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Enhanced strength-ductility synergy in stainless steel 316L through hierarchically tailored microstructure via laser-based repair deposition

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ABSTRACT

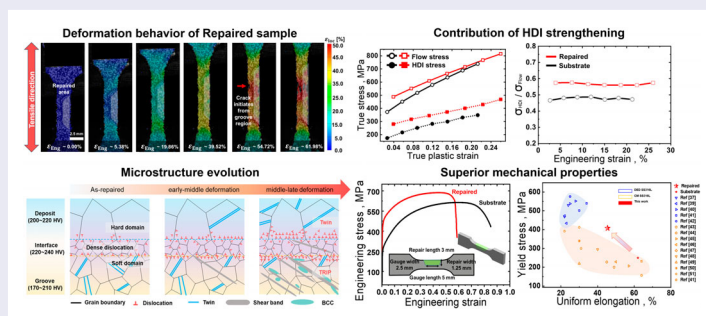
Repair offers significant time—cost savings and durability by reusing existing parts efficiently. Prior to implementation of repaired parts, it is crucial to clarify mechanical properties and deformation mechanisms to ensure proper performance. In this study, a repair deposition was performed on the groove of wrought stainless steel 316L using the same material powder. The repaired sample exhibited an excellent combination of ductility and strength (T.E.L: 58%, YS: 407 MPa, UTS: 673 MPa) compared to the wrought counterpart. With heterogeneity in microstructure, the hetero-deformation-induced strengthening of the repaired sample resulted in a synergy of strength-ductility combinations.

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Direct energy deposition; repair; deposition in a groove; microstructural heterogeneity; mechanical property



1. Introduction

Metal repair is recognized as an efficient strategy for restoring damaged components, thereby saving material cost, time, and effort, as opposed to manufacturing entirely new products [1, 2]. Among a variety of techniques, Direct Energy Deposition (DED) stands out as advanced solution for repairing damaged components with intricate geometries through direct deposition onto damaged surfaces [3, 4]. Saboori et al. [5] and Shamsaei et al. [6] have summarized

the DED repair process in various industrial applications, demonstrating its versatility across in different metallic materials. To properly estimate repairability, it is essential to demonstrate the integrated properties of the pre-existing part and repaired deposit including their geometric conformity. However, numerous studies predominantly focus on achieving defect-free deposits to enhance mechanical properties, indicating a substantial need for further research in this area.

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In essence, the DED process applies lasers directly to the pre-existing materials, inevitably inducing microstructural evolution on the surface [7, 8]. Significant changes in mechanical properties can be achieved by adjusting process parameters and tailoring the microstructural heterogeneity [9, 10]. DED-fabricated stainless steel 316L (SS316L) possesses higher strength but lower ductility compared to its conventionally processed counterpart, indicating that the microstructure consists of austenite γ and δ -ferrite [11, 12]. These heterostructured materials exhibit extra hardening potential and enhanced mechanical performance greater than those expected by the rule of mixtures during tensile deformation [13]. The underlying mechanism of hetero-deformation-induced (HDI) strengthening is revealed to achieve a superior strength-ductility combination [14, 15]. As a result, repair deposition enables damaged parts to be restored to their original dimensions with microstructural heterogeneity between the repaired and pre-existing structure. Therefore, we introduce enhanced strength ductility synergy in repaired SS316L compared to its original properties to achieve a better understanding of repair application in industrial areas.

2. Materials and method

For the deposition process, commercial gas-atomized SS316L powders were used, and their morphologies were presented using field-emission scanning electron microscopy (FE-SEM). The shape of SS316L powders is spherical with satellites, as shown in Figure 1a, and its average particle size is $\sim 103.6 \mu\text{m}$ measured by laser particle-size analyzer (Malvern Mastersizer 2000, Malvern, UK). The chemical composition of the SS316L powders is 67.62Fe-2.5Mo-1.12Mn-0.62Si-11.03Ni-17.09Cr-0.022C (wt%).

Through wire cutting, a 3 mm depth groove with a trapezoidal shape of 135° was made, as seen in Figure 1b. Based on the process parameters shown in Table 1, a total of 31 layers were deposited in a bi-directional pattern, with a rotation angle of 90° . For the mechanical testing, plate-type and sub-sized tensile specimens were extracted with a 5 mm gauge length and a 2.5 mm gauge width in the substrate. The repaired tensile specimen includes a deposit area with a 1.25 mm in gauge width and a 3 mm in gauge length, as shown in Figure 1b.

The dog-bone specimens for the tensile test were polished up to 1200 grit with SiC papers.

The repair deposition was carried out using the DED machine (MX-lab, InssTek Co., Republic of Korea) composed of a laser module, and 3-axis movement. The deposition was made into the grooves with a laser diameter of $400 \mu\text{m}$, where the hatching space is $300 \mu\text{m}$ and layer height is $100 \mu\text{m}$. Ar gas was used for the gas flow system, where each of its flow values was controlled with 3 L/min of coaxial gas, 8 L/min of shielding gas, and 8 L/min of powder feeding gas. The SS316L powders were fed by a nozzle through an Ar gas stream, where the speed of the laser module was adjusted at 850 mm/min.

Microstructure analysis was conducted using an FE-SEM (JSM-7100F, JEOL, Japan) equipped with an electron backscatter diffraction (EBSD) detector with the step size of $1.5 \mu\text{m}$. The EBSD data were processed by using TSL-OIM7 analysis software. All the specimens for EBSD were ground and polished up to $0.04 \mu\text{m}$ with a colloidal silica suspension. Microhardness profiling was carried out using a micro Vickers hardness machine (HM-220B, Mitutoyo, Japan) to verify the hardness of the specimen, including the repaired and adjacent areas. A total of 1,440 data points were obtained through hardness measurements conducted at intervals of 0.1 and 0.15 mm on a surface area measuring 4.0 mm in depth and 5.4 mm in width to analyze the hardness distribution across the surface. The tensile tests were conducted with a universal testing machine (Instron 1361, Instron, USA). All the tensile tests were repeated at least 3 times for reproducible results. During the tensile test, the digital image correlation (DIC) method was applied to measure the tensile strain accurately with the machine (Aramis 5M, GOM, Germany).

3. Result and discussion

Figure 1c illustrates the initial microstructure of a wrought substrate, along with the EBSD inverse pole figure (IPF) map characterized by uniformly shaped grains. As depicted in the magnified IPF map of Figure 1d, the overall grains appear equiaxed, with an average grain size of $\sim 23.51 \mu\text{m}$. In contrast, Figure 1f shows a variation of microstructure in the initial state of the repaired sample (See more in Fig. S1). The microstructure of the groove exhibits the typical characteristics of a wrought sheet (equiaxed grains, grain size: $\sim 20.99 \mu\text{m}$).

Table 1. Process parameter of DED repair deposition.

Deposition strategy	Laser power	Travel speed	Hatching space	Scanning pattern
Contour	220 W for Filling	850 mm/min	0.3 mm	Bi-directional scanning with a 90° rotation for the next layer
Filling	280 W for Contour			
Contour				

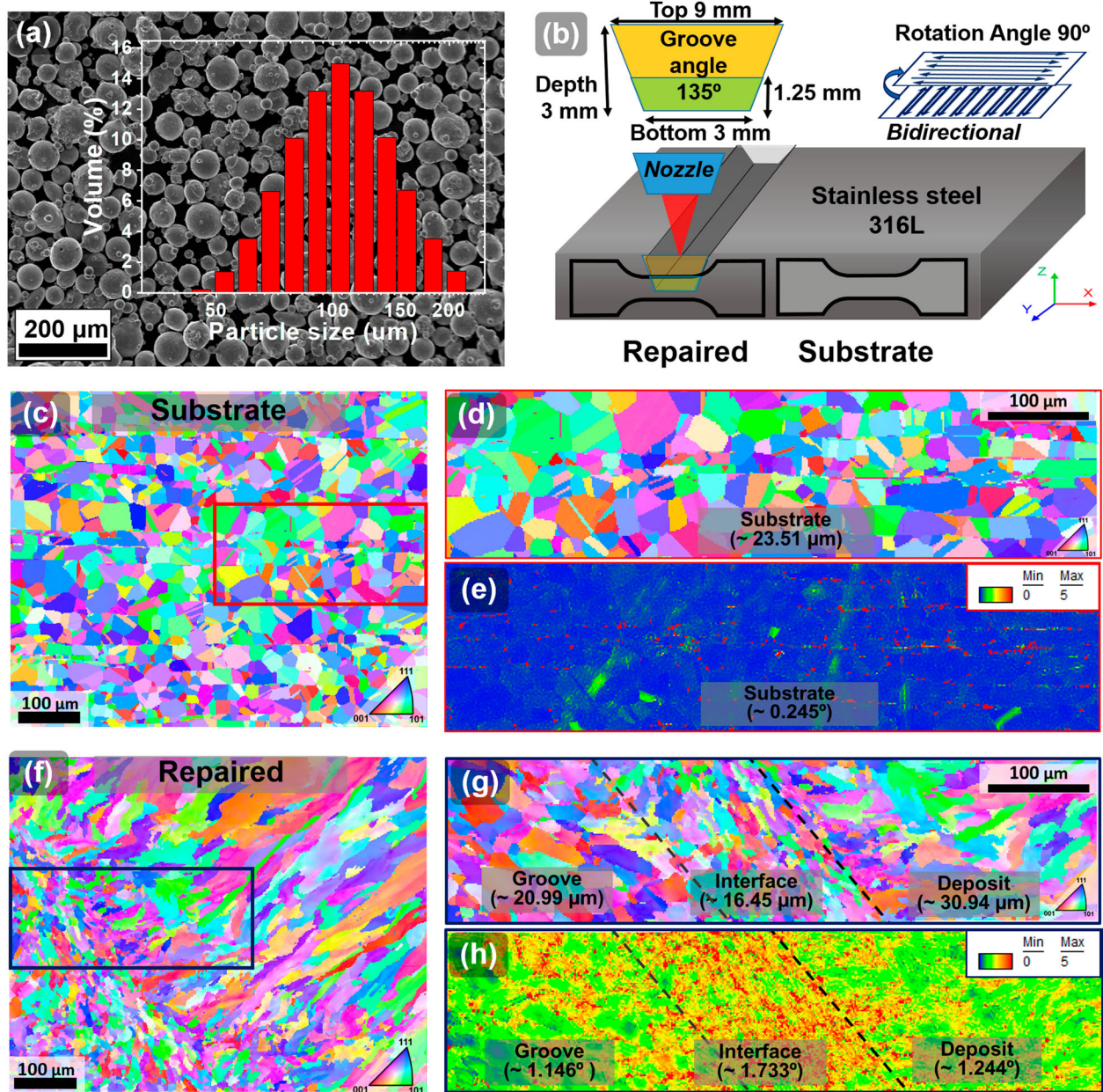


Figure 1. (a) The shape of SS316L powder (SEM image) with its size distributions (d_{10} : 67 μm , d_{90} : 163 μm), (b) Schematic of fabrication of tensile specimen, repair deposition process in the groove with an angle of 135° and depth of 3 mm. Initial substrate microstructure (c-e), repaired sample (f-h), containing magnified IPF map showing average grain size with KAM map indicating high dislocation density in the repaired sample.

During repair deposition, the microstructure in the heat-affected zone of the groove (interface) varies due to the influence of laser energy (grain size, $\sim 16.45 \mu\text{m}$). Initially, the grain size of the deposit area was $\sim 30.94 \mu\text{m}$ and progressively enlarged during the repair deposition. Notably, the grains in the deposit area show epitaxial growth in a diagonal direction toward the center of the deposit, which is in the opposite direction to the heat flux dissipation. During the deposition, grains grow in the opposite direction of the heat energy flux (i.e. in the thermal gradient direction) [10], as heat energy flows

from high to low temperature regions, creating a thermal gradient. This results in the promotion of a diagonal columnar shape because the growth direction is determined by the movement of heat energy from hotter to cooler regions [16]. Moreover, during the DED process, the previous layer is partially remolten, and solidification is often dominated by epitaxial growth combined with directional solidification in the building direction. The repeated heat cycles cause the grains to grow epitaxially from the substrate, and the continuous remelting and solidification process allows the grains to enlarge

progressively. This is driven by the thermal gradient and the movement of the solidification front, which promote the alignment and growth of the grains in a specific direction [11]. Consequently, the accumulation of heat from repeated thermal cycles leads to epitaxial growth and causes grain enlargement.

As observed in Figure 1f, significant variations in grain size occur across the entire region, resulting in overall heterogeneity. The initial columnar structure near the interface exhibits fine grain ($\sim 16.45 \mu\text{m}$), attributed to the higher cooling rates experienced in the initial deposit area, which promotes the formation of smaller columnar grains [12]. The reduction in grain size corresponds to an increase in the length of grain boundary within the microstructure, leading to the hindrance of the dislocation glide along slip planes. The strengthening mechanism of the refined microstructures was attributed to the Hall-Petch relationship, which correlates the strength to the grain size [11, 17]. Moreover, small grains undergo repeated melting and solidification during the cooling process with continuous accumulation of dislocations caused by repetitive thermal cycles [18, 19]. In Figure 1h, the kernel average misorientation (KAM) distribution in the initial state of the repaired sample is illustrated, indicating higher dislocation density at the interface region ($\sim 1.733^\circ$) compared to the substrate's ($\sim 0.245^\circ$) in Figure 1e, and the deposit ($\sim 1.244^\circ$). The

high dislocation density in the microstructures, which serves as the hardening mechanism well established by Taylor hardening, is due to the rapid solidification in the DED process where thermal contraction is continuously generated [20–22]. This confirms that the repaired sample experiences tension and compression due to localized thermal stress cycles induced by laser input, leading to a dense dislocation network [23, 24].

Figure 2a presents a hardness profile of the repaired sample. The deposit area exhibits higher hardness values compared to the groove region. Notably, the interface area on both sides shows the highest hardness value, clearly distinguishing the repaired deposit region from the groove region. Generally, the higher hardness and strength are generated in DED SS316L by reduced grain size and a dense dislocation network from the AM Process. Specifically, the repeated thermal cycles and rapid solidification during DED result in finer grain structures and increased dislocation densities within the deposit [25, 26]. These finer grains and higher dislocation densities enhance the hardness of the material. Possessing the relatively hard domain of the deposit/interface area and the soft domain of the groove simultaneously, the repaired sample demonstrates its intrinsic heterogeneity in mechanical properties.

The tensile properties of the repaired specimens, as plotted in Figure 2b, exhibit high yield strength (YS)

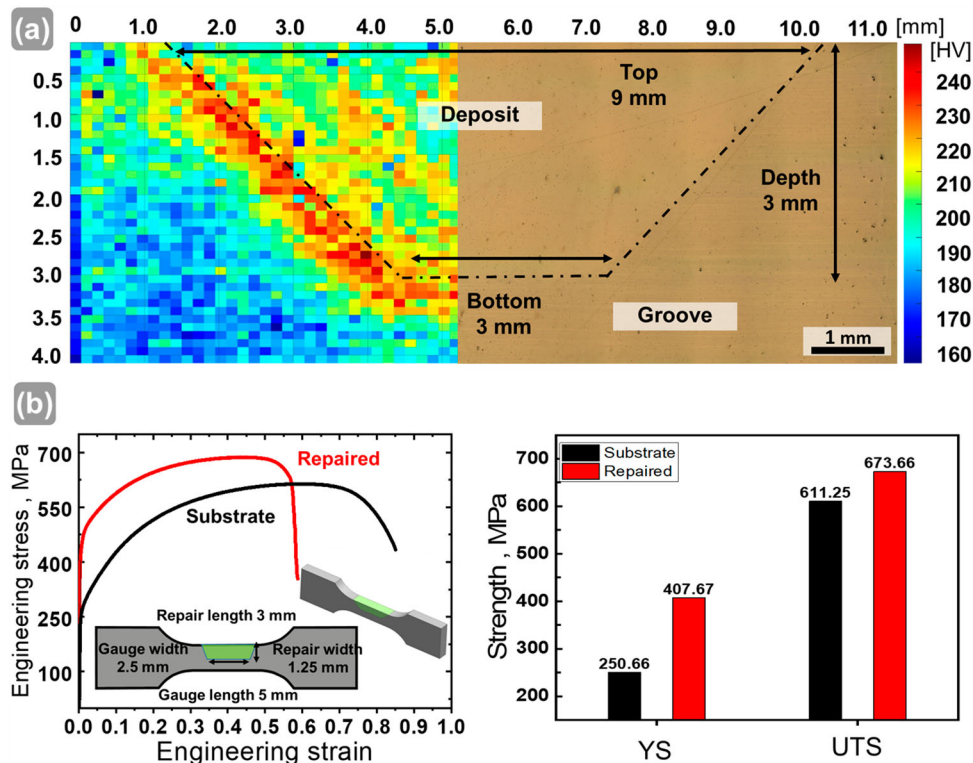


Figure 2. (a) OM image with hardness profile of the repaired sample showing the distribution of the deposit and the adjacent area, (b) Stress-strain curves of the substrate and repaired sample.

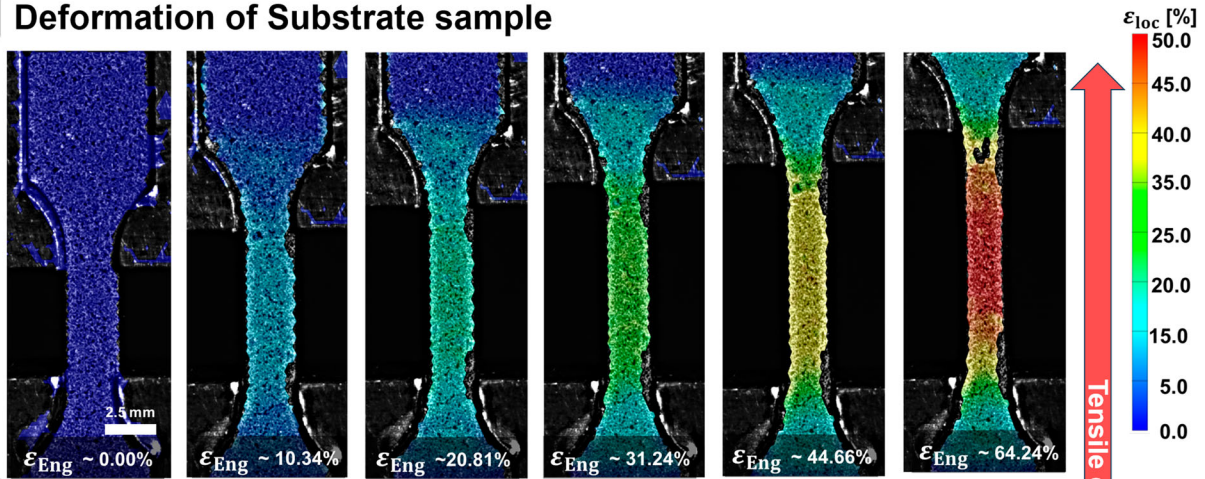
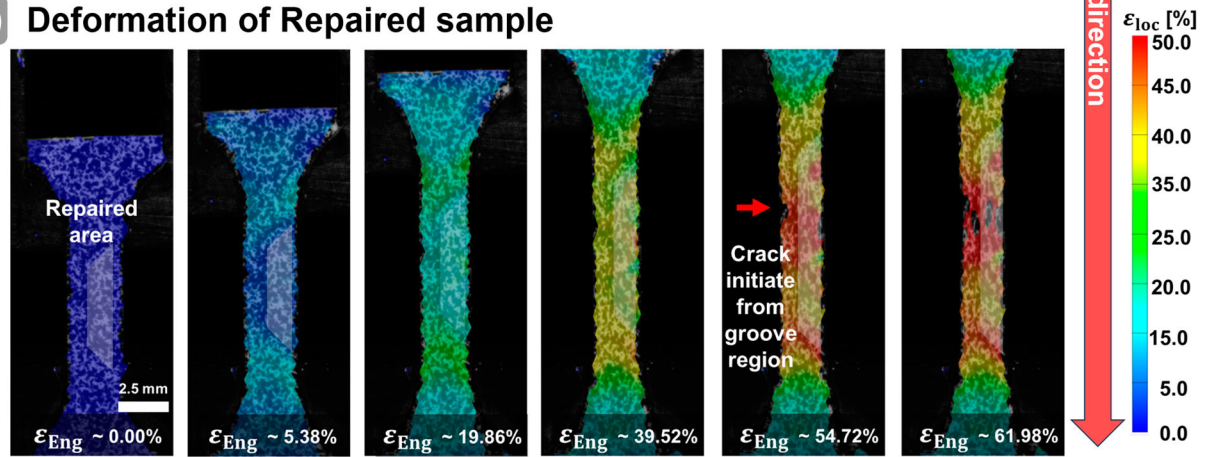
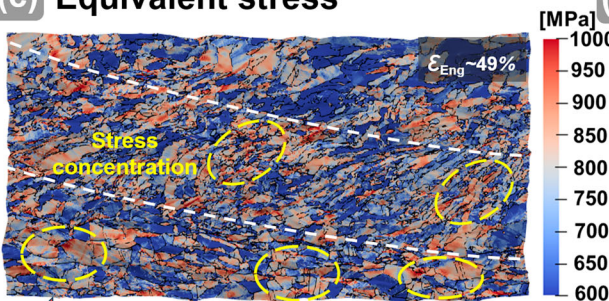
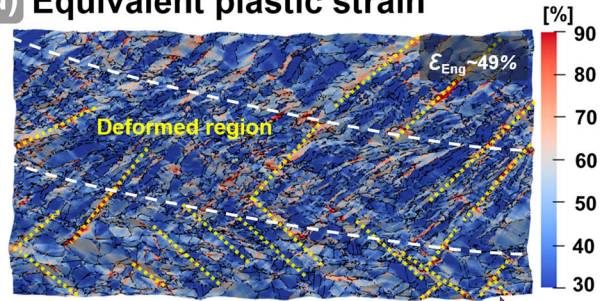
(a) Deformation of Substrate sample**(b) Deformation of Repaired sample****(c) Equivalent stress****(d) Equivalent plastic strain**

Figure 3. (a) Deformation behavior of the substrate sample exhibiting a uniform distribution, (b) that of the repaired sample showing strain partitioning (the white trapezoidal box indicating the deposit area, the red arrow presenting the crack initiation from the groove region), CPFEM results of the repaired sample (c) ES and (d) EPS at $\epsilon_{Eng} \sim 49\%$ indicating the concentration of stress and strain in the interface and groove.

and ultimate tensile strength (UTS), with a slight loss in ductility (total elongation, T.EL: $\sim 58.77\%$ in the repaired sample and $\sim 84.99\%$ in the substrate). Despite the slight loss in elongation, the repaired specimen exhibited an excellent combination of strength and ductility and achieved higher values in YS of ~ 408 MPa and UTS of ~ 674 MPa than those of the substrate (YS: 251 MPa, UTS: 611 MPa).

In Figure 3a, the plastic strain distribution in the substrate sample with a homogenous microstructure demonstrates a deformation behavior where strain is uniformly distributed as tensile testing progresses. Contrastingly, Figure 3b illustrates the deformation behavior of the repaired sample, including local deformation distribution. The trapezoidal area (highlighted in white, which indicates the deposit area) can be clearly recognized in

the sample, where the deformation is initially concentrated in the groove region, while the deformation in the deposit area is suppressed, leading to distinct strain partitioning. This occurs due to the presence of distinct microstructures in the repaired sample, composed of the deposit area, interface, and groove region. Interestingly, cracks initiate from the groove area at a local strain of $\sim 54.72\%$ for the repaired sample, finally leading to fracture. This can be attributed to the stress exceeding the UTS of the groove area (~ 611 MPa), extracted from the substrate.

The equivalent stress (ES) and equivalent plastic strain (EPS) within the microstructures were obtained at $\epsilon_{\text{Eng}} \sim 49\%$ by the crystal plasticity finite element simulation method (CPFEM) as shown in Figure 3c and d, respectively. Resulting from overlap with the stress concentration, the deformation region is mainly retained in the interface and groove. Therefore, in the terminal deformation stage of the repair sample, the damage occurs within the groove rather than from the interface, and cracks also begin to sprout, corresponding to Figure 3b ($\epsilon_{\text{Eng}} \sim 54.72\%$). A superior strength-ductility combination of

the repaired specimen can be explained by the HDI hardening effect. The repaired specimen consists of three parts: the hard domain for the deposit/interface area and the soft domain for the groove area (Figure 3a). Typically, in heterostructured alloys, geometrically necessary dislocations (GNDs) are produced to accommodate the incompatibility between hard and soft domains at the interface of these domains [27]. The accumulated GNDs serve as obstacles and effectively restrict the dislocation further movement, thereby delaying the yielding of the soft domain, the groove region.

To analyze deformation behavior, EBSD characterization was conducted at 25% and 50% of local strain areas, captured by the DIC technique. Figure 4a-d depicts the IPF map, the density of GND, the formation of phases, and twin boundary maps after 25% and 50% global deformations for both the substrate sample and the repaired sample.

In the substrate sample, as shown in Figure 4a and c, phase transformation from face-centered cubic (FCC) austenite to body-centered cubic (BCC) martensite phase during plastic deformation is observed, referred to

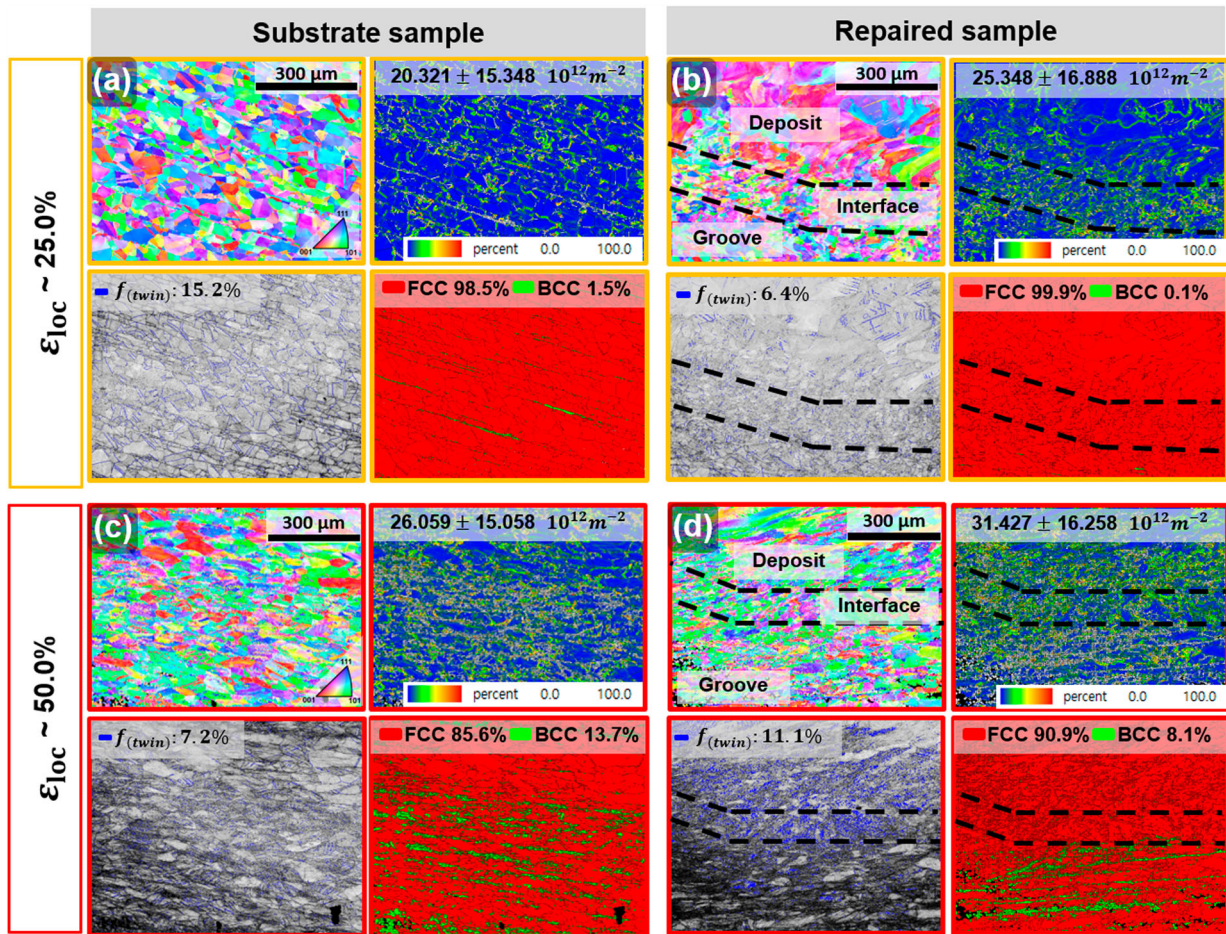


Figure 4. IPF and evolution of the GND density, twin activity, and phase fraction during tensile deformation at local strains of $\sim 25.0\%$ and $\sim 50.0\%$ in the substrate (a, c) and repaired samples (b, d).

as transformation-induced plasticity (TRIP). The XRD results after the tensile test (presented in Fig. S2 of the Supplementary material) confirm the formation of the BCC phase after the tensile test. Relatively abundant martensite appears after the tensile test with a fraction of BCC 13.7% while FCC 86.3%. This transformation is quite common in deformed SS316L, which raises work hardening ability and strength at highly deformed areas to postpone the occurrence of necking [28]. Phase transformation from FCC to BCC is predominantly observed in the substrate sample and partially occurs in the repaired sample in the groove region. Moreover, deformation twinning emerges during plastic deformation, yet the fraction of twins in the substrate decreases as the shear band area increases because twin boundaries act as initiation sites for BCC transformation along the path of shear bands. Therefore, the fraction of BCC phase increases by consuming the deformation twins [29, 30]. As a result, the evolution of shear bands with the interaction of twins during subsequent deformation leads to the occurrence of the TRIP phenomenon, which explains the ductile behavior of the substrate sample.

In the repaired sample, as presented in Figure 4b and d, similar tendencies can be observed in the groove region, which is a relatively soft domain in the repaired sample. Phase transformation from FCC to BCC with a fraction of $\sim 8.1\%$ occurs but is restricted to the soft groove area that shows the appearance of shear bands, while the hard deposit region shows prominent development of twin. The formation of shear bands in the deposit area is suppressed even at the same strain level because the deposit and interface area (hard domain) inhibit plastic deformation. Moreover, in the deposit area, a prominent development of deformation twins is observed ($f_{twin} \sim 11.1\%$) due to its relatively large grain size and dislocation network from the deposition process [31, 32]. These deformation twin boundaries in the deposit area restrain deformation effectively by acting as barriers to the dislocation movement [33]. Consequently, the deposit area exhibits increased strength rather than ductility owing to the enhanced twinning activity compared to the groove region and substrate area during tensile deformation, which is consistent with the results of tensile tests.

Despite generating high twinning activity during deformation, the repaired sample exhibits reduced ductility compared to the substrate sample, which is attributed to the high dislocation density of the deposit area. During the repair deposition, repetitive thermal energy generates an accumulation of dislocations, which subsequently leads to a decrease in the ductility of the repaired sample [34, 35]. As a result, the uniform elongation of the repaired sample ($\sim 44.14\%$) is moderately reduced compared to that of the substrate ($\sim 65.69\%$).

Both samples show accumulation of GND during tensile deformation. The GND densities of the substrate sample at 0%, 25%, and 50% are lower than those of the repaired sample. The repaired sample contains a trapezoidal heterostructure, which is more favorable for facilitating accumulation of GND. The initial GND density of the repaired sample (See Fig. S3) is low ($\sim 19.4 \times 10^{12} \text{ m}^{-2}$), but as the sample deforms during tensile straining, GNDs significantly accumulate and exhibit high values at 25% deformed stage ($\sim 25.3 \times 10^{12} \text{ m}^{-2}$) and 50% deformed stage ($\sim 31.4 \times 10^{12} \text{ m}^{-2}$). The accumulated GNDs obstruct the movement of dislocation and generate HDI stress, which produces non-uniform hardness and plastic deformation, leading to an extra hardening effect. As aforementioned, microstructural features are crucially correlated with the high cooling rate that the material undergoes during the DED repair, leading to the high strength-ductility combination.

To estimate the HDI strengthening effect on tensile flow stress, the loading-unloading-reloading (LUR) test was conducted with the tensile specimen of each sample. The result of the LUR test was plotted in Figure 5a. The HDI stress (σ_{HDI}) was calculated from LUR curves according to the established procedure [36]. The σ_{HDI} value of the repaired sample is higher than that of the substrate during the overall LUR test, demonstrating that the hard domain of interface/repaired (compared to groove area) effectively facilitates the pile-up of dislocations leading to a high HDI stress contribution to flow stress that is greater than in the substrate sample (Figure 5b). In Figure 5c, the ratio of the HDI stress to the flow stress ($\sigma_{HDI}/\sigma_{flow}$) reveals that the HDI contribution in the repaired specimen ($\sim 58\%$) which is higher than in the substrate ($\sim 49\%$). Due to the heterogeneous grain size and high density of dislocations, higher HDI hardening is achieved.

Figure 5d schematically illustrates the strengthening mechanism in the repaired sample. During deformation, the plastic incompatibility between the hard and soft areas in the repaired specimen escalates, since the hard-deposit area becomes harder from the profusely formed deformation twins, which leads to a moderate loss of elongation with proper work hardening behavior. As a result, the repaired sample shows an enhanced strength-ductility synergy compared to previously researched DED-manufactured SS316L [9, 37–42] and conventionally manufactured (CM) SS316L [43–51], as displayed in Figure 5e and f. The repaired sample had excellent performance in the combination of ductility and strength, which is closely related to the toughness of the material. However, the bonding strength of the interface remains uncertain regarding the geometry of the groove, which will be further investigated.

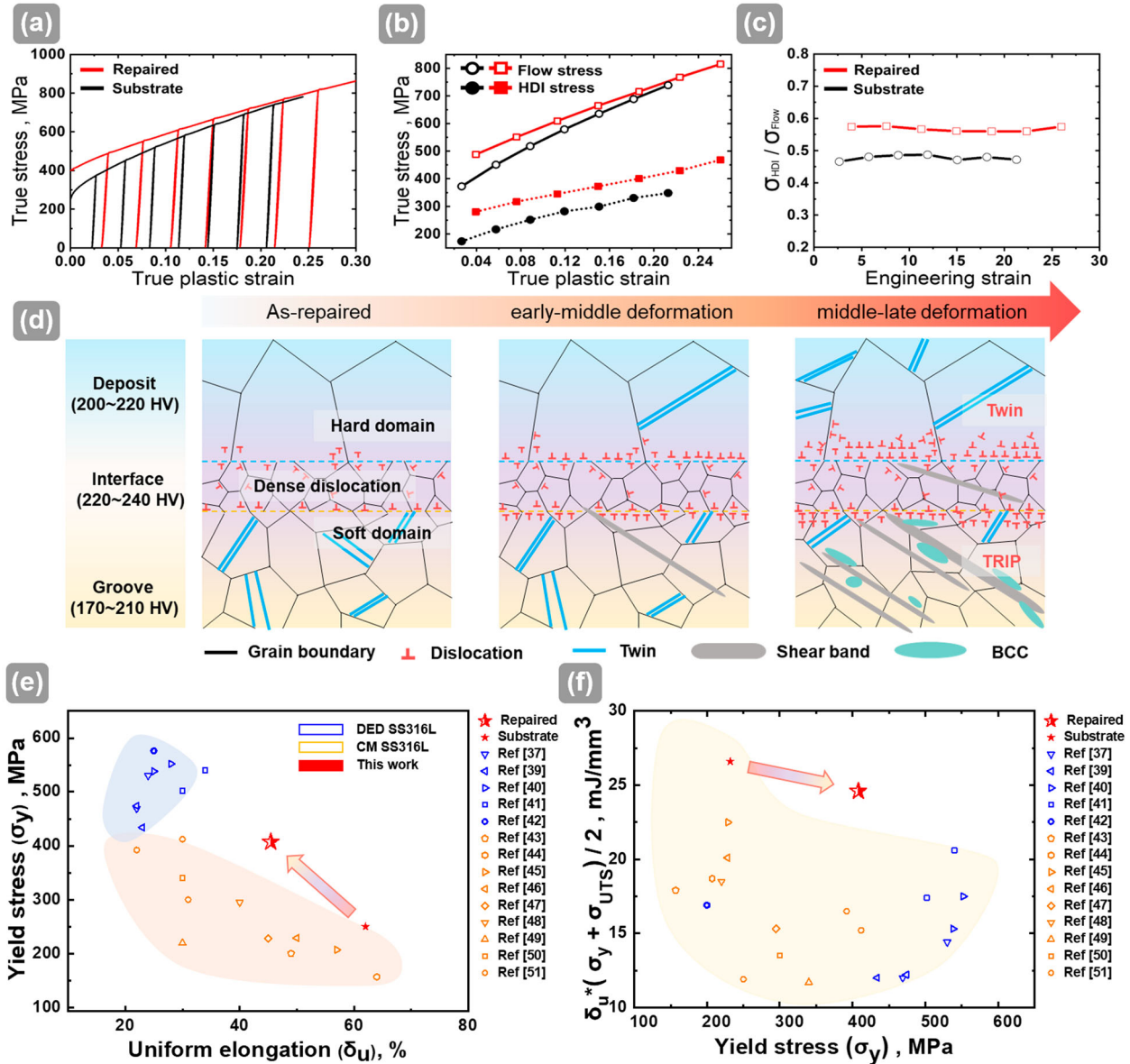


Figure 5. (a) LUR curves, (b) Evolution of HDI stress with increasing true plastic strain, (c) Contribution of HDI stress to the flow stress during the LUR test, and (d) Schematic of the strengthening mechanism in the repaired sample. (e) Ashby plot of the repaired-SS316L with CM-processed and DED-manufactured SS316L, showing a superior strength-ductility synergy and (f) excellent toughness with a high yield strength of the repaired sample.

4. Conclusions

In summary, the heterogeneity of the microstructure in SS316L through repair deposition results in a superior combination of ductility and strength (T.EL: 58%, YS: 407 MPa, UTS: 673 MPa), compared to the base material. The enhanced strength is explained by the high density of GND provoked by the incompatibility of deformation between the base material and the repair deposit. The deposit area (hard domain) obstructs the yielding of the groove area (soft domain), increasing the overall yielding point. Moreover, during tensile deformation, the

twinning activity in the deposit area enforces mechanical incompatibility. These factors contribute to the promotion of GND piling up profusely at the interface, resulting in an excellent combination of strength and ductility in the repaired sample (Figure 5f). Based on the findings of this investigation, repair deposition could be applied in various industries, such as power plants and aerospace, to achieve mechanical properties superior to those of the base material as well as shape repair. Further research could investigate the long-term service performance of repaired components.

CRedit authorship contribution statement

Man Jae SaGong: Investigation, Conceptualization, Data Curation, Visualization, Validation, Writing—Original Draft, Formal analysis. **Rae Eon Kim:** Investigation, Data curation, Visualization. **Jeong Ah Lee:** Investigation, Data curation, Visualization. **Hyojin Park:** Investigation, Data curation, Visualization. **Soung Yeoul Ahn:** Investigation, Data curation. **Shuai Xu:** Data curation. **Haiming Zhang:** Data curation. **Renhao Wu:** Funding, Supervision, Validation, Writing—Review & Editing. **Hyoungeop Kim:** Project administration, Funding, acquisition, Conceptualization, Supervision, Writing—Review & Editing.

Disclosure statement

No potential conflict of interest was reported by the author(s).

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