



Short Communication

Insights into crack prevention and property improvement for additively manufactured ultra-high-strength steel structures with complex geometries

Jun Wang^{a,*}, Mark Taylor^b, Chenglei Diao^a, Ed Pickering^b, Jian Qin^a, Yao Lu^a,
Sonia Martins Meco^a, Jialuo Ding^a, Stewart Williams^a

^a Welding and Additive Manufacturing Centre, Cranfield University, Cranfield MK43 0AL, UK

^b Department of Materials / Henry Royce Institute, The University of Manchester, Manchester M13 9PL, UK



ARTICLE INFO

Keywords:

300 M ultra-high-strength steel
Wire-based direct energy deposition
Post heat treatment
Cracking prevention
material performance

ABSTRACT

Hybrid wire-arc directed energy deposition (WDED), in which complex features are deposited onto a forged base, offers a cost-effective solution for manufacturing geometrically complex ultra-high-strength steel components, particularly for aerospace applications. However, cracking at the base forging/build interface during post-build heat treatment limits its widespread application. This study investigates the underlying causes of interfacial cracking, highlighting microstructural inhomogeneity, elemental segregation and transformation stresses as likely key contributing factors. A modified three-step post-build heat treatment incorporating a normalisation step was developed to mitigate some of these issues. The optimised process successfully suppressed cracking by refining prior-austenite grains before the application of a conventional quenching step. This enhanced tensile performance beyond AMS6419K standards, supporting the industrial implementation of hybrid WDED in aerospace structures.

1. Introduction

300 M ultra-high-strength steel (UHSS) is known for its excellent combination of high tensile strength, fracture toughness, and fatigue resistance [1]. These properties make it ideal for critical load-bearing structures such as aircraft landing gear, which must perform reliably under dynamic loading and harsh environments [2]. As the aerospace industry moves towards stronger, lighter, and more cost-effective structural solutions, the demand for optimised manufacturing processes for 300 M continues to grow [3]. Additive manufacturing (AM) demonstrates significant advantages over traditional forging and casting, offering improved material efficiency, geometric flexibility, and reduced lead times. In particular, AM enables near-net-shape fabrication of complex features, minimising machining waste, especially beneficial in landing gear production [4]. Among AM technologies, wire-arc directed energy deposition (WDED) is especially attractive for large-scale structural steel components, owing to its high deposition rate, cost-effectiveness, and promising mechanical performance [5].

Recent studies have demonstrated the feasibility of manufacturing 300 M UHSS using WDED, achieving promising mechanical properties through optimised processing and post-build heat treatment [6–8].

Wang et al [6] investigated gas metal arc WDED of 300 M UHSS and identified that using a shielding gas of 98 % Ar + 2 % O₂ provides optimal process stability, deposition quality, and mechanical performance. Xiong et al [7] demonstrates that interlayer temperature significantly influences the microstructure evolution and mechanical performance of cold metal transfer (CMT) WDED 300 M UHSS, highlighting its critical role in tailoring heat accumulation and phase transformation during deposition. Xiong et al [8] also demonstrates that an optimised two-step heat treatment with austenitizing at 970 °C significantly refines the microstructure and enhances the mechanical properties of CMT WDED 300 M UHSS, achieving superior strength and toughness through carbon supersaturation and fine ε-carbide precipitation. However, the use of fully-AM'd components in aerospace remains limited by certification and cost challenges. As a practical alternative, hybrid manufacturing, in which complex features are deposited onto a forged base, combines the structural reliability of forged 300 M with the flexibility of WDED. This hybrid strategy is particularly effective for features such as lugs and flanges, aiming to reduce material removal, improve sustainability, and enhance design integration. A key challenge in hybrid WDED is the fusion interface between the base forging and the built material. This zone often exhibits

* Corresponding author.

E-mail address: jun.wang.123@cranfield.ac.uk (J. Wang).

<https://doi.org/10.1016/j.addlet.2025.100307>

Received 14 June 2025; Received in revised form 14 July 2025; Accepted 19 July 2025

Available online 20 July 2025

2772-3690/© 2025 The Authors. Published by Elsevier B.V. This is an open access article under the CC BY license (<http://creativecommons.org/licenses/by/4.0/>).

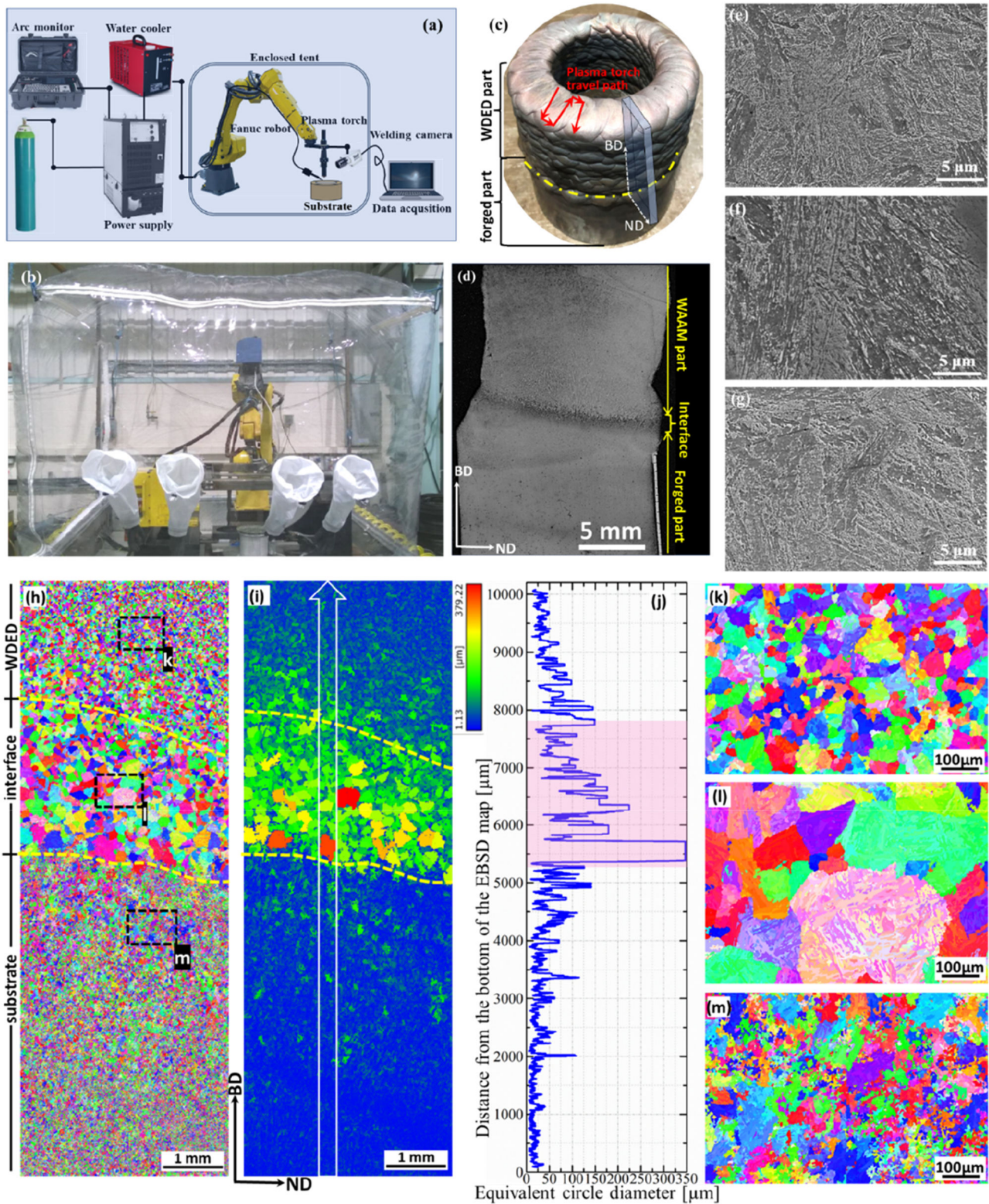


Fig. 1. (a) Schematic diagram for PTA-WDED experimental setup; (b) overview of the setup showing the global shielding condition; (c) view of 300 M hollow cylinder deposit; (d) macro morphology taken from the BD-ND plane; (e, f, g) typical microstructure taken from the WDED part, interface, and forged part, respectively; (h) large-area reconstructed prior austenite grain (PAG) EBSD orientation map; (i) corresponding PAG size distribution map; (j) variation of reconstructed PAG equivalent circle diameter; (k, l, m) magnified reconstructed PAG EBSD maps taken from the WDED part, interface, and forged part, respectively.

grain coarsening, microstructural heterogeneity, residual stress accumulation, and elemental segregation, making it prone to metallurgical defects during deposition or post-build heat treatment. Eimer et al [9] investigated the mechanical performance of the substrate–deposit

interface in WDED aluminium alloy and identified it as a potential weak zone due to hardness mismatch, limited metallurgical bonding, and crack susceptibility, particularly when the substrate and wire alloys were not well matched. Their study further demonstrated that inter-pass

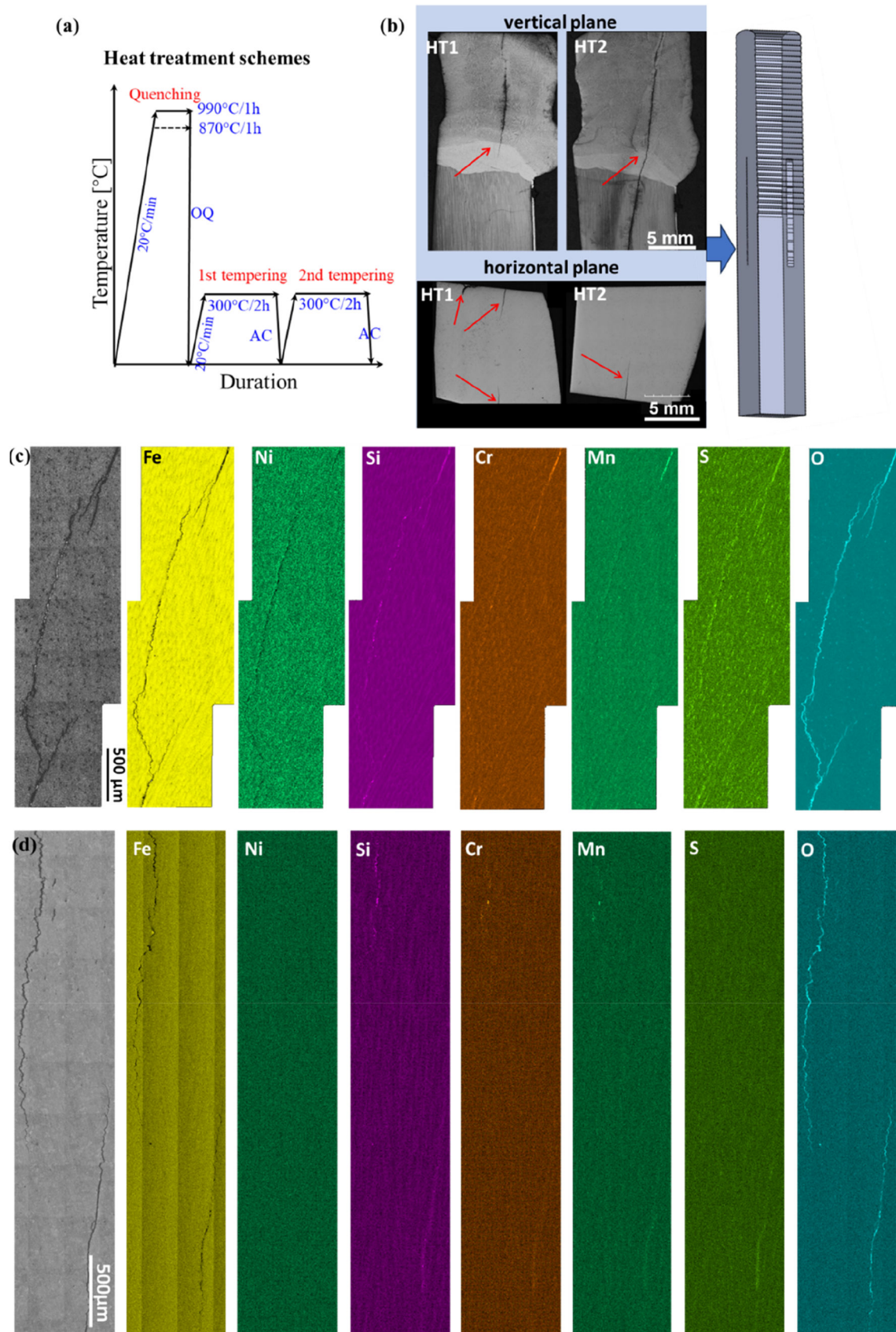


Fig. 2. (a) Post heat treatment schemes applied in WDED 300 M samples; (b) macro-morphology taken from vertical and horizontal plane showing the cracks locations and propagation; (c, d) Band contrast maps with corresponding EDS mappings showing cracks and their surroundings taken from HT1 and HT2 samples, respectively.

rolling and post-deposition heat treatment can significantly improve microhardness uniformity and eliminate interface defects. Zhang et al [10] revealed that fatigue cracks initiated at the interface between wrought and WDED Ti6Al4V tend to propagate into the wrought substrate, driven by differences in microstructure and residual stress, highlighting the interface as a mechanically weaker zone. In UHSS like 300 M, such vulnerabilities in interface are especially detrimental due to their limited tolerance to microstructural discontinuities.

This work aims to address these concerns by investigating interface behaviour and crack formation mechanisms during post-build heat treatment of a hollow cylinder 300 M feature deposited by WDED. The hollow cylindrical components and interfaces investigated in this work represent typical features of complex-shaped load-bearing structures in practical engineering applications. Such geometries inherently involve curved surfaces, internal cavities, and substrate–deposit interfaces, which introduce additional challenges in both fabrication and performance. Therefore, insights gained from these non-planar and interface-rich structures are directly relevant to the manufacturing and optimisation of ultra-high-strength steel components with complex geometries. The findings offer practical guidance for the industrial implementation of hybrid AM in aerospace-grade steel components and contribute to establishing a reliable path toward the certification of hybrid additively manufactured aircraft landing gear.

2. Feature deposition and characterisation

300 M UHSS wire with a diameter of 1.2 mm was used as the feed-stock, and a forged hollow 300 M rod (inner diameter: 58 mm; outer diameter: 88 mm; height: 55 mm) served as the substrate; both were supplied by Airbus Landing Gear Group. The chemical composition was measured by ICP-OES analysis, with the following values in wt. %: C – 0.43, Si – 1.67, Mn – 0.77, P – <0.005, S – 0.004, Cr – 0.85, Mo – 0.40, Ni – 1.81, Cu – <0.01, V – 0.07, and balance Fe. Figs. 1(a, b) illustrate the experimental setup for the plasma-transferred arc (PTA) WDED process. An EWM (Alpha Q552) power source with a wire feeder (Phoenix Drive 4) provided the arc and wire feed. The substrate was clamped to a workbench, and the torch motion was controlled by a 6-axis FANUC robot. The setup was placed in an enclosure filled with pure argon, with oxygen maintained below 800 ppm. A novel oscillation strategy was adopted to compensate for the ± 20 % path density difference between inner and outer diameters, ensuring minimal layer height variation. The wire feed rate, torch travel speed, current, plasma gas flow, and inter-pass temperature were set at 2.35 m/min, 3.3 mm/s, 220 A, 0.8 L/min, and 150 °C, respectively, for stable metal transfer. The deposited ring (Fig. 1(c)) exhibited smooth, consistent layers along the build direction (BD), with a clear transition to the forged substrate (Fig. 1(d)). One cross-section sample was extracted from the normal-building (ND-BD) plane for interface study at both macro- and micro-scales. The specimens were polished and subsequently etched using a 2 % Nital solution for microstructure analysis using a scanning electron microscope (SEM, TESCAN VEGA 3) at 15 kV with a working distance of 10 mm. Electron backscatter diffraction (EBSD) mapping with simultaneous energy-dispersive X-ray spectroscopy (EDS) was conducted using a Thermo Scientific Apreo SEM with an Oxford Symmetry-2 detector at 20 kV, 51 nA, 70 ° tilt, a step size of 1 μ m for large-area mapping and 0.3 μ m for small-area mapping. Prior austenite grains (PAGs) were reconstructed using the Nishiyama-Wasserman orientation relationship, with non-indexed pixels filled via Oxford Instruments' AZtecCrystal software (Version 3.1).

The as-deposit microstructure featured three distinct regions: WDED deposit, interface, and forged substrate, which are displayed in Fig. 1(e), (f), (g), respectively. A lath martensitic structure was observed in the WDED deposit, formed due to the high cooling rate from high temperature immediately following deposition, which favoured the formation of martensite directly from austenite, which is commonly observed in AM UHSS [11]. The microstructure in the interface also appeared

martensitic for similar reasons, with indication of some carbides from tempering by the heat from the WDED thermal cycles. The microstructure in the forged substrate away from the interface appears to be refined tempered martensite, which may also have been tempered somewhat by the thermal cycling from the WDED layers.

A large-area EBSD scan was performed in the BD-ND plane, encompassing the WDED region, the interface, and the forged substrate. As indicated by the reconstructed PAGs orientation map (Fig. 1(h)) and magnified PAG maps (Fig. 1(k, l, m)), the microstructure throughout the scanned region is composed of equiaxed PAGs, even in the WDED region. This is primarily due to the accumulated thermal cycles caused by the deposition of successive layers, which break up coarse as-solidified columnar PAGs through repeated re-austenitisation [6]. However, the grain size varies significantly across different regions. The corresponding grain size map (Fig. 1(i)) and the variation of equivalent circle diameter (ECD) with vertical distance (Fig. 1(j)) further quantify this gradient. The PAGs in the WDED region are relatively fine (~50–150 μ m) and uniformly distributed. The forged region exhibits even finer PAGs (~50–100 μ m), resulting from the conventional processing of the material. Interestingly, the interface zone corresponds to the coarsest and most heterogeneous PAG size distribution, with ECD values exceeding 250 μ m. The coarse PAGs in this region are likely a result of the nature of the thermal cycling experienced by in the lower region of the interface (i.e., base material at the top of the forging). This material will have experienced a high-temperature re-austenitisation in the HAZ of the first deposited WDED pass, leading to the formation of coarse PAGs. Subsequent passes can refine these coarse-grain structures by repeated re-austenitisation in their overlapping HAZs. However, the material in the HAZ of the first deposited layers will not have been subject to the same level heat and refinement from subsequent passes – this material will have experience fewer, lower-temperature cycles than regions above. Hence, there is a sharp transition from the fine-grained forged material to coarse grains at the bottom of the interface, as would be seen in the HAZ of a typical single-pass weld. The PAGs size then refines progressively up the BD owing to increased repeated austenitisation caused by thermal cycling from overlapping HAZs [12]. Importantly, despite this gradient in grain size, the EBSD maps reveal continuous grain boundary networks and no evidence of microstructural discontinuities or defects (e.g., cracks) in the as-deposit condition.

3. Two-step heat treatment and cracking analysis

Post-build heat treatment is essential for relieving residual stresses, homogenising microstructure, and achieving the desired balance of strength, toughness, and dimensional stability in WDED 300 M builds. Conventional heat treatment of wrought or cast 300 M steel involves austenitizing at 870 °C for 1 h followed by oil quenching, and double tempering at 300 °C for 2 h each, with air cooling after each cycle [13]. The tempering relieves quenching stress, causes carbide precipitation and forms new martensite from retained austenite decomposition [14]. However, due to the microstructural anisotropy and non-uniform thermal history in AM processed materials, a higher austenitizing temperature may be required to ensure complete austenitisation and uniform transformation across the deposited regions. Therefore, two different heat treatment schemes (HT1: 870 °C/1h/oil quench (OQ) + 300 °C/2h/air cooling (AC)/twice; HT2: 990 °C/1h/OQ + 300 °C/2h/AC/twice) were designed and applied in this study, shown in Fig. 2(a). However, cracks are visible in samples on both the vertical and horizontal sections subjected to these post-build heat treatments, shown in Fig. 2(b). In vertical plane, it can be seen that cracks are predominantly observed to initiate at or near the fusion interface and propagate along the build direction. This region is known to experience the most complex thermal history during the WDED process, including partial reheating, high residual stresses, and grain coarsening [15,16]. In the horizontal cross-sections, crack initiation was predominantly observed at the sample surface and extended inward, attributed to the steep

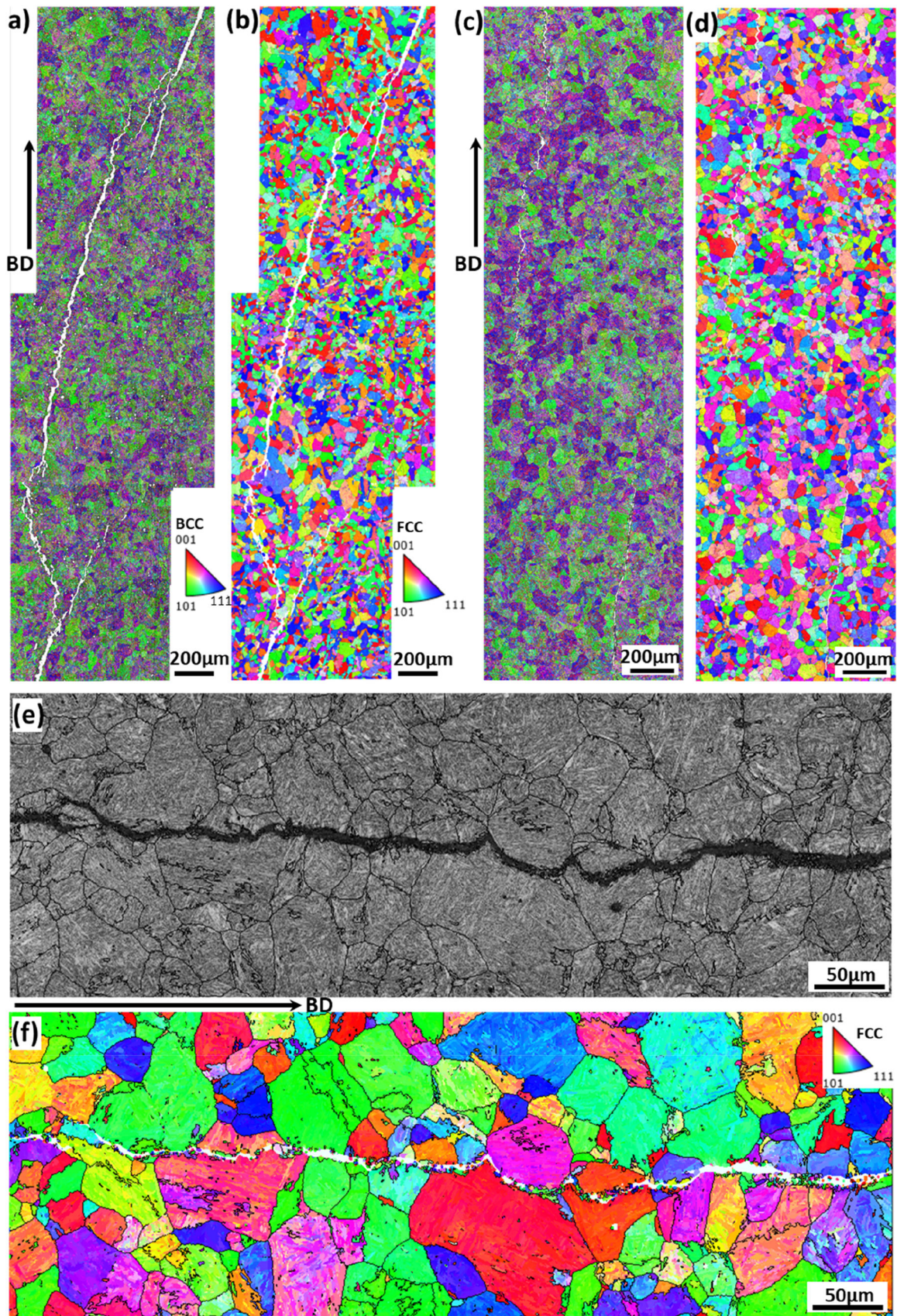


Fig. 3. EBSD orientation maps and corresponding reconstructed PAGs orientation maps: (a,b) taken from region depicted in Fig. 2(c), (c, d) taken from region depicted in Fig. 2(d); (e, f) enlarged band contrast map overlaid with reconstructed PAG boundaries, and corresponding reconstructed PAGs orientation map showing the cracking propagation path and surrounding PAGs.

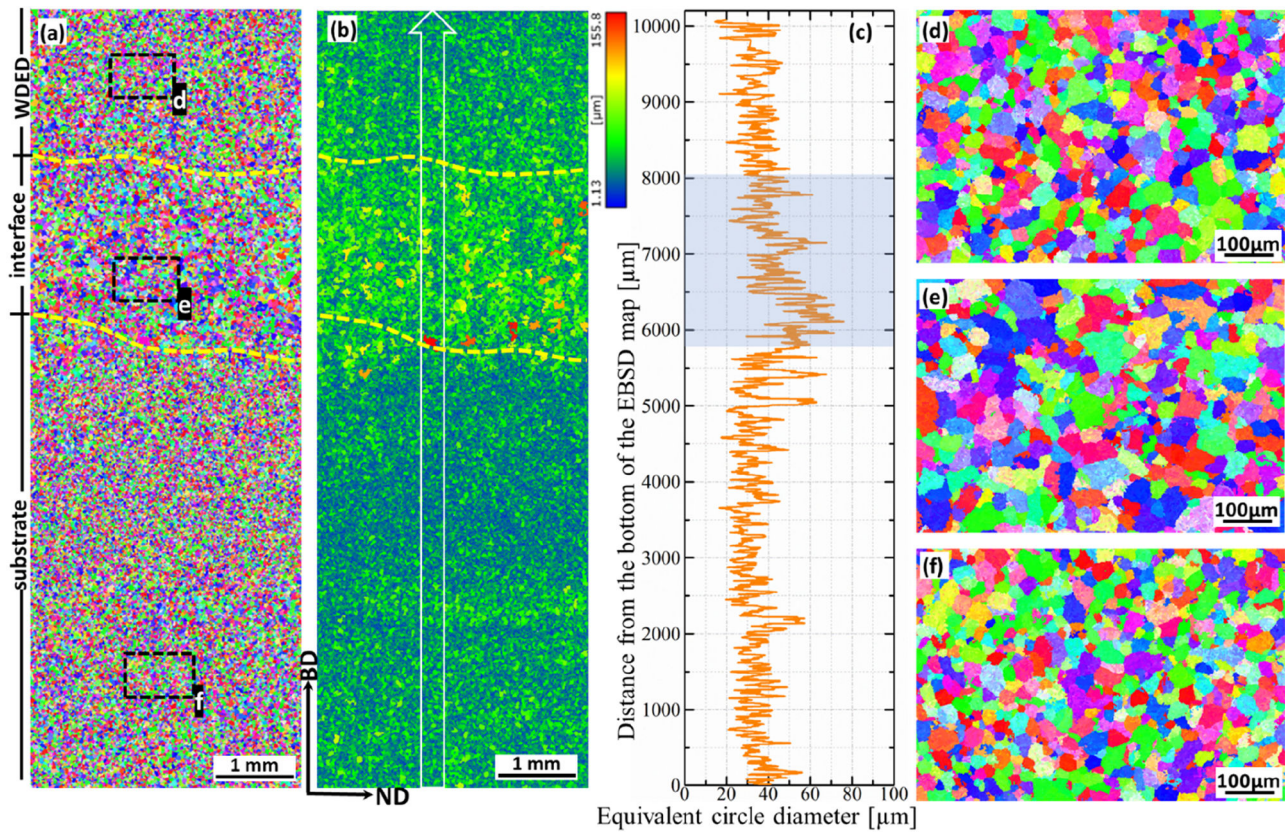


Fig. 4. (a) Large-area reconstructed PAG EBSD orientation map of 300 M subject to HT3 across the interface; (b) corresponding PAG size distribution map; (c) variation of reconstructed PAG equivalent circle diameter across the interface; (d, e, f) magnified reconstructed PAG EBSD maps taken from the WDED part, interface, and forged part after HT3, respectively.

thermal gradients and associated quench stresses concentrated at the surface region. Such factors contribute to localised stress concentration and microstructural mismatch, making this region highly susceptible to quench cracking.

EDS elemental mappings around the observed cracks (Fig. 2(c, d)) revealed pronounced elemental microsegregation in the heat-treated WDED 300 M samples. Despite undergoing full quenching and tempering cycles, local enrichment of alloying elements such as Ni, Cr, Mn, and Si remains evident. Although the EDS shows enrichment in S, this may be due to peak overlap with Mo. However, some S-rich particles are also enriched in Mn, suggesting the possible presence of non-metallic inclusions, which may act as crack initiation sites during heat treatment [17]. The crack paths are also accompanied by a slight enrichment of S and O, however the oxygen signal in particular may have resulted from surface topography or possibly material trapped within the crack after fracture during sample preparation. More importantly, the crack paths consistently coincide with these segregated zones, suggesting that elemental segregation significantly weakens grain boundary cohesion and acts as a preferred path for crack initiation and propagation. Such segregation likely originates from the non-equilibrium solidification during deposition and is not fully homogenised by the in-built thermal cycling nor the post-heat treatment. These findings reveal that, beyond thermal stresses, chemical inhomogeneity is likely to play a role in post-build treatment cracking behaviour in WDED 300 M components, particularly at the interface.

Fig. 3(a–b) and (c–d) present the BCC-martensite EBSD orientation and reconstructed PAG maps taken from the crack-containing regions previously identified in Figs. 3(c) and (d). Following heat treatment, the PAGs appear largely homogenised, with an average equivalent circle diameter of approximately 45 μm, and exhibit random orientation. Despite this apparent uniformity, cracks are observed to propagate along

the build direction (BD), with a clear tendency to follow PAG boundaries (PAGBs), as shown in the magnified maps in Figs. 3(e) and (f). This behaviour aligns with previous reports on hydrogen-induced cracking [18], where PAGBs are known to serve as preferential crack paths due to their structural and chemical vulnerabilities. Although hydrogen embrittlement is not the dominant mechanism in this study, the underlying principles remain relevant: PAGBs consist of high-angle martensitic sub-boundaries that inherently possess elevated grain boundary energy and are prone to impurity segregation, particularly of sulphur (S). There is some indication of higher S content in Fig. 2 from segregation of more likely sulphide inclusions as previously mentioned. Segregated S may reduce boundary cohesion alternatively sulphide/-oxide inclusions may act as preferential sites for cracking, especially during austenitization and quenching [19,20]. These inclusions may further embrittle PAGBs by acting as local stress concentrators or nucleation sites for microcracks [21]. The observed crack morphology is predominantly intergranular or quasi-cleavage, with cracks favouring propagation along PAGBs over intrapacket martensitic boundaries. This behaviour is further promoted in regions containing coarse PAGs, which offer fewer obstacles to crack deflection and serve as zones of local stress concentration.

Thermal gradients and volume changes due to rapid martensitic transformation during quenching can generate substantial internal stresses that are well known to drive cracking, particularly at microstructurally heterogeneous regions such as the AM-deposited interfaces and fusion boundaries. These stresses arise from a combination of thermal contraction mismatch, transformation-induced volume expansion, and constrained deformation across dissimilar microstructures, which collectively produce high local tensile stresses. In combination with brittle, untempered martensitic structures, quench cracking can be commonplace even in conventionally manufactured high-strength steel

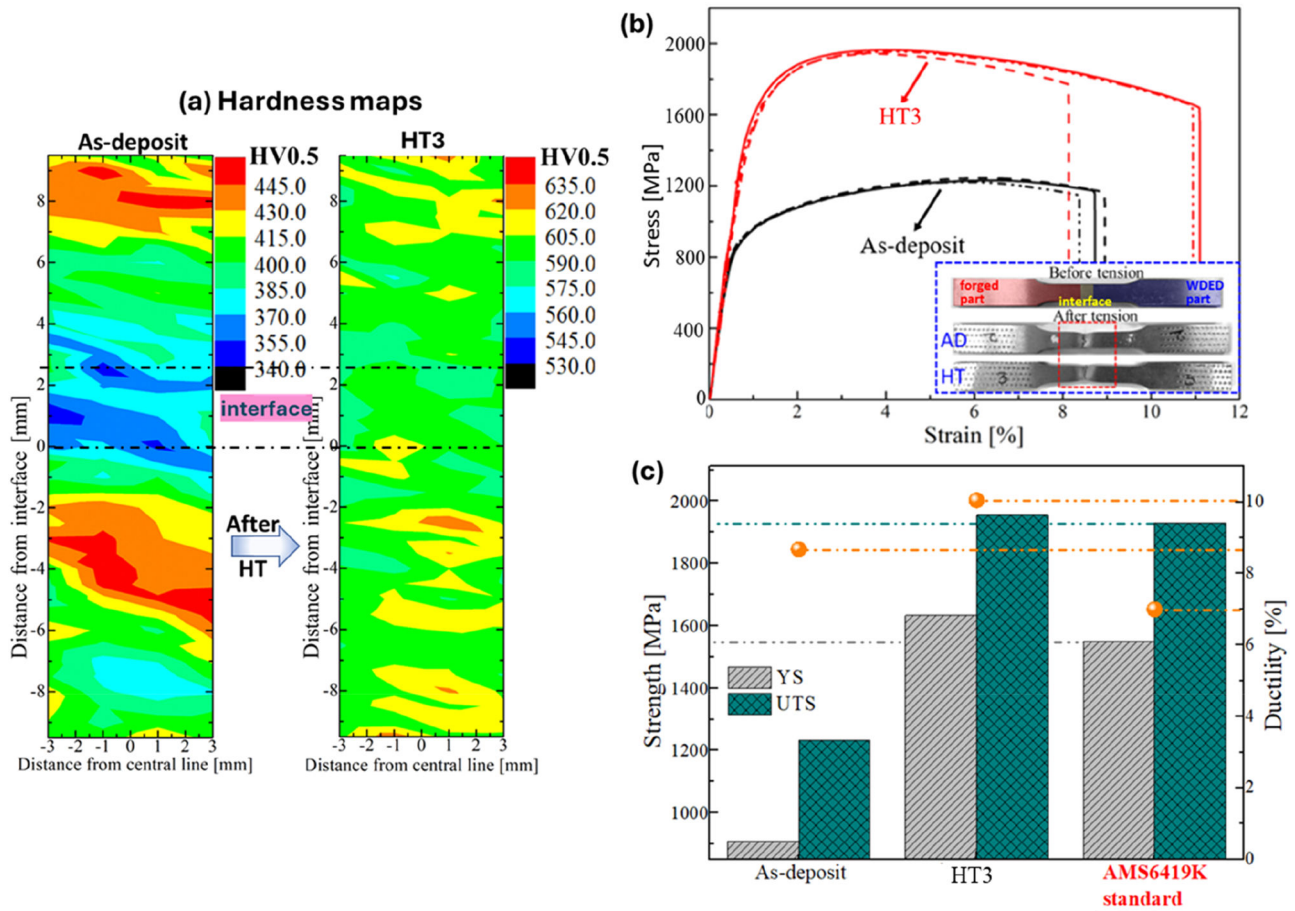


Fig. 5. Mechanical performance comparison in as-deposit and HT3 conditions: (a) Hardness maps across the interface; (b) tensile strain-stress curves and (c) corresponding strength and elongation values.

parts. This risk is further exacerbated in AM builds due to residual stress accumulation from layer-wise deposition and the presence of segregation or prior austenite grain boundary networks acting as stress concentrators [22,23]. Therefore, addressing quench-induced cracking in AM-wrought hybrid structures requires careful control of both thermal history and microstructural evolution during post-build heat treatments.

4. Cracking prevention and property improvement via modified heat treatment

To address the suspected quench cracking observed after the above HT schemes, a modified heat treatment procedure was developed by introducing a normalisation step at 950 °C for 1 h, followed by furnace cooling, prior to the quenching (870–990 °C) and double tempering cycles. This additional normalisation step, in particular the slow furnace cool it comprised, aimed to reduce the magnitude of the thermal gradients and internal stresses generated by the martensitic transformation [24], whilst at the same time refining the PAG structure as any austenitisation would be expected to do. This heat treatment is referred to as HT3.

Fig. 4(a) presents the reconstructed PAGs orientation map of a representative WDED-interface-substrate region following the application of HT3 post build. The PAGs are generally refined and equiaxed throughout the section, with no visible coarse or elongated grains near the interface. The PAG size map (Fig. 4(b)) and corresponding grain size profile (Fig. 4(c)) reveal a slight grain coarsening in the interface region, but the difference is marginal and significantly reduced compared to the as-deposited conditions. This refinement in comparison to HT1 or HT2 was attributed to the additional cycling between austenite and

martensite during the normalisation. It is likely that the slower cooling rate from the normalisation step prevented quench cracking in the first step, while the refined PAG structure further prevented cracking during or after oil quenching from the austenitisation (second) step. The magnified PAGs orientation maps (Fig. 4(d–f)) confirm that no abnormal grain growth or alignment is present, and the PAGs remain randomly oriented and well-distributed. Most importantly, no cracks or intergranular defects were observed in any of the inspected regions, suggesting that HT3 successfully suppressed the crack-driving mechanisms previously identified, and can therefore be considered as an improved heat treatment. These findings demonstrate that a post-build heat treatment including an initial normalization step is an effective strategy for preventing cracking in WAAM-fabricated 300 M steel components, particularly in hybrid AM/forged components.

The mechanical properties in both as-deposit and HT3 conditions were then evaluated to assess the effectiveness of the modified HT scheme. The microhardness transitions were measured using a Zwick/Roell Hardness Tester with a load of 500 g and spacings of 0.5 mm in vertical and 3 mm in horizontal. Flat tensile specimens with a gauge length of 24 mm to effectively cover the interface region, following ASTM E8 standard, were tested on an Instron 5500r machine at room temperature with a constant strain rate of 1 mm/min. Three sets of tensile coupons were taken from each condition to ensure data reliability. The hardness map shown in Fig. 5(a) across the WDED-interface-forged substrate region reveals significant improvement in terms of homogeneity after HT3 heat treatment. In the as-deposited sample, the hardness exhibits strong heterogeneity, ranging from ~340 HV0.5 in the interface zone to over 440 HV0.5 in the WDED and forged region, reflecting microstructural inhomogeneity. Notably, the interface zone

appears as a localised soft region, consistent with previous observations of coarser grain structures. After the modified heat treatment (HT3) the hardness becomes much more uniform across the interface, with values ranging between ~530–635 HV0.5. This confirms that the new heat treatment including normalising promotes microstructural homogenization and mechanical consistency. The tensile performance (Fig. 5(b, c)) shows a dramatic improvement after the improved heat treatment. Compared to the as-deposit condition, which exhibits a yield strength (YS) of ~1100 MPa, ultimate tensile strength (UTS) of ~1450 MPa, and low ductility (~4.5 % strain), the HT3 heat-treated sample demonstrates a significantly increased YS of ~1500 MPa, a UTS exceeding 1900 MPa, and an elongation exceeding 9 %. Therefore, after the modified heat treatment (HT3), it not only meets but exceeds the strength and ductility targets set by the AMS6419K aerospace standard.

5. Conclusions

This study provides new insights into cracking prevention and property improvement in hybrid WDED 300 M steel components, with conclusions shown below.

- Conventional (austenitisation + quenching + tempering) heat treatments led to interfacial cracking, likely due to the interplay between quench cracking stresses, coarse PAGs and residual microsegregation. Cracks were found to propagate along PAGBs and were seen to interact somewhat with the microsegregation.
- A modified heat treatment incorporating a first normalisation step was introduced. After such modified heat treatment to address these issues, cracking is eliminated, hardness uniformity is improved, and tensile properties exceed AMS6419K standards. This success can be attributed to the slower cooling rate in the normalisation step preventing quench cracking during cooling, and the high-temperature hold during normalization refining the PAG structure sufficiently to further prevented cracking during or after oil quenching in the subsequent conventional austenitisation + quenching steps.
- These findings demonstrate that tailored heat treatment strategies are essential for post-build processing of hybrid AM components with complex geometry to meet the standard requirement, offering a reliable route towards the qualification and adoption of WDED 300 M in aerospace applications.

CRedit authorship contribution statement

Jun Wang: Writing – review & editing, Writing – original draft, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **Mark Taylor:** Writing – review & editing, Methodology, Investigation, Formal analysis, Data curation. **Chenglei Diao:** Writing – review & editing, Investigation, Data curation. **Ed Pickering:** Writing – review & editing, Resources, Methodology, Funding acquisition. **Jian Qin:** Writing – review & editing, Validation, Investigation, Data curation. **Yao Lu:** Writing – review & editing, Methodology, Formal analysis, Data curation. **Sonia Martins Meco:** Writing – review & editing, Project administration. **Jialuo Ding:** Writing – review & editing, Supervision, Funding acquisition. **Stewart Williams:** Writing – review & editing, Supervision, Resources, Funding acquisition, Conceptualization.

Declaration of competing interest

The authors declared that they have no conflicts of interest and have no commercial or associative interest that represents a conflict of interest in connection with the work submitted.

Acknowledgement

This work is financially supported by the ‘Hybrid Direct Energy Deposition Sprint’ project (NO. 113345) funded by the Aerospace

Technology Institute (ATI) and ‘Landing Gear Industrial Breakthroughs (I-Break)’ (10003486) funded by Innovate UK. The authors would also like to acknowledge the raw material support from Airbus. The authors would also like to acknowledge facilities access and support from the Henry Royce Institute through EPSRC grants EP/R00661X/1, EP/S019367/1, EP/P025021/1, and EP/P025498/1. The authors would also like to thank the technical support from Nisar Shah and Steve Pope.

Data availability

Data will be made available on request.

References

- [1] J. Dang, et al., Deformation behavior and microstructure evolution of 300M ultrahigh strength steel subjected to high strain rate: an analytical approach, *J. Mater. Res. Technol.* 25 (2023), <https://doi.org/10.1016/j.jmrt.2023.05.203>.
- [2] P. Skubisz, J. Sinczak, Properties of direct-quenched aircraft forged component made of ultrahigh-strength steel 300M, *Aircr. Eng. Aerosp. Technol.* 90 (5) (2018), <https://doi.org/10.1108/AEAT-12-2015-0253>.
- [3] A. LaTour, M. Charles, N. Raghunathan, M. Ahlfors, Towards production of landing gear materials with laser powder bed fusion, *Int. J. Refract. Met. Hard Mater* 91 (2020), <https://doi.org/10.1016/j.ijrmhm.2020.105272>.
- [4] D. Gu, X. Shi, R. Poprawe, D.L. Bourell, R. Setchi, and J. Zhu, “Material-structure-performance integrated laser-metal additive manufacturing,” 2021. doi: 10.1126/science.abg1487.
- [5] C.R. Cunningham, J.M. Flynn, A. Shokrani, V. Dhokia, and S.T. Newman, “Invited review article: strategies and processes for high quality wire arc additive manufacturing,” 2018. doi: 10.1016/j.addma.2018.06.020.
- [6] J. Wang, et al., Investigation of 300M ultra-high-strength steel deposited by wire-based gas metal arc additive manufacturing, *Int. J. Adv. Manuf. Technol.* 129 (7–8) (2023), <https://doi.org/10.1007/s00170-023-12566-9>.
- [7] Y.B. Xiong, D.X. Wen, Z.Z. Zheng, J.J. Li, Effect of interlayer temperature on microstructure evolution and mechanical performance of wire arc additive manufactured 300M steel, *Mater. Sci. Eng. A* 831 (Jan. 2022), <https://doi.org/10.1016/j.msea.2021.142351>.
- [8] Y.B. Xiong, D.X. Wen, Z.Z. Zheng, C.Y. Sun, J. Xie, J.J. Li, Effect of heat treatment on microstructure and mechanical properties of directed energy deposition-Arc 300M steel, *Mater. Charact.* 198 (2023) 112756, <https://doi.org/10.1016/j.matchar.2023.112756>.
- [9] E. Eimer, S. Williams, J. Ding, S. Ganguly, B. Chehab, Mechanical performances of the interface between the substrate and deposited material in aluminium wire direct energy deposition, *Mater. Des.* 225 (2023), <https://doi.org/10.1016/j.matdes.2023.111594>.
- [10] J. Zhang, et al., Crack path selection at the interface of wrought and wire + arc additive manufactured Ti-6Al-4V, *Mater. Des.* 104 (2016), <https://doi.org/10.1016/j.matdes.2016.05.027>.
- [11] W. Fan, S.J.L. Wang, Y.J. Peng, H. Tan, Y.X. Wang, Y. Qi, Z. Feng, F.Y. Zhang, B. He, X. Lin, Microstructure and mechanical properties of an ultra-high strength steel fabricated by laser hybrid additive manufacturing, *Mater. Sci. Eng.* 885 (2023) 145594, <https://doi.org/10.1016/j.msea.2023.145594>.
- [12] C. Barr, S. Da Sun, M. Easton, N. Orchowski, N. Matthews, M. Brandt, Influence of delay strategies and residual heat on in-situ tempering in the laser metal deposition of 300M high strength steel, *Surf. Coat. Technol.* 383 (Feb. 2020), <https://doi.org/10.1016/j.surfcoat.2019.125279>.
- [13] Y. Tomita, Development of fracture toughness of ultrahigh strength, medium carbon, low alloy steels for aerospace applications, *Int. Mater. Rev.* 45 (1) (2000), <https://doi.org/10.1179/095066000771048791>.
- [14] E. Abbasi, Q. Luo, D. Owens, Microstructural characteristics and mechanical properties of low-alloy, medium-carbon steels after multiple tempering, *Acta Metall. Sin. (Engl. Lett.)* 32 (1) (Jan. 2019) 74–88, <https://doi.org/10.1007/s40195-018-0805-6>.
- [15] B. Ahmad, X. Zhang, H. Guo, M.E. Fitzpatrick, L.M.S.C. Neto, S. Williams, Influence of deposition strategies on residual stress in wire + arc additive manufactured titanium Ti-6Al-4V, *Metals* 12 (2) (2022), <https://doi.org/10.3390/met12020253>.
- [16] J. Zhang, X. Wang, S. Paddea, X. Zhang, Fatigue crack propagation behaviour in wire+arc additive manufactured Ti-6Al-4V: effects of microstructure and residual stress, *Mater. Des.* 90 (2016), <https://doi.org/10.1016/j.matdes.2015.10.141>.
- [17] F. Qayyum, et al., Influence of non-metallic inclusions on local deformation and damage behavior of modified 16MnCr5S steel, *Crystals* 12 (2) (2022), <https://doi.org/10.3390/cryst12020281>.
- [18] Y. Momotani, A. Shibata, D. Terada, N. Tsuji, Effect of strain rate on hydrogen embrittlement in low-carbon martensitic steel, *Int J Hydrog. Energy* 42 (5) (2017), <https://doi.org/10.1016/j.ijhydene.2016.09.188>.
- [19] Y. Tomita, Low fracture toughness of desulphurised 0.4C–Cr–Mo–Ni ultrahigh strength steel and its improvement, *Mater. Sci. Technol.* 7 (2) (1991) 97–104, <https://doi.org/10.1179/mst.1991.7.2.97>.
- [20] Y. Tomita, Development of fracture toughness of ultrahigh strength low alloy steels for aircraft and aerospace applications, *Mater. Sci. Technol.* 7 (6) (1991) 481–489, <https://doi.org/10.1179/mst.1991.7.6.481>.

- [21] S. Huh, Y. Heo, J.S. Lee, C.H. Yim, The role of sulfide growth on the segregation behavior of sulfur at the grain boundary in a high-nickel steel, *Mater. Charact.* 199 (2023) 112845, <https://doi.org/10.1016/j.matchar.2023.112845>.
- [22] G. Krauss, Martensite in steel: strength and structure, *Mater. Sci. Eng.* 273-275 (1999) 40–57, [https://doi.org/10.1016/S0921-5093\(99\)00288-9](https://doi.org/10.1016/S0921-5093(99)00288-9).
- [23] O. Falodun, S. Oke, M. Bodunrin, A comprehensive review of residual stresses in carbon steel welding: formation mechanisms, mitigation strategies, and advanced post-weld heat treatment techniques, *Int. J. Adv. Manuf. Technol.* 136 (2025) 4107–4140, <https://doi.org/10.1007/s00170-025-15088-8>.
- [24] Metallurgical modelling of welding, *Mater. Charact.* 34 (4) (1995), [https://doi.org/10.1016/1044-5803\(95\)80084-0](https://doi.org/10.1016/1044-5803(95)80084-0).



DSPACE

<https://dspace.org/>

Insights into crack prevention and property improvement for additively manufactured ultra-high-strength steel structures with complex geometries

Wang, Jun; Taylor, Mark; Diao, Chenglei; Pickering, Ed J.; Qin, Jian ...

2025-07-01

Elsevier

<https://doi.org/10.1016/j.addlet.2025.100307>

Wang J, Taylor M, Diao C, et al., (2025) Insights into crack prevention and property improvement for additively manufactured ultra-high-strength steel structures with complex geometries. Additive Manufacturing Letters, Volume 14, July 2025, Article number 100307