the exact local stress values in both cases remain uncertain, the necked region would experience higher stress than the non-necked region. Here, we consider the elastoplastic history of S-S curves during a loading/unloading process. The specimen initially experiences elastic deformation until reaching a yield point. After yielding, plastic deformation occurs, involving both plastic and elastic strain components ($\epsilon_{ij}^P + \epsilon_{ij}^E$ in stages ① to ②). When the specimen is unloaded at stage ②, the stress would decrease linearly along the elastic modulus line (③), retaining only plastic strain (ϵ_{ij}^P). This suggests that the strain between the stages ② and ③

corresponds to the elastic strain component at ②. Therefore, unknown local stresses in the non-necked region and the necked region can be obtained by measuring strains between ② and ③ ($\sigma_{ij} = C_{ijkl} \epsilon_{kl}^E$). (In reality, however, the situation might not be so simple, which will be argued in the later part of this paper.) The DIC technique was used as a strain-visualizing tool, and obtained results are compared with finite element method (FEM) simulation. It is important to note that while we explain the concept of stress measurement using the example of simple uniaxial tensile testing, this methodology is also applicable to further complicated-shaped materials and microstructural heterogeneities.

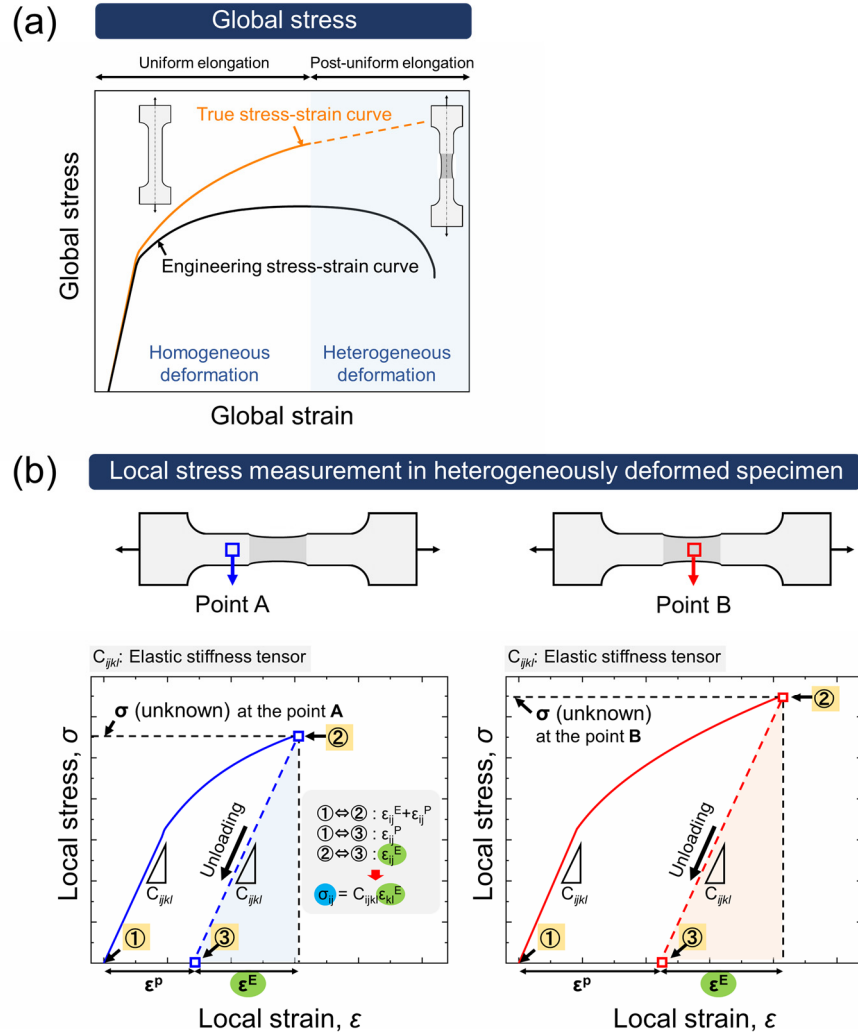


Figure 1 Schematic concept of local stress visualization. (a) Sketch showing engineering stress-strain (S-S) curve and true S-S curve obtained from uniaxial tensile testing. (b) Illustration of local stress measurement in a heterogeneously deformed specimen after necking. Unknown local stress values at the necked region and non-necked region can be estimated by measuring elastic strain during the unloading process.

A fully martensitic-structured low-carbon steel (Fe-8Ni-0.1C: mass%) was used for the macroscopic local stress measurement by the method proposed above. The yield strength and tensile strength were 928 MPa and 1287 MPa, respectively, and elastic modulus of the specimen was measured to be 192 GPa. The present study uses a single-notched specimen, capable of introducing complex stress conditions via uniaxial tensile loading. **Figure 2** (a) shows an image of the single-notched specimen having a gage length of 20 mm, width of 10 mm and thickness of 1 mm, involving one-side half-circle notch with a circle radius of 2 mm. Random-dot pattern was introduced on the surface of the specimen for the precise DIC-strain analysis. **Figure 2** (b) exhibits the load-engineering strain curve of the specimen during loading/unloading. The specimen was subjected to tensile loading to stage ②, followed by unloading (③). The second loading was then performed to verify a linear increase in stress (i.e., to confirm elastic behavior) in the stages between ② and ③. The DIC-strain distributions at stages ② ($\epsilon^P + \epsilon^E$) and ③ (ϵ^P) are shown in **Fig. 2** (c) and (d), respectively. The colors of DIC-strain maps represent the magnitude of the normal strain component (ϵ_{11}) in the tensile direction, as indicated by the key strain color bar. In reality, it is noteworthy that strain at stage ③ (**Fig. 2** (d)) is not a perfect plastic strain component, but involves a small amount of elastic strain (residual stress). Here, however, for the first attempt to visualize stress distributions using the method, the stage ③ is assumed to be a residual stress-free state. A strain difference between stages ② and ③, deemed to elastic strain component (ϵ^E), is depicted in **Fig. 2** (e). The elastic strain seems to exhibit a notable change around the notch.

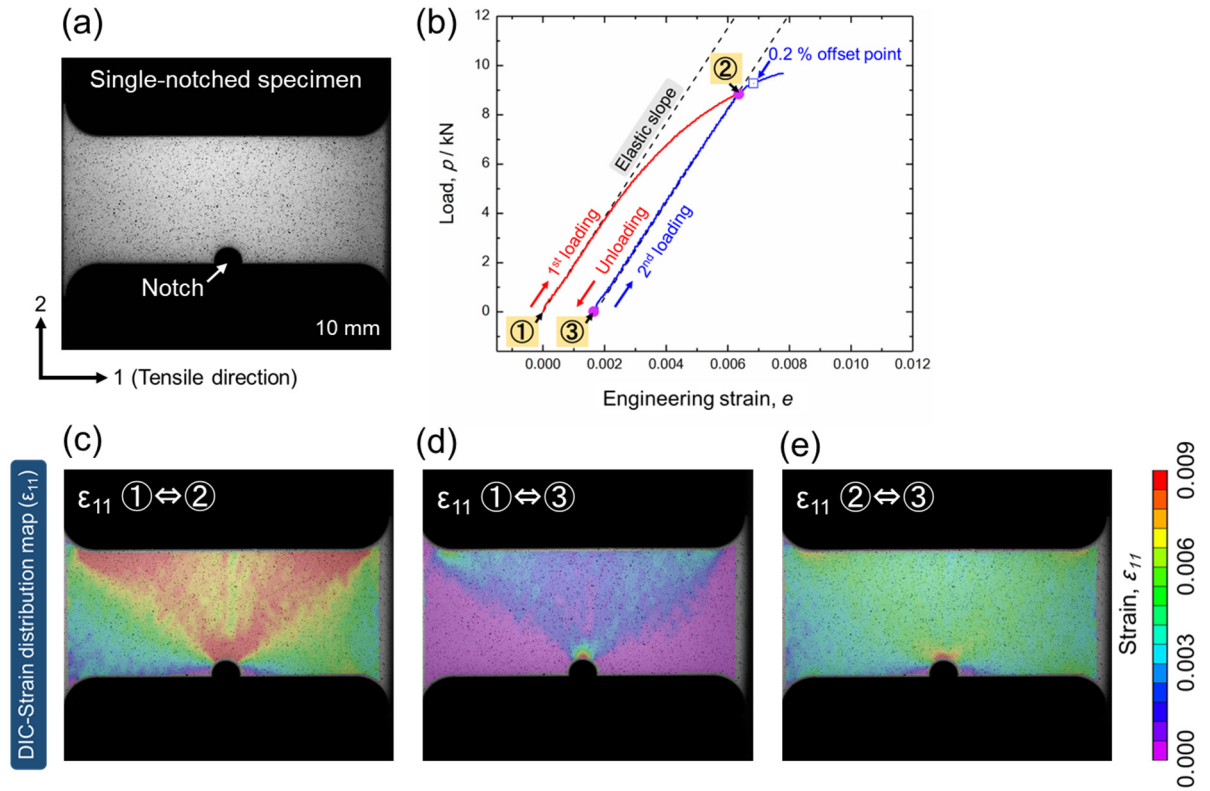


Figure 2 DIC analysis for obtaining local strain distribution in a single-notched tensile specimen. (a) Preparation of the single-notched specimen with random-dot patterns on the surface as markers for the DIC analysis. (b) Load-strain curves of the notched specimen during loading/unloading process. Points ①, ② and ③ on the graph correspond to deformation states before deformation, under loading and after unloading, respectively. (c-d) DIC-local strain distribution maps for ϵ_{11} component in the strain range of (c) ① $\leftrightarrow$ ② and in (d) ① $\leftrightarrow$ ③. (e) Local strain distribution in the strain range of ② $\leftrightarrow$ ③.

Stress distribution mapping was attempted with this elastic strain distribution result. Stress calculation was conducted under a plane-stress condition ($\sigma_{33} = \sigma_{13} = \sigma_{23} = 0$) because the DIC-analyzed region was limited to the surface of the specimen. **Figure 3** (a) shows stress distributions for σ_{11} , σ_{22} and σ_{12} , obtained by our present approach. σ_{11} and σ_{22} represent the normal stress components in tensile and transverse directions, respectively, and σ_{12} denotes shear stress. Positive (+) and negative (-) values in the normal stress (σ_{ii}) indicate tensile and compressive stresses, respectively. **Figure 3** (b) displays stress distributions obtained by FEM simulation. The simulation employed the compulsory-displacement boundary condition for a model having an elastic modulus of 192 GPa and a Poisson's ratio of 0.3. In addition, a true stress-strain curve obtained from a real tensile test was used as an input to simulate plastically

deformable nature of the specimen. It was found that the experimentally obtained stress distribution (**Fig. 3 (a)**) coincided well with the results obtained by FEM simulation (**Fig. 3 (b)**). In the σ_{11} stress map, significant stress concentrations around the notch root are observed as expected. The σ_{22} stress map indicates that most areas show compressive (negative) stresses, but, interestingly, the tensile (positive) stress concentration is recognized at the notch root. In the σ_{12} shear stress map, stress distributions with opposite signs are observed on the right and left sides of the notch, achieving a static balance in the whole specimen. A quantitative comparison of the local stress distributions was conducted by investigating the stress line profile (**Fig. 3 (c)**). Different stress components (σ_{11} , σ_{22} , σ_{12} and σ_H (hydrostatic stress: $\Sigma\sigma_{ii}/3$)) were investigated along Line 1 and Line 2. Local stress values obtained through the present method exhibit quantitatively good agreements with those obtained by the FEM calculation. While almost stress values match well overall, the σ_{22} value on Line 1 is somewhat lower than that obtained from the FEM simulation, particularly in the distance range of 1 mm to 3 mm from the notch root, which will be mentioned in the discussion part later.

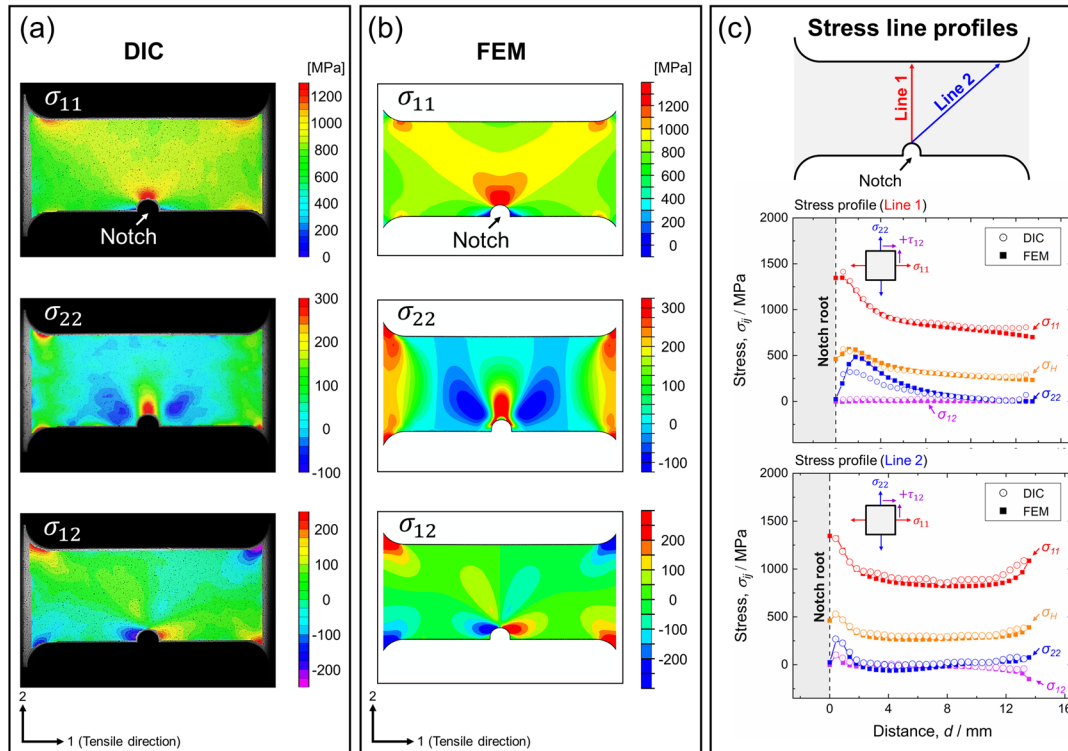


Figure 3 Local stress distributions of different stress components (σ_{11} , σ_{22} and σ_{12}) obtained from (a) our proposed approach using DIC technique and (b) FEM simulation. (c) Stress profiles of different stress components (σ_{11} , σ_{22} , σ_{12} and σ_H (hydrostatic stress)) along Line 1 and Line 2, both initiating from the notch root.

The feasibility of the present method was also assessed by examining the energy balance between the internal energy (total elastic strain energy) of the specimen and the external work performed by the tensile machine. Total elastic strain energy was calculated by the equation, $E^{in} = \int 1/2 \times \sigma_{ij} \epsilon_{ij}^E dv$ ($= \int 1/2 \times C_{ijkl} \epsilon_{kl}^E \epsilon_{ij}^E dv$), where v represented the gage volume of the specimen. In gage volume calculation of total strain energy, stresses/elastic strains were assumed to have the same value along the thickness direction of the specimen. External work was calculated as $W^{out} = f \times u_1$, where f and u_1 denoted load measured at a loadcell and stroke displacement, respectively. **Figure 4** (a) shows the elastic strain energy density (MJ/m^3) map at a load of 8.84 kN. The elastic strain energy is predominantly concentrated around the notch. **Figure 4** (b) shows the plots of total elastic strain energy and external work as a function of engineering strain during the second loading. The total elastic strain energy increased with deformation, and remarkably, its value closely matched that of the external work ($E^{in} \approx W^{out}$). Based on the results presented in **Fig. 3** and **4**, our current approach demonstrates considerable potential as an effective technique for visualizing stress distributions in bulky materials.

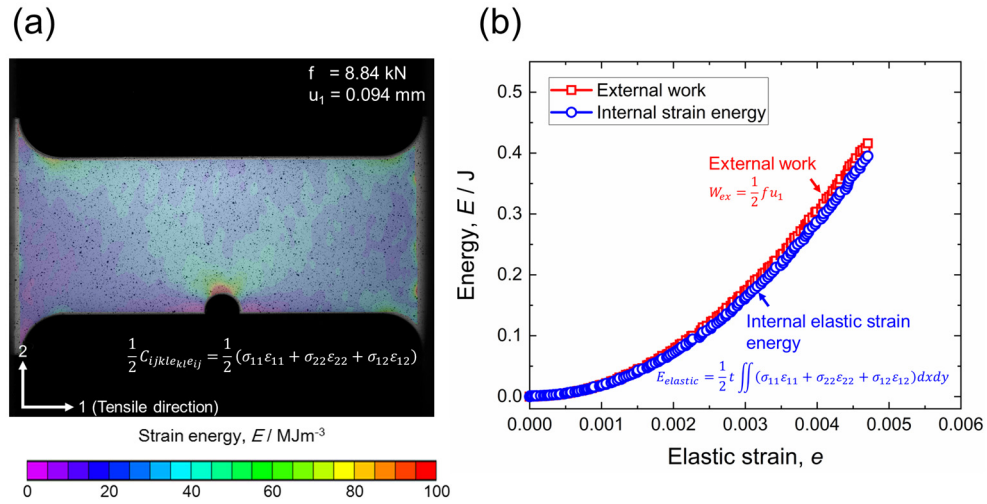


Figure 4 (a) Elastic strain energy density map of the single-notched tensile specimen at a load of 8.84 kN and stroke-displacement of 0.094 mm. (b) Plots of the external work achieved by tensile machine (red) and the internal elastic strain energy in the specimen (blue), as a function of the global tensile elastic strain.

We have tried stress visualization also at a microstructural scale. A commercial AZ31 Mg alloy (Mg-3.1Al-1.1Mg-0.3Mn: mass%) with a hexagonal close-packed (HCP) crystal structure was selected for microstructural stress visualization, because of its low elastic modulus (~ 45 GPa) and anisotropic slip behavior limited to basal slip that could provide large elastic strain and large plastic anisotropy [22-24] leading to stress/strain localizations at grain boundaries. The feature of single-phase would have merit for the current analysis in terms of minimizing residual stress. **Figure 5** (a) shows the EBSD-orientation color map, elastic stiffness (C_{1111}) map and Schmid factor map for (0001) $\langle 1-210 \rangle$ basal slip, all observed from the normal direction (ND) of the hot-rolled specimen. The specimen exhibited strong (0001) basal texture on the ND plane. No significant variation in elastic stiffness was recognized. While most grains had low Schmid factors for (0001) basal slip, some grains had high Schmid factors for the basal slip. Grains A and B, indicated on the orientation color map, will be referenced in the stress line profiles shown later. **Figure 5** (b) illustrates distributions of local stresses (σ_{11} , σ_{22} , σ_{12} and σ_{eq} (von-Mises equivalent stress)) and elastic strain energy density (E) at an engineering strain of 1.24%. The local stresses are computed by employing individual elastic stiffness components (C_{ijkl}) on each grain with corresponding elastic strains. Local stress exhibited a variation depending on grains, and stress localization tended to occur at grain boundaries. The elastic strain energy was localized in specific grains with high Schmid factors, suggesting that slip deformation influenced elastic strain energy, although the underlying reason for this phenomenon remained unclear. Additionally, stresses and elastic strain energy density were quantified through line profiles between Grains A and B (**Fig. 5** (c)). The black straight line indicates the position of the grain boundary (GB) between Grains A and B. The value of stress σ_{11} tended to change significantly in the vicinity of GB, where σ_{22} reached its lowest value. Moreover, the equivalent stress σ_{eq} and elastic strain energy E reached their peak value at GB. Although our analysis is limited to a highly localized area bridging only two grains, the results suggest that the grain boundary acts as a distinct site exhibiting stress transitions or critical stress values.

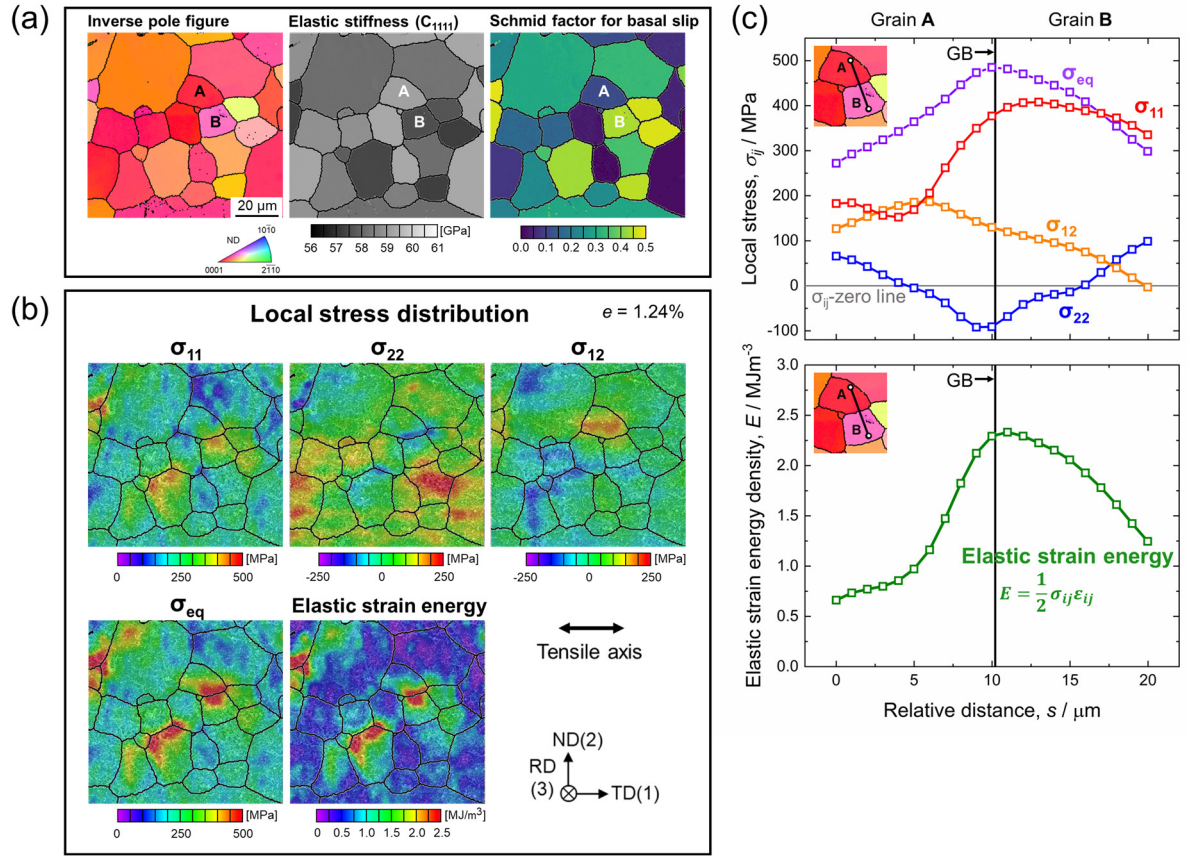


Figure 5 Stress distribution in the microstructure of AZ31 Mg alloy. (a) The EBSD-orientation color map, elastic stiffness (C_{1111}) map and Schmid factor map for (0001) basal slip with respect to the horizontal tensile axis. (b) Local stress distribution (σ_{11} , σ_{22} , σ_{12} and σ_{eq} (von-Mises equivalent stress)) and elastic strain energy (E) distribution at a tensile strain of 1.24% in AZ31 Mg alloy. (c) Stress and energy line profiles in the local region across Grain A and Grain B shown in (a). The black vertical line on the graph is the position of the grain boundary (GB) between Grains A and B.

Our initial attempt to measure local stress using the DIC technique offers notable advantages in terms of easiness in measurements and broad scalability suitable for both macroscale and microscale applications. Furthermore, this approach requires only one material parameter, i.e., elastic modulus (C_{ijkl}), as the preliminary input data, eliminating the need for additional information like tensile curves. However, it should be noted that the current approach also has limitations. Firstly, this method cannot take into account residual stress at the unloaded stage (Stage ③ in Fig. 1 or Fig. 2). In reality, materials exhibit residual stress more or less due to deformation inhomogeneity, caused by several factors, such as metalworking history and microstructural features like texture and phase characteristics. The

1 difference in the stress values, particularly in σ_{22} , observed between our approach and FEM
2 simulation (**Fig. 4 (d)**) is considered attributed to the absence of considering residual stress in
3 the present approach. While the notched specimen used in the current study shows overall
4 good agreement, it may be unsuitable for the case with extremely high residual stresses.
5 Combining other residual stress measurement techniques can be one option to overcome such
6 a limitation. At the microstructural level, the FIB (focused ion beam) ring-core milling
7 technique with the DIC method [25, 26], capable of measuring relaxation strain by locally
8 cutting around the region of interest using FIB, can be combined with the current approach.
9 Another solution is to combine the EBSD-Wilkinson method. The improvement through such
10 combinations will be confirmed in forthcoming studies. Secondly, in real metallic materials,
11 load-release during unloading does not perfectly follow a linear path as ideally illustrated by
12 unloading dashed lines in **Fig. 1 (b)**, but exhibits a slight deviation from the straight line at
13 lower stress levels. Deviations become significant in composite materials [27, 28] or materials
14 accompanying twinning [29], so it would be necessary to avoid such materials in the use of the
15 present method so far. This is another issue to be addressed in future using correcting
16 techniques like the interpolation method.

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18 The current approach newly proposed by our group for visualizing stress distributions
19 holds value as an initial estimation tool that can be easily applied without any constraints
20 regarding length scale. This would be positioned as a simple stress-visualizing technique, also
21 serving as a foundational technique for further practical analysis, and it can subsequently be
22 refined through the incorporation of residual stress measurement techniques to improve
23 accuracy and deepen understanding of materials deformations.
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Author contributions

M-H.P. carried out most of the experiments and analysis and M-H.P., A.S. and N.T. discussed the research. S.Y. performed microstructural stress visualization. The alloys were processed by Y.T. M-H.P. and N.T. wrote the paper. All authors discussed the results, and checked and revised the manuscript.

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