

Investigation of the residual strain and deformation mechanisms in laser-welded Eurofer97 steel for fusion reactors

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ABSTRACT

A fusion power plant requires not only the control of high energy plasma but also advanced techniques for maintenance and assembly to generate electricity consistently and safely. Laser welding is a promising technique for cutting and joining pipes and in-vessel components made of Eurofer97, a European baseline structural material. However, the substantial residual strain induced during post-weld cooling degrades the mechanical properties and reduces the lifespan of engineering components. Establishing the underpinning mechanistic connection between residual strain, microstructure, and tensile behaviour is critical to lifetime assessments of engineering components. Here, the heterogeneous strain evolution in laser-welded Eurofer97 joint is quantitatively evaluated using *in situ* neutron diffraction at the lattice-scale, nanoindentation at the microscale, and digital image correlation (DIC) at a macroscale. The residual strain in the loading direction is characterised via neutron diffraction and validated using a plasma-focused ion beam (PFIB-DIC) ring-core method. Superimposing the microstructural strengthening, the highest residual tensile strain ($0.6 \times 10^{-3} \mu\epsilon$) accelerates the accumulation of tensile deformation around the fusion line (FZ/HAZ interface), whereas residual compressive strain ($-1 \times 10^{-3} \mu\epsilon$) hinders the tensile strain evolution around the heat-affected zone and the base material interfaces, increasing the localised yield strength to 506 MPa. Residual strain is the primary strengthening mechanism during the initial deformation stage, although the microstructural strengthening then dominates as deformation increases. This work reveals the critical role of residual strain, and the results provide insight into managing structural integrity and developing predictive tools for lifetime assessment.

1. Introduction

Residual strain is one of the vital assessment factors of structural integrity. In this context it is the locked-in strain induced by welding processes after the removal of all external loads. It is well-known that residual strain associated with complex microstructures notably affects the mechanical properties [1–7]. Residual strain is highly detrimental, as it can be responsible for various engineering problems such as distortion, stress corrosion cracking, fatigue, and degradation of tensile strength [8]. To address the deleterious residual strain related effects, and ensure safety and longevity, it is crucial to establish a fundamental understanding of the deformation mechanism associated with residual

strain. This is especially challenging in complex materials systems, such as those in the nuclear power industry.

Nuclear fusion is a potential source of green power that could end humanity's reliance on fossil fuels and the well-known environmentally damaging carbon emissions that come with them. In the quest for this new source of energy, a series of fusion tokamaks have been or are being investigated, from the Joint European Torus (JET) and the International Thermonuclear Experimental Reactor (ITER), to the DEMOnstration power plant (DEMO) [9]. To meet the demands of low radiation activities, high-temperature mechanical performance, and appropriate thermal conductivity, reduced-activation ferritic/martensitic (RAFM) steels were developed. They are based on Grade 91 steel with Mo and Nb

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replacing lower activation elements like W and Ta [4,10–12]. One such RAFM is Eurofer97 which is currently the European baseline structural material for the pipe and in-vessel components (e.g., breeding blanket and divertor cassette) for the EU-DEMO [13–15]. The assembly and maintenance of these in-vessel components presents difficulties because of the internal grid structures [5]. Rapid reliable welding and joining techniques are required to maintain the structural integrity of these breeding blanket joints.

Laser welding is considered one of the most promising techniques because its equipment is compact and portable, and it enables full penetration in single-pass welds in thick sections. The low heat input of the laser welding process leads to a narrow weld and heat-affected zone. Combining laser welding with a robotic device could enable cutting and welding operations in complex working conditions [3]. Previous studies have demonstrated the feasibility of remote laser tools to manufacture and maintain the in-vessel components in fusion plants by laser welding [3,6,16]. However, the non-uniform thermal expansion and phase transformations give rise to residual strain and local distortion [17–19]. Such residual strain can negatively affect mechanical properties, such as yield strength, fracture toughness and ultimate tensile strength (UTS), which are indispensable for the design protocol of nuclear fusion components. Heterogeneous residual strain over the sub-regions of the weldment in association with microstructure aggravates crack initiation and propagation at the interface of two sub-regions.

Efforts have been made to quantify multi-scale residual stress and strain experimentally in laser-welded Eurofer97 joints using neutron diffraction, neutron Bragg edge imaging, plasma focused ion beam-digital image correlation (PFIB-DIC) and nanoindentation, and predictions of residual stress and strain have been achieved via a finite element analysis (FEA) simulation [20–22]. Whilst previous studies established the interactions of residual strain with microstructure substantial changes in micro-hardness [21], establishing micro-hardness alone is insufficient for a comprehensive analysis of residual strain effects on deformation mechanism. The tensile behaviour is usually required to demonstrate the yield strength for lifetime assessments of engineering components. Investigations on the tensile deformation mechanism have only been achieved with correlation to microstructure in a Eurofer97 joint. For example, the strain evolution of laser-welding Eurofer97 was characterised using a DIC technique and compared with that of the post-welded heat treatment (PWHT), where the microstructure is restored [23]. Nevertheless, exactly how the residual strain distributes in the loading direction, and its effect on the tensile deformation remains unclear. This incomplete understanding prevents a full appreciation of the underpinning deformation mechanisms operating in laser-welded Eurofer97 steel.

Here, the deformation mechanism of the laser-welded Eurofer97 joint is studied experimentally by comparing the tensile strain evolution in an as-welded joint with a PWHT joint, measured by *in situ* neutron diffraction and a DIC technique. Prior to loading, neutron diffraction was applied to quantify the residual strain distribution across the weldment, and the microstructural distribution was investigated using electron backscatter diffraction (EBSD). As the diffraction technique is vulnerable to microscale residual strain and compositions, PFIB-DIC was used to validate that the lattice residual strain is representative of the macroscale residual strain. By studying the tensile behaviour in the fusion zone (FZ), heat-affected zone (HAZ), and base material (BM) regions, the interactions with residual strain and microstructure were explored. This investigation establishes an underpinning understanding of the deformation mechanism, which provides insight into optimising the remote welding processes for the fusion power plant components so as to improve the longevity of the fusion reactor.

2. Materials and methods

2.1. Materials and microstructural characterisation

Eurofer97 steel made by Böhler Austria GmbH, with a composition of Fe-0.11C-8.82Cr-1.08W-0.13Ta-0.48Mn-0.20 V (in wt.%), was used in the present investigation. The detailed fabrication, rolling processes, and heat treatment of Eurofer97 steel were described in a previous study [24]. The single laser welding process was carried out by The Welding Institute (TWI) using a 5 kW Yb-fibre laser source, achieving a welding speed of 1.2 m/min. The post-weld heat treatment (PWHT) was applied for residual stress relaxation and re-distribution, which involved annealing at 760 °C for 1.5 h under vacuum conditions, followed by furnace cooling [22]. For the tensile tests, flat dog-bone specimens were designed based on the ASTM E8 standard size with a gauge length of 25 mm and a cross-section of $6 \times 6 \text{ mm}^2$ and cut from the laser-welded Eurofer97 plate using electrical discharge machining. To provide a stress-free (d_0) reference for neutron diffraction analysis, as-welded sample and PWHT samples were cut into a comb-shape, where the teeth of comb are in dimension of $4 \times 1.5 \times 6 \text{ mm}^3$, to release the residual strain in the loading direction. The microstructure was characterised along the centre line of the laser-welded Eurofer97 joint using EBSD, and further residual stress characterisation was carried out via neutron diffraction and PFIB-DIC, as shown in Fig. 1(a). The EBSD maps were recorded using a Joel JSM-7100F with an accelerating voltage of 20 kV equipped with a Thermo Fisher Lumis EBSD detector using a step size of 500 nm. The post-data treatment was achieved using the MTEX 5.2.8 MATLAB toolbox [25], and the average grain size in the FZ, HAZ, and BM regions was measured from the EBSD orientation maps using the mean linear intercept method [26].

2.2. Lattice residual strain and lattice strain evolution characterised using neutron diffraction

The strain characterisation was performed using the SALSA monochromatic neutron diffraction instrument at the Institut Laue-Langevin, France, with a gauge volume of $0.6 \times 0.6 \times 2 \text{ mm}^3$ to obtain high-resolution strain data in the sub-regions of the weldment (Fig. 1(a)). The grain family whose crystal plane is normal to $\langle 110 \rangle$ lies along the loading direction is designated as $\langle 110 \rangle // LD$. A wavelength of 1.685 Å was used to measure the $\langle 110 \rangle // LD$ grain family in the loading direction. Prior to the tensile loading, neutron diffraction measurements were performed to measure the residual strain distribution. This was achieved by taking line-scan measurements across the weldment at 0.3 mm intervals, characterising the material at five different depths (from 1 mm to 5 mm below the top surface). The uniaxial tensile load was then applied in a step-load controlled manner where the maximum loading stress applied (610 MPa) was below the ultimate tensile stress (UTS). At each load step of the loading and unloading procedure *in situ* neutron diffraction measurements were carried out by performing a line scan across the weldment at the mid-thickness position, as illustrated in the inset figure in Fig. 1(a). To ensure the quality of the diffraction patterns, 1500 neutron statistics were acquired at each datapoint. The diffraction spectra were then analysed to obtain scattering angle, 2θ , for lattice strain calculation and the full width at half maximum (FWHM) for microstructure analysis [27].

2.3. Macroscopic strain evolution using DIC

The lab-based uniaxial tension experiments were carried out using the same geometry dog-bone samples, and the tensile strain evolution was evaluated using DIC. Tensile loading was applied until failure with a universal tensile instrument (Instron 5500 R) at a constant displacement rate of 0.02 mm/s. A raster speckle pattern was designed and printed with a dot size of 0.05 mm and a density of 75% to achieve a high-quality correlation [28,29]. DIC software (Vic-2D v6, Correlated Solutions, US)

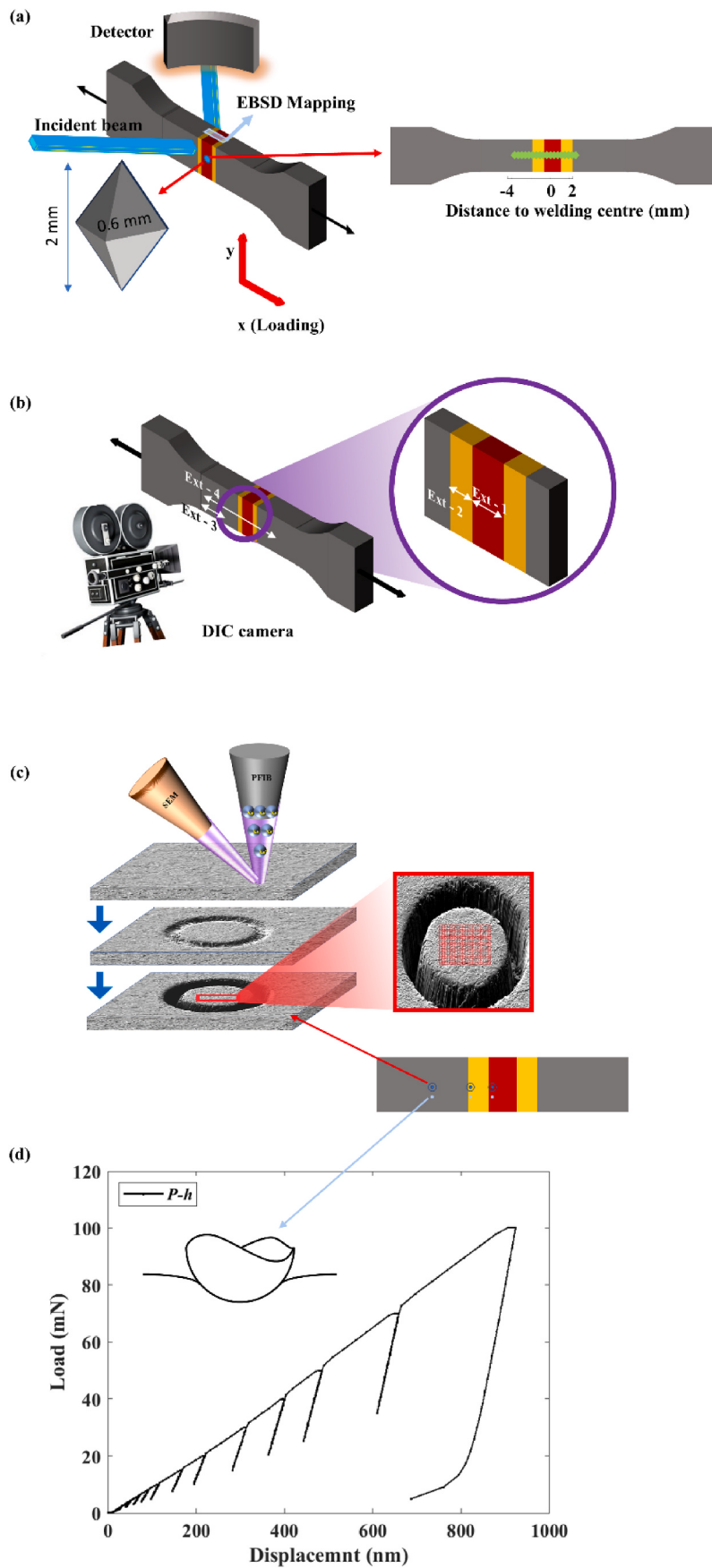


Fig. 1. Schematic figures of the experimental setup. (a) *In situ* neutron diffraction lattice strain measurements and residual lattice strain measurements prior to loading. Inset figure shows the line scans were performed at 300 μm intervals, covering the three sub-regions during uniaxial tension. (b) Lab-based uniaxial tensile experiments were conducted, and strain evolution was analysed using Vic-2D DIC system. The virtual extensometers (Ext-1-4) are highlighted for macro strain measurements in different regions. (c) Incremental PFIB milling procedures during which residual strain is released. Sequence of secondary electron images were acquired for residual strain analysis using MATLAB-based DIC method. (d) Nanoindentation experiments were performed using a spherical indenter with 13 cycles of load and unload procedures at the highlighted position in fusion line, HAZ/BM interface, and BM region.

[30] was used to obtain the tensile deformation with a subset of 51 pixels and a step size of 5 pixels. Engineering strain was calculated from the displacement measurements, removing outliers with a 15-pixel Gaussian smoothing decay filter. The strain evolution in the sub-regions of the weldment was evaluated with virtual extensometers (nominated as Ext-1, 2 and 3), and the Ext-4 was welded to measure the global strain evolution for laser-welded and PWHT samples (Fig. 1(b)).

2.4. Residual strain characterisation using PFIB-DIC

The PFIB-DIC ring-core method was used to quantify the residual strain at the fusion line (FZ/HAZ interface) and HAZ/BM interface. These regions usually contain dramatic microstructural changes and significant microscale residual strain which affects the peak position. Neutron diffraction is vulnerable to these microscale strains. Using the PFIB-DIC technique can validate that the residual lattice strain derived from neutron diffraction measurements is minimally affected by the insufficient microscale strain and can be representative to macroscale residual strain in loading direction. Fig. 1(c) shows the PFIB-DIC measurement position and how the incremental material removal was adopted to fabricate a ring-core with an inner diameter of 30 μm that contains enough grains to represent the macroscale residual strain given the pillar dimension is approximately three times the grain size [31]. The PFIB milling procedure was performed using a TESCAN MIRA 3 PFIB-SEM (University of Surrey, UK). When the milling depth reached the diameter of the ring-core, the residual strain is assumed to be fully released [32]. High-quality low-noise secondary electron images were obtained after each material removal step, with a spot size of 10 nm. A specifically developed automated program by TESCAN s.r.o. Repeatedly conducted the incremental milling and image acquisition procedures. As reported elsewhere, MATLAB-based DIC software was used to visualise the strain relaxation using a resolution of 20 pixels and a subset size of 31 pixels for each increment [31,33].

2.5. Determination of stress-strain curve using spherical nanoindentation tip

The nanoindentation was conducted at the position of the solid light blue dot in Fig. 1(d) to determine the stress-strain curve for the sub-regions of the laser-welded sample. Load control was used to conduct the indentation in a multicycle mode using a spherical diamond indenter with a ball radius of 4.25 μm from Ultra Indentation Tester (UNHT), CSM. The stepwise force were applied in 13 loading cycle until increasing to 100 mN with unloading of the half of maximum load of the cycle. A dwell time of 60s was used at the maximum load in each cycle to minimise the thermal drift. The load-displacement data were analysed with the Oliver and Pharr method [34] by fitting the unloading curve of each cycle to a power law to obtain contact stiffness, S , which was used to calculate the contact depth (h_c) and area (A). The spherical indentation tip was selected because of its advantages of enabling a continuous change of the mean pressure and characteristic strain as the indenter advances into the material [35]. A wide range of analytical formulations have been developed to calculate effective values of indentation stress and strain, which could depend on the instrument, materials, and test procedures. In previous studies [34], it has been found that the geometrical definition of contact radius provides a better fit of elastic properties compared with the Hertz' contact mechanics. Therefore, in this work, the contact radius was calculated using the indenter radius (R_i) standard geometric construction of a spherical cap ($a_c = \sqrt{2h_c R_i - h_c^2}$). The indentation strain and stress are defined by the widely used expression originally presented by Tabor, where the empirical factors are employed ($\epsilon_{IT} = 0.2 \frac{a_c}{R_i}$ and $\sigma_{IT} = \frac{L}{2.8A}$) [36].

3. Results

3.1. Microstructures

The stitched EBSD orientation maps in inverse pole figure (IPF) colouring, pole figures (PFs), and IPFs are shown in Fig. 2. Based on the microstructural characterisation of phases, grain size, and the nature of any precipitates, the weldment is conventionally divided into three sub-regions, i.e., FZ, HAZ, and BM regions. The selected single laser welding technique induced narrow FZ and HAZ regions with widths of ~ 1.2 mm and 1.4 mm, respectively. The average dimension of the grain size across the weldment in the loading direction varies, being 18.1 ± 1.1 μm in the FZ region and 10.7 ± 0.6 μm in the HAZ region. And 12.3 ± 0.7 μm in the BM region. The high thermal input leads to full austenitisation and grain growth in the FZ region, whilst partial austenitisation and subsequent martensitic phase transformations are evident in the HAZ region. The crystal dimension changes dramatically around the fusion line, although there is a narrow transition area. In the BM region, the grain size resembles the as-received materials. The PF indicates a weak imprecise texture for the measured (110) crystal plane, and the IPF implies isotropic properties in the loading direction in both the FZ and HAZ regions.

3.2. Macro-scale tensile strain evolution characterisation using DIC

The tensile strain evolution was measured using DIC while the laser-welded and PWHT samples were deformed. As shown in Fig. 1(c), the virtual extensometers were applied to quantify strain evolution in sub-regions of the weldment (Ext-1,2 and 3) and the global strain evolution of the whole gauge volume (Ext-4). The engineering stress-strain curves, before necking, are shown in Fig. 3(a) and Fig. 3(c). Yield stress was measured with a standard 0.2% offset on the stress-strain curves. The FZ region shows a higher yield strength of 539 MPa than the HAZ region (519 MPa). The global yield stress of the laser-welded sample (484 MPa) is higher than that of the PWHT sample (444 MPa). The inset figure in Fig. 3(a) shows that once the laser-welded sample commenced plastic deformation, the FZ and HAZ regions deformed less than the BM region. Fig. 3(b) visualises the accumulation of heterogeneous localised strain in the low-strength BM region at different applied loads during tensile deformation. The position where the local strain accumulates significantly indicates the possible fracture occurrence in the corresponding region. Closer inspection of the weld region in the early stages of deformation shows that the strain evolution around the fusion line is as significant as that in the BM region, whereas minor strain evolution is observed in the FZ centre and at the HAZ/BM interface. Looking at Fig. 3(d), only a small inhomogeneous tensile strain evolution was observed in the PWHT samples, and the highest strain evolution was observed around the HAZ/BM interface.

3.3. Micro-scale strain evolution characterisation using spherical indenter

Fig. 4(a) shows the real-time load-displacement curves during the multiple cyclic loading and unloading. Compared to the fusion line and HAZ/BM interfaces, the deepest penetration was disclosed in the BM region at the same load. The scatter in Fig. 4(b–d) shows the indentation stress and strain derived by the equation introduced above. The curves present the representative plastic deformation behaviour which was expressed by fitting the scattering using a power law equation ($\sigma_{IT} = K\epsilon_{IT}^n$) via the strength coefficient (K) and indentation work hardening coefficient (n). The representative elastic tensile behaviour was described in dash line (Fig. 4(b–d)) by employing the values of Young's modulus obtained from nanoindentation test. The indentation stress at the yield point at the interfaces and BM region was obtained from the intercept between the elastic and plastic fitting curves, which are 1396 MPa, 1527 MPa and 622 MPa, respectively. The indentation yield stress is higher than that measured from tensile test, which is likely arisen from

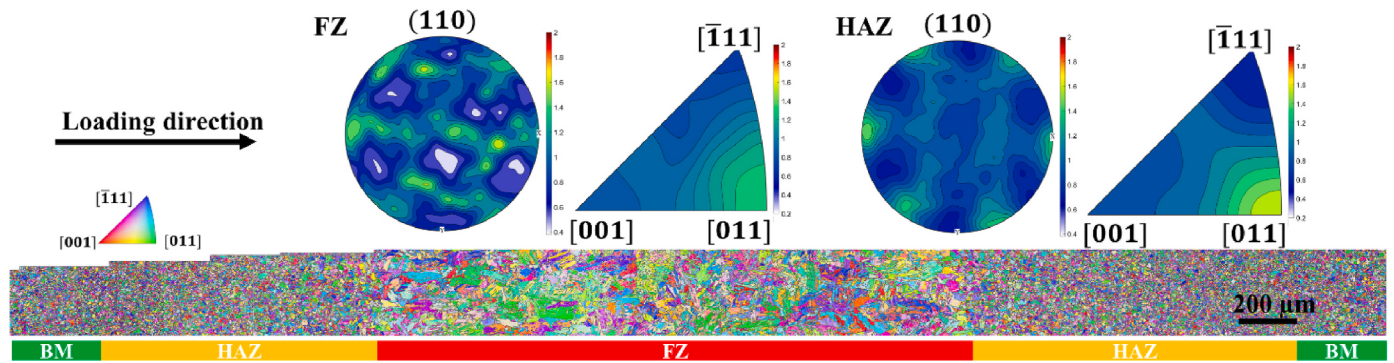


Fig. 2. EBSD orientation map in IPF colouring across the weldment. On the top of the EBSD orientation map is the PF of (110) crystal plane and IPF in the loading direction for the FZ and HAZ regions, respectively.

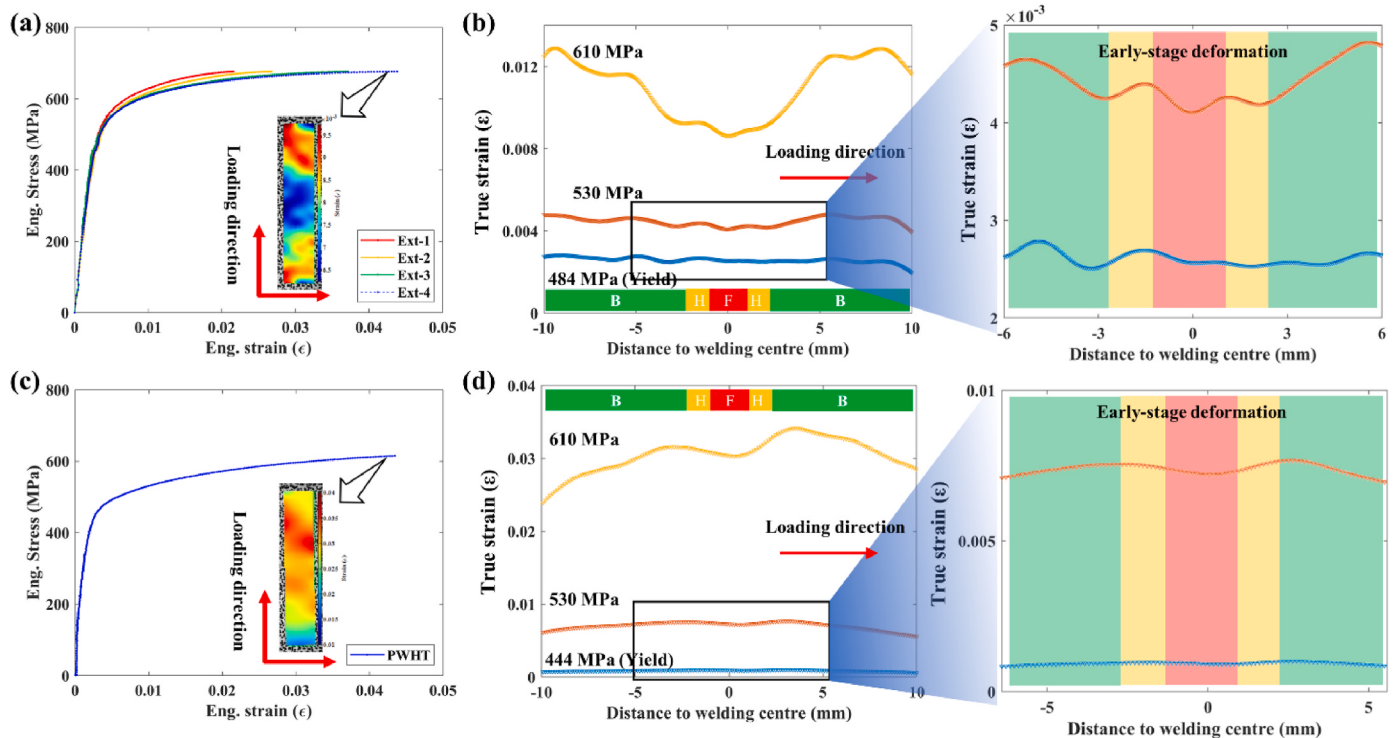


Fig. 3. Tensile performance of laser-welded and PWHT samples derived from the lab-based uniaxial tension. (a) & (c) Engineering stress–strain curves for the loading direction. The inset figure shows the 2D heterogeneous strain evolution maps across the sample. (b) & (d) Heterogeneous strain evolution occurs in sub-regions of the weldment at different applied stress. The colour bars in (b) and (d) show the sub-region of the weldment, where the F is FZ region, H is HAZ region and B is BM region. (For interpretation of the references to colour in this figure legend, the reader is referred to the Web version of this article.)

indentation size effects, as the radius of $4.25 \mu\text{m}$ only cover one or a couple of grain that is of the order of $\sim 18 \mu\text{m}$ in FZ region and $\sim 12 \mu\text{m}$ in HAZ and BM regions. In addition, the definition of indentation strain-stress parameters (i.e., the true indenter radius as a function of depth) and the yield point under indenter could be oversimplified here. However, acquiring detailed elastic and plastic properties in different sub-regions is desirable to establish predictive tools for lifetime assessment of the engineering joints. The high-resolution and high-precision provide this technique unique advantages in measuring the narrow sub-regions of the weldments, and new and more rigorous methods are under development.

3.4. Lattice-scale strain evolution during uniaxial tension

To study the deformation of the laser-welded Eurofer97, *in situ* neutron diffraction measurement during tensile deformation was used to

quantify the lattice strain evolution and investigate the interaction behaviour between residual strain and tensile deformation. The average measurement uncertainty of the neutron diffraction was approximately $\pm 30 \mu\epsilon$ in the base material rising to approximately $\pm 70 \mu\epsilon$ in the FZ and HAZ [17]. As shown in Fig. 5(a), the lattice strain increases during tensile deformation, and a slightly lower lattice strain was observed in the FZ and HAZ regions at large deformation state (i.e. $\sigma = 610 \text{ MPa}$). A small degree of compressive lattice residual strain is induced in the $\langle 110 \rangle$ /LD grain family after unloading. To investigate the residual strain effect on tensile deformation, the global and local lattice strain evolution is extracted separately from the area highlighted by the rectangles in Fig. 5(a), and the lattice strain/applied stress curves in the four different regions are summarised in Fig. 5(b). In the elastic regime the lattice strain evolution is linear as a function of the engineering stress. Upon entering the plastic regime, it starts deviating from linearity, and a compressive shift is observed. The HAZ/BM interface (Area-2) shows the

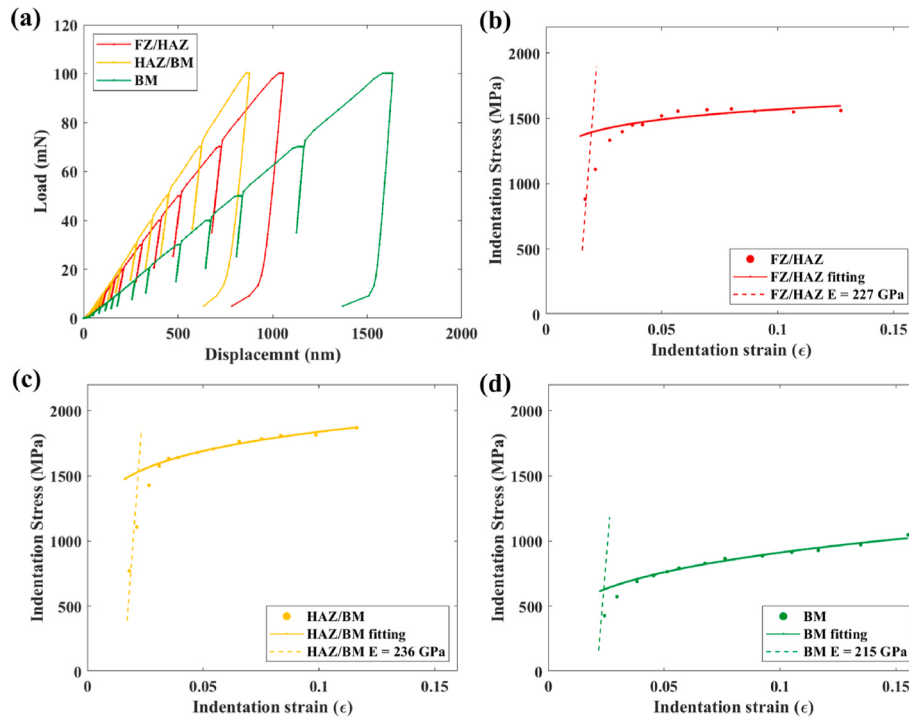


Fig. 4. Mechanical performance derived from sub-regions of laser-welded sample. (a) Load-displacement curves. (b–d) Scatter of indentation stress and strain, plastic behaviour curves fitted using power function and elastic behaviours described using values of Young's modulus obtained from indentation test. The interception between the two curves indicates the indentation yield stress.

highest yield strength (506 MPa) compared to the other areas (449 MPa), and a compressive shift in lattice strain is observed for the fusion line and HAZ/BM interface area. Comparing the slopes of the unloading curve, a similar stiffness is observed in all four areas.

The lattice strain evolution of PWHT sample during tensile deformation can be observed in Fig. 6(a). The full relief of the residual strain after PWHT in the loading direction is manifest, and the lattice strain distribution is homogeneous across the weldment under tension (Fig. 6(a)). A slightly lower applied strain was found in the welding region at the highest applied stress ($\sigma = 600$ MPa), which indicate the microstructures were insufficiently restored. As was the case with the laser-welded sample, residual compressive strain is induced in the PWHT samples after unloading. The global lattice strain evolution of the PWHT sample is plotted as a function of engineering stress (Fig. 6(b)), indicating the linearity until 449 MPa. Although the tensile strength is similar with the global lattice strain evolution of the laser-welded sample, the stiffness of the PWHT sample is lower. As shown in Figs. 5(c) and Fig. 6(c), in contrast to the constant FWHM during the tensile deformation (0.52°) in the BM regions, a mild decrease in the FWHM is observed in the FZ and HAZ regions in the plastic regime, which may indicate a lower dislocation density. After the PWHT, the FWHM is similar across the sample and remains the same as in the BM region.

3.5. Residual strain characterisation in loading direction using neutron diffraction and PFIB-DIC

The residual strain distribution in the loading direction of laser-welded and PWHT Eurofer97 measured via neutron diffraction and the PFIB-DIC ring-core method is shown in Fig. 7. The residual lattice strain of $\langle 110 \rangle$ //LD grain families was characterised by neutron diffraction on laser-welded and PWHT samples. Fig. 7(a) shows a peak of residual tensile strain (with a magnitude of $\sim 0.6 \times 10^{-3} \epsilon$) in the FZ region, next to a trough of residual compressive strain ($\sim -1 \times 10^{-3} \epsilon$) in the HAZ region. The PWHT reduces the residual strain substantially, as shown in Fig. 7(b). The measurement using neutron diffraction

indicates that the residual strain is of a similar magnitude along the thickness direction. To overcome limitations of neutron diffraction in microscale strain, the PFIB-DIC ring-core method was employed to validate the results derived from neutron diffraction. The residual strain relaxation in the loading direction is characterised at the fusion line and at the HAZ/BM interface (Fig. 7(c)) using the PFIB-DIC method. Inverting the sign of the strain relaxation is the residual strain. The microscale strain is supposed to be maximum at measured positions due to the significant changes of microstructures and phase constituents. Additionally, the peak tensile/compressive residual strain is observed in these regions via neutron diffraction technique. The residual tensile strain derived using the PFIB-DIC technique is consistent with the results from neutron diffraction in the FZ region, although the residual compressive strain shows a slightly higher value ($\sim -1.6 \times 10^{-3} \epsilon$) in the HAZ region, which validates reliability of using residual lattice strain of $\langle 110 \rangle$ //LD grain families to represent macroscale residual strain. Both techniques indicate there is almost negligible residual strain in the BM region.

4. Discussion

In body centred cubic steels, the $\{110\}$ family of planes are weakly affected by intergranular strains and can, therefore, be considered to represent the bulk mechanical properties of the material [37]. Here, the $\langle 110 \rangle$ //LD grain family is selected to study the residual strain and strain evolution with neutron diffraction. The residual strain distribution of laser-welded and PWHT Eurofer97 steel was investigated using neutron diffraction. As the narrow FZ and HAZ regions induced by the laser welding, a high-resolution measurement (0.6 mm) was applied to evaluate the residual strain distribution in the sub-regions. The 0.3 mm spacing between each gauge centre position results in 0.3 mm overlaps to achieve a high spatial resolution residual strain measurement in the loading direction. The FZ region shows residual tensile strain due to solidification shrinkage of the molten steel balanced by compressive strain in the HAZ regions [38]. The magnitude of residual strain is lower

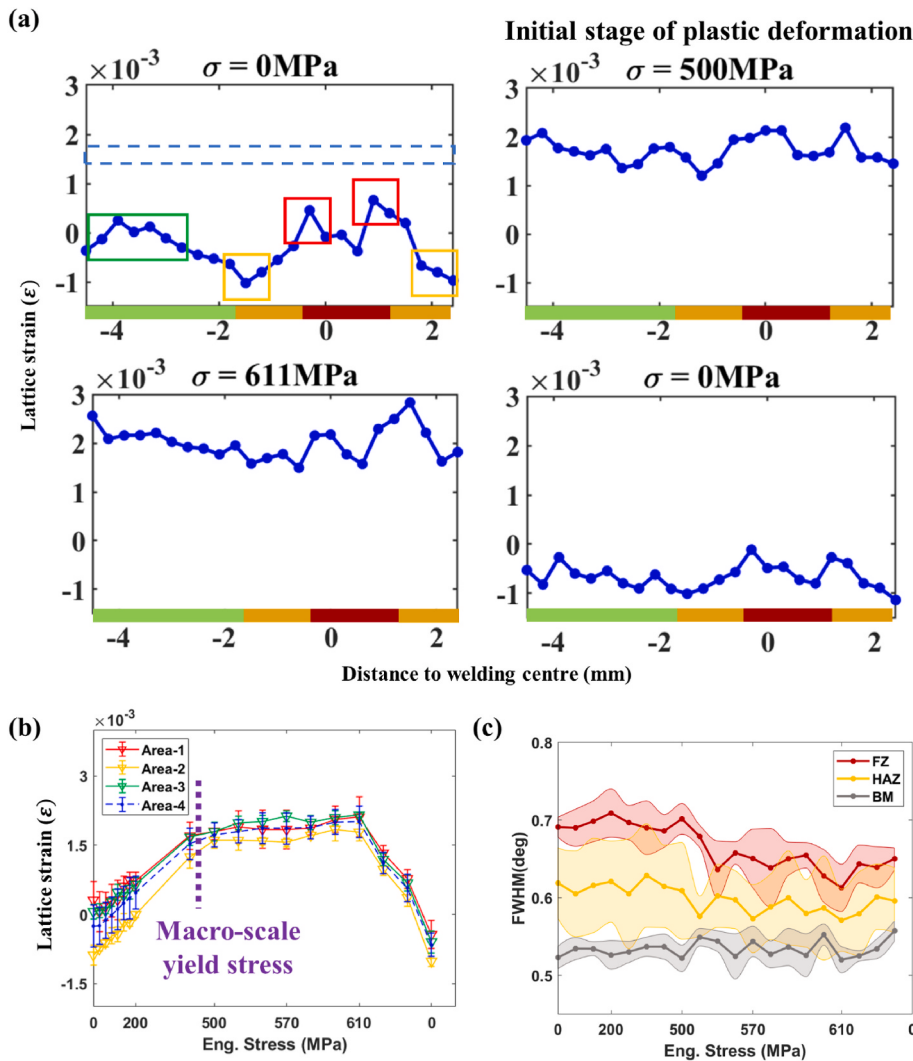


Fig. 5. Interaction between residual strain and applied strain for laser-welded sample. (a) Lattice strain distribution across the weldment at specific loads, e.g. initial residual strain state, initial plastic deformation, highest applied stress, and after unloading. The coloured bars indicate the sub-regions of the weldment. (b) Average lattice strain evolution at different regions is highlighted by the rectangles in (a), and the macro-scale yield stress is indicated. (c) FWHM evolution during the load and unload procedure.

around the centre of the FZ region because of the offset of compressive strain from the martensitic phase transformation [39]. The measured residual lattice strain distribution trend is consistent with the neutron diffraction measurements on autogenous laser-welded Grade 91 steel [17,18,40,41]. The similar magnitude of residual strain distribution in the thickness direction indicates a full weld penetration and a symmetrical thermal history. As the neutron diffraction is affected by the microscale residual strain and deviation of composition, resulting in challenges in measuring the residual strain in the small sub-regions, validating the residual strain measurement is necessary. The PFIB-DIC ring-core method was applied to validate the residual strain in the loading direction to ensure the microscale strain in both laser-welded and PWHT Eurofer97 does not affect the neutron diffraction measurement. As the texture was weak in the loading direction (see inset IPF of Fig. 2), the micro-strain (inter/intra-granular residual strain) was in self-equilibrium. Therefore, the PFIB-DIC measurement evaluated the macro-scale residual strain [31], which can be directly compared with the neutron diffraction measurements. The consistency of residual strain between the two techniques confirms that the uncertainty of the neutron diffraction measurements caused by the application of the reference lattice spacing is in a reasonable range, validating the precision of the measurements arising from the neutron diffraction.

The global lattice strain curve of $\langle 110 \rangle$ //LD grain family began to yield at 449 MPa (Area-4 in Figs. 5(b) and Fig. 6(b)), which is slightly earlier than observed on the macro-scale. This is attributed to other

grain families continuing to deform elastically. Furthermore, it is reported that the stiffness of the grain families decreases in the order $\{222\}$, $\{110\}$, $\{321\}$, $\{211\}$, $\{310\}$, and $\{200\}$ in Grade 91 steel (body centred cubic structure) [41]. The stiffer grain families also bears more load, which results in a slight compressive shift of the $\langle 110 \rangle$ //LD grain family relative to the elastic regime [41]. The slight compressive shift of lattice strain evolution in the plastic regime is consistent with the gradual stress transfer from stiffer crystal orientations to the more compliant ones [41–43]. The FWHM value of 0.52° in the BM region is consistent with previous studies of laser-welded Eurofer97 steel carried out using neutron diffraction [17]. The decrease of FWHM in the FZ and HAZ regions might be attributed to the elimination of free dislocations and lath boundaries (sub-grain boundaries) [44,45]. The FZ region contained more martensitic lath boundaries than the HAZ region, which might explain the large decrease in the FWHM value.

For a reliable lifetime assessment of welded components, knowledge of the tensile strain evolution is usually required to demonstrate the yield strength mismatch between sub-regions in weldments. The sub-region with the lowest strain evolution carries more load than the surrounding regions once the sample enters the plastic regime, leading to the elastic-plastic strain mismatch. This mismatch alters the intensity of local plastic deformation, consequently accelerating the initiation of cracks and affecting fracture toughness [46–48]. Using conventional mechanical tests, it is almost impossible to measure the strain evolution in such a small sub-region. Applying nanoindentation technique and

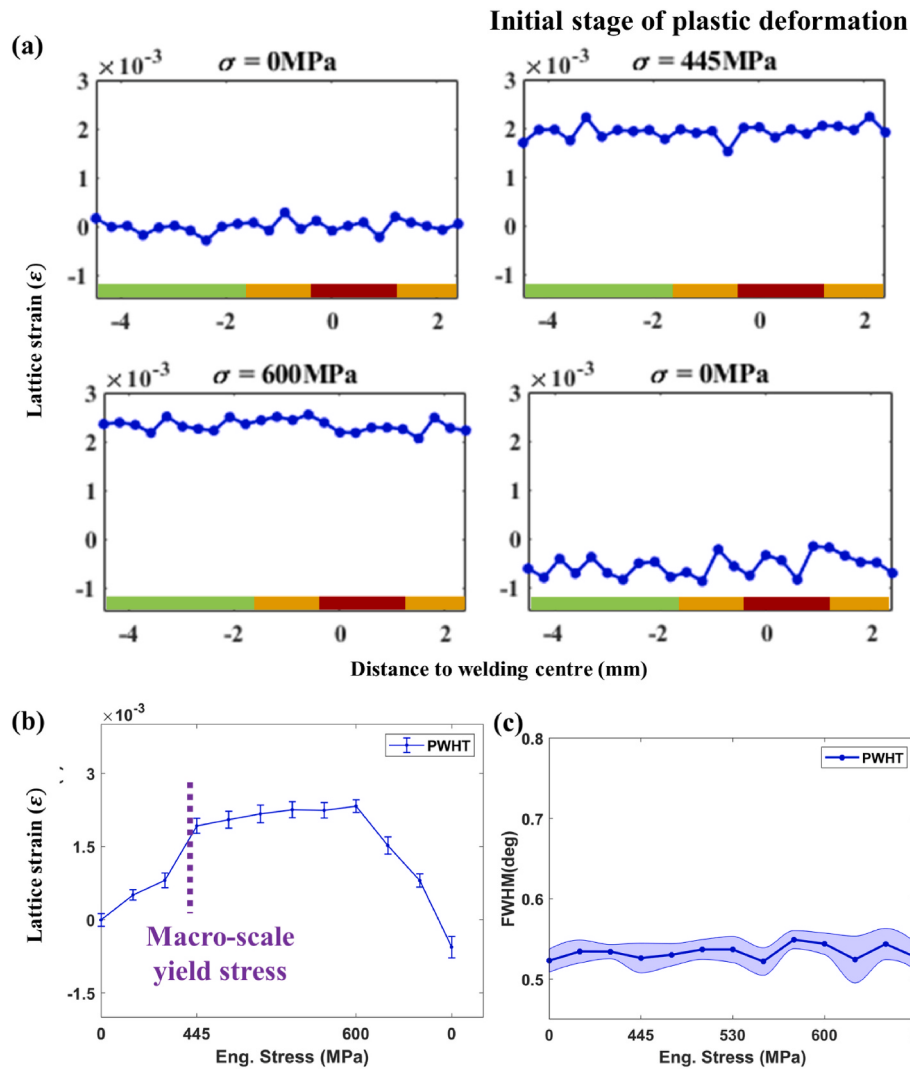


Fig. 6. Interaction between residual strain and applied strain for PWHT sample. (a), (b) & (c) are plotted in the same manner for the PWHT sample.

combining uniaxial tension tests with the *in situ* neutron diffraction and lab-based DIC enables the high-resolution characterisation of the microscale, lattice-scale and macroscale strain evolution in the loading direction, respectively. The heterogeneous tensile properties of the laser-welded sample in Fig. 3(a) indicates localised strain is accumulated in the BM regions, implying that deformation occurs more extensively than in the remaining material through necking and on to fracture. Less deformation was observed in the FZ and HAZ regions (Figs. 3(b) and Fig. 4(b–d)) due to the presence of martensite, which enables more load to be carried than in the other regions in the plastic regime [22,49]. This is demonstrated with multi cycle indentation test, where FZ and HAZ regions have higher yield strength than BM region, together with indicative stress-strain evolution. The increase in carbon in solution, owing to carbide dissolution in the FZ region, is another cause of strengthening [22,41]. The refined grains strengthen the material in the HAZ according to the Hall-Patch effect [50,51]. The microstructural strengthening gradually diminishes while moving to the BM region, where the microstructure is very similar to that of the parent material.

The heterogeneous strain distribution in the magnified part of Fig. 3 (b) implies that the significant tensile deformation is driven by the residual tensile strain at the fusion line. By contrast, the residual compressive strain at the HAZ/BM interface hinders local tensile deformation. Similar conditions were also found from the $\langle 110 \rangle // LD$ grain family (Fig. 5(a) and (b)) and microscale indentation stress-strain

curves (Fig. 4(b–d)). The quantitative strain evaluation characterisation via *in situ* neutron diffraction indicates an increase in yield strength to 506 MPa at the HAZ/BM interface (Area-2). Conversely, the residual tensile strain degraded the yield strength at the fusion line (Area-1) to the same as that in the BM region. Additionally, the residual strain effect is substantial during the elastic deformation and initial stage of plastic deformation (magnified figures in Fig. 3(b) and (d)). The increasing plastic deformation, however, limits the residual strain effect in the total strain evaluation because the induced tensile deformation changes the residual strain states. This implies that the microstructure dominated local strengthening in the FZ and HAZ regions at large plastic deformation. A small degree of compressive lattice residual strain was found in the $\langle 110 \rangle // LD$ grain family after unloading for both laser-welded and PWHT samples, which might be attributed to the reverse plastic flow on unloading [8]. A comparison with a previous evaluation of tensile strain evolution carried out on Grade 91 steel joint [23] shows that the present results display a comprehensive underpinning understanding of the deformation mechanism due to the establishment of the correlation between residual strain, microstructure, and tensile properties in the sub-regions of the weldment.

The PWHT lowered the global yield strength to around 444 MPa by inducing tempering of martensite and formation of fine precipitates, which is consistent with previous investigations [22,23]. The FWHM dropped in the FZ and HAZ regions, indicating that the tensile strength

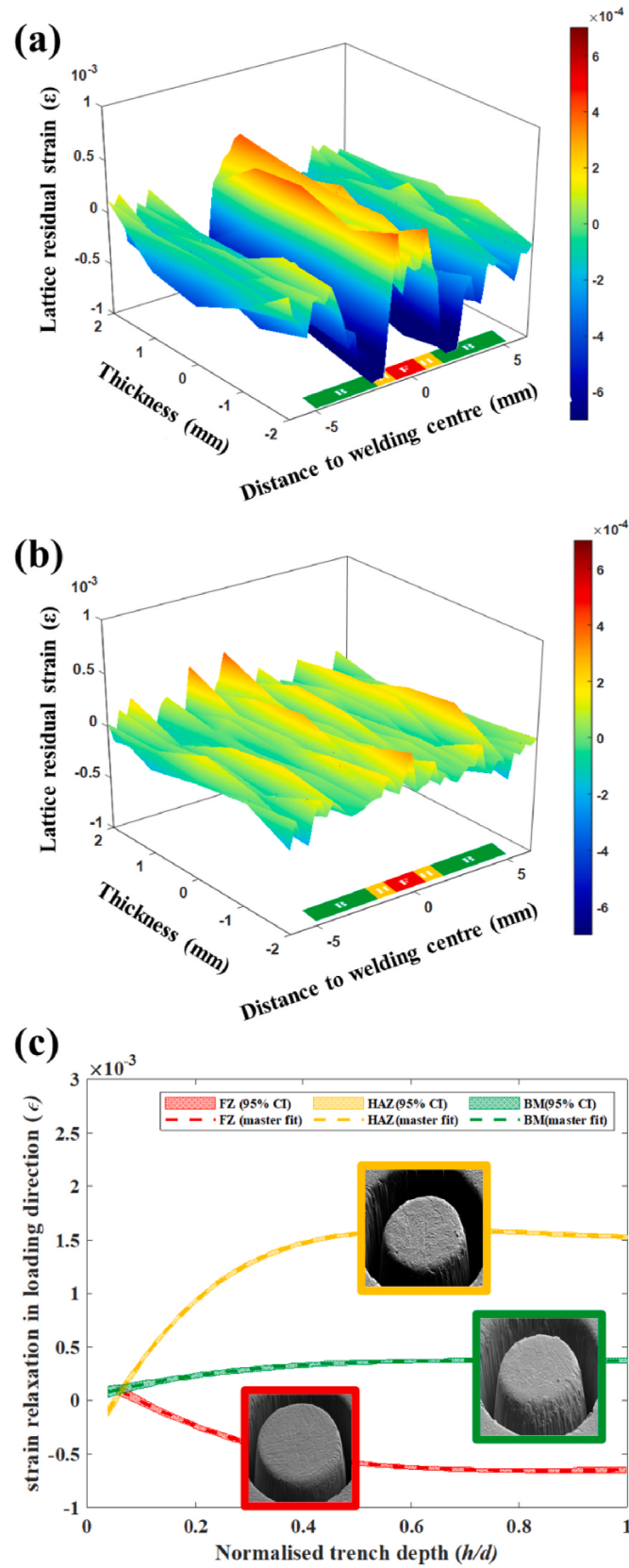


Fig. 7. (A) Residual strain distribution of $\langle 110 \rangle$ /LD grain family of laser-welded sample. (b) Residual strain distribution of $\langle 110 \rangle$ /LD grain family of PWHT sample. (c) Strain relaxation in loading direction measured by the PFIB-DIC ring-core method at FZ line (red), HAZ/BM interface (yellow) and BM (green) region, respectively. (For interpretation of the references to colour in this figure legend, the reader is referred to the Web version of this article.)

degraded due to a decrease in the dislocation density [52]. Although the microstructure was not fully restored and a slightly heterogeneous strain evolution (Figs. 3(d) and Fig. 6(a)) was found in the weld region when high plastic deformation was applied, the PWHT eliminated residual strain in the loading direction entirely and reduced the strain mismatch around the fusion line and at the HAZ/BM interface significantly. In addition, without the prevention of residual compressive residual strain, the strain accumulation and fracture tended to occur at the HAZ/BM interface as a result of the residual mismatch between the two microstructures.

5. Conclusions

The residual strain distribution in the loading direction has been obtained using neutron diffraction and the PFIB-DIC ring-core method. The tensile strain evolution in a laser-welded and PWHT Eurofer97 steel during tensile deformation has been measured using *in situ* Neutron diffraction and lab-based tensile-DIC. The principal conclusions of this study are.

1. The residual strain of laser-welded sample measured by neutron diffraction established that the highest residual tensile and compressive strain in the loading direction is found around the fusion line ($0.6 \times 10^{-3} \mu\epsilon$) and at the HAZ/BM interface ($-1 \times 10^{-3} \mu\epsilon$). The three PFIB-DIC ring-core measurements from around the fusion line, the HAZ/BM interface, and the BM region confirmed the findings of the neutron diffraction.
2. The *in situ* neutron diffraction and lab-based tensile-DIC strain evaluation showed that the FZ and HAZ region of laser-welded sample withstand more load once the applied stress exceeds the yield point of the material because of the microstructural strengthening. The residual compressive strain then induced the superimposed strengthening effect on the local tensile strength to 506 MPa near the HAZ/BM interface, which was distinctly higher than in the surroundings. Such mismatch accelerates crack initiation and affects the fracture toughness.
3. The selective PWHT is effective in mitigation of residual strain and elimination of elastic-plastic mismatch between sub-regions of the weldment.
4. The residual strain affects the tensile behaviour significantly at the initial deformation stage, and the microstructural strengthening mechanism dominates in the later stages of deformation.

Author contribution

Bin Zhu: Methodology, Investigation, Formal analysis, Writing – original draft, Writing – review & editing.

Nathanael Leung: Investigation, Writing – review & editing.

Yiqiang Wang: Supervision, Funding, Conceptualisation, Methodology, Writing – review & editing.

Hannah, Zhang: Investigation, Writing – review & editing.

Jirí Dluhoš: Methodology, Writing – review & editing.

Thilo Pirling: Investigation, Writing – review & editing.

Michael Gorley: Supervision, Writing – review & editing.

Mark J. Whiting: Supervision, Writing – review & editing.

Tan Sui: Supervision, Funding, Conceptualisation, Methodology, Writing – review & editing.

Data availability

Data will be made available on request.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence

the work reported in this paper.

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