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Microstructure Property Relationships in Directed Energy Deposited Bohler K340 Tool Steel under Different Post- Processing Heat Treatments

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Abstract

Directed Energy Deposition (DED) offers significant potential for the additive manufacturing and localized repair of high-performance industrial components. However, the complex thermal history of the process induces severe microsegregation and phase heterogeneities in high-chromium tool steels. This thesis presents a systematic experimental and computational investigation into the microstructural evolution, phase transformation kinetics, and mechanical response of Böhler K340 cold-work tool steel processed via DED. The research was strategically divided into two phases: first, a comparative evaluation of the deposition zone in bulk specimens fabricated under Argon (Ar) and Nitrogen (N₂) gas atmospheres; and second, a comprehensive interfacial and comprehensive multi-zone analysis of repaired groove geometries (2–5 mm depth and 4,5,8 mm width) encompassing the substrate (SU), heat-affected zone (HAZ), and DED deposit. Post-deposition behavior was optimized under two distinct heat treatment routes: Direct Triple Tempering (DTT, 2 × 525°C + 1 × 560°C) and Quenching and Triple Tempering (QTT, austenitization at 1070°C followed by oil quenching and triple tempering).

Characterization of the bulk deposition zones revealed that, In the as-built (AB) condition, rapid directional solidification prompted the epitaxial growth of columnar prior austenite grains containing a refined cellular-dendritic substructure, backed by an inherent inter-layer heat treatment (IHT) effect. For the repaired geometries, a complete cross-sectional analysis across the SU/HAZ/DED interfaces was performed. Advanced Electron Backscatter Diffraction (EBSD) phase mapping quantified an exceptionally high fraction of retained austenite (RA) within the repaired DED zone (55.3% for AB and 53.4% for DTT). This was caused by solute trapping and severe microsegregation of Cr and Mo into intercellular networks, which mechanically and chemically stabilized the austenite and restricted martensite growth.

To elucidate these phenomena, non-equilibrium Scheil-Gulliver simulations were performed via Thermo-Calc, successfully modeling the solute enrichment in the final liquid. Correlative EDS and morphological analyses identified three distinct carbide populations across the repaired zones: primary blocky Cr-rich M_7C_3 carbides, highly stable primary Nb/V-rich MC particles, and fine, film-like secondary $M_{23}C_6$ carbides precipitated during tempering. TC-PRISMA kinetic simulations verified that the high supersaturation trapped within the matrix during quenching dramatically accelerated secondary carbide precipitation in QTT conditions compared to DTT, driving a powerful secondary hardening response.

Mechanical profiling via detailed Micro-Vickers hardness mappings highlighted a severe ~300 HV gradient between the soft as-built repair zone (450–550 HV) and the hardened substrate (800 HV). Application of the DTT route promoted secondary hardening (600–

700 HV) within the deposit but failed to eliminate the strong <001> solidification texture or the retained austenite. Crucially, DTT induced a detrimental localized softening dip (524 HV) in the partially austenitized HAZ (HAZ-P) due to over-tempering and carbide coarsening. Conversely, the QTT route successfully triggered a complete microstructural reset, fully dissolving the cellular-dendritic boundaries, dropping the retained austenite to an isotropic 0.77%, and eliminating the directional texture. Macroscopic evaluation of the repair configurations indicated defect-free bonding in wider tracks (8 mm), whereas narrow grooves (4 mm) suffered from lack of fusion. Ultimately, the QTT treatment successfully bridged the mechanical gap across the repair interfaces, yielding a uniform hardness level of 750–830 HV and ensuring microstructural and structural continuity for industrial tooling restoration.

1 Introduction

1.1 Research context

In modern manufacturing, the demand for high-precision and durable metallic parts has grown steadily, driven by global trends toward higher productivity, reduced waste, and improved energy efficiency. A significant portion of industrial components, particularly in automotive, aerospace, and heavy machinery, are produced by cold and hot forming operations. With nearly 30% of global greenhouse gas emissions originating from material processing, industries face increasing pressure to reduce energy consumption, extend component lifetime, and implement repair-oriented strategies aligned with circular-economy principles [1]. In fact, repairing high-value engineering components is often far more economical than producing new replacements, mainly because manufacturing apart from raw materials requires significantly more energy and cost. Studies show that repair operations use only about 15% of the energy needed for full production. In practical terms, the same amount of material or energy required to fabricate one new component can instead be used to restore approximately 7 to 11 damaged parts [2]. On the other word, recent socio-economic trends, emphasizing flexibility, production efficiency, time and cost reduction, waste minimization, and environmental sustainability, have intensified the demand for advanced tooling materials. These materials must not only withstand severe mechanical and thermal stress but also align with the growing focus on environmentally conscious manufacturing practices, driving both academic and industrial[3].

Tooling used for metal forming is one of the crucial elements involved in the production of metal components for various industries. As can be seen, for instance, in the automotive industry, the percentage of production of metal parts by this industry reached 44% of the total production of metal parts globally in 2015. Moreover, the rise in the production of vehicles in countries such as the US, China, Japan, and Germany has resulted in increased demands for high-quality forming tools. Apart from this particular sector, there has been steady growth in the global market for stamping and punching tools. The reason behind this is the increasing use of metals in aerospace, transportation, automotive, and precision industries. In this regard, tool steels are among the strategically crucial materials, with a worldwide market value of about USD 5.5 billion in 2023 and a projected value of USD 7.3 billion in 2030. Over 70% of the high-production forming and molding tools are made of hot-work or cold-work tool steels, characterized by their hardness, wear resistance, and thermal stability. However, some recent industrial studies reveal that on average the lifespan of such tools has dropped by 15–20% in the last ten years owing to more challenging conditions. [4], [5].

Tool steels play a central role in this context. These alloys are widely used for manufacturing tools for punching, blanking, cold rolling, extrusion, deep drawing,

bending, and similar forming operations where components are shaped at room temperature under high compressive stresses. To perform reliably under such demanding conditions, Tool steels must provide a unique combination of high hardness, abrasive and adhesive wear resistance, compressive strength, toughness, and dimensional stability [6].

1.2 State-of-the-art

1.2.1 Tool Steel

The tool steels are complicated alloys of iron metal with carbon and carbide forming alloying metals, which include chromium, vanadium, molybdenum, tungsten, and even cobalt in some cases. The ratio of carbon to other alloying metals determines characteristics such as high hardenability, high wear resistance, low temper brittleness, and good toughness. The modern tool steels are described based on their ability to have all these qualities in one alloy. In general, these properties are influenced by specific compositional and microstructural factors [7], [8]. Compared with structural steels, tool steels generally have higher carbon contents and, except for certain types such as water-hardening steels, significantly higher alloying additions [9].

The higher levels of carbon and alloying elements will help in the formation of hard crystal structures such as martensite and carbides that form the basis for the high strength and hardness. In most cases, tool steel that has been hardened will have a microstructure made up of tempered martensite with carbide structures. Carbides play a central role in determining the mechanical properties of tool steels [10]. Their type and distribution are strongly influenced by the alloying elements present, most notably chromium, vanadium, tungsten, and molybdenum. Six major carbide types are typically found in tool steels; each associated with a distinct crystal structure and chemical formula [7], [9].

Since these steels are carefully engineered, every microstructural feature contributes to their final performance under service conditions in different applications [7], [11]. The application areas of such alloys include more than just cutting, since there are many types of tool steels that are used in situations where wear resistance, strength, toughness, and thermal stability are required. Wear resistance can be achieved by increasing the volume of carbides so as to make the material resistant to wear and erosion, thus extending its lifetime in challenging environments. Toughness, on the other hand, can be related to low carbon content, giving the material resistance to breaking and chipping under impact loading. Attainable hardness, determined by heat treatment and controlled alloy design, allows them to reach hardness values far exceeding most other steels. Red hardness, or hot hardness, describes the ability of steel to maintain hardness at elevated temperatures. Retaining hardness under such extreme conditions prevents plastic deformation of the tool [7], [8]. By applying carefully controlled heat treatment, their compositions are optimized to deliver the necessary combination of

properties for industrial use[9], [12]. Their versatility of applications allows them to be employed in metal forming[10], [13], [14] (dies, blanking, piercing, drawing and thread rolling, trimming), machining operations[15], [16], wear-critical components (bearings, bushings, mold inserts), and high-temperature service (springs, fasteners, valves). With growing industrial demands for higher performance, the applications of modern tool steels continue to expand steadily[8].

Tool steel can be classified in several ways; the American Iron and Steel Institute (AISI) classification system is most widely applied. It divides tool steels into seven groups according to their alloying elements, heat treatment response, and applications: water-hardening steels, Low-alloy special-purpose steels, mold steel, shock-resisting steel, hot-work steel, cold-work steel, and high-speed steel [17]. Among them established categories of tool steels, cold-work, hot-work, and high-speed steels constitute the three most widely used groups.

Hot-work steels (Group H) are designed to withstand softening and maintaining strength, hardness, and impact resistance at elevated service or heat treatment temperatures. Their combination of low-to-medium carbon (0.30–0.40 wt.%) and moderate-to-high alloy content provides good toughness and hot hardness, though wear resistance remains moderate[9] [17].

Chromium-based grades achieve resistance to softening and thermal fatigue through the precipitation of fine chromium and vanadium -rich carbides. Tungsten-alloyed types offer excellent hot-strength stability via abundant W–Cr carbides but with reduced toughness. Molybdenum-based steels provide similar high-temperature stability, supported by Mo-, Cr-, V-, and W-containing carbides that promote durability in prolonged high-temperature service [9].

High-speed steel (HSS) are designed to maintain hardness at high temperatures while performing cutting at increased speed. Usually, HSS consists of alloying elements amounting to no more than about 20 wt%. Molybdenum or/and tungsten, chromium, vanadium, cobalt, and carbon can be considered as major components of high-speed steels. Initially, HSS was used only in cutting tools, although some HSS grades are used for cold work as well [17].

Tungsten-based grades (T-series) include about 18% tungsten or more and have excellent red hardness and wear resistance due to large amounts of stable alloy carbides. Molybdenum-based grades (M-series) partially replace tungsten by molybdenum and show similar characteristics with somewhat better toughness but higher sensitivity to decarburization [9].

In contrast to hot work steels, cold work steels are unable to withstand softening at elevated temperatures; therefore, they cannot be used at temperatures exceeding approximately 200–260°C. There exist three kinds of cold work steels: A (air-hardening),

D (high carbon, high chromium), and O (oil-hardening). Regardless of different alloying elements, all these steels contain relatively small amounts of alloying elements, thus lacking high-temperature capability [17].

In oil-hardening steels the high hardness is provided by high carbon martensite and dispersed carbide precipitates; oil quenching allows achieving adequate hardenability. Air-hardening steels are alloyed in such a way that it is possible to obtain martensite after air quenching, avoiding any distortions and increasing dimensional stability. Steels with high carbon content and high chromium content are characterized by extremely good wear resistance due to high content of alloy carbides and tempered martensite. However, high hardness makes machining and grinding difficult [9].

Although possessing excellent characteristics, tool steels can be subjected to softening, wear, and cracking because of heat and mechanical stresses, chemical environment, and other factors accelerating wear, thermal cracking, plastic deformation, and fatigue failure [18]. The latter problem can be solved by appropriate heat treatment providing improvement of mechanical properties. Heat treatment of tool steels will influence mainly the microstructure of martensitic matrix, type and distribution of carbides, and, hence, hardness, toughness, and wear resistance [19].

Ramesh et al.[12] demonstrated for D2 tool steel that increasing the annealing soaking time promotes the transformation of carbide morphology from irregular to more rounded and uniformly distributed carbides within a pearlitic matrix, improving ductility while reducing overall strength. Farahany [20] investigated the effect of triple tempering temperature on K340 cold-work tool steel and observed that tempering in the range of 500–580 °C resulted in tempered martensite containing irregular M_6C and spherical secondary carbides. The study showed that hardness reached a maximum of 63.4 HRC at 520 °C, while impact toughness increased steadily with tempering temperature up to 580 °C, and abrasive wear was highest at the upper tempering limit, indicating a strong interdependence between microstructural evolution, carbide coarsening, and mechanical performance. Studies on high-carbon tool steels have shown that incorporating cryogenic treatments between hardening and tempering can significantly influence microstructure and mechanical performance. For AISI D2, adding a liquid nitrogen cryogenic step (77 K) increased the volume fraction of fine carbides ($<1\ \mu\text{m}$), altered the carbide types, and transformed retained austenite into martensite, resulting in higher ultimate tensile strength, improved ductility, and an increased elastic modulus compared to conventional heat treatment. Fracture analysis indicated a transition from cleavage to mixed cleavage–ductile fracture, and stress–strain behavior revealed enhanced strain-hardening due to a more uniform distribution of fine carbides, which reduces active dislocation length and increases flow stress [21]. In Böhler K340 cold-work tool steel, deep cryogenic treatment (DCT) added prior to triple tempering after high-temperature hardening produced a finer and more homogeneous martensitic

structure compared to conventional triple tempering. This refinement improved impact toughness while maintaining comparable hardness and enhancing wear resistance [22]. Further studies involving K340 reveal that both shallow and deep cryogenic treatments contribute to refining of the laths of martensite, decreasing retained austenite to below 1.8 vol.% in case of shallow cryogenic treatment and to below 1.0 vol.% in case of deep cryogenic treatment, and increasing the amount of fine spherical secondary carbides. All of these effects taken together result in a more uniform matrix, improved dimensional stability, and increased wear resistance. It becomes obvious how important it is to control the annealing, austenitizing, tempering, and cryogenic processes in order to ensure proper carbide morphology, phase balance, and mechanical properties of high carbon cold work tool steels [6]. Moreover, the artificial neural network (ANN) model was developed for optimizing heat treatment of hot work tool steels. The model identified key alloying effects, vanadium as the most beneficial for improving both hardness and toughness, while Si, Mn, and Cr increased hardness but reduced toughness, and Mo had the opposite effect. The optimized composition and heat treatment conditions were determined to maximize performance [18].

1.2.2 Additive Manufacturing

There are several metallurgical methods for producing tool steels, each having advantages and drawbacks. In conventional method, tool steel is made from melting in electric arc furnace, followed by secondary melting either in argon oxygen decarburization process (AOD) or vacuum oxygen decarburization process (VOD). This is economical but results in small amounts of impurities [11]. Purer steels can be produced using electroslag remelting (ESR) or vacuum arc remelting (VAR) technologies; however, this increases costs [11], [23]. Forging and hot rolling result in breaking up coarse carbides and increased toughness, but conventional processes have problems like segregation, high energy use, and uneven microstructure [23].

In response to these problems, other manufacturing processes have been developed. One of these is curved continuous casting (C3), which provides a process that is simple, economical in energy, and less wasteful and costly while increasing yield [11], [23]. Another process that allows for a more uniform and inclusion-free microstructure with better toughness is powder metallurgy (PM). This involves using gas atomized powders and is HIPped together, but its high cost and inability to scale up have made it problematic [24], [25]. Other processes under development include metal injection molding (MIM) for making small and intricate parts. Spray forming (Osprey) produces dense billets with PM-like microstructures through fewer steps and higher deposition rates, though yield and size limitations remain [11], [26], [27].

Despite the wide use of conventional techniques, issues related to wastage of materials, long processing time, and the difficulty in forming complex geometry have made people turn towards more modern manufacturing processes [28], [29].

These problems can be addressed through the introduction of new technologies that go by the names of rapid prototyping (RP) [30] or, generally speaking, additive manufacturing (AM)[30]. AM involves building up components in a layer-by-layer manner starting from the bottom up by joining the material together. This method allows achieving complex designs with great flexibility and at the same time saves materials, cuts the production process time, and does not require any special tools. AM has several benefits when compared to conventional manufacturing techniques. First of all, AM allows building functional parts on demand which decreases the manufacturing cost and reduces logistic difficulties connected with storing, transporting, and timely delivering crucial replacement parts [11], [28], [31], [32], [33].

Although AM technology has already gained much popularity in such developed areas as aerospace[27], [34], and medical engineering[35], it is actively evolving towards new industries like tooling, building construction, energy, oil and gas industries, etc. This broadening scope highlights the growing importance of additive manufacturing as a key enabler of innovation, customization, and sustainable production across diverse industrial domains [36].

Additive manufacturing technologies may work with different types of materials such as polymers, ceramics and metals, but the most interest among both scientific and business communities have recently attracted the AM technologies for metals owing to the unique opportunity to merge the flexibility of AM with the enhanced characteristics of metal alloys. Besides typical additive advantages such as speed and precision, there are also environmentally friendly features of metal AM such as low energy demand, less pollutant generation and high material utilization [31].

1.2.2.1 Metal Additive Manufacturing

Research activity has also primarily focused on a narrow subset of metal AM processes rather than the full spectrum of available technologies [29]. Metal AM processes are divided by ASTM International Committee [37], [38], into seven major categories depending on the kind of feedstock and the energy sources [30], [39]. Vat Photopolymerization (VPP), Material Extrusion (ME), Powder Bed Fusion (PBF), Directed Energy Deposition (DED), Material Jetting (MJ), Binder Jetting (BJ), and Sheet Lamination (SL). Schematic depiction of each metal AM process type can be found in Fig. 1.2-1. Various forms of feedstock, including powder, wire, filament, sheet, or liquid, are possible, as well as different kinds of thermal trajectories and energy supply, depending on the chemistry and energetics of materials transformation[30]. All seven categories have been implemented commercially in certain degree; however, only five are widely utilized for the creation of metal components. Every process category has its own strengths and weaknesses regarding material properties, surface finish, processing speed, and cost of manufacturing [31], [40]. Powder Bed Fusion (PBF) and Directed

Energy Deposition (DED) are the main processes used for metal additive manufacturing because they can produce strong, precise, and near-finished parts.

Powder Bed Fusion (PBF) is an additive manufacturing process in which a concentrated thermal energy source selectively fuses regions of a powder bed to build parts layer by layer[30]. As shown in *Figure 1.2-2c*, once a layer is completed, the build platform lowers and a new layer of powder spread across the surface, which is then selectively melted or sintered to the previous layer[31], [41].

PBF represents the dominant group of metal AM processes, accounting for approximately 54% of the metal AM market in 2020[42]. The most common variants include Selective Laser Melting (SLM), Direct Metal Laser Sintering (DMLS), Direct Metal Laser Melting (DMLM), and Electron Beam Melting (EBM). Although the general principles of these technologies are similar, minor differences exist in energy source, processing parameters, and material compatibility. SLM (or Laser Metal Fusion, LMF) is typically used for single-component metals such as aluminum, whereas DMLS is preferred for metal alloys like titanium and aluminum alloys. DMLM, closely related to DMLS, uses ultra-thin layers and refined melt pools to achieve better surface quality and reduced porosity. In contrast, EBM utilizes a high-energy electron beam instead of a laser, operating in a vacuum environment to fuse the metallic powders[31], [43].

The main advantages of PBF include the ability to produce highly complex geometries without support structures, a wide range of compatible materials, and minimal post-processing for support removal. However, the process also has several limitations, such as high equipment and operating costs, size constraints, potential thermal distortions, and surface finish sensitivity to powder particle size[31].

Typical materials used in PBF include stainless steels, tool steels, nickel- and cobalt-based superalloys, titanium and aluminum alloys, bronze, and various metal-ceramic composites, making it one of the most versatile technologies in the metal AM domain[31].

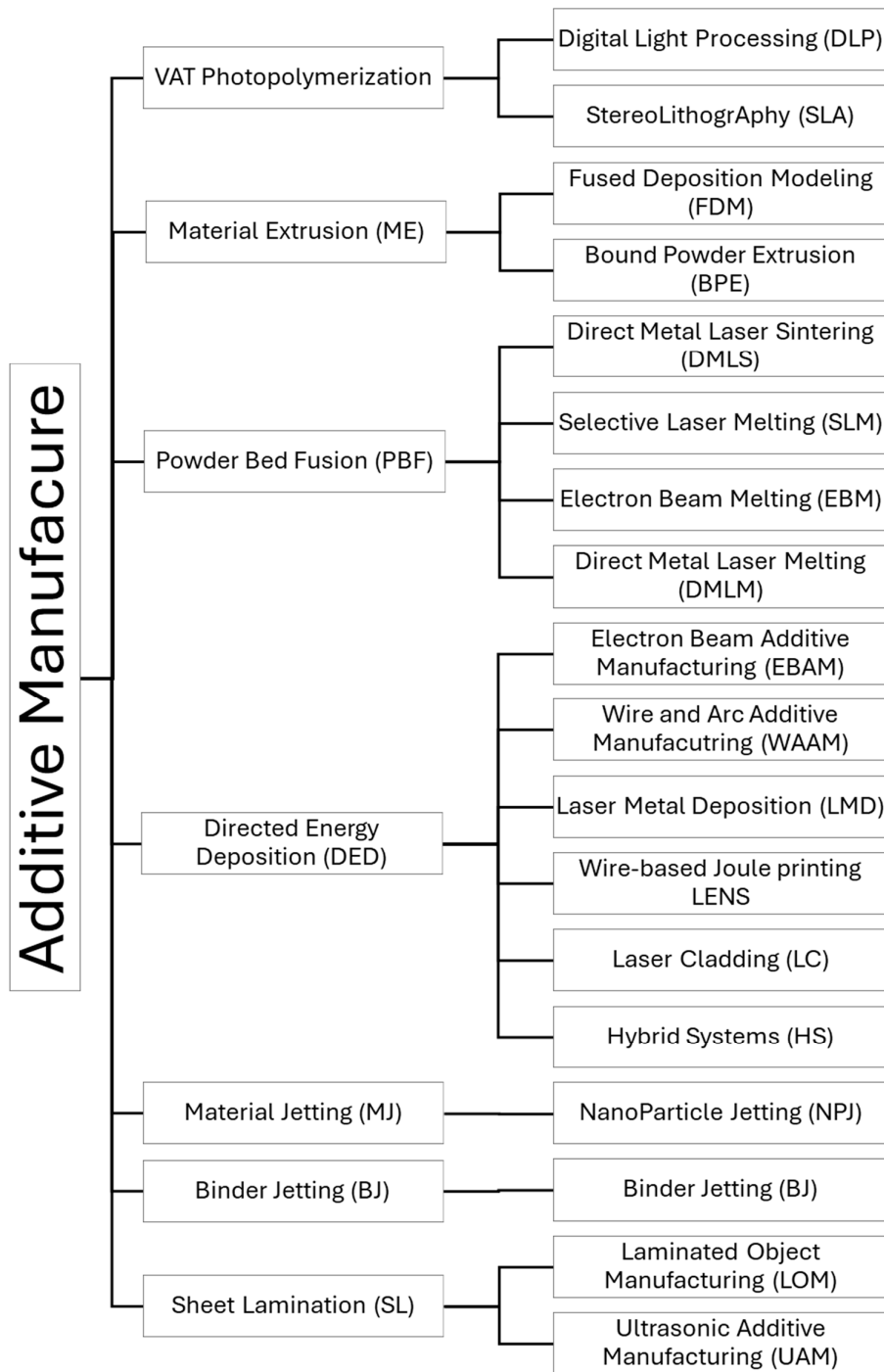


Figure 1.2-1 Classification of Additive Manufacturing

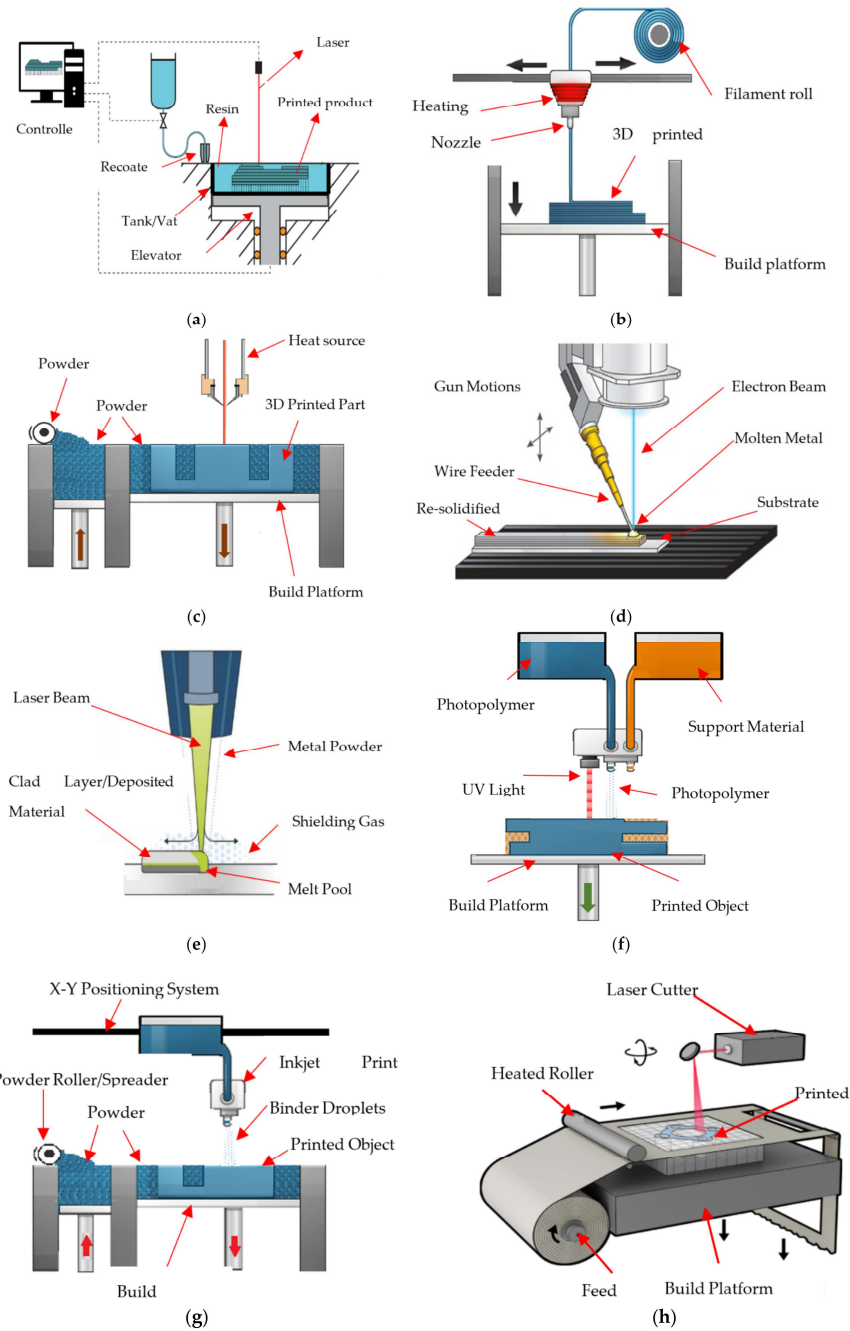


Figure 1.2-2. Metal AM Processes; (a) VAT Photopolymerisation System, (b) Material Extrusion Process, (c) Powder Bed Fusion Process, (d) Wire DED Process, (e) Powder DED Process, (f) Material Jetting Process, (g) Binder Jetting Process, and (h) Sheet Lamination Process

1.2.2.2 Directed Energy Deposition

Directed Energy Deposition (DED) is an additive manufacturing process in which focused thermal energy is used to fuse materials by melting them as they are being deposited[30]. This is typically achieved using a nozzle combined with a multi-axis robot that selectively adds material to a substrate, providing enough energy to form a new layer[31].

The Directed Energy Deposition (DED) process, often referred to as 3D welding, is a versatile additive manufacturing technique capable of producing near-net-shape metal components using either powder or wire as the feedstock, with a laser serving as the primary energy source. Wire feedstock enables the fabrication of large three-dimensional components at very high deposition rates. However, due to the use of a non-focused heat source and rapid material deposition, the parts produced typically exhibit near-net shapes with limited dimensional accuracy and poor surface finish ($R_a \approx 200 \mu\text{m}$)[44], [31], [45]. Compared with powder-based DED, wire offers higher catchment efficiency, defined as the ratio of the material successfully deposited to the total material supplied, and faster build rates, making it a more cost-effective option for large-scale manufacturing[44].

Nevertheless, the applicability of wire is constrained by the limited range of metallic wires commercially available, which restricts material versatility. Although the combination of powder feedstock and electric arc heating has been conceptually explored, it remains relatively under-investigated in current research and industrial practice[44].

Metallic powders are the most employed feedstock form. Owing to the fact that the DED machine uses pneumatic transport for feeding the powder into the system, the particle size of the powder is relatively higher than that used in PBF processes. Larger particles are preferred in DED machines since they promote ease in handling and minimize the tendency to form clumps during processing. They are also able to overcome the melt-pool surface tension due to their greater momentum, making them more effective in catching. On the other hand, fine particles create smooth surfaces, but they do not flow well. However, the particle size selection in DED requires careful optimization. Excessively small particles can decrease deposition efficiency and lead to material loss, whereas overly large particles may increase roughness and reduce part resolution [44].

DED can be divided into two primary categories based on the energy source: cold spray and thermal energy. Cold spray, also known as kinetic energy deposition, involves propelling fine particles onto a substrate with sufficient velocity to create a dense coating or layer[46]. Thermal energy DED systems, on the other hand, rely on a laser, electron beam, plasma, or arc to selectively melt the wire or powder feedstock and deposit it onto the build platform layer by layer. These technologies are mainly applied in metal component fabrication, utilizing robotic welding techniques to achieve high deposition rates, though with lower resolution[31], [47].

Common DED processes include Electron Beam Additive Manufacturing (EBEAM or EBAM), Wire and Arc Additive Manufacturing (WAAM), wire-based Joule printing, Laser-Engineered Net Shaping (LENS), Laser Cladding (LC)[48], and Hybrid Systems (HS)[49] (Figures 1.2-2 d-e) [31].

In EBEAM, an electron beam melts wire feedstock, whereas WAAM uses an electrical arc to melt wire, with deposition performed layer by layer.[47] WAAM systems may employ welding methods such as Gas Metal Arc Welding (GMAW), Gas Tungsten Arc Welding (GTAW), or Plasma Arc Welding (PAW)[45].

Blown Powder Technology (BPT) represents another DED approach in which the substrate is locally melted, and metal powder is deposited into the molten pool using a focused laser beam. Unlike powder bed methods, BPT focuses both the laser and powder onto the substrate to produce near-net-shape parts with high precision. An inert gas shroud protects the molten metal and powder from oxidation[50]. LENS is one of the BPT techniques that melts metallic powder precisely where it is being applied. Some of the modifications of the BPT process include Laser Metal Deposition (LMD), Laser Solid Forming (LSF), Direct Laser Metal Deposition (DLMD), Direct Laser Deposition (DLD), Direct Light Fabrication (DLF), Laser Deposition Welding (LDW), Powder Fusion Welding (PFW), Electron Beam Direct Manufacturing (EBDM), and Direct Metal Tooling (DMT). LMD technology deposits a stream of metallic powders suspended in an inert gas with the help of a powerful laser. This method is very popular because of its small heat affected zone[31], [51].

LC, or Laser Cladding[48], deposits pre-placed material onto a substrate with a laser heat source and can operate using either a one-step process, where cladding material and substrate are melted simultaneously, or a two-step process, where powder is first deposited and then irradiated with a laser [49].

Hybrid systems combine DED with subtractive processes like milling to achieve tighter tolerances and multifunctional manufacturing[49]. This system merges additive and machining processes within a single machine, enabling the production of high-quality metallic components by compensating for the weaknesses of each individual process[28].

Typical feedstock materials include titanium alloys, stainless steels, nickel alloys, and cobalt alloys for powders, while wires may consist of titanium and its alloys, Inconel 600, 625, 718, copper, nickel alloys, stainless steels of the 300 series, aluminum alloys, alloy steels, and magnesium alloys[31]

DED is widely applied for both repairing damaged parts and building new metal components. One of the most significant applications of DED is the fabrication of Functionally Graded Materials (FGMs)[52] that exhibit tailored mechanical and thermal properties across a component. Another emerging research direction involves the

production of Metal Matrix Composites (MMCs)[53], with current efforts focusing primarily on feedstock design and optimization. Additionally, DED allows the efficient processing of nickel-based superalloys, making it particularly suitable for high-temperature and corrosion-resistant applications.[44]

DED has also found use in the manufacturing of grid-stiffened structure[54], which are lightweight yet capable of carrying high loads. Such structures, commonly used in aerospace [34], offer superior vibration damping, high bending stiffness, and reduced structural weight while maintaining convenient mounting configurations. A notable example is the isogrid structure[55], characterized by a repetitive triangular rib pattern that enhances both mechanical performance and structural integrity[44].

Recent research has highlighted the critical influence of key DED process parameters and their post-processing, such as laser power, powder feed rate, scanning speed, and layer thickness, on the resulting deposition quality and mechanical behavior of tool steels. Studies by Omar (2022)[56], Colombo et al. (2024)[57], and Shim et al. (2016)[58] on D2 and M4 tool steels deposited on annealed D2 substrates demonstrated clear correlations between energy input and microstructural evolution. The outcomes indicated that DED-produced D2 has greater hardness values compared to annealed wrought D2 and lower hardness when compared with quenched D2. Nevertheless, after tempering processes, the hardness values of the produced samples became similar to those of quenched D2, proving that the post-process heat treatment is necessary for obtaining optimal mechanical behavior of additively manufactured cold work tool steels.

In relation to M4 high-speed steel fabricated on the surface of D2, another major finding of the study was that the phase formation, the morphology of carbides, and fusion interface characteristics depended on energy density parameters. All of these findings are relevant to the development of a proper understanding of the process-structure relationships in the case of multi-material components made of tool steels. Moreover, the conducted studies on different approaches to determine the thickness of layers demonstrated that the conventional technique causes errors related to poor geometry and extra metal deposition especially when the laser power is relatively high. On the other hand, feedback-controlled procedures improve the geometric accuracy but lead to uneven micro-hardness because of the variation of heat generation. The developed technique for choosing thicknesses guarantees both constant melt pool and uniform micro-hardness and geometric characteristics.

Recent investigations on K340 cold-work tool steel were explored by Kenevisi et al. Its processability and post-processing behavior using Directed Energy Deposition (DED), particularly for repairing damaged components such as blanking dies. Depositing double-track geometries, near-fully dense structures were achieved; however, the repeated thermal cycles inherent to DED induced microstructural heterogeneity, including variations in martensitic lath structure, retained austenite content, and carbide

distribution [59]. Systematic investigation into the effects of heat treatments indicated that austenitization temperature, tempering temperature, and tempering cycles were key factors affecting hardness and amount of retained austenite, where hardness depended mostly on interactions between the two temperatures and retained austenite was mostly affected by the tempering process [19]. Topography and microstructural examinations also showed that laser power, nozzle speed, and flow of carrier gas had a great impact on surface roughness, dilution (between 5% and 36%), and porosity; the higher laser power and lower nozzle speed resulted in better surface properties. Taken together, these results imply that whereas near-defect-free dense coatings can be obtained through DED of K340 tool steel, optimization of processing parameters and post-coating heat treatments is required to ensure uniformity of microstructure, carbide shape, and mechanical properties [60].

According to comparative research on DED-fabricated tool steels including H13 and D2, the difference between deposits and wrought materials may be significant because of different solidification structures. Depositions of H13 usually display greater hardness when compared with wrought material, whereas as-deposited D2 displays lower hardness owing to retained austenite and uneven carbide distribution. Post-deposition tempering leads to microstructure homogenization by transformation into tempered martensitic structure with fine dispersion of carbides and increased hardness and reduced anisotropy. It is clear from the example above that proper heat treatment is able to mitigate some of the microstructure heterogeneities caused by DED process and thereby enhance wear resistance of the part [61].

The thermal stability of AM-processed tool steels is another factor affecting service performance. Studies comparing the softening resistance of various DED-processed tool steels, a cold work steel (V4E) and a high-boron steel (HBS), have shown that the retention of high-temperature at 550 and 600 °C, hardness depends heavily on the stability of carbide populations. Alloys rich (V4E) in thermally stable MC carbides display superior resistance to softening during tempering, whereas steels containing borides (HBS) or low-melting eutectic phases tend to degrade more quickly at elevated temperatures. Since the accumulation of thermal damage is closely tied to crack initiation during service, microstructural stability after heat treatment directly contributes to extending the lifetime of tools such as stamping dies and high temperature forming components [62].

Additive manufacturing techniques, especially those of Directed Energy Deposition (DED), allow for melting and solidification at local sites, providing a chance to tailor the microstructure through process parameter and alloy adjustment. Yet, because of fast cooling and large temperature gradient, residual stresses, anisotropy, cracking, and non-equilibrium microstructure pose significant problems to obtaining maximum mechanical properties.

Various studies on the subject have focused on cracks in additive manufactured tool steels, which tend to be hard and alloyed and are generally difficult to weld and experience severe segregation issues. One type of steel extensively studied in this field is D2 steel, prone to cracking when laser-deposited. Studies on direct laser deposition of D2 have shown that cracking is predominantly concentrated within inter-dendritic regions enriched with eutectic austenite and M_7C_3 carbides, where repeated thermal cycling during layer-wise deposition promotes liquation of low-melting eutectic films. As reported through in-situ thermal monitoring, recurrent reheating above the solidus temperature accelerates liquation cracking, leading to longer and more continuous cracks as the build height increases. This behavior highlights the fundamental role of thermal history and segregated microconstituents in governing crack susceptibility in high-hardness tool steels fabricated by AM [63].

The issue is not limited to D2; similar mechanisms have been identified in other alloy systems. For instance, investigations on matrix high-speed steels produced by DED revealed that rapid solidification and solute segregation lead to retained austenite in inter-dendritic regions and α' -martensite in dendrites. Segregation depresses the local solidification temperature and promotes the formation of low-melting eutectic carbides, thereby triggering both solidification and liquation cracking. Moreover, high residual tensile stresses inherent to the DED process exacerbate cracking by promoting cold crack initiation in martensitic regions adjacent to the hot cracks, effectively linking microcracks into long macroscopic longitudinal cracks. These observations collectively demonstrate that cracking arises from a coupled interaction between segregation-driven microstructure evolution, localized melting phenomena, and the development of residual stresses during additive manufacturing [64].

Efforts to improve the processability of tool steels have therefore focused on alloy design and process optimization. Previous work with laser powder bed fusion of high-carbon tool steels has revealed that the lowering of the Martensite Start Temperature or the carbon content can be used to substantially reduce susceptibility to cracking. Hot work steels having low to medium alloy content can be manufactured using only slight pre-heating, while high-carbon steels need to be subjected to greater thermal conditioning in order to prevent thermal shock and the development of a brittle structure. Alterations of the alloying of standard steels like 1.2344 have proven helpful in this regard. These advances illustrate the necessity of integrating material development with process control to overcome the intrinsic AM-related cracking challenges in tool steels [41].

1.2.3 Repairing by Additive Manufacturing

During service, tool components are exposed to impacts, corrosion, and fluctuating thermal-mechanical loads that create surface and subsurface defects, especially microcracks, which propagate under cyclic loading and contribute to progressive material loss [65], [66]. Wear is therefore a critical degradation mechanism in tool steels,

originating within a thin, dynamically formed “active layer” consisting of oxides, transformed phases, heat-affected regions, and the adjacent matrix; this layer continuously evolves under stress, temperature, sliding velocity, frictional heating, and environmental conditions, showing that wear results from the entire friction system, not just the material [67], [68].

Tool wear typically progresses through three stages: abrasive wear, followed by adhesive wear through local micro welding, and finally fatigue wear as cyclic loading drives crack growth. Although alloying additions (W, V, Mo), optimized carbide populations, and compressive residual stresses can slow this progression, steels with similar hardness may still behave differently highlighting that microstructure, more than hardness, governs wear resistance. High-carbon, high-chromium, carbide-rich tool steels are designed to resist these mechanisms, yet their performance depends on carbide type, size, distribution, and matrix stability rather than hardness alone [69].

From a metallurgical perspective, abrasion resistance improves with fine martensitic structures and well-distributed fine carbides, whereas carbide coarsening or matrix softening (e.g., at high tempering temperatures) accelerates wear. Surface engineering methods such as nitriding can further enhance wear resistance, though excessively thick compound layers may become brittle [70]. Ultimately, key microstructural features, MC-type carbides, secondary carbide precipitation, matrix hardness, and high-temperature deformation behavior, control transitions between mild adhesion, severe adhesion (seizure), and interfacial plastic deformation. Higher MC-carbide fractions reduce adhesion and micro welding, while matrix stability at elevated temperature determines seizure resistance, confirming that wear behavior is fundamentally dictated by microstructure across all loading conditions [71].

Hot-work tool steels such as H13 experience severe surface degradation during cyclic thermo-mechanical loading, which accelerates crack initiation and wear, reducing tool life. Traditional repair techniques like welding or thermal spraying often introduce additional thermal stresses, reduce metallurgical compatibility, and provide limited-service life extension. In contrast, laser-based cladding and deposition processes have emerged as more effective repair methods. Studies using CO₂ laser cladding with CPM-9V powders demonstrate the formation of microstructures containing hard vanadium carbides embedded within martensitic and retained-austenite matrices, significantly increasing surface hardness—often up to four times that of the substrate. Additionally, these processes generate beneficial compressive residual stresses, which help suppress crack propagation and thereby enhance die life [72].

Industrial case studies reinforce these findings. For example, Toyota’s transition from conventional tungsten inert gas (TIG) welding to powder-blown Directed Energy Deposition (DED) for repairing aluminum die-casting molds drastically improved die longevity. Dies repaired via DED achieved lifetimes comparable to brand-new tooling,

eliminating unplanned downtime and emergency repairs. Life-cycle assessments further demonstrated that DED repair significantly reduces environmental impact and material waste compared to traditional repair technologies, aligning with modern sustainability and energy-efficiency goals [73].

1.2.4 Heat treatment of tool steel

Heat treatment plays a central role in defining the mechanical performance of tool steels, hardening of tool steels generally involves three fundamental steps: austenitizing, quenching, and tempering. Each stage introduces major microstructural changes that ultimately determine the final mechanical properties. The microstructure changes during the process of austenitization from a spheroidized ferrite-carbide microstructure to austenite with some amount of undissolved carbides. The dissolution of carbon and alloying agents like Cr and Mo into the austenite will influence factors like the martensite start temperature (M_s), hardenability, amount of retained austenite, and possibility of secondary hardening after tempering. High austenitizing temperatures lead to high amounts of carbon and alloying agent dissolved into the austenite, leading to low M_s temperatures and higher retained austenite content, thus resulting in lower quench hardness unless subzero treatment is performed [9], [74].

Fonseca et al. (2022) [75] investigated the tempering behavior of additively manufactured AISI H13 hot-work tool steel and concluded that the cellular microsegregated structure formed during the LPBF process strongly affects the decomposition of retained austenite and the phase transformation sequence during tempering. The results showed that at temperatures below 650 °C, the retained austenite remained stable during tempering and transformed into fresh martensite after cooling, while at higher temperatures, the retained austenite transformed into ferrite and carbides, accompanied by the dissolution of the cellular microsegregation structure. They also reported that the inhomogeneous distribution of alloying elements caused the phase transformations to follow local equilibrium conditions rather than the overall equilibrium of the system.

Uniform heating during austenitization is crucial, especially for highly alloyed tool steels that contain high carbide fractions and exhibit low hot ductility. Preheating steps are often required to minimize thermal gradients, avoid localized dimensional changes, and prevent cracking. Moreover, grain growth must be controlled because carbides that normally inhibit grain-boundary motion can dissolve or coarsen at high temperatures, leading to excessive austenite grain size and a subsequent loss of toughness after quenching [9], [74].

Upon quenching, the austenite transforms to martensite, accompanied by a characteristic volume expansion. If cooling is too slow, transformations to ferrite and carbides may occur instead, producing dimensional changes and reducing hardness. Quenching media selection and cooling rate must therefore balance the need for

martensitic transformation against the risk of quench cracking and distortion. Air hardening is generally preferred for highly alloyed tool steels, although oil quenching or gas quenching under pressure may be used depending on the application. Retained austenite is commonly present after quenching, especially in high-carbon or high-alloy steels, and needs to be controlled through tempering or cryogenic treatments [9], [19], [74].

Tempering stabilizes the martensitic structure, reduces internal stresses, and precipitates fine secondary carbides, enhancing wear resistance and improving toughness. In tool steels repeated tempering cycles are often necessary to fully transform retained austenite and promote a strong secondary-hardening response. The precipitation of alloy carbides during tempering significantly contributes to the high hardness values required for demanding cold-work applications [9], [74].

Deirmina et al.[76] investigated the effect of heat treatment on the microstructure and properties of SLM-fabricated H13 hot-work tool steel and concluded that the as-built microstructure consisted of partially tempered martensite with a high amount of retained austenite, which produced a stronger secondary hardening effect during tempering compared to quenched steel. The results showed that quenching removed the cellular/dendritic structure formed during the SLM process and reduced the local hardness heterogeneity. They also reported that the tempering behavior of the as-built samples was significantly different from that of the quenched samples. In addition, both tempered and quenched-tempered samples exhibited promising fracture toughness, and despite the higher hardness, the tempered samples showed fracture toughness comparable to the quenched and tempered condition.

Furthermore, adjustments in heat-treatment parameters directly influence retained austenite, dimensional stability, and resistance to cracking, all of which are vital for the longevity of dies and tooling exposed to cyclic mechanical and thermal loads. Surface treatments such as nitriding or PVD coatings are often used alongside conventional heat treatments to further improve wear resistance, although the underlying microstructure remains the primary factor governing performance [69].

Several researchers have investigated the heat treatment behavior and phase transformations of tool steels, particularly additively manufactured steels, using different simulation and modeling approaches. Chou et al. (2021) [77] studied the influence of solidification structure and microsegregation on the austenite-to-martensite transformation in LPBF-manufactured hot-work tool steel using computational thermodynamics and kinetics. Their results showed that alloying element segregation in intercellular regions reduced the martensite start temperature (M_s) and stabilized retained austenite.

Schaffnit et al. (2015)[78] developed a modified Scheil–Gulliver model for solidification simulation of steels. Using the CALPHAD approach, the model accurately predicted liquidus and solidus temperatures and provided a better understanding of solidification behavior and alloying element redistribution.

Deng and Ju (2018)[79] investigated the quenching and tempering process of steels through mathematical modeling and numerical simulation. Their model successfully predicted microstructural evolution, hardness distribution, carbide precipitation, and retained austenite decomposition under both isothermal and non-isothermal conditions. The simulation results showed good agreement with experimental data.

Hou et al. (2016)[80] studied carbide precipitation during tempering of martensitic Fe–Cr steel using the Kampmann–Wagner model combined with CALPHAD thermodynamic databases. Their results demonstrated that quantitative modeling could successfully predict the formation and growth of M₇C₃ and M₂₃C₆ carbides and could be used as an effective tool for heat treatment optimization and materials design.

In addition, Jin (2001) [81] developed a finite element model for simultaneous simulation of temperature, stress, microstructure evolution, and distortion during quenching and tempering processes. The results indicated that the model could accurately predict residual stresses, distortion, and phase transformations, making it a useful tool for heat treatment design of steels.

Overall, these studies demonstrate that thermodynamic, kinetic, and numerical simulations play an important role in understanding heat treatment behavior, phase transformations, and optimization of tool steel properties, especially in additive manufacturing applications.

1.3 Aims and Scopes

Despite significant advancements in research conducted in relation to the additive manufacturing of tool steels, there is an ongoing lack of systematic investigation regarding DED process in either manufactured tool steel parts or repairable areas using this approach. Specifically, effects of post-processing heat treatments in terms of carbide phase transformations, martensitic behavior, and hardness properties have not received sufficient attention. There are only scarce data available on carbide phases types, distributions, sizes, and their transformation in heat treated DED deposits due to specific thermal history caused by the DED process. Moreover, no information exists on the changes of retained austenite, kinetics of martensitic transformation, tempering behaviors, etc. in DED tool steel. To our knowledge, there have been no comparative studies of heat-treated bulk samples and those obtained by means of repair with the use of DED. Hence, optimal schedules for heat treatments of DED tool steel have yet to be found.

This work aims to fill in this gap via experimental and numerical analysis of DED produced Böhler K340 cold work tool steel deposits. The research is designed to clarify how different heat-treatment procedures influence the microstructural features and mechanical performance of both bulk deposits and DED-repaired regions. By examining the evolution of carbides, martensitic transformation behavior, and retained austenite content under controlled heat-treatment conditions, this work aims to develop a comprehensive microstructure–property framework for DED-processed K340.

The overarching objective of the study is therefore to evaluate, in a comparative manner, the microstructure and hardness of as-built, quenched-and-tempered, and tempered bulk DED samples, and to extend the same analysis to grooves filled via DED repair. To support the experimental findings, thermodynamic simulations of the heat-treatment procedures are performed using Thermo-Calc software to predict phase stability and carbide evolution. Furthermore, comprehensive hardness testing is systematically conducted across all conditions, encompassing both the direct as-built DED states and all subsequent heat-treated specimens.

1.4 Thesis Structure

To ensure logical flow and clarity, the thesis is organized into the following chapters:

Chapter 1 provides the overall background, introducing tool steels, additive manufacturing technologies, and the principles of Directed Energy Deposition (DED). This chapter also offers a comprehensive literature review, identifies existing knowledge gaps, and outlines the motivation and objectives of the study.

Chapter 2 describes the materials used in this work, including the DED feedstock, substrate material, and processing parameters. The defined heat-treatment procedures applied to both bulk and repaired samples are also presented. In addition, explain details the characterization techniques employed throughout the study. These include preparation process of samples for fractography, optical microscopy, SEM, EDS, EBSD, and hardness measurements.

Chapter 3 presents and analyzes the experimental results for bulk DED samples. Microstructural evolution, carbide morphology, phase transformations changes under different heat-treatment conditions are examined and discussed.

Chapter 4 focuses on the repaired samples, evaluating the quality of the deposited layers, the substrate–deposit interface, and the microstructural and mechanical response and simulation before and after heat treatment.

Chapter 5 conclusion of effect of heat treatment on microstructural and mechanical properties of K340 in bulk and repaired samples

2 Materials and Methodology

2.1 Material Selection

The choice of the correct tool steel is essential in maximizing the mechanical properties not only of the entire piece but also in repair situations. The products function in extreme environments requiring both ductility, toughness, fatigue resistance, wear resistance, and corrosion resistance in order to achieve maximum longevity. At the same time, other aspects associated with manufacture, including machinability and polishability, have to be taken into account to guarantee functional effectiveness and workability. It can thus be seen that the choice of the correct material is crucial in eliminating any defect origins. Among the various grades of tool steels available for industrial applications, the choice of the specific alloy strongly depends on the balance between hardness, toughness, wear resistance, and heat-treatability required for the intended service conditions.

Cold-work tool steels are primarily employed in applications involving high compressive stresses and abrasive wear at relatively low operating temperatures such as punching, blanking, cold rolling, extrusion, deep drawing, bending, and pressing[12], [14], [82]. Consequently, these steels must exhibit high hardness, strong abrasive and adhesive wear resistance, elevated compressive strength, good toughness, and excellent dimensional stability.

K340 is one of the contemporary types of tool steel that has been developed to attain an ideal balance between hardness, wear resistance, and toughness. The main alloying elements of this high-chromium steel include chromium, molybdenum, vanadium, and niobium. This alloy has excellent mechanical properties such as strength, dimensional stability, and high resistance to abrasion and adhesion. These attributes make this type of tool steel very useful in industrial processes involving high contact pressure and abrasive environment. Examples of applications include stamping and forming dies, blanking and shearing tools, extrusion dies, precision tools, and wear resistant structures[6], [22], [59].

Superior performance of K340 comes from its optimized chemical composition which comprises about 8 wt.% of chromium and 1.1 wt.% of carbon. The above composition causes the formation of a hard martensite structure with precipitated carbides within it. In this regard, segregation during solidification leads to the formation of carbide phases such as MC, M_2C , and M_7C_3 . Some of these carbides do not dissolve even after being subjected to austenitizing. Heat treatment further enhances the formation of secondary carbides such as $M_{23}C_6$ [19], [20].

Table 2.1-1 Nominal chemical composition of K340

	C	Cr	Mo	Si	Mn	V	AL	Nb
K340	1.10	8.30	2.10	0.90	0.40	0.50	+	+

2.2 Feedstock and Specimen Preparation

To investigate the influence of additive manufacturing parameters and subsequent heat treatments on the microstructural and mechanical behavior of tool steel, a series of specimens were fabricated using the Directed Energy Deposition (DED) technique. The fabrication process and feedstock selection were carefully designed to ensure reproducibility and to represent realistic conditions for both bulk fabrication and repair applications. The following subsections describe the feedstock characteristics, sample geometries, and deposition configurations adopted in this study.

Two sets of specimens were produced using the Directed Energy Deposition (DED) process with K340 Isodur powder as the feedstock material. The powder was gas-atomized using argon(Ar) and nitrogen(N₂) as atomizing gases, yielding a particle size distribution between 45–125 µm ($D_{50} \approx 77\text{--}88 \text{ µm}$), which ensured a stable flow and uniform deposition quality.

Deposition and fabrication of multi-layer bulk and repaired samples were carried out using a Laserdyne 430 laser metal deposition system (Prima Additive Co.). The DED setup is equipped with a 1 kW Yb laser source and a coaxial powder-delivery nozzle. For sample production, in this study, using a laser power of 660 W, nozzle travel speed of 530 mm/min, carrier gas flow rate of 6.3 l/min, and an overlapping of 50% were made.



Figure 2.2-1 Prima Additive Laserdyne 430 DED machine

2.2.1 Bulk Samples

A total of six cubic samples with a size of $15 \times 15 \times 10 \text{ mm}^3$ were fabricated directly on K340 base plates to study the bulk material behavior. These samples were deposited under identical processing parameters using both Argon- and Nitrogen-atomized powders to evaluate the effect of atomization gas on microstructure and mechanical properties. Each sample exhibited full density and uniform bead geometry suitable for subsequent heat treatment and characterization.

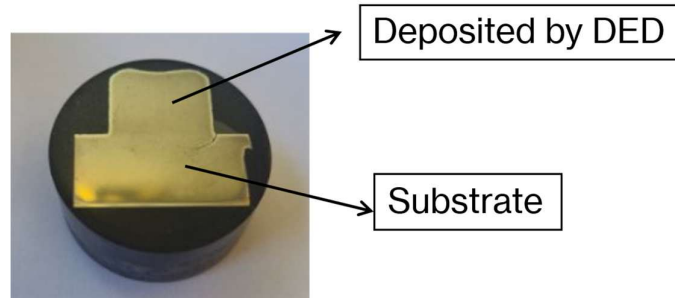


Figure 2.2-2 Cubic Bulk Sample (mounted)

Table 2.2-1 Characterizations of Bulk Samples, Atomized by Ar and N₂

Sample	C	Cr	Mo	V	Mn	Si	Al	Nb	N
Ar- Atomized	1.08	7.86	1.52	0.40	0.41	0.61	0.68	0.12	0.001
N ₂ - Atomized	1.06	8.03	1.60	0.36	0.38	0.65	0.74	0.13	0.009

2.2.2 Repair Samples

In addition to the bulk specimens, twelve samples mimicking the repairing of K340 cold-work tool steel were produced. Grooves of varying geometries, depths of 2, 3, 4, and 5 mm, and widths of 4, 5, and 8 mm, were machined into K340 substrates and subsequently refilled using the DED process with Argon as the shielding and carrier gas. The primary aim was to assess the integrity of the fusion line, dilution effects, and microstructural continuity between the deposited and base material.

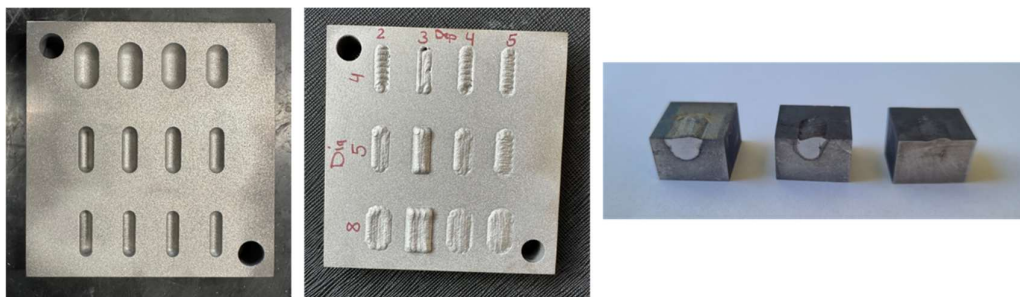


Figure 2.2-3 12 Twelve different groove dimensions before and after being filled by DED, and three repaired samples after cutting, polishing, and etching.

Table 2.2-2 Code, Dimensions of repaired samples

Code of Samples	24	25	28	34	35	38	44	45	48	54	55	58
Depth (mm)	2			3			4			5		
Width (mm)	4	5	8	4	5	8	4	5	8	4	5	8

All deposited specimens were sectioned for characterization. Standard metallographic preparation was performed for Light Optical Microscopy (LOM), Scanning Electron Microscopy (SEM), Electron BackScatter Diffraction (EBSD), and Energy-dispersive spectroscopy (EDS). Mechanical properties were assessed through microhardness measurements. This configuration allows direct comparison of how different heat treatment procedures affect the microstructural features and mechanical performance of K340 tool steel fabricated or repaired via DED.

2.3 Heat Treatment Procedures

Because of its high carbon content and multiple alloying elements, K340 tool steel exhibits a complex microstructure, consisting of primary and secondary carbides that form during solidification and evolve through subsequent heat treatments. The material's overall performance is strongly governed by the morphology, distribution, and stability of these carbides.

Therefore, precise optimization of heat-treatment parameters, including austenitizing, quenching, and tempering, is essential to control carbide refinement and transformation behavior. Properly selected conditions directly influence key mechanical properties such as hardness, fracture toughness, and resistance to abrasive wear, ensuring that the steel performs reliably under demanding service environments.

To investigate microstructural evolution, a comprehensive heat-treatment strategy was applied to both bulk and repaired samples. Heat treatments were performed using TAV MiniJet HP235 and Nabertherm LH 60/14 furnaces. The heat-treatment variables considered were austenitizing temperature, tempering temperature, and number of tempering cycles.

Two distinct heat treatment routes were established to evaluate the microstructural evolution: Direct Triple Tempering (DTT) (Figure 2.3-1) and Quenching and Triple Tempering (QTT) (Figure 2.3-2). While the DTT route subjects the as-built samples directly to the tempering cycles to relieve residual stresses, the QTT route incorporated a complete hardening phase. In this phase, samples were austenitized at 1070 °C for 30 minutes and subsequently oil-quenched to achieve uniform hardening and minimize retained austenite. Both routes shared a common triple-tempering sequence, consisting of two stages at 525 °C and a final stage at 560 °C, with each step lasting 2 hours followed by air cooling to promote carbide refinement and a stabilized martensitic structure.

The application of heat treatments varied based on the sample configuration. For the bulk specimens, a comprehensive study was conducted covering the As-built, DTT, and QTT conditions for both powder types. However, for the repaired specimens, the heat treatment strategy was specifically assigned based on the repair depth. While all repaired samples were initially analyzed in the As-built state, the DTT process was exclusively applied to the 4 mm depth samples, specifically codes 44, 45, and 48. In contrast, the QTT protocol was reserved for the 5 mm depth repair samples, including codes 54, 55, and 58, to evaluate the impact of a full re-austenitization cycle on the deeper repair zones.

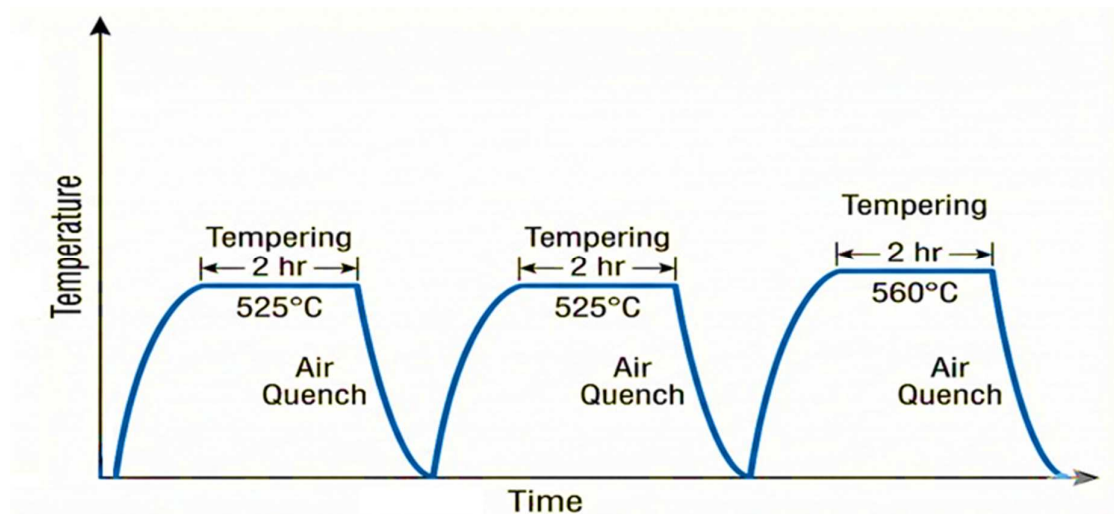


Figure 2.3-1 Schematic illustration of the DTT heat treatment recipes

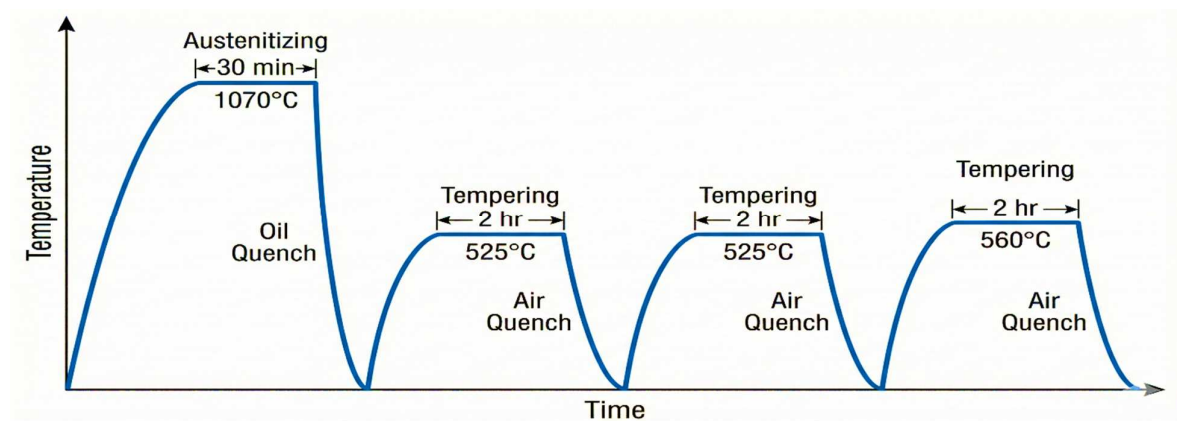


Figure 2.3-2. Schematic illustration of the QTT heat treatment recipes

2.4 Preparation Techniques

2.4.1 Metallography

Metallography relies on careful specimen preparation including cutting, mounting, grinding, polishing, etching to enable accurate observation of microstructural characteristics such as grain morphology, phase distribution, porosity, and defects. These features directly influence mechanical performance, making proper preparation essential for reliable interpretation.

For metallographic preparation, the samples were sectioned using a Brillant 220 precision cutter operated at low cutting speeds to prevent any alteration of the microstructure. The specimens were then hot mounted in conductive resin using a Remet IPA 30 mounting system.

Further surface preparation was conducted using a Mecotech Z34 and Z32 machines. Grinding began with SiC coarse abrasive papers (P320) and progressed through increasingly finer grit sizes (from #180 up to #4000 grit), with each step performed at 150–200 rpm under continuous water flow to prevent overheating. The grinding direction was alternated by 90° between steps to ensure complete removal of previous scratches. Afterward, polishing was performed at reduced speeds (50–80 rpm) using 3 µm and 1 µm diamond and colloidal silica solution with appropriate lubricants to produce a smooth, reflective and a mirror-like finish surface suitable for optical analysis.

For microstructural etching aimed at revealing austenite grains, grain boundaries, and melt pool morphology, the polished samples were immersed in a ferric chloride etchant ($HCl + FeCl_3 + H_2O$) for approximately 5 seconds, rinsed, dried, and examined under an optical microscope.

2.4.2 Characterization Techniques

A combination of optical microscopy (LOM), scanning electron microscopy (SEM), energy-dispersive spectroscopy (EDS), and stereomicroscopy were employed to characterize the fabricated specimens. These complementary techniques provided insights into surface morphology, internal structural features, and elemental composition.

2.4.2.1 Stereomicroscopy

Stereomicroscopy was performed using a Leica EZ4W to obtain an initial macroscopic assessment of the samples prior to detailed microstructural analysis. This technique provides a large field of view, making it suitable for identifying surface features such as layer boundaries, cracks, porosity, and morphology variations inherent to DED-processed K340 tool steel.

2.4.2.2 Light Optical Microscopy

A Leica optical microscope equipped with a digital imaging system and magnifications from 50× to 1000× was used for microstructural analysis. Brightfield imaging mode was used throughout the study. Samples were positioned with careful orientation, and multiple overlapping images were captured across the surface to ensure complete coverage and fine-detail visualization.

2.4.2.3 Scanning Electron Microscopy (SEM) and EDS

SEM observations were carried out using a Zeiss EVO 15 scanning electron microscope (SEM) equipped with Ultim Max EDS detector by Oxford Instruments. After venting the chamber and mounting the sample, the stage position was adjusted to ensure safe clearance. Once vacuum conditions stabilized, secondary electron (SE) and backscattered electron (BSE) detectors were used to obtain high-resolution images. SE imaging was used to reveal surface topography, while BSE imaging highlighted compositional contrast and subsurface morphology.

For higher-resolution analysis and chemical characterization, SEM was coupled with energy-dispersive X-ray spectroscopy (EDS). This allowed qualitative and semi-quantitative detection of alloying elements, identification of precipitates, and construction of elemental maps. Although EDS has spatial and sensitivity limitations, it remains a valuable tool for understanding phase chemistry and distribution.

2.4.2.4 Electron Backscatter Diffraction (EBSD)

Electron Backscatter Diffraction (EBSD) was employed to characterize the grain size distribution, grain morphology, and crystallographic orientation of the material. The analysis was carried out using an Oxford EBSD system, which determines local crystal orientation by detecting backscattered electron diffraction patterns that follow Bragg's law.

2.4.3 Hardness Measurements

Mechanical properties were characterized primarily through microhardness measurements using a Vickers hardness tester (HNVS-1000DX). The tests were carried out under standardized conditions, applying a load of 0.2 kg with a dwell time of 15 s, in accordance with ASTM E384. In addition, micro-instrumented indentation tests were conducted using force-controlled loading, holding, and unloading cycles to capture the hardness response under different processing conditions. Prior to testing, all specimens were polished to a mirror-like finish to minimize errors associated with surface roughness and to ensure reliable indentation measurements.

Micro-Vickers hardness measurements were performed along the build direction of the specimens in order to systematically cover all relevant regions, including the deposited material, the heat-affected zone, and the base material. For the repaired samples, three conditions were investigated: as-built, DTT, and QTT. For each condition, a total of 120

indentations were performed, distributed into four series of 30 measurements. In each series, half of the indents were placed in the DED region and the other half in the base material, allowing a direct and comprehensive comparison. The spacing between adjacent indentations, both within and between series, was maintained at more than three times the indentation diagonal length, corresponding to approximately 100 μm , in accordance with standard testing recommendations.

To visualize the spatial distribution of hardness, hardness maps were generated using OriginPro software. This approach enabled a clear representation of hardness gradients and local variations across the samples, highlighting the influence of the deposition process and microstructural heterogeneity on the mechanical response.

2.4.4 Thermo-Calc software

Thermo-Calc was employed to predict carbide types and phase stability because it provides thermodynamic and kinetic calculations based on the exact chemical composition of the alloy. These simulations allow the identification of stable and metastable carbide phases at different temperatures, supporting interpretation of complex microstructures in high-alloy tool steels like K340. The predicted phases and their temperature ranges were used to validate and complement the qualitative and semi-quantitative results obtained from EDS analysis. Since EDS alone cannot fully distinguish carbide types with similar elemental signatures or accurately determine phase equilibria, Thermo-Calc offers an essential theoretical reference. Together, the two techniques provide a robust and reliable framework for confirming carbide identity and understanding their evolution during heat treatment.

3 Results and Discussion: Bulk Samples

This chapter provides a detailed characterization of bulk K340 cold-work tool steel specimens produced by the Directed Energy Deposition (DED) process, with the aim of elucidating the relationship between processing conditions, microstructural evolution, and mechanical behavior. The analysis focuses on a comparative evaluation between the as-built condition and samples subjected to post-process heat treatments involving quenching and tempering.

Subsequently, the influence of heat treatment on phase transformations and carbide evolution is investigated. Elemental distribution and segregation behavior are analyzed through energy-dispersive spectroscopy combined provides insight into elemental distribution and carbide formation. This integrated approach enables a systematic understanding of how microstructural features variations in DED-fabricated K340 tool steel.

3.1 Macroscopic Analysis

A macroscopic examination of the specimens was first carried out using stereomicroscopy to assess surface features and large-scale defects in the printed parts. This method allowed the detection of surface-related imperfections, including lack-of-fusion areas, surface irregularities, and other defects visible at the millimeter scale. Representative stereomicroscope images of all six samples are shown in Figure 3.1-1, providing an overall view of cross sections of the additively manufactured components.

From a macrostructural perspective, distinct melt pool boundaries and HAZ regions are clearly visible in the as-built and tempered conditions, reflecting the layered nature of the DED process. In contrast, the quenched-and-tempered condition leads to a more homogeneous microstructure, in which the layer bands and HAZ boundaries are no longer discernible due to extensive phase transformation and diffusion during heat treatment.

Macroscopic examination reveals cracking in the base material of all samples, with cracks consistently initiating at the interface of the first deposited layer. The origin of these cracks is mainly associated with the severe thermal gradients imposed on the substrate at the beginning of the DED process. The combination of high laser energy input and the relatively low thermal conductivity of K340 tool steel causes rapid surface heating against a cold bulk, generating high tensile thermal stresses that exceed the local strength of the material. In addition, rapid self-quenching in the substrate HAZ promotes the formation of brittle, untempered martensite, further increasing the susceptibility to crack initiation.

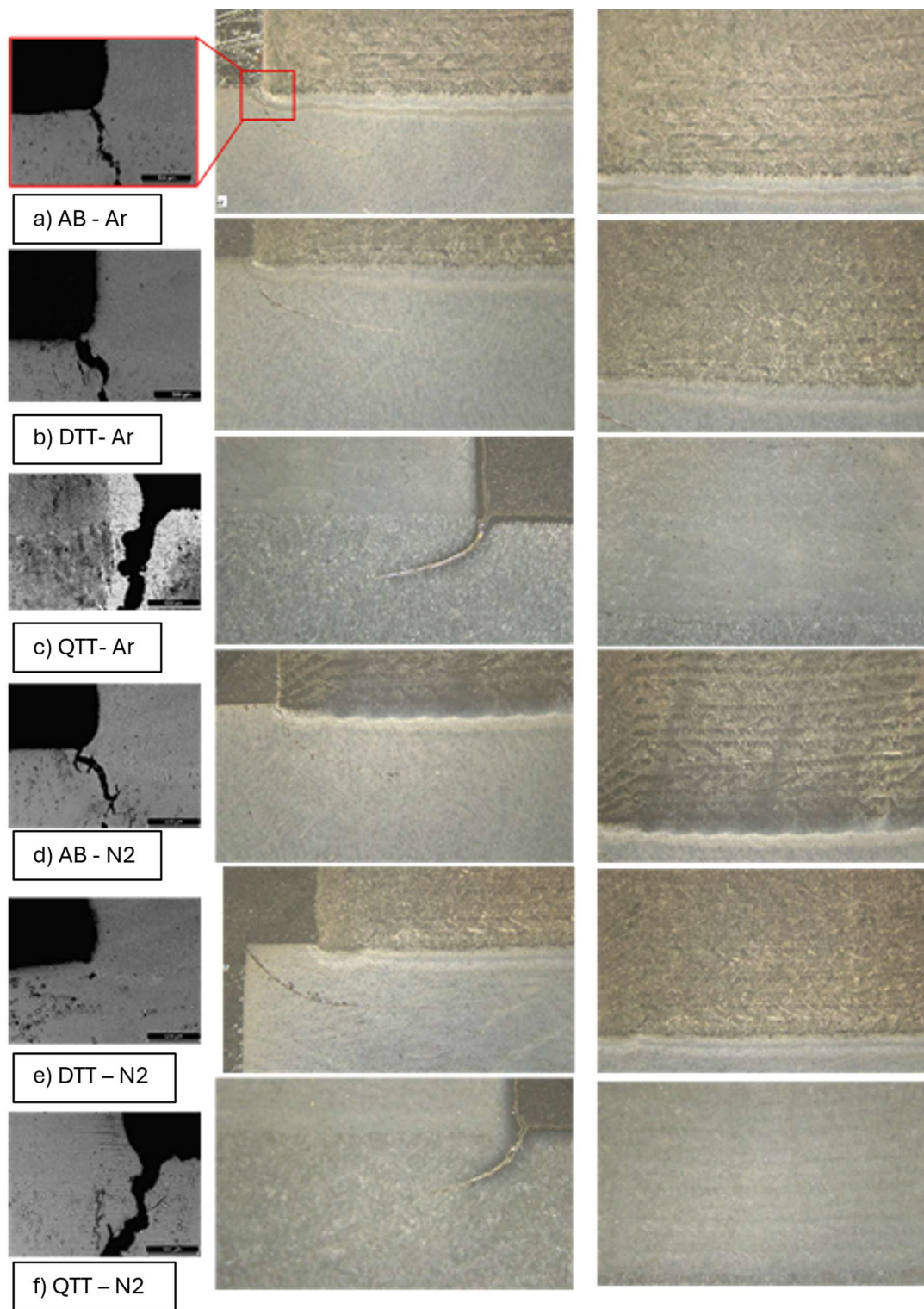


Figure 3.1-1 Stereomicroscope(x30) and microscopic(x50) macrographs of the cross-sections of the Bulk samples related both N2-and Ar- Atomized a) AB-Ar. b) DTT-Ar c) QTT-Ar d) AB-N2 e) DTT-N2 f) QTT-N2

Post-process heat treatment appears to promote the propagation of pre-existing cracks through the redistribution of residual and thermal stresses. Since hardened K340 tool steel has relatively low fracture toughness and high notch sensitivity, these stresses tend to concentrate at geometric discontinuities, particularly near the groove corners, allowing existing cracks to extend further into the substrate. Although most cracks initiate from the groove corner, the DTT-N2 specimen shows an apparent shift in the crack initiation location, which is most likely related to the sectioning plane rather than a different cracking mechanism.

3.2 Microstructural analysis of As-Built DED samples

In the DED process, the thermal history of the deposited material plays a decisive role in defining microstructural characteristics such as grain morphology and size. This influence arises from the inherently high heating and cooling rates, steep temperature gradients, and the progressive increase in bulk temperature during deposition. Achieving optimal mechanical performance, therefore, requires careful control of these thermal conditions[83].

The microstructural cross-sectional optical micrographs of the as-built K340 samples fabricated under Argon (AB-Ar) and Nitrogen (AB-N2) shielding gas atmospheres, are illustrated in figure 3.2-1. The low-magnification images reveal a layer-by-layer structure with a regular and overlapping melt pool morphology with distinct arc-shaped interfaces, which indicate stable processing conditions and consistent track deposition during fabrication.

In terms of densification, the AB-Ar sample displays a highly dense structure with no significant macroscopic defects. In contrast, while the AB-N2 specimen maintains a generally high relative density, localized porosities are scattered within the deposited layers. Upon increasing the magnification, the internal grain structure becomes apparent in both conditions; the microstructure is dominated by columnar prior austenite grains that have grown epitaxially across the fusion boundaries, extending continuously along the build direction.

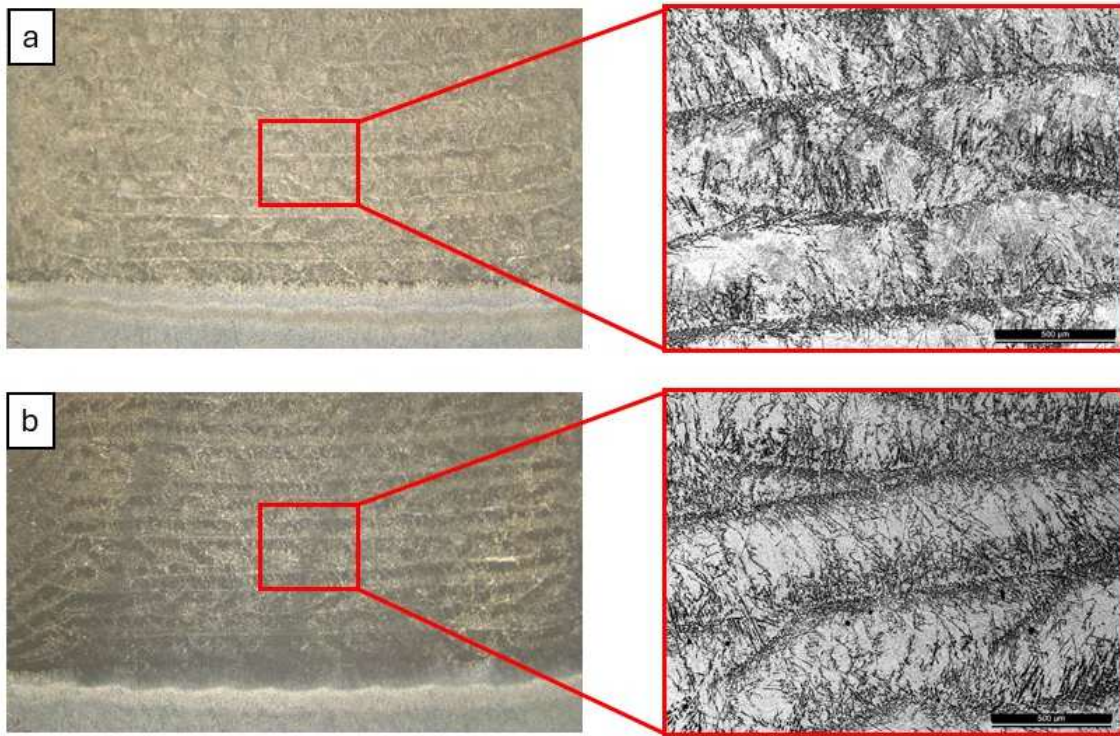


Figure 3.2-1. LOM micrographically of layers of As-Built Samples a) Ar-atomized (x50) b) N2-Atomized (x50)

Upon increasing the magnification in figure 3.2-2, the internal structure of melt pool becomes apparent, displaying a transition from the fusion line boundary toward the center of the deposited tracks. High-magnification imaging further exposes a refined cellular-dendritic substructure encased within the primary grains, suggesting a rapid solidification mode typical of laser-based additive manufacturing.

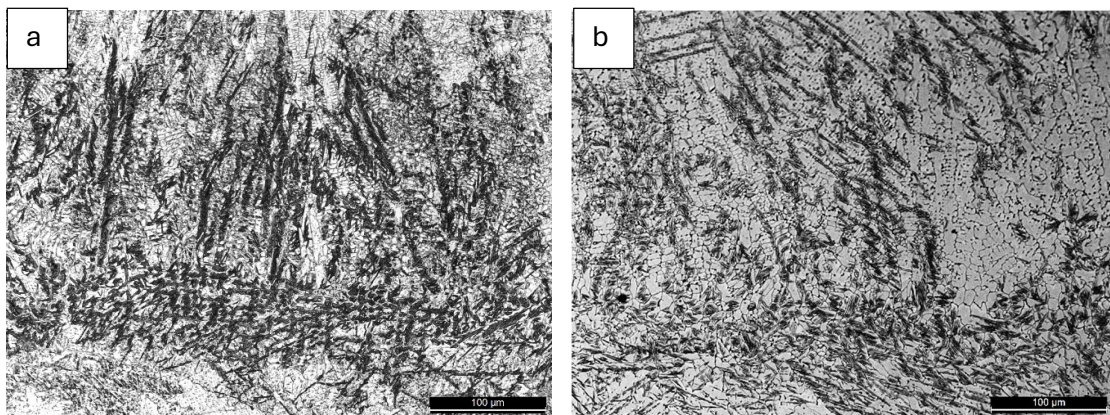
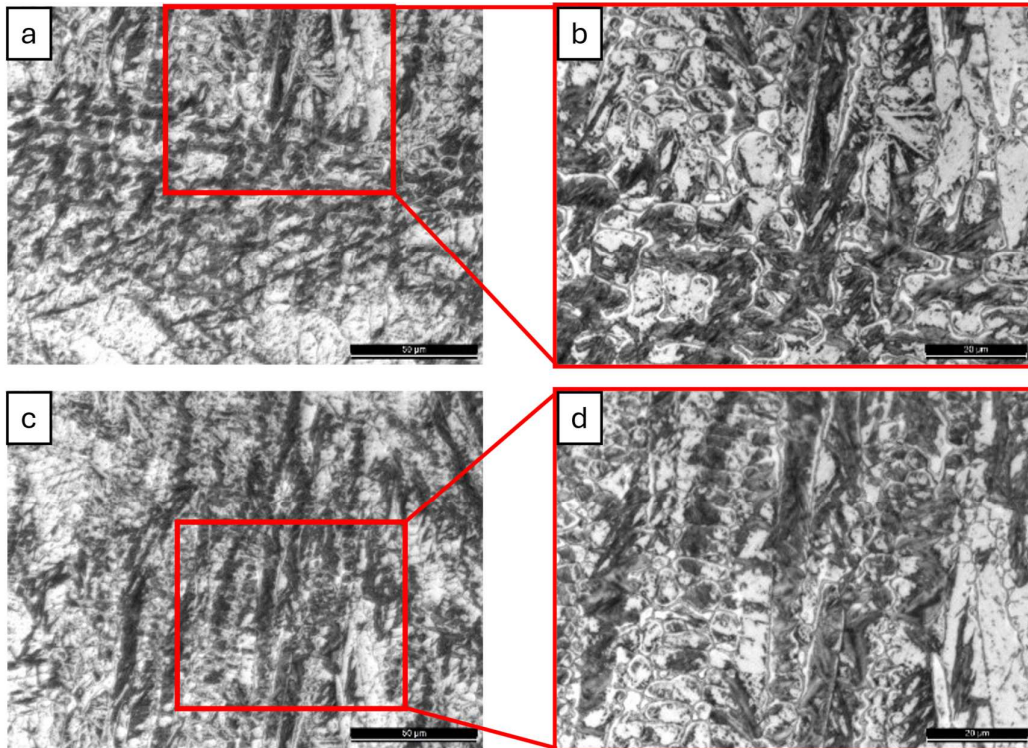


Figure 3.2-2. LOM micrographically of a layer of As-Built samples a) AB-Ar b) AB-N2 (x200)

When a molten pool solidifies on a solid substrate of previous layer and under steep thermal gradients, epitaxial growth of columnar grains tends to grow directly from the substrate into the liquid, where the solidifying material inherits the crystallographic orientation of the underlying solid. Consequently, columnar prior austenite grains (PAGs) nucleate at the fusion boundary and grow preferentially along the build direction, parallel

to the maximum heat flow, perpendicular to the local solid–liquid interface[11]. In the early stages of solidification, different grains compete, and those whose orientations are better aligned with the direction of heat flow grow faster and remain, while the others gradually disappear[4]. The columnar grains do not grow straight but gradually curved as the thermal field evolves, not only because of the heat source in additive manufacturing is continuously moving, the direction of heat flow changes during deposition, but also the solid–liquid interface becomes more curved, this curvature due to the high laser power and low scanning speed of DED (Figure 3.2-3) [4], [11], [56], [84], [85].



*Figure 3.2-3 The curvature of columnar grains (AB-Ar)
a, b) After fusion boundary (x500, x1000) c, d) Middle of layer (x500, x1000)*

A schematic representation of this mechanism, shown in figure 3,2-4, illustrates the evolution of columnar grain structures during solidification on a solid substrate, where the dashed line denotes the fusion boundary.

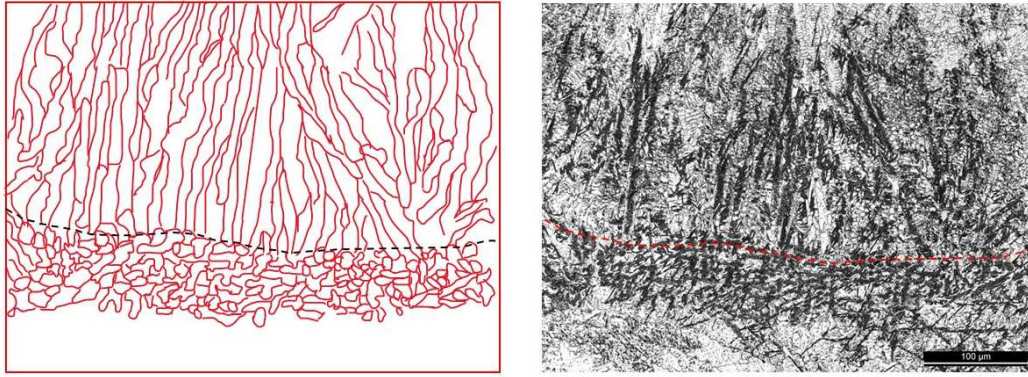


Figure 3.2-4. Sketch texture of evolution of columnar grain structures

Generally, the solidification behavior of the melt pool is governed by the thermal gradient at the solid–liquid interface (G) and the solidification growth rate R . The specific morphology of these grains is dictated by the G/R ratio; Near the substrate, heat is efficiently extracted through the base material, resulting a high ratio at the melt pool bottom favors cellular growth, appears comparatively fine and locally equiaxed. In contrast, toward the upper part of the building, the effective thermal gradient decreases due to the increasing distance from the substrate and the accumulation of thermal energy, promotes constitutional supercooling ahead of the solidification front, leading to interface instability and the development of columnar–dendritic structures oriented along the build direction through Ostwald ripening, a phenomenon common in heat-affected zones near the melt boundary. [56], [58], [86], [87]. As shown in the SEM image in Figure 3.2-5 and 3.2-6, the As-Built K340 tool steel exhibits both cellular and dendritic microstructures formed during directional solidification in the DED process.

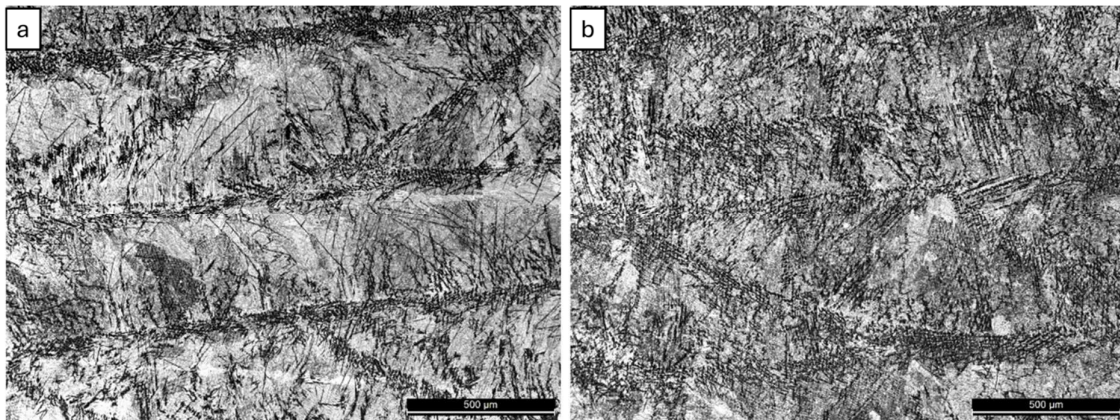


Figure 3.2-5 LOM of Solidification behavior of melt pool a) Near substrate, b) Toward upper part (x50)

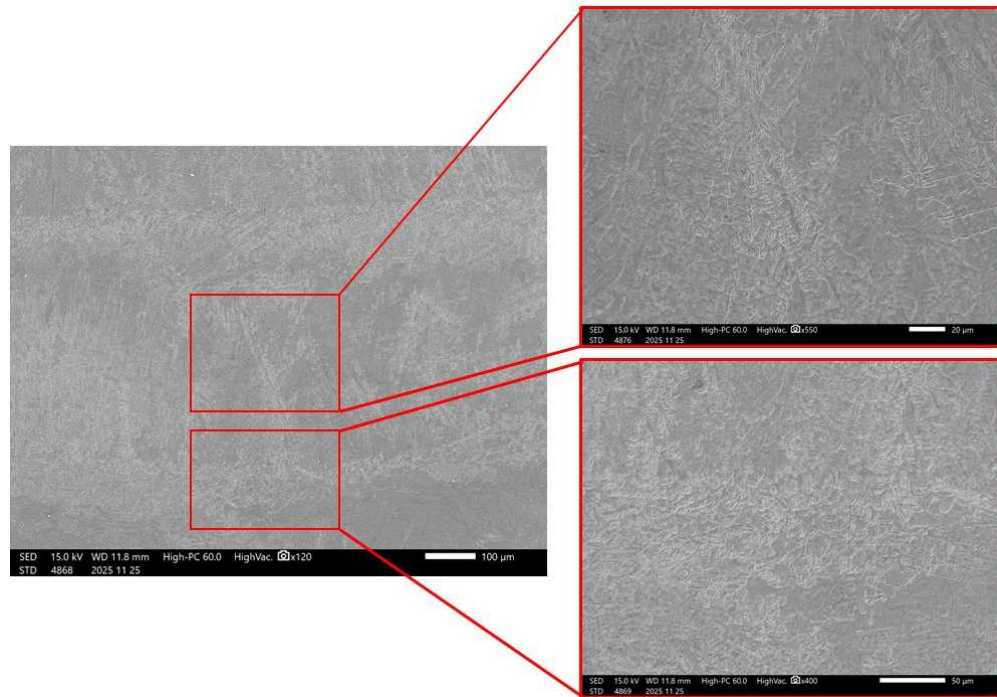


Figure 3.2-6 SEM of Solidification behavior of melt pool

In the LOM of figure 3.2-7, the effect of Inter-layer Heat Treatment (IHT) is visible as a subtle change in contrast within the previously deposited layers. Just below the interface with the newly deposited layer, the tempered martensite appears slightly lighter and less strained compared to the central part of the layer. This indicates that the upper portion of the previous layer has experienced localized reheating and tempering due to the deposition of the next layer. The underlying martensitic laths are partially recovered, fragmented, or rearranged, forming intragranular subdivisions that resemble small equiaxed regions, while the central part of the layer retains its original, highly strained appearance. This observation confirms that thermal cycling during DED primarily affects the upper part of each layer, leaving the lower portion closer to the substrate less altered [69], [88], [89].

The non-uniform microstructure in DED-fabricated parts is mainly related to the complex thermal history during processing, including rapid cooling and the intrinsic heat treatment (IHT) caused by repeated reheating from subsequent layers and passes. Zheng et al [90] . reported that different regions of the component experience different thermal cycles. In particular, the bottom and middle sections are exposed to more reheating compared to the top layers, leading to slower cooling rates and consequently different microstructures and mechanical properties. Similarly, Jandric et al [90]. Showed that the thermal conditions vary significantly along the build direction.

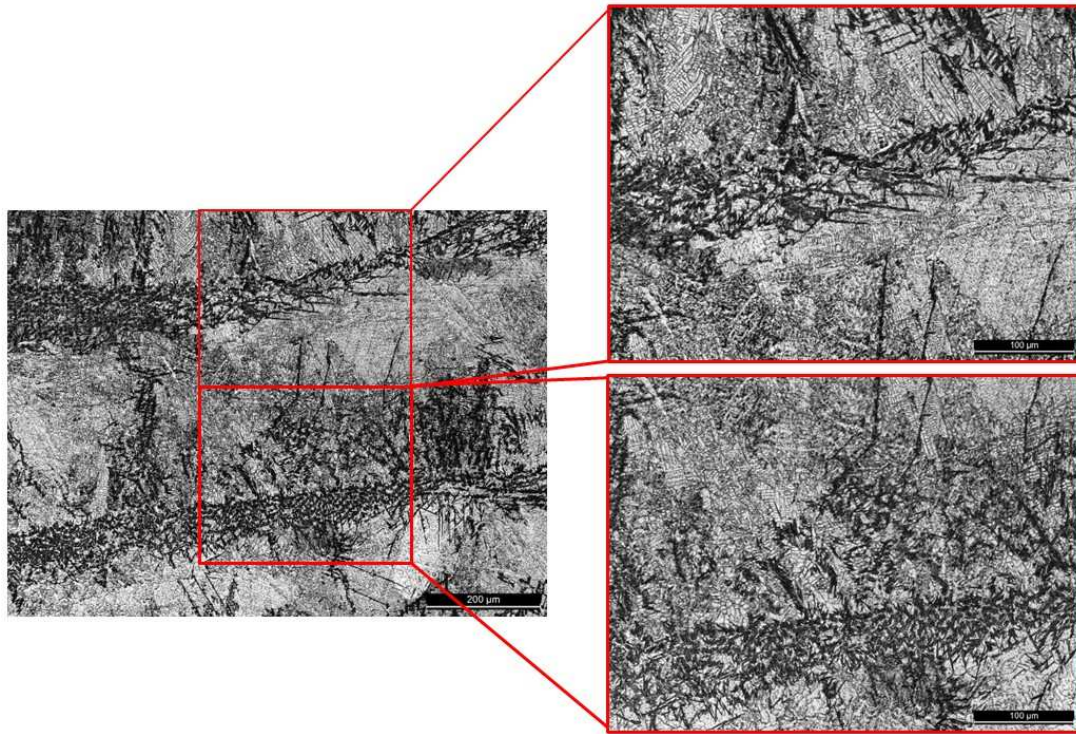


Figure 3.2-7 Effect of Inter-layer Heat Treatment (IHT)

3.3 Phase Transformation of As-Built DED samples

Previous sections described how DED process influences the micro- and macrostructure of K340 tool steel during solidification. However, after solidification, the material continues to cool and undergoes solid-state phase transformations. In addition, reheating previously deposited layers during DED can further change the phases and modify residual stress and distortion. This section therefore focuses on the solid-state phase transformations that occur during DED processing.

Figures 3.3-1 present SEM and LOM images of the microstructure of DED-fabricated K340 tool steel in the as-built condition. The material exhibits a hierarchical microstructure. After etching, At the microscale, SEM observations reveal a martensitic substructure embedded within retained austenite (RA), together with a cellular–dendritic solidification morphology and chemically distinct interdendritic regions.

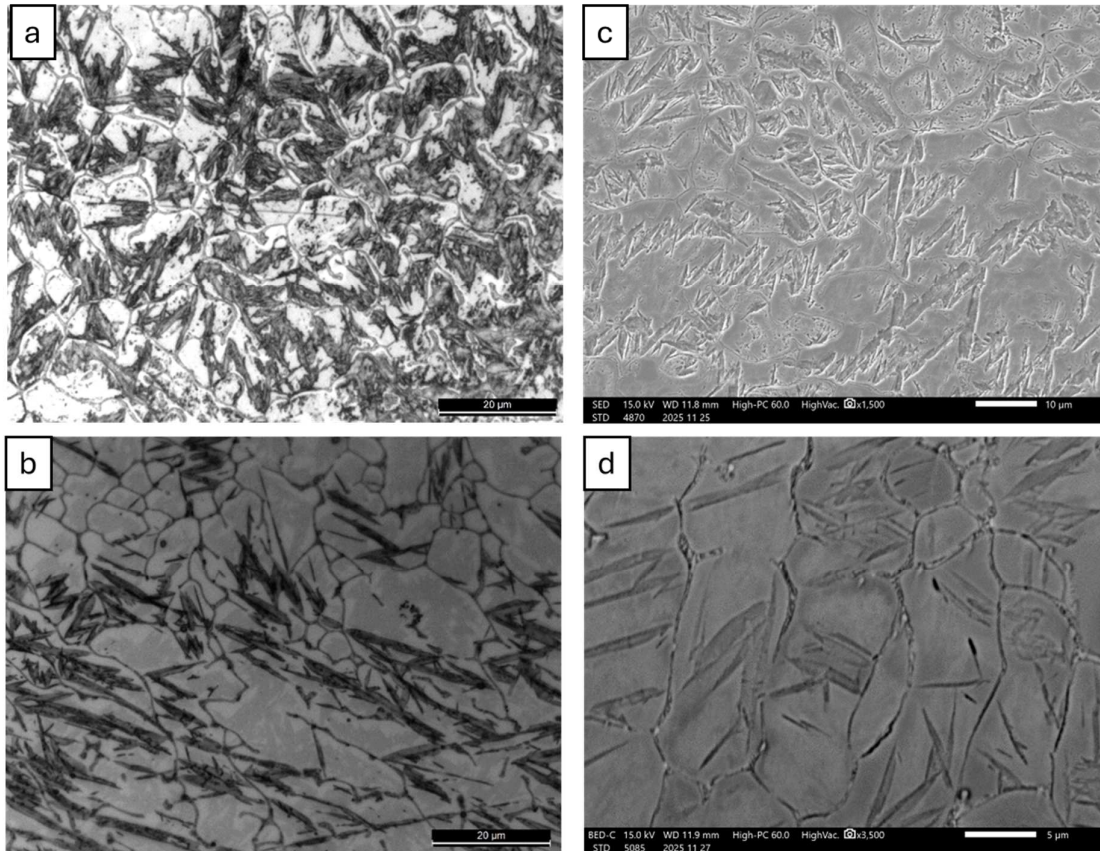


Figure 3.3-1 SEM & LOM of fine martensite blocks were surrounded by retained austenite. A,c) AB-Ar b,d)AB-N2

The prior austenite grains are relatively coarse due to the solidification conditions during the DED process. Within this solidification microstructure, the martensitic phase transformation takes place during cooling. As a result, the martensitic lath blocks, appearing as dark zigzag needle-like features, predominantly form inside individual solidification cells. In many regions, these martensitic blocks are confined within the cells by retained austenite (RA) located along the cell boundaries, which appears as bright regions in etched cross-sections, (Figure 3.3-1). The presence of retained austenite at these boundaries is associated with local chemical segregation during rapid solidification, presence of a continuous, solute-rich interdendritic network composed of eutectic carbides and retained austenite, continuous narrow lines along the solidification cell boundaries, as observed in the etched micrographs, which stabilizes the austenite and limits the growth of martensite across the cell interfaces. However, this confinement is not uniform throughout the material, and in some areas martensitic blocks may partially extend beyond the cell boundaries (Figure 3.3-1, 3.3-2) [4], [91]. Unlike low-alloyed systems where martensitic laths may traverse multiple dendritic boundaries.

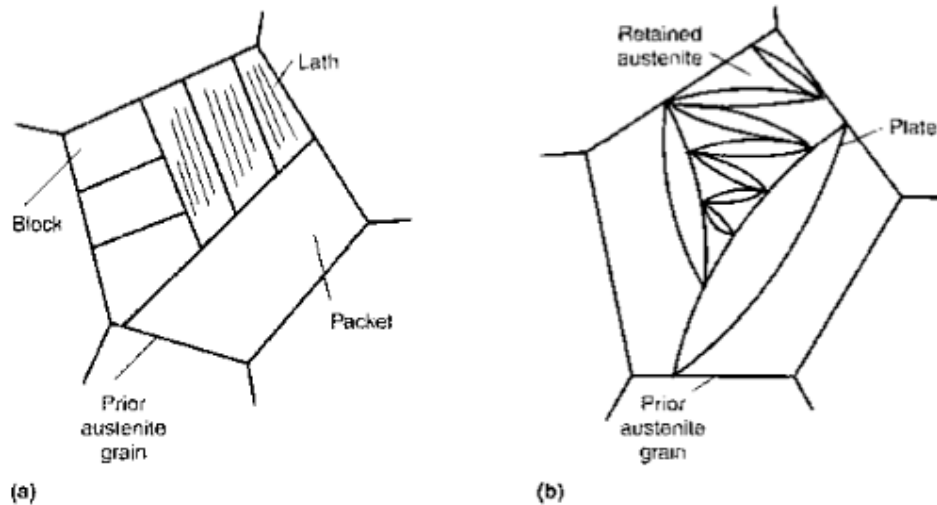


Figure 3.3-2 Morphology of a) Lath martensite b) Plate martensite [11]

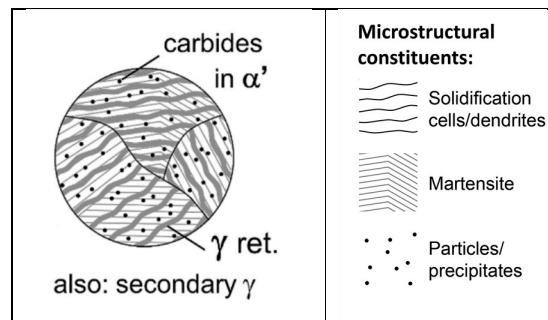


Figure 3.3-3. Schematic representation of the microstructure of tool steels manufactured by AM [4]

Beyond the morphological variations, the rapid solidification inherent to the DED process also has a pronounced effect on solute redistribution and carbide formation within the microstructure (Figure 3.3-3). Due to the extremely high cooling rates, alloying elements do not have sufficient time to diffuse uniformly during solidification, which leads to a higher number of secondary dendrite arms. At the same time, solute elements are progressively rejected from the growing dendrites and accumulate in the interdendritic regions, increasing local chemical inhomogeneity, where they lower the local melting temperature and promote the formation of low-melting eutectic carbides. Due to the limitations of diffusion time these carbides cannot become rounded, which explains why long, rod-like carbides are often observed in the interdendritic areas[9], [64].

This rapid solidification also causes chemical segregation and nonequilibrium composition that results in dual-phase distribution forming in different regions. The dendritic regions, which solidify first and are relatively poor in alloying elements, tend to transform into martensite during cooling. In contrast, the interdendritic regions solidify later and become enriched in elements that stabilize austenite, such as carbon and other alloying elements. As a result, austenite can remain stable at room temperature in these interdendritic zones, while martensite forms mainly within the dendrites[9], [64].

Although EDS analysis provides the elemental composition needed to distinguish different carbides, the microstructural appearance gives further confirmation. The type and distribution of carbides in table 3.3-1[7], the compositional characteristics of alloy carbides in tool steels (Table 3.3-2) and the Fe–Cr–C invariant reactions summarized in table 3.3-3 [17]. The shape and morphology of the precipitates are also important for identifying the carbide phases. Compact and blocky particles are typically different from continuous or semi-continuous film-like carbides formed along grain boundaries. For example, the change from primary blocky carbides inside the matrix to elongated carbide networks at the grain boundaries is associated with the transformation from MC or M_7C_3 carbides to secondary $M_{23}C_6$ carbides [92]. This observation supports the thermodynamic predictions related to solidification and heat-treatment processes.

Table 3.3-3-1 The type and distribution of carbides [7]

Type	Element distribution in M	
	Most	Least
M_3C	Fe, Mn, Cr	W, Mo, V
$M_{23}C_6$	Cr	Mo, V, W
M_6C	Fe, Mo, W	Cr, V, Co
M_7C_3	Cr	...
M_2C	W, Mo	Cr
MC	V	W, Mo

Table 3.3-2 Characteristics of alloy carbides in tool steels[17]

Type of carbide	Lattice type	Remarks
M3C	Orthorhombic	A carbide of the cementite (Fe_3C) type. M may be Fe, Mn, Cr, with a little W, Mo, V.
M7C3	Hexagonal	Mostly found in Cr alloy steels. Resistant to dissolution at higher temperatures. Hard and abrasion resistant. Found as a product of tempering high-speed steels.
M23C6	FCC	Present in high-Cr steels and all high-speed steels. The Cr can be replaced with Fe to yield carbides with W and Mo.
M6C	FCC	A W- or Mo-rich carbide. May contain moderate amounts of Cr, V, Co. Present in all high-speed steels. Extremely abrasion resistant.
M2C	Hexagonal	W- or Mo-rich carbide of the W_2C type. Appears after temper. Can dissolve a considerable amount of Cr.
MC	FCC	V-rich carbide. Resists dissolution. Small amount that does dissolve reprecipitates on secondary hardening.

Table 3.3-3 Invariant liquidus reactions in the C-Cr-Fe system [17]

System	Temperature (°C)	Reaction
C-Fe	1493	$L + \delta\text{Fe} \leftrightarrow \gamma\text{Fe}$
	1153	$L \leftrightarrow \gamma\text{Fe} + \text{C}$
C-Cr	1490	$L \leftrightarrow (\text{Cr}) + \text{Cr}_{23}\text{C}_6$
	1520	$L + \text{Cr}_{7}\text{C}_3 \leftrightarrow \text{Cr}_{23}\text{C}_6$
	1780	$L + \text{Cr}_3\text{C}_2 \leftrightarrow \text{Cr}_7\text{C}_3$
	1895	$L + \text{C} \leftrightarrow \text{Cr}_3\text{C}_2$
Cr-Fe	None	
C-Cr-Fe	1275	$L + \delta\text{Fe} \leftrightarrow \text{M}_{23}\text{C}_6 + \gamma\text{Fe}$
	1255	$L + \text{M}_{23}\text{C}_6 \leftrightarrow \text{M}_7\text{C}_3 + \gamma\text{Fe}$
	1160	$L + \text{M}_7\text{C}_3 \leftrightarrow \text{M}_3\text{C} + \gamma\text{Fe}$
	1156	$L + \text{M}_3\text{C} \leftrightarrow \gamma\text{Fe} + \text{C}$
	1230	$L + \text{M}_7\text{C}_3 + \text{C} \leftrightarrow \text{M}_3\text{C}$
	1585	$L + \text{M}_3\text{C}_2 \leftrightarrow \text{M}_7\text{C}_3 + \text{C}$

The as-built microstructure shows carbides enriched in chromium and molybdenum and in third place rich in vanadium. Instead of a continuous eutectic network, which is sometimes reported in similar tool steels, the microstructure is mainly characterized by chains of interdendritic carbides distributed along the cellular or dendritic boundaries.

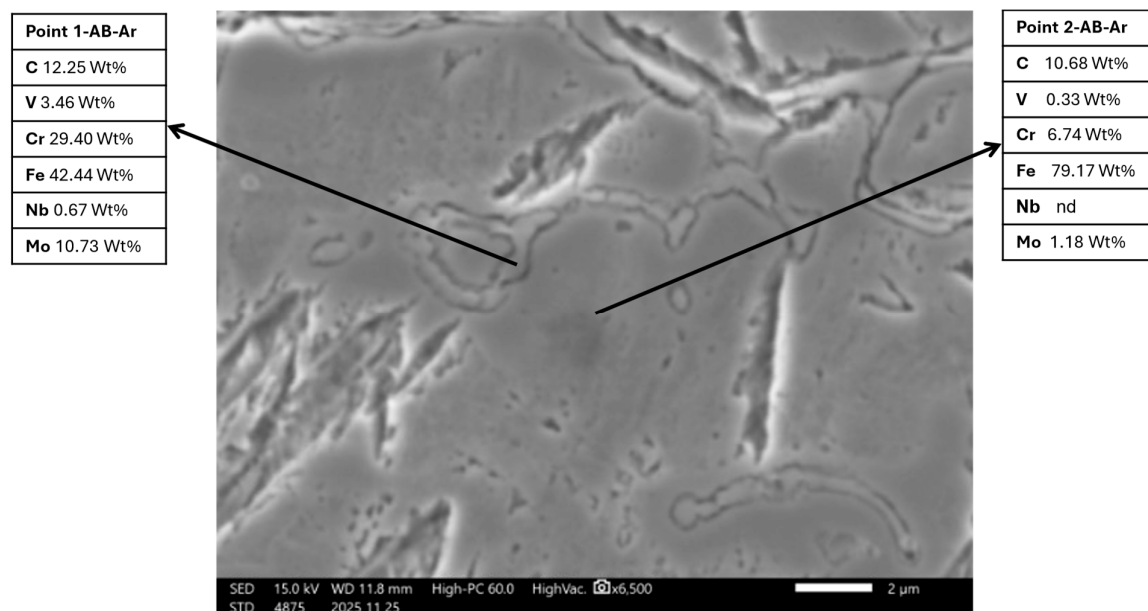


Figure 3.3-4 EDS Point analysis of AB-Ar sample

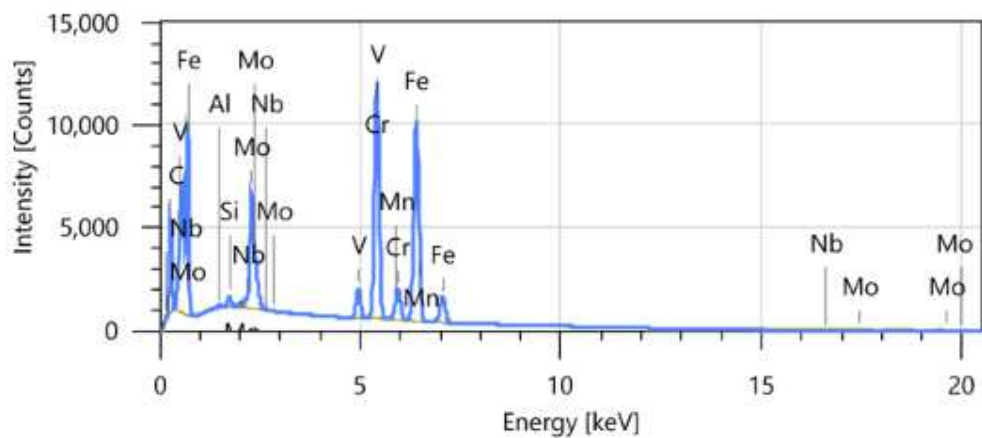


Figure 3.3-5 EDS of Point 1-AB-Ar

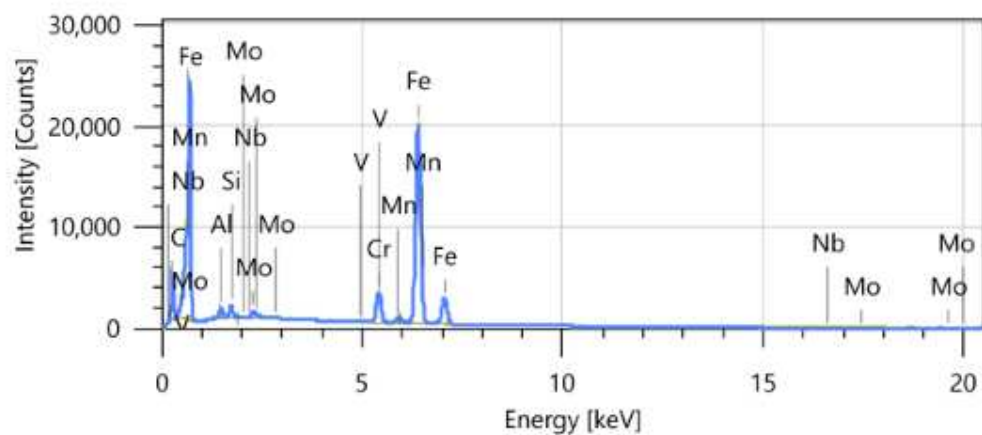


Figure 3.3-6 EDS of Point 2-AB-Ar

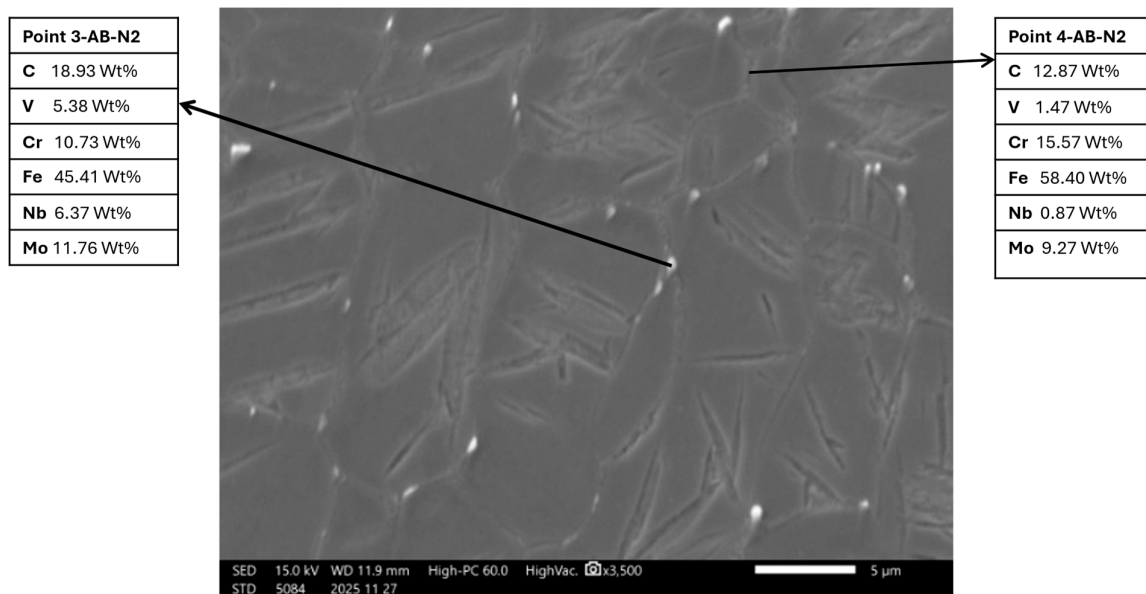


Figure 3.3-7 EDS Point analysis of AB-N2 sample

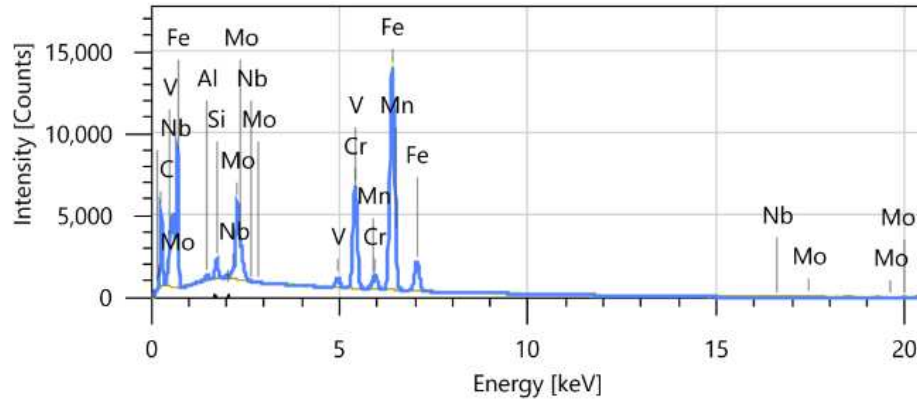


Figure 3.3-8 EDS of Point 4-AB-N2

SEM-EDS mapping and point analyses (Figures 3.3-4 to 3.3-8, Table 3.3-4) reveal the presence of at least two chemically distinct carbide populations in the as-built DED-processed K340 steel. After solidification, coarse carbides tend to decorate the cell or dendrite boundaries, forming partial networks (as points 1 and 4 in Figures 3.3-4 and 3.3-7, respectively), while finer carbides are also observed within the dendritic regions (Point 3-AB-N2, Figure 3.3-7).

Table 3.3-4 EDS of Element distribution in As Built Bulk sample

Point	C	AL	Si	V	Cr	Mn	Fe	Nb	Mo
1-AB-Ar	12.25	0.06	0.3	3.46	29.4	0.69	42.44	0.67	10.73
2-AB-Ar	10.68	0.77	0.74	0.33	6.74	0.39	79.17	nd	1.18
3-AB-N2	18.93	0.4	0.73	5.38	10.73	0.3	45.41	6.37	11.76
4-AB-N2	12.87	0.24	0.79	1.47	15.57	0.52	58.4	0.87	9.27

According to the type and distribution of carbides in table 3.3-1, to the compositional characteristics of alloy carbides in tool steels (Table 3.3-2) and to the Fe–Cr–C invariant reactions summarized in Table 3.3-3, Comparing the martensitic matrix (Point 2-AB-Ar: Fe-rich, ~79 wt% Fe and ~6–7 wt% Cr) with coarse carbides in the cell or dendrite boundaries (Point 1-AB-Ar and Point 4-AB-N2), such carbides exhibit pronounced enrichment in Cr (≈ 15 –29 wt%) together with significant Mo (≈ 9 –11 wt%) and minor V, such Cr-rich are primary carbides, most likely the M_7C_3 type, which are known to form during solidification in high-C, high-Cr cold-work tool steels. Their location along interdendritic regions and cellular/dendritic boundaries further supports their identification as primary solidification carbides formed due to microsegregation of Cr and C during rapid directional solidification in the DED process. For the nominal composition of K340 (~8 wt% Cr, ~1.1 wt% C), the solidification path in the Fe–Cr–C system involves the eutectic-type reaction $L \rightarrow \gamma + M_7C_3$ within the temperature range of approximately 1160–1255 °C, which is consistent with the observed interdendritic carbide morphology [17], [19], [69].

In contrast, finer intragranular particles (Point 3-AB-N2) show a different chemical signature, characterized by noticeable enrichment in Nb (~6 wt%), V (~5 wt%), and Mo (~11 wt%) enrichment in addition to moderate Cr levels (~10 wt%). Based on the known carbide-forming tendencies of transition elements (Table 3.3-3), Nb and V have a strong affinity for carbon and preferentially form MC-type carbides. Therefore, these fine particles are more reasonably interpreted as complex Nb/V-rich MC-type carbides rather than simple Cr_7C_3 . The absence of ideal stoichiometric compositions is attributed to the multicomponent alloy chemistry and the limited accuracy of EDS for carbon quantification. Furthermore, the rapid solidification and repeated thermal cycling inherent to the DED process promote microsegregation and non-equilibrium carbide formation, preventing full homogenization and leading to the coexistence of Cr-rich primary carbides and Nb/V-rich complex carbides in the as-built microstructure[17], [19], [69].

The elemental map (Figure 3.3-9) obtained from the AB-N2 (as same as AB-Ar) region shows a matrix-dominated composition with slight enrichment of Cr and Mo compared to the bulk matrix values, indicating interdendritic segregation rather than a continuous carbide phase. No significant Nb enrichment is observed at the map scale, suggesting that Nb-containing carbides are localized and discrete rather than forming extended networks.

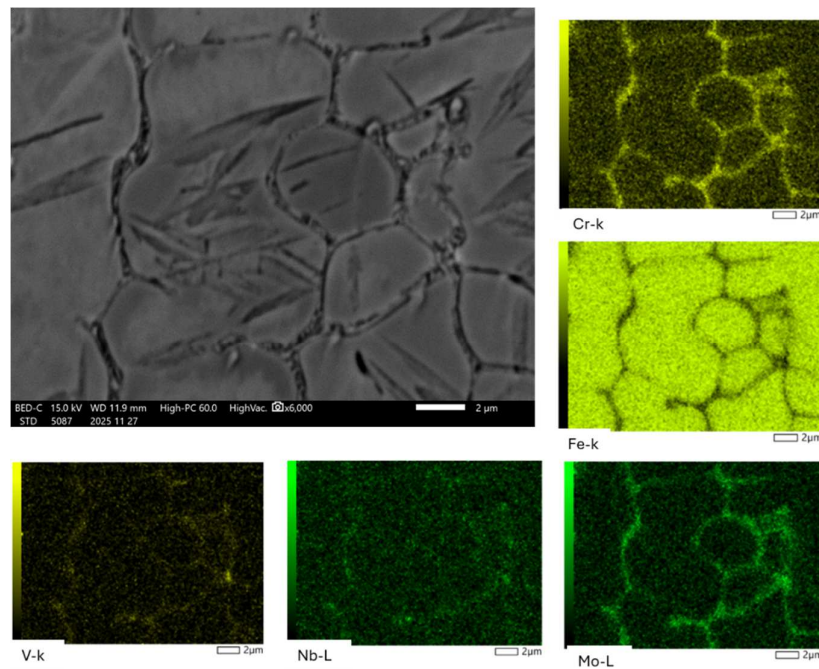


Figure 3.3-9 MAP analysis of AB-N2 sample

3.4 Effect of Heat Treatment on Microstructure and Carbide Evolution of DED Samples

Following the detailed characterization of the as-built microstructure of the DED-processed K340 tool steel, the focus is now shifted toward evaluating the effect of post-processing heat treatments. Although the as-built microstructure represents the inherent microstructural development and thermal history associated with the DED processing, it is anticipated that further heat treatment processes will drastically influence the phase structure, carbide structure, and microstructural stability. Hence, the objective of the next section will be to conduct systematic studies for understanding the microstructural development associated with various heat treatment procedures.

3.4.1 Microstructural Response to post processing of heat treatment

It is an accepted fact that the main problem in the heat treatment of tool steels is to find a proper compromise in hardening and strengthening on the one side, and preserving toughness and eliminating retained austenite on the other. In order to increase toughness and minimize retained austenite, high strength tool steels should be tempered several times at relatively high temperatures. The effect of such a procedure will be a transformation of retained austenite and development of tougher tempered martensitic structure. Retained austenite is detrimental since it lowers toughness, hardness and fatigue strength, and in addition, may transform into martensite during its use. As a result, the wear resistance decreases, dimensional stability becomes poor due to volume changes, and local stresses appear due to differences in phase volumes.[24], [93], [94].

For this reason, common heat treatment routes are applied, conventional quenching and tempering treatment and triple tempering heat treatment. The austenitizing temperature is carefully selected to prevent excessive dissolution of primary carbides, which would increase the carbon content of the austenite. Too much carbon leads to the formation of very hard but brittle martensite and increases the amount of retained austenite after quenching[24].

Figures 3.4-1(a-d) illustrate the microstructural response of the K340 specimens, fabricated using Argon- and Nitrogen-atomized feedstocks, to two distinct heat treatment strategies: Direct Triple Tempering (DTT) and Quenching and Triple Tempering (QTT). The layer-wise nature of the DED process remains visible after both heat treatment processes, conducted under gas-shielded atmospheres. In the QTT condition, the fusion boundaries between successive deposited layers become significantly less pronounced, resulting in a more homogeneous and integrated microstructure across the build height. In contrast, DTT condition also exhibits overall refinement and partial homogenization; however, subtle microstructural heterogeneities remain visible, indicating an approximately homogeneous but still locally differentiated microstructure.

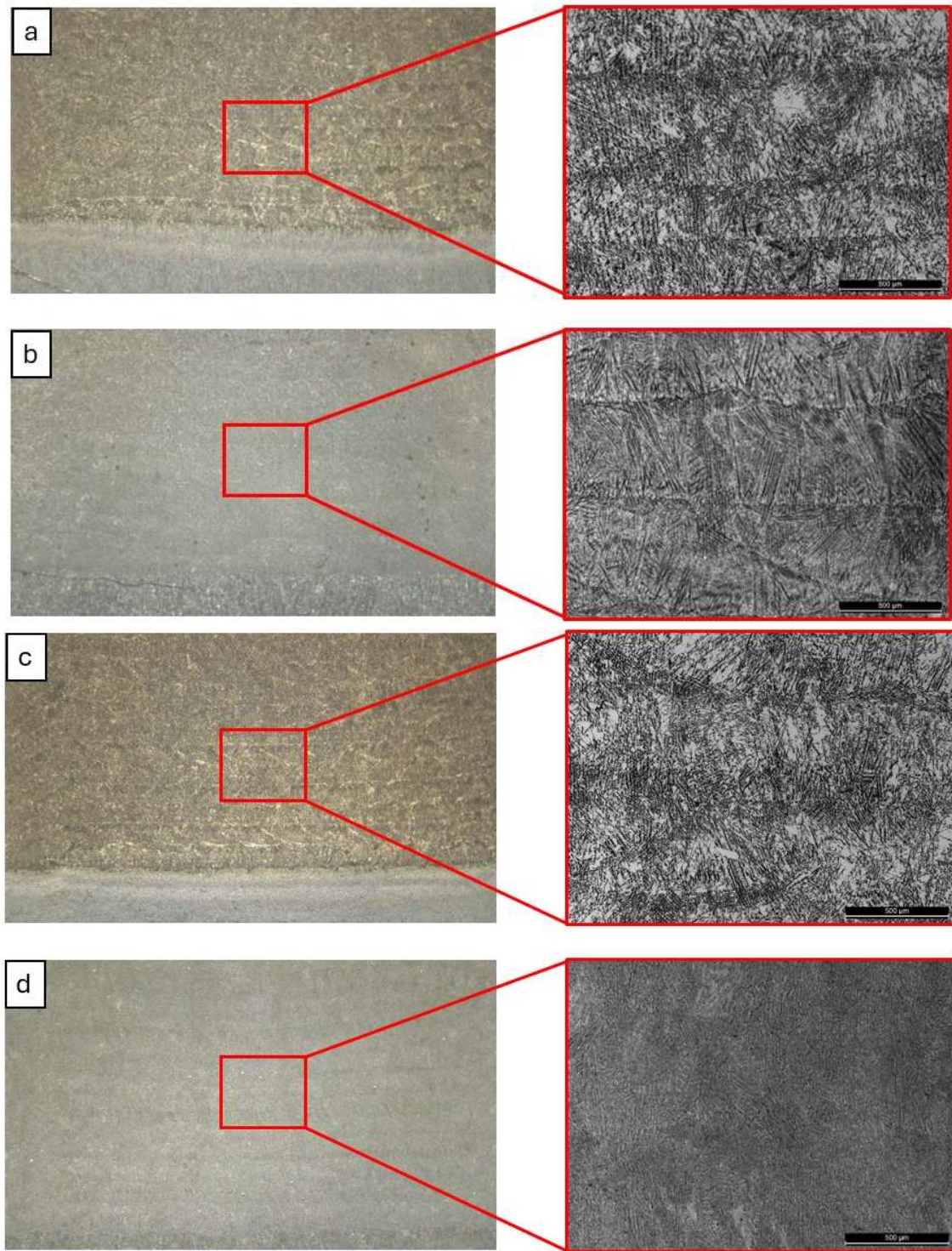


Figure 3.4-1 Stereomicroscope and LOM of Microstructural of heat treatment bulk samples
a)DTT-Ar b)QTT-Ar c)DTT-N2 d)QTT-N2

In addition, like the as-built samples fabricated under nitrogen shielding, the presence of porosity is also observed in the heat-treated specimens. This confirms that the detected pores are primarily process-induced during deposition and are not eliminated or generated by the subsequent heat treatment routes.

In the Direct Triple Tempered (DTT) condition, both argon- and nitrogen-atomized DED samples exhibit a very similar chemical and microstructural response. No significant differences in elemental distribution or carbide chemistry were detected between the two shielding atmospheres.

The DTT can be classified as a controlled softening heat treatment, is conducted at temperatures below the A_{c1} temperature and therefore does not induce any type of austenitization. Figure 3.4-2(a-d) is shown LOM with higher magnification after triple tempering, under these conditions, the as-built martensitic matrix undergoes stress relief and partial recovery, accompanied by a gradual reduction of residual stress, partial decomposition of retained austenite (RA), and carbide (C) precipitation. However, since the temperature remains below A_{c1} , the primary solidification-induced architecture of the DED process, including layer boundaries, cellular/dendritic features, and epitaxial columnar morphology, is largely preserved [69], [94], [95].

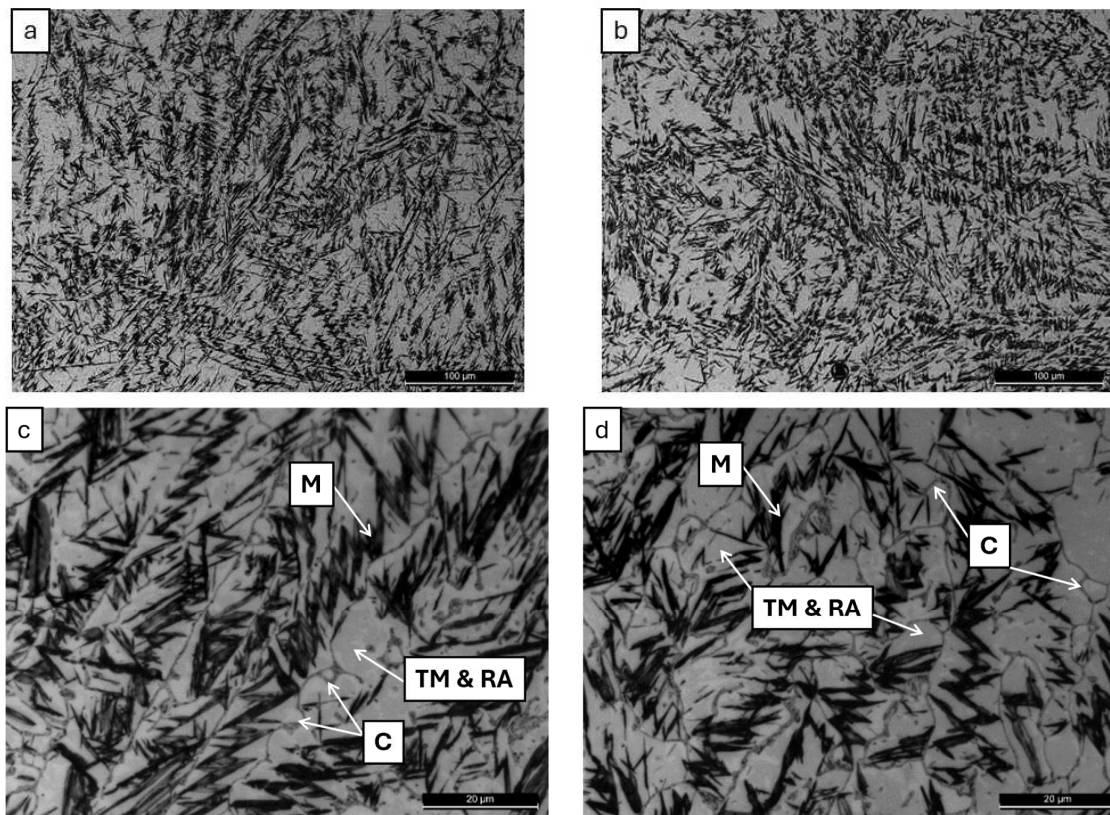


Figure 3.4-2 LOM of microstructural of DTT samples x200 & x1000 a, c) DTT-Ar b, d) DTT-N2
M; Martensite, TM: Tempered Martensite, RA: Retained Austenite, C: carbide

The martensitic lath structure becomes more distinct and uniformly distributed, while the previously observed circular or chaotic features at layer interfaces disappear. The bright matrix corresponds to a tempered martensitic (TM)–retained austenite (RA) mixture, within which faint linear features represent residual segregation traces or fine tempering-induced carbides rather than solidification boundaries. Consequently, the

microstructure after triple tempering remains structurally similar to the as-built condition, with phase heterogeneity and layer-wise features still clearly distinguishable.

In the LOM images, due to the limited spatial resolution of LOM, very fine carbides cannot be clearly distinguished and are only manifested as local light–dark contrast variations within the tempered martensitic matrix.

In the SEM images (Figure 3.4-3), martensitic laths appear as elongated dark plates, grouped into blocks with similar crystallographic orientation. Fine carbides are observed as bright particles and thin films, preferentially precipitated along lath and block boundaries.

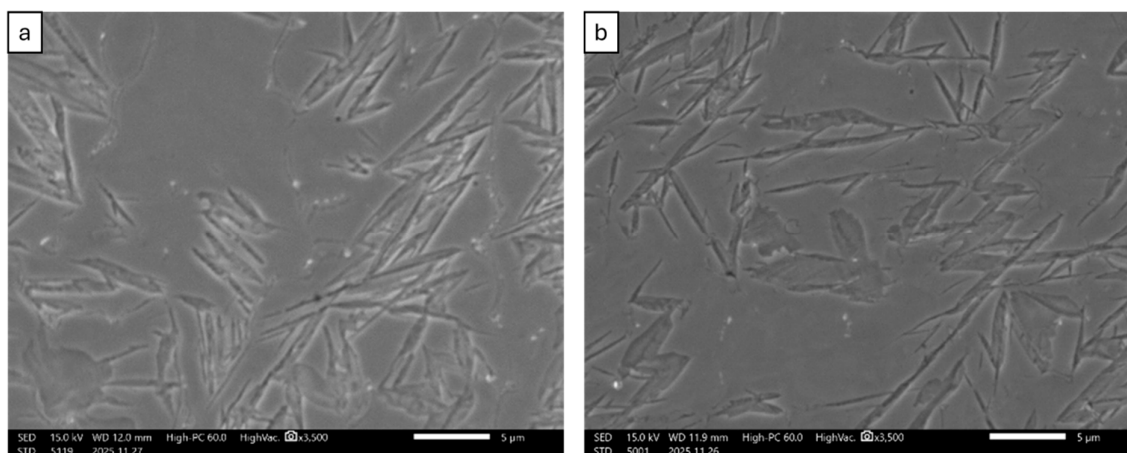


Figure 3.4-3 SEM of Microstructure of DTT bulk samples a)DTT-Ar b)DTT-N2

In the Quenched and Triple Tempered (QTT) condition, both argon- and nitrogen-atomized samples exhibit a generally similar microstructural appearance. In figure 3.4-4 related samples after Quenching and Triple Tempering routes, it is clear that the DED-processed K340 steel exhibits a predominantly tempered lath martensitic microstructure. Generally, after QTT the austenite will be transformed into tempered martensite[56]. The martensite is organized into well-defined blocks and packets within each deposited layer, indicating significant stress relaxation and microstructural homogenization compared to the as-built and triple tempering conditions. The prior cellular/dendritic solidification features are no longer evident, although the interlayer fusion boundaries remain visible due to the layer-wise thermal history.

The lighter-etching background primarily corresponds to tempered martensite with reduced internal strain, while a limited amount of retained austenite may persist as thin films between martensitic laths. Fine, dispersed carbides precipitated during tempering are also observed, contributing to refinement of the microstructure and influencing the stability of the remaining retained austenite. Due to the high alloying content in cold-work tool steels, the martensite start (Ms) and martensite finish (Mf) temperatures are reduced. As a result, the transformation of austenite to martensite during quenching to room temperature is incomplete, leaving a certain amount of retained austenite (RA) in

the microstructure. The as-quenched structure therefore consists mainly of martensite, undissolved primary carbides, and sometimes bainite. For this reason, multiple tempering treatments are applied to reduce retained austenite, improve toughness, and promote alloy carbide precipitation. In addition, the tempering temperature strongly affects the stability of retained austenite and the precipitation behavior of carbides within tempered martensite, thereby controlling the final mechanical properties [19], [94], [95].

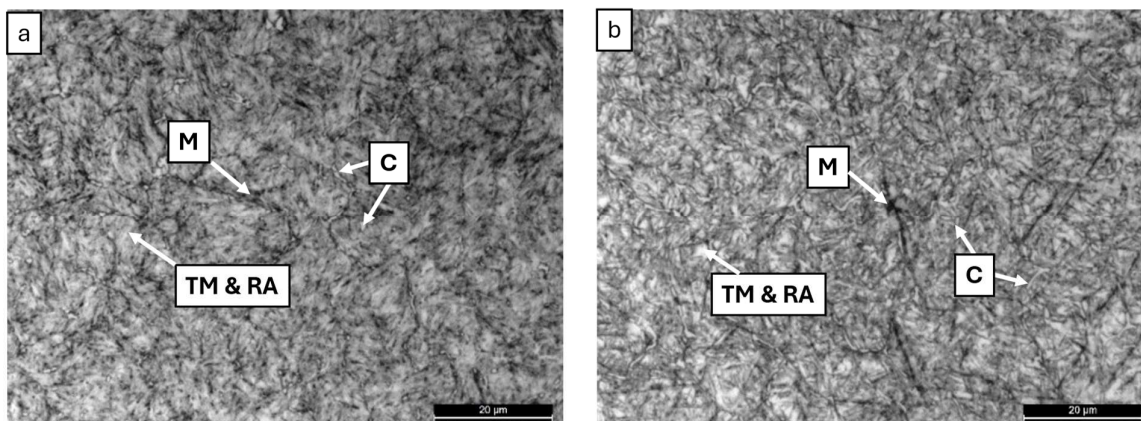


Figure 3.4-4 LOM of microstructural of QTT samples x1000 a) QTT-Ar b) QTT-N2
M; Martensite, TM: Tempered Martensite, RA: Retained Austenite, C: carbide

Optical microscopy (LOM) reveals a relatively homogeneous structure across the deposited layers. Chain-like bright features are observed within the matrix, which are associated with carbide-rich regions formed during quenching and subsequent tempering. Due to the limited spatial resolution of LOM, very fine carbides precipitated within martensitic laths cannot be individually resolved and instead appear as local light-dark contrast variations within the matrix.

SEM observations (Figure 3.4-5) provide clearer insight into the microstructural constituents. In SEM images, the chain-like features observed in LOM are more distinctly resolved and appear as carbide collection or carbide-decorated regions. Additionally, fine bright particles are visible, which are attributed to secondary carbides precipitated during tempering.

The matrix exhibits a dense network of fine, interwoven lath-like features, characteristic of tempered martensite formed after full austenitization and quenching. Compared to the triple-tempered condition, the QTT microstructure shows a higher degree of homogenization, with no clear remnant of the original DED solidification morphology.

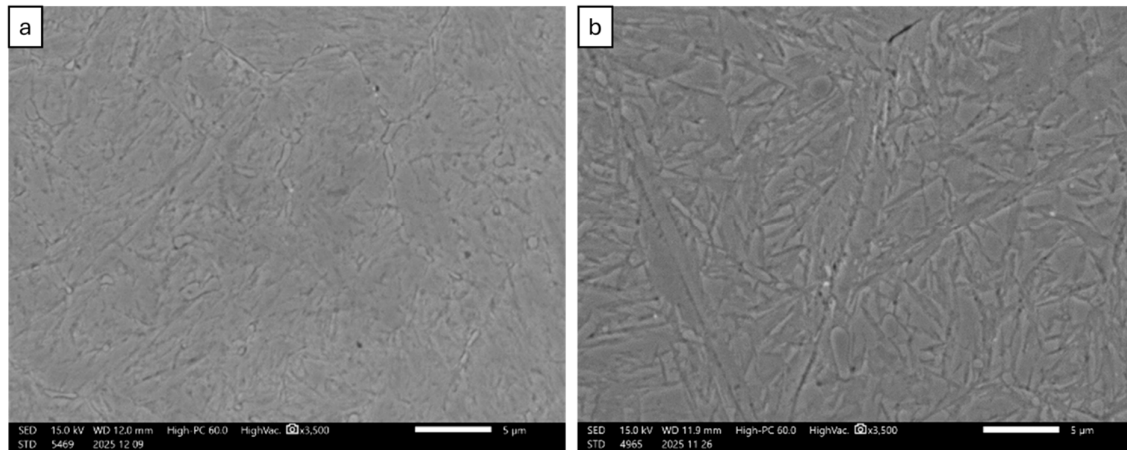


Figure 3.4-5 SEM of Microstructure of QTT bulk samples a)QTT-Ar b)QTT-N2

The QTT treatment involves heating above the A_{c3} temperature, leading to complete austenitization and an almost full microstructural reset. This process eliminates the original solidification morphology and dissolves a significant fraction of the previously formed carbides, resulting in a chemically and structurally homogenized austenite prior to quenching. Upon cooling, a new martensitic microstructure forms, governed primarily by local chemical composition rather than inherited solidification features. As a result, the fusion lines and dendritic boundaries become indistinct, and the microstructure appears significantly more homogeneous [56], [69].

When oil quenching is applied, the high cooling rate causes austenite to transform into martensite. However, the martensite finish temperature is lower than room temperature, part of the austenite does not transform and remains as retained austenite in the microstructure. Elemental segregation at grain boundaries or in interdendritic regions further lowers the martensite start temperature, which promotes the formation of retained austenite. In etched microstructures, retained austenite typically appears as brighter regions[19], [93], [94].

While QTT eliminates the solidification-induced architecture and resets microstructure, the subsequent martensite formed through conventional quenching and tempering exhibits limited refinement and weaker texture due to the indirect and diffusion-assisted decomposition of retained austenite[96].

Although quenching and tempering promotes an overall homogenization of the microstructure, the upper deposited layers still exhibit clearly distinguishable martensitic lath and block morphologies. This is attributed to the asymmetric thermal history inherent to the DED process, where the upper layers experience fewer reheating cycles during fabrication, resulting in higher retained solidification-induced strain and more pronounced martensitic substructures after QTT [56].

Under QTT conditions, samples processed with nitrogen shielding show better microstructural uniformity compared to those processed under argon. This improvement

is linked to the role of nitrogen as an interstitial element, which helps stabilize austenite and reduces the visible effects of compositional gradients after transformation. However, this benefit comes at the cost of increased microporosity, caused by nitrogen absorption in the melt pool and its limited solubility during the rapid solidification characteristic of the DED process [69], [84].

3.4.2 Evolution and Redistribution of Carbides during Heat Treatment

Following the microstructural characterization of the heat-treated DED samples, the discussion is extended to the evolution of carbides induced by the applied heat treatment routes. In K340 tool steel, carbides are strongly affected by both quenching and tempering parameters and play a key role in controlling martensitic morphology, retained austenite stability, and overall microstructural homogeneity. Therefore, a focused discussion on carbide evolution is essential to fully interpret the microstructural changes observed after quenching and tempering.

In case of DTT, according to point-wise EDS chemical analysis was performed on selected microstructural features, as marked in figures 3.4-6 and 3.4-7, and the corresponding quantitative results are summarized in table 3.4-1 and reference tables 3.3-1 to 3.3-3 and shape of carbides. The analyses reveal the presence of chemically distinct regions within the tempered microstructure.

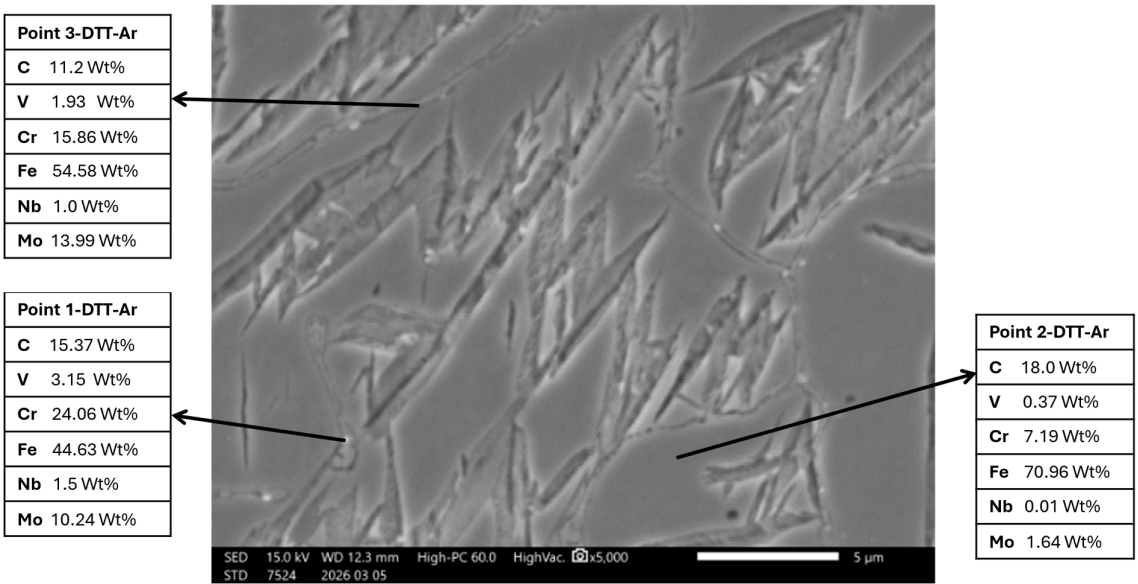


Figure 3.4-6 EDS Point analysis of DTT-Ar Bulk sample

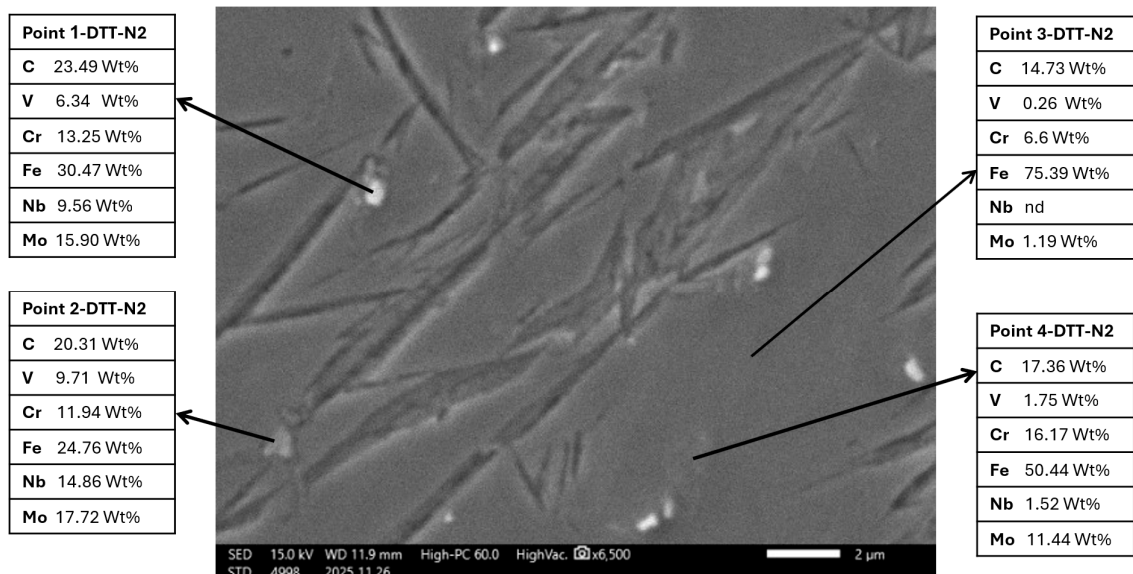


Figure 3.4-7 EDS Point analysis of DTT-N2 Bulk sample

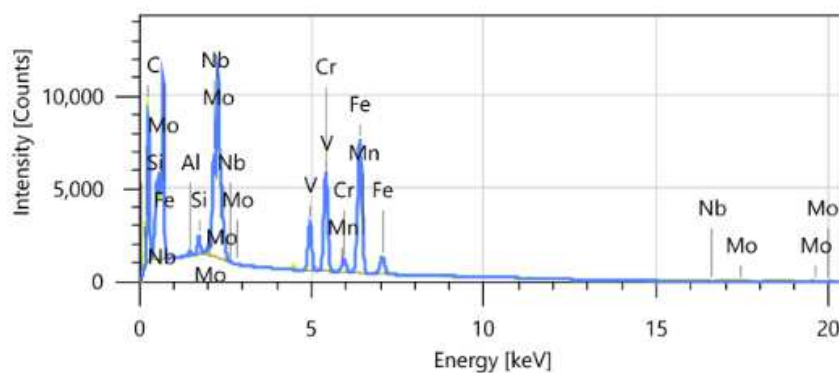


Figure 3.4-8 EDS of Point 1-DTT-N2

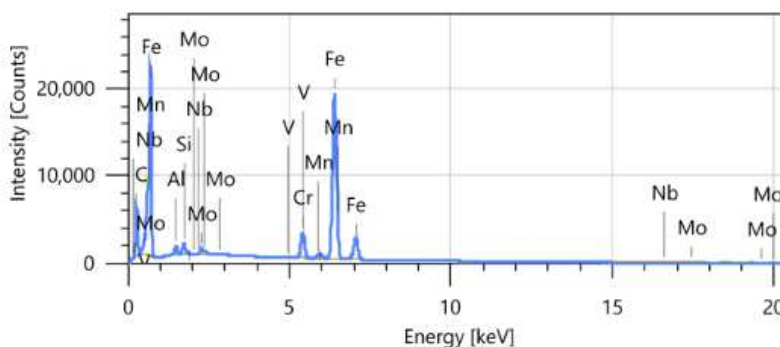


Figure 3.4-9 EDS of Point 3-DTT-N2

By comparison between received data and tables 3.3-1 to 3.3-3, and evaluating the geometric shape of the precipitates, a clear correlation can be established to identify the carbide phases. Specifically, in the DTT-Ar sample, Point 1 exhibits a pronounced enrichment in Chromium (~24 wt.%) and Molybdenum (~10.24 wt.%). When coupled with

its approximately massive and blocky shape, this chemical signature strongly pointates the presence of primary M_7C_3 carbides. In contrast, point 3 (DTT-Ar), despite also being enriched in Cr and Mo, displays a distinct elongated, film-like morphology. Based on this film shape and its specific alloying proportions, this feature can be conclusively identified as secondary $M_{23}C_6$ carbides that precipitated along the boundaries. On the other hand, point 2 (DTT-Ar) is heavily dominated by Iron (~71 wt.%) with minimal alloying content, which confirms it represents the tempered martensitic matrix region.

A similar phase identification principle applies to the DTT-N2 sample; points showing substantial enrichment in strong carbide-forming elements like Vanadium and Niobium (such as Points 1 and 2), while lacking high Cr content, display a compact geometry characteristic of primary MC carbides. Lastly, point 4 (DTT-N2) exhibits a Cr-rich signature (~16 wt.%) with an elongated network, correlating again with the $M_{23}C_6$ type.

Table 3.4-1 EDS of Element distribution in DTT Bulk sample

Point	C	AL	Si	V	Cr	Mn	Fe	Nb	Mo
1-DTT-Ar	15.37	0.13	0.37	3.15	24.06	0.54	44.63	1.5	10.24
2- DTT-Ar	18.00	0.67	0.79	0.37	7.19	0.38	70.96	0.01	1.64
3- DTT-Ar	11.20	0.20	0.66	1.93	15.86	0.58	54.58	1.00	13.99
1- DTT-N2	23.49	0.16	0.57	6.34	13.25	0.27	30.47	9.56	15.9
2- DTT-N2	20.31	0.12	0.38	9.71	11.94	0.21	24.76	14.86	17.72
3- DTT-N2	14.73	0.77	0.76	0.26	6.60	0.30	75.39	nd	1.19
4- DTT-N2	17.36	0.13	0.67	1.75	16.17	0.52	50.44	1.52	11.44

Elemental EDS maps (Figure 3.4-10, 3.4-11) further confirm this interpretation by showing localized clustering of chromium, manganese, and molybdenum at specific sites, particularly along martensitic lath and block boundaries with their film-like shape indicating $M_{23}C_6$ secondary carbides and partly massive shape showing M_7C_3 primary carbides. In addition, some bright spots in Nb are probably MC primary carbides. These elemental accumulations coincide with the bright features observed in SEM images, confirming that they represent carbide-rich regions rather than retained austenite or etching artifacts.

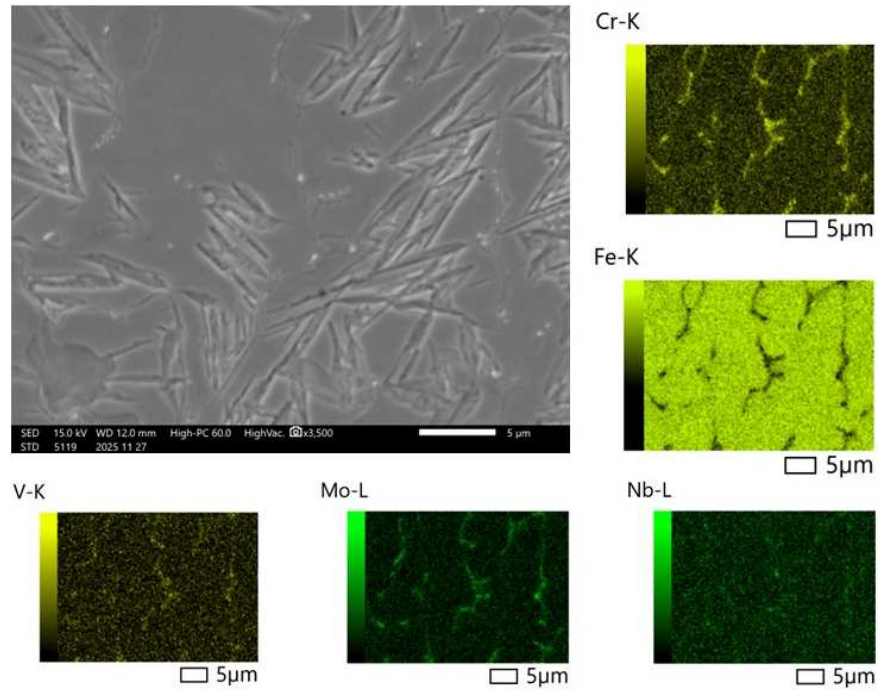


Figure 3.4-10 MAP of DTT-Ar Bulk Sample

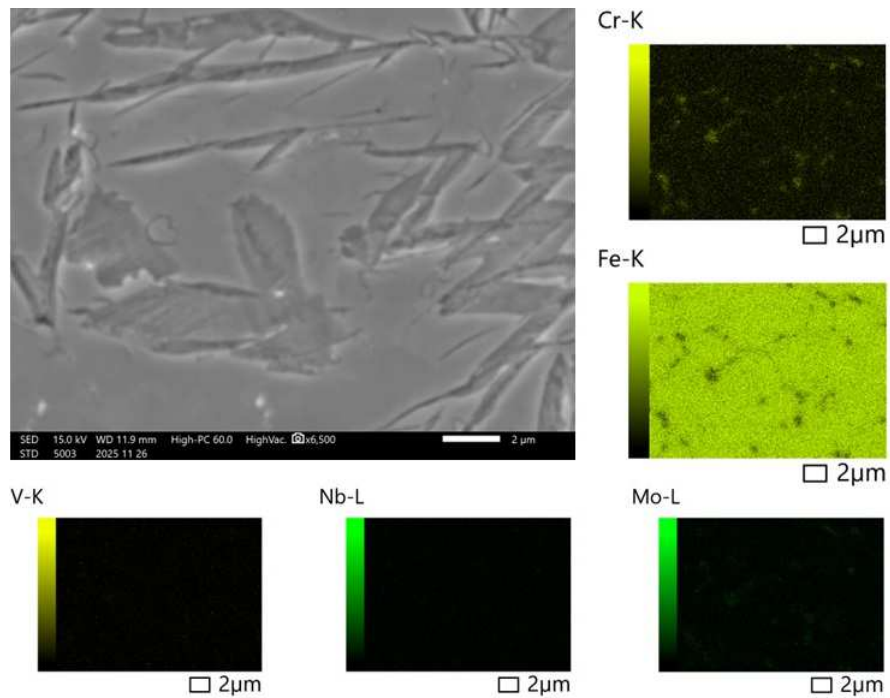


Figure 3.4-11 MAP of DTT-N2 Bulk sample

A similar analytical approach was applied to the Quenched and Triple Tempered (QTT) samples. As in the tempered condition, no meaningful differences were observed between argon- and nitrogen-atomized specimens in terms of chemical composition or elemental partitioning.

Point EDS analyses conducted in selected regions, indicated in Figures 3.4-12 and 3.4-13 and reported in Table 3.4-2, reveal the presence of chemically heterogeneous carbide populations. Several analyzed particles are enriched mainly in chromium, while others show additional enrichment in molybdenum. These compositional variations indicate the formation of different types of alloy carbides during quenching and subsequent tempering.

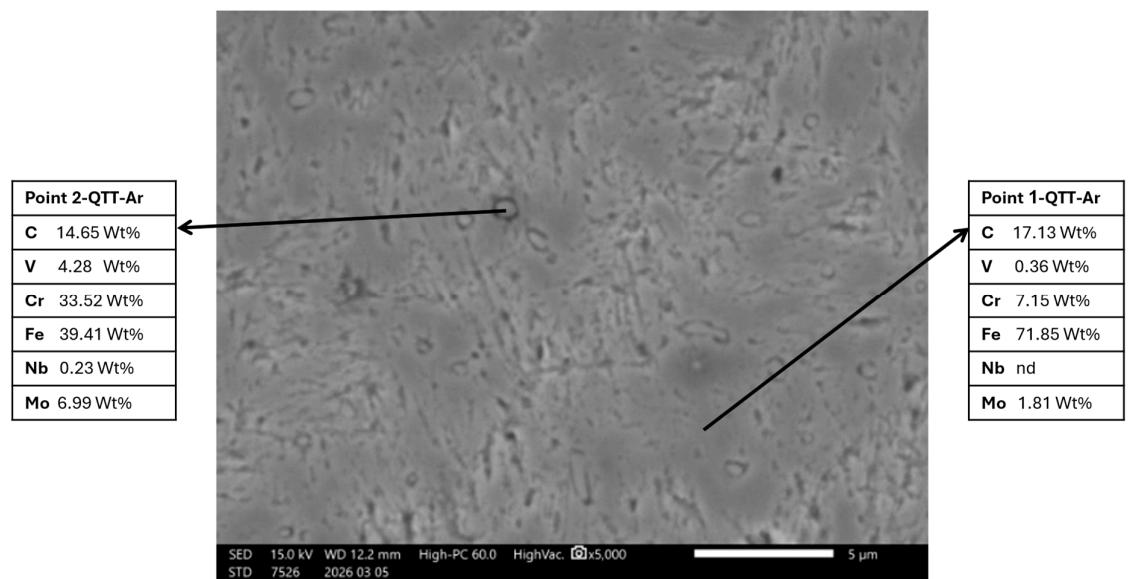


Figure 3.4-12 EDS Point analysis of QTT-Ar Bulk sample

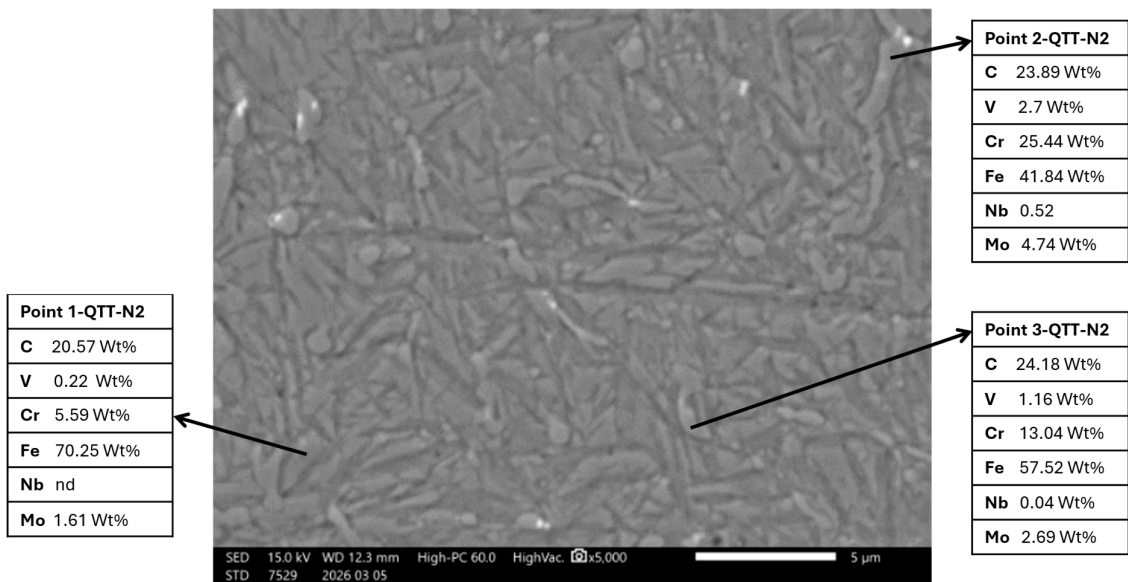


Figure 3.4-13 EDS Point analysis of QTT-N2 Bulk sample

Table 3.4-2 EDS of Element distribution in QTT Bulk sample

Point	C	AL	Si	V	Cr	Mn	Fe	Nb	Mo
1-QTT-Ar	17.13	0.68	0.71	0.36	7.15	0.32	71.85	nd	1.81
2-QTT-Ar	14.65	0.17	0.24	4.28	33.52	0.5	39.41	0.23	6.99
1-QTT-N2	20.57	0.7	0.79	0.22	5.59	0.28	70.25	nd	1.61
2-QTT-N2	23.89	0.2	0.28	2.7	25.44	0.4	41.84	0.52	4.74
3-QTT-N2	24.18	0.48	0.56	1.16	13.04	0.32	57.52	0.04	2.69

The matrix regions analyzed in the QTT samples contain only minor amounts of alloying elements, consistent with a tempered martensitic matrix formed after full austenitization and quenching. EDS mapping clearly shows that alloying elements are preferentially concentrated in discrete regions, forming chain-like or clustered carbide distributions, while the matrix remains comparatively chemically uniform.

The EDS point analyses (Table 3.4-2) obtained from the quenched and triple tempered samples processed under argon and nitrogen atmospheres indicate the presence of both matrix regions and Cr-rich carbide particles. In the QTT-Ar sample, point 1 shows a high Fe content (~72 wt%) with relatively low Cr and Mo, which is consistent with the tempered martensitic matrix. The relatively high carbon signal is most likely related to the presence of very fine carbides within the interaction volume of the electron beam or adjacent carbide particles. In contrast, point 2-QTT-Ar exhibits a significantly higher Cr content (~33 wt%) together with elevated carbon, indicating the presence of a Cr-rich carbide. According to the carbide composition ranges reported in the two reference tables previously discussed, this composition possibly consists of M_7C_3 -type carbides rather than $M_{23}C_6$ -type due to its shape commonly observed in high-chromium tool steels.

A similar trend is observed in the QTT-N2 sample. Point 1-QTT-N2 shows a composition dominated by Fe (~70 wt%), indicating that the analysis was taken from the tempered martensitic matrix. However, point 2-QTT-N2 presents a considerable enrichment of Cr (~25 wt%) and carbon, suggesting the presence of a Cr-rich carbide phase. Point 3-QTT-N2 exhibits intermediate values between the matrix and the carbide particles, which likely indicates that the analysis point was located at the interface between a carbide particle and the surrounding matrix or that the electron beam interaction volume included both phases. To conclusively distinguish between the Cr-rich carbide types (M_7C_3 versus $M_{23}C_6$), their geometric morphology acts as a decisive factor; while the massive, blocky primary carbides correspond to the M_7C_3 type, the finer, continuous film-like networks observed at the grain boundaries confirm the precipitation of secondary $M_{23}C_6$ during the triple tempering cycles.

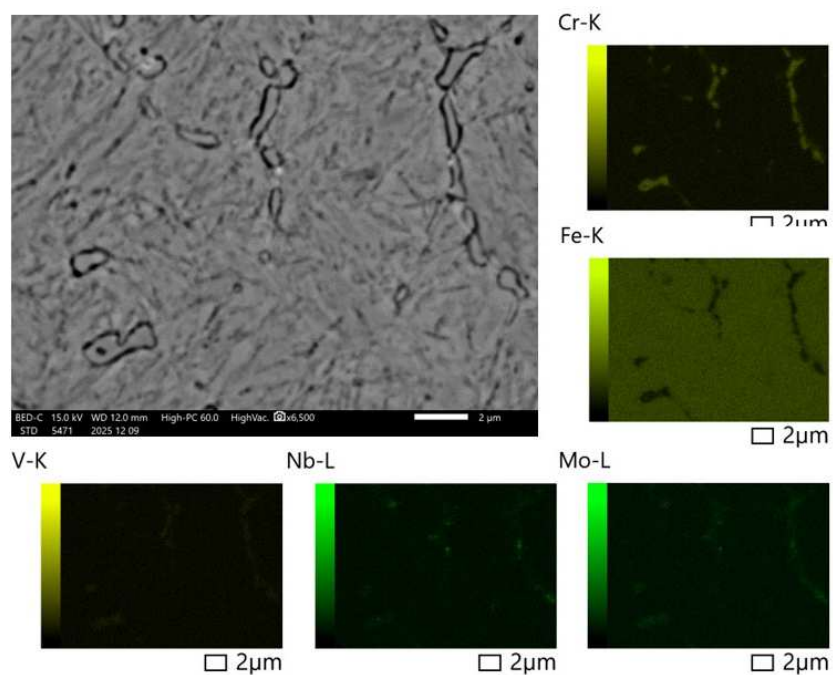


Figure 3.4-14 MAP of QTT-Ar Bulk Sample

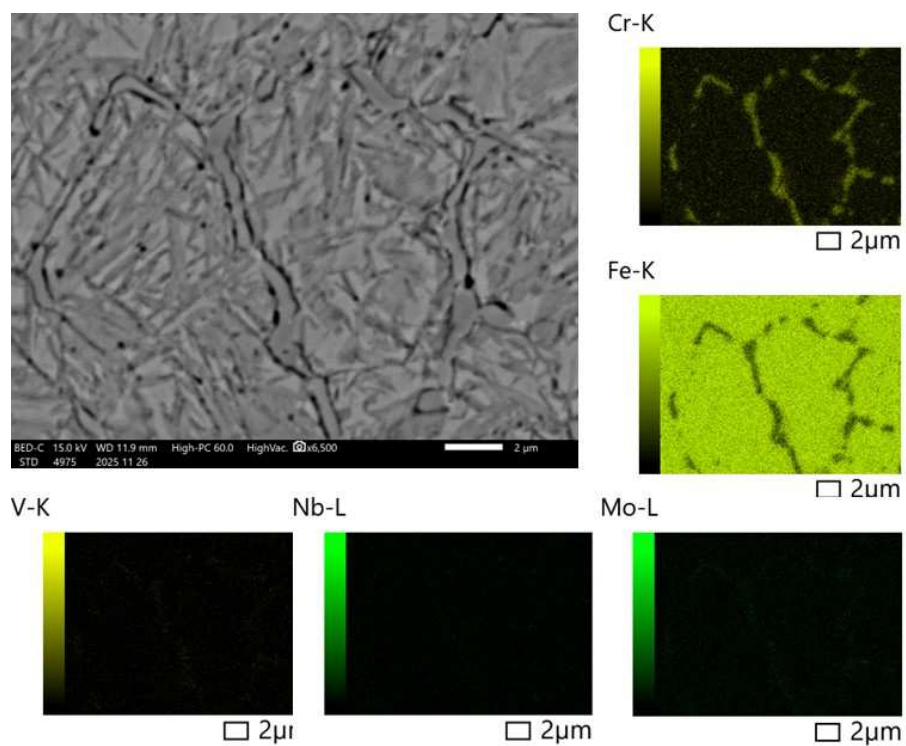


Figure 3.4-15 MAP of QTT-N2 Bulk Sample

Based on the elemental EDS mapping of the QTT-Ar and QTT-N2 sample (Figures 3.4-14 and 3.4-15), distinct morphological features can be correlated with specific alloy carbide phases. The chromium (Cr) map reveals two prominent configurations: narrow, elongated network lines and larger, massive regions.

According to the established carbide morphological criteria, the narrow and elongated Cr-rich lines that follow the grain and cell boundaries indicate the film-like precipitation of secondary $M_{23}C_6$ carbides during the tempering cycles. Conversely, the larger, massive Cr-rich areas correspond to the primary M_7C_3 carbides that formed during the initial solidification stage.

Furthermore, the elemental map highlights highly localized, bright discrete points strongly enriched in Niobium (Nb). These features possess a compact, blocky geometry free of chromium, which definitively classifies them as primary MC carbides. This spatial and morphological distribution confirms a heterogeneous carbide population dispersed within the tempered martensitic matrix.

A comparison between the two feedstock conditions indicates a noticeable difference in carbide morphology. In the N2-atomized DED specimens, the Cr-rich film-like carbides appear to form a more continuous interdendritic network, whereas in the Ar-atomized specimens the same carbide network is generally more discontinuous and fragmented. This observation was consistently found in the analyzed regions and suggests that nitrogen atomization promotes a more interconnected carbide distribution. Such continuous carbide networks are generally considered detrimental to fracture toughness because they provide preferential paths for crack initiation and propagation, while a discontinuous carbide distribution is expected to interrupt these brittle paths and improve crack resistance. Nevertheless, this interpretation is based on the analyzed fields and should be regarded as a microstructural trend rather than direct proof of the bulk mechanical behavior.

4 Results and Discussion: Repair Samples

In this section, the results and discussion related to the repaired samples are presented. The repair experiments were performed on a K340 tool steel substrate that had previously undergone a conventional quenching and tempering (Q&T) heat treatment, providing a hardened and tempered base material representative of real industrial components.

As summarized in Table 2.2-3 (section 2.2.2 Feedstock and specimen preparation), repair grooves with three different widths were machined at four different depths (2, 3, 4, and 5 mm). The selection of different groove depths and widths was intentional, aiming to evaluate the filling capacity of the DED process and to assess the susceptibility to potential defects such as lack of fusion or incomplete bonding. In addition, this design allows investigation of the effect of deposited material volume and the intensity of repeated thermal cycles associated with increasing repair depth.

After deposition, different heat treatment conditions were applied depending on the groove depth. The samples with 2 and 3 mm depth were analyzed in the as-built condition. For the 4 mm depth, the samples were examined both in the as-built condition and after Direct Triple Tempering (DTT). Similarly, the deepest repair (5 mm) was investigated in the as-built condition as well as after a complete Quenching and Triple Tempering (QTT) heat treatment. This approach enables a direct comparison between untreated and heat-treated repair states in identical geometrical conditions.

A comprehensive microstructural characterization was carried out across three distinct regions: the DED deposit, the heat-affected zone (HAZ), and the base material. The primary objective of this comparison is to evaluate the effectiveness of post-deposition heat treatments in promoting microstructural homogenization, reducing the extent and heterogeneity of the HAZ, and improving the structural continuity across the DED/HAZ/base interfaces.

In addition to microstructural observations and simulation of heat treatment processes by Thermo-Calc software, hardness measurements were performed in all three regions for both as-built and heat-treated conditions. This combined analysis allows correlating microstructural evolution and carbide behavior with local mechanical response. The methodology used for microstructural and carbide analysis follows the same approach adopted for the bulk samples, enabling a consistent and direct comparison between bulk and repaired conditions.

4.1 Macroscopic Evaluation of Repaired Samples

Macroscopic observations were first performed on the repaired samples to assess the overall quality of the deposited material and the integrity of the repair region. This analysis focuses on the presence of macroscopic defects such as lack of fusion, cracks, and porosity, as well as their dependence on groove geometry and repair depth.

Representative macroscopic images of As-Built (AB), Direct Triple Tempered (DTT) , and Quench Triple Tempered (QTT) samples of the repaired grooves at different depths and widths are shown in Figure 4.1-1. The observations indicate that the occurrence and severity of defects are strongly influenced by the repair geometry.

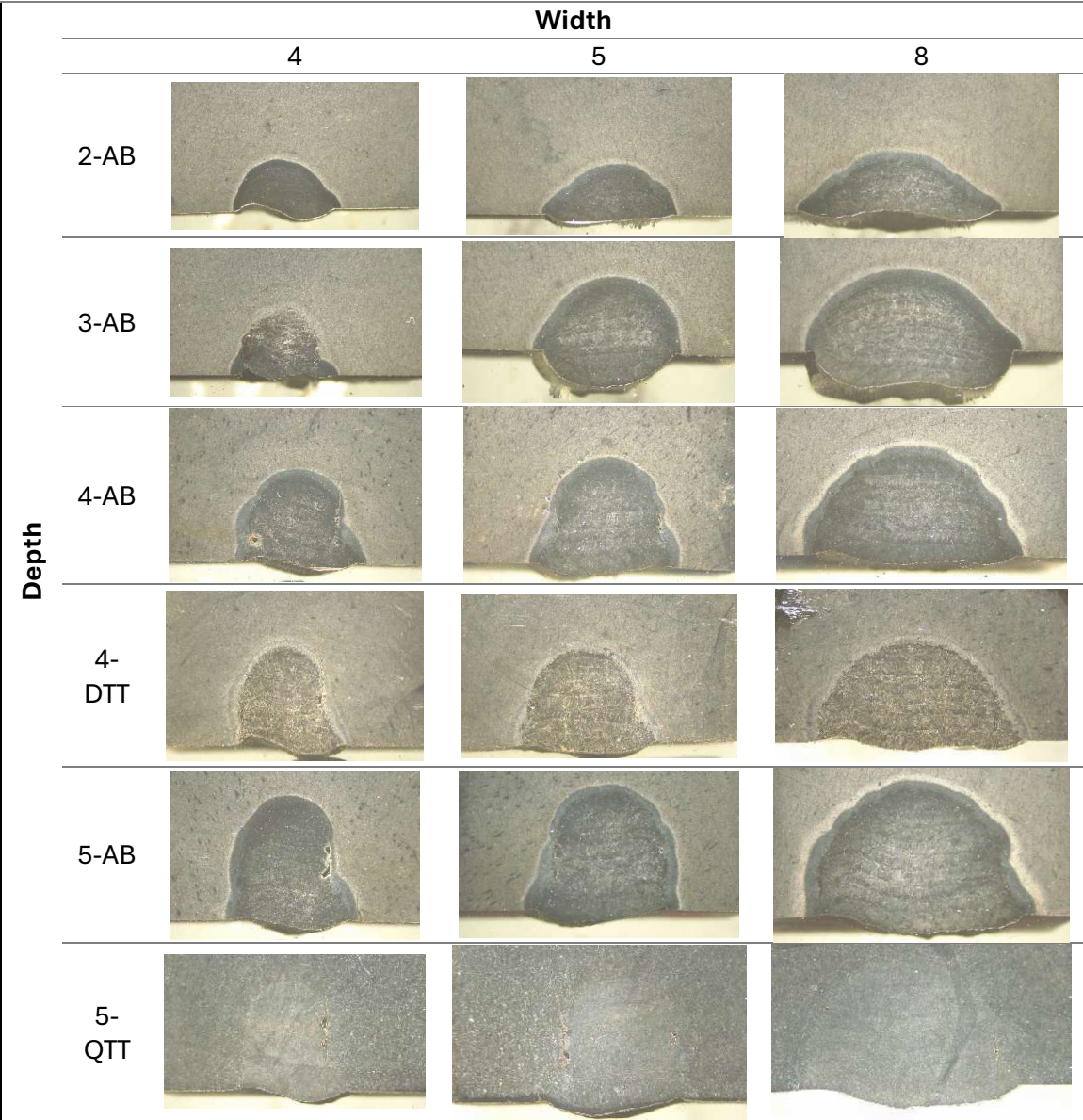


Figure 4.1-1 Macroscopic Images of As-Built (AB), Direct Triple Tempered (DTT), Quench Triple Tempered (QTT) in different Width and Depth

The macroscopic observations show that several grooves were successfully repaired using the DED process without visible defects. Grooves with a width of 8 mm showed very good results, including strong bonding between the deposited material and the substrate, and defect-free filling at all examined depths. The larger groove width allows better laser access and a more stable melt pool, which helps ensure proper melting of both the

powder and the groove side walls. Likewise, grooves with a shallow depth of 2 mm showed no lack of fusion or porosity at the interface.

In contrast, defects such as lack of fusion and local porosity were observed at the groove side walls when the groove depth increased to 3–5 mm in grooves with widths of 4 and 5 mm. These defects consistently appeared at the interface between the deposited material and the substrate. The problem was most severe in the narrowest grooves (4 mm wide), where increasing the depth from 4 to 5 mm led to more pronounced defects.

This behavior is mainly related to the higher depth-to-width ratio of the grooves. In deep and narrow geometries, the groove side walls become almost parallel to the laser beam, which limits laser irradiation and prevents complete melting of these areas. As a result, proper metallurgical bonding cannot form, creating so-called dead zones that lead to lack of fusion and trapped porosity.

Following the macroscopic evaluation of the as-built samples, the heat-treated conditions were examined in order to compare the overall appearance of the repaired regions.

For the Direct Triple Tempered (DTT) condition, heat treatment was carried out on samples with a groove depth of 4 mm and widths of 4, 5, and 8 mm. In comparison with the as-built state, the deposition region appears brighter in optical contrast. The boundary between the deposited material and the substrate is still distinguishable; however, the transition zone appears visually narrower. The general geometry of the repaired grooves remains unchanged.

Lack of fusion defects are still observed at the groove side walls in the same locations identified in the as-built samples. These defects are particularly visible in the narrower grooves. In some sections, the apparent size of the defects differs slightly compared to the as-built condition. However, since the heat-treated samples were sectioned opposite to the as-built samples and the cutting blade thickness introduces a small material offset, minor variations in defect appearance should be interpreted with caution.

For the Quenched and Triple Tempered (QTT) condition, heat treatment was performed on samples with a groove depth of 5 mm for all three widths. In this condition, the contrast difference between the deposited region and the substrate is significantly reduced. The transition zone is difficult to clearly distinguish at the macroscopic scale, and the individual deposition layers are no longer clearly visible. The repaired area appears more uniform across the entire cross-section.

Similar to the DTT condition, defects such as lack of fusion at the groove side walls may still be observed. As in the previous case, their apparent size and visibility can be influenced by the sectioning position and cutting conditions.

4.2 Microstructural Evolution in Repaired Samples

4.2.1 Substrate Zone

The substrate (SU) region serves as the reference microstructure in this study. The substrate (SU) microstructure, shown in the LOM image (Figure 4.2-1), consists of large and some fine carbide particles embedded in a fine martensitic matrix with retained austenite. This structure reflects the prior quenching and triple tempering treatment applied to the substrate before the DED process was applied, resulting in a hardened tempered martensitic matrix with dispersed carbides.

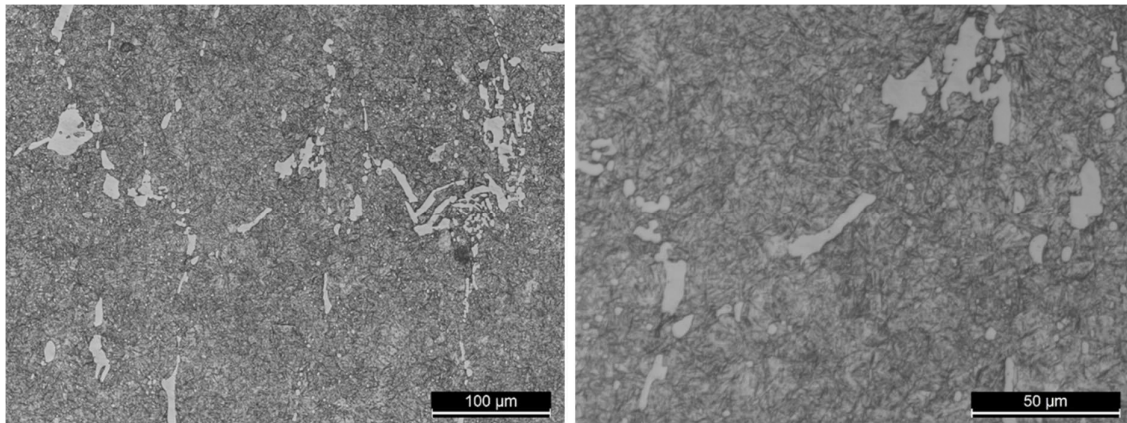


Figure 4.2-1 LOM image of Substrate (SU) of Repaired sample (x200, x500)

SEM observations confirm the presence of lath martensite and clearly distinguishable carbide particles with different sizes and morphologies (Figure 4.2-2). The larger carbides appear relatively blocky and uniformly distributed, while finer carbides are dispersed within the martensitic matrix.

The EDS point analyses (TABLE 4.2-1) obtained from the as-built substrate (SU-AS) region of the repaired samples indicate the presence of different carbide types together with the martensitic matrix typical of high-alloy tool steels.

The substrate (SU) region corresponds to the original base material that had already undergone quenching and tempering prior to the DED process. Therefore, from a metallurgical perspective, no significant change in the type of carbides is expected between the AB, DTT, and QTT conditions. The EDS results confirm this assumption, as the overall compositional trends remain very similar in all three cases. Any minor variations observed in the elemental concentrations are mainly related to the exact location of the analysis point (carbide vs. matrix) rather than a real microstructural transformation.

Points characterized by high Fe content (approximately 70–80 wt%) with moderate amounts of alloying elements such as Cr and Mo (e.g., 1-SU-AB and 1-SU-DTT) correspond to the tempered martensitic matrix of the substrate. This composition is

consistent with the typical microstructure of quenched and tempered high-alloy tool steels.

Table 4.2-1 EDS of Element distribution in Substrate (SU) of Repaired sample in different condition

*AB: As-Built, DTT: Direct Triple Tempered, QTT: Quench and Triple Tempered									
Point	C	AL	Si	V	Cr	Mn	Fe	Nb	Mo
1-SU-AB	10.23	0.95	0.89	0.31	7.05	0.42	77.83	nd	2.23
2-SU-AB	18.15	0.06	0.07	4.5	37.67	0.65	32.92	0.04	5.94
3-SU-AB	30.53	0.07	0.09	3.9	0.97	0.09	1.5	56.94	5.91
2-SU-DTT	16.01	0.09	0.09	2.76	36.05	0.65	41.13	0.002	3.19
1-SU-DTT	11.92	0.4	0.4	0.26	6.07	0.38	78.77	nd	1.79
1-SU-QTT	19.3	0.08	0.07	4.46	37.55	0.59	32.35	nd	5.61
2-SU-QTT	25.43	0.08	0.07	4.38	1.16	nd	2.21	55.95	10.72

In several other points, a high Cr concentration (around 35–38 wt%) is observed together with significant carbon and moderate Fe and Mo contents (e.g., 2-SU-AB, 2-SU-DTT, and 1-SU-QTT). Comparing our results with the reference tables (3.3-1 to 3.3-3) shows that these initial massive carbides are most likely M_7C_3 . This classification is strongly supported by both their blocky shape and chemical signature, which features high chromium content, negligible niobium, and only minor amounts of other alloying elements.

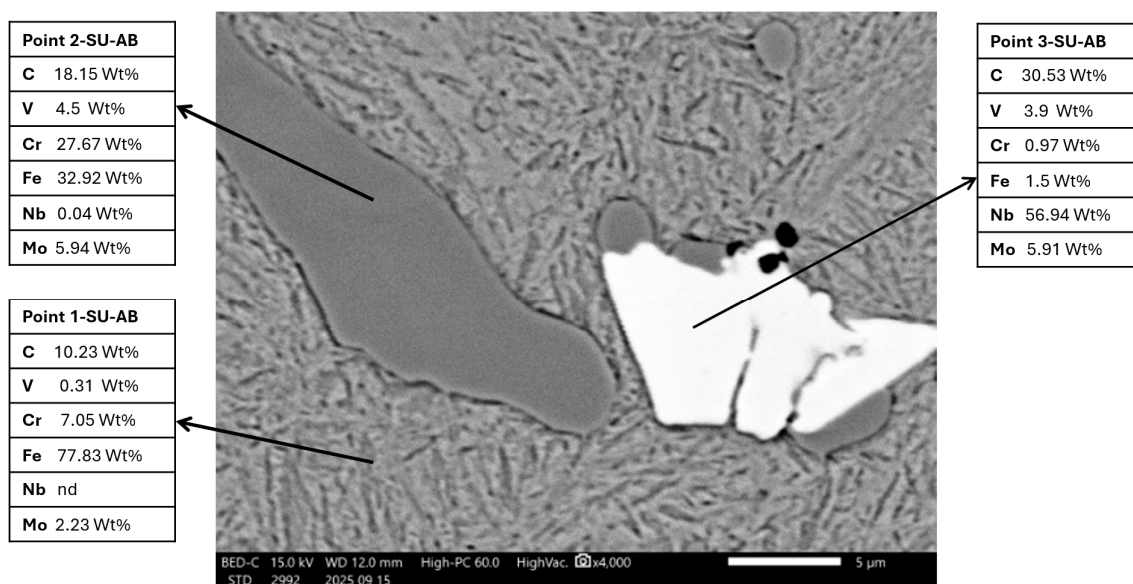


Figure 4.2-2 EDS Point analysis of SU-AB repaired sample

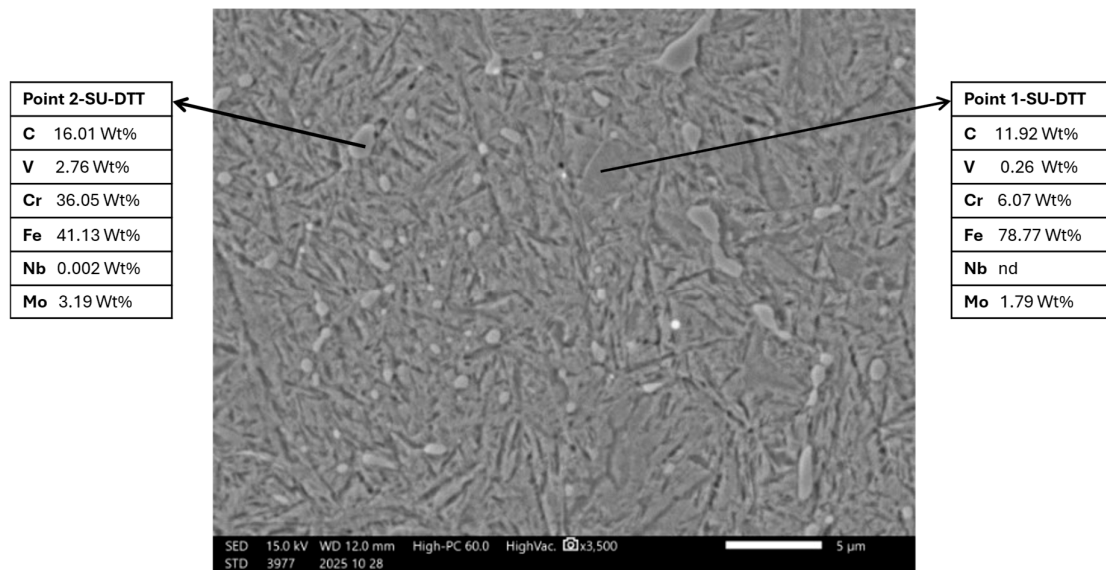


Figure 4.2-3 EDS Point analysis of SU-DTT Repaired sample

The selected SEM field shown in Figure 4.2-3 (in the DTT substrate) contains predominantly fine carbides and is not fully representative of the entire substrate. Coarser primary carbides, comparable to those observed in the as-built condition, were also identified in other regions of the substrate but are not included in this figure.

Another distinct type of particle is observed in points with very high Nb content ($\approx 55\text{--}57$ wt%) and very low Fe and Cr contents (e.g., 3-SU-AB and 2-SU-QTT). This composition clearly corresponds to Nb-rich MC-type carbides. These carbides are thermodynamically very stable and typically remain undissolved even after heat treatment, persisting as primary carbides within the microstructure.

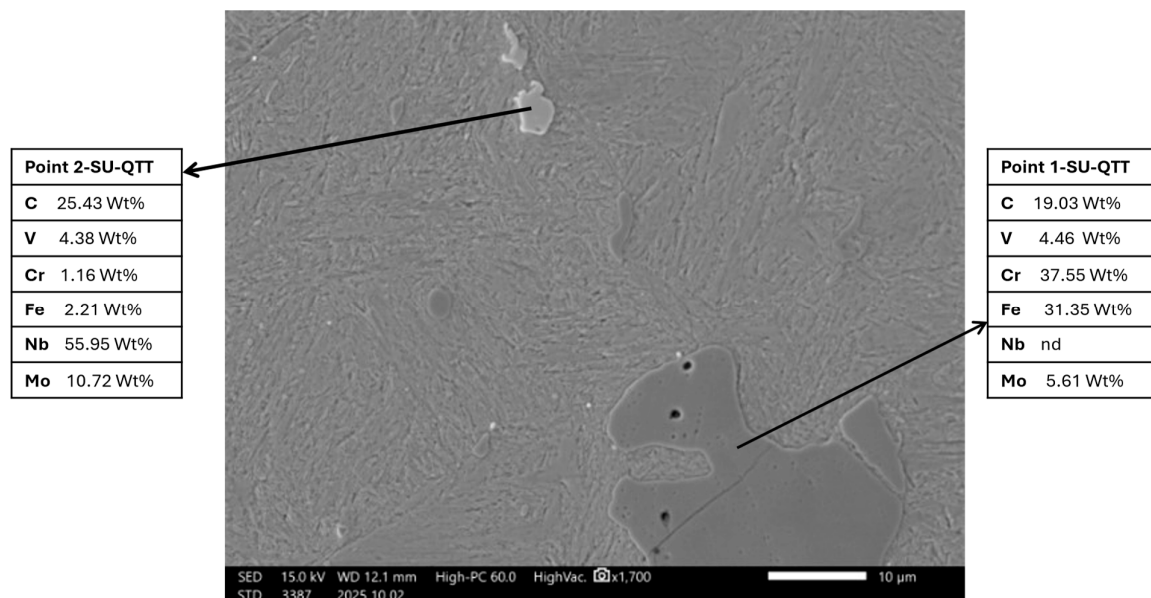


Figure 4.2-4 EDS Point analysis of SU-QTT Repaired sample

4.2.2 Heat Affected Zone (HAZ)

Figure 4.2-5 presents the LOM images of the interface region in the as-built (AB), direct triple tempered (DTT), and quenched and triple tempered (QTT) conditions.

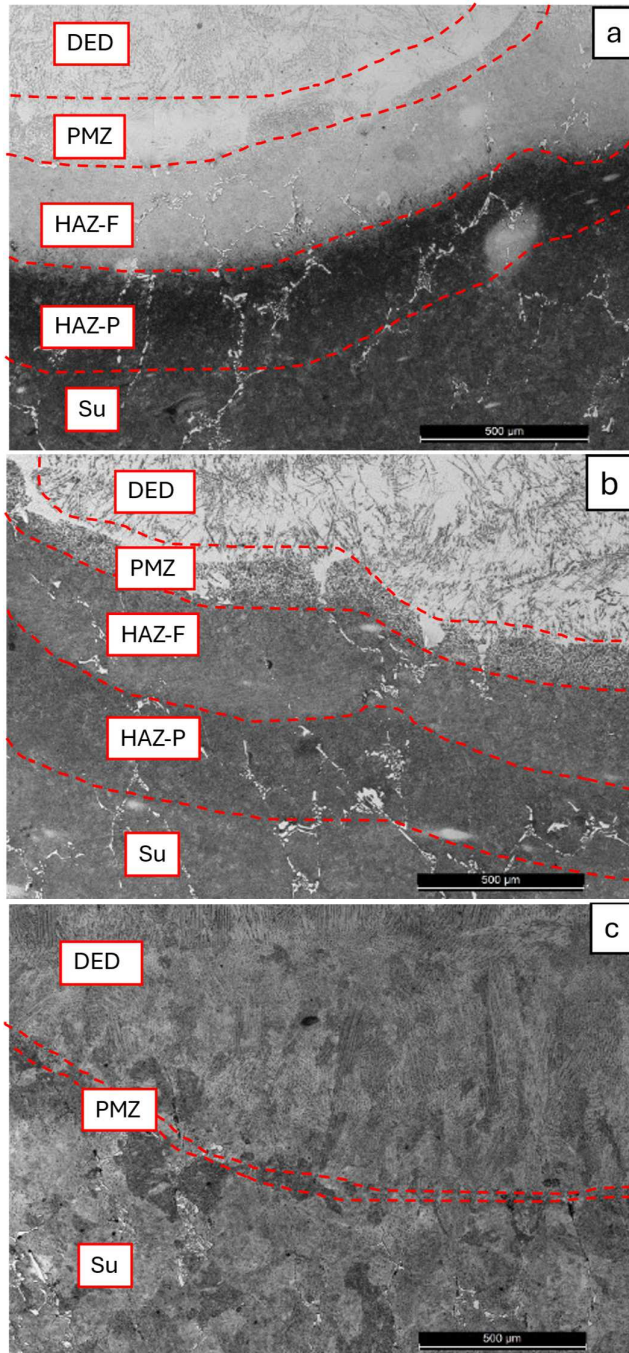


Figure 4.2-5 LOM images of the interface region
a) as-built (AB), b) Direct Triple Tempered (DTT),
c) Quenched and Triple Tempered (QTT) conditions

In the As-Built (AB) sample (Figure 4.2-5-a), the substrate (SU), partially melted zone (PMZ), heat-affected zone (HAZ), and deposited (DED) regions are clearly distinguishable, with sharp microstructural contrast across the fusion boundary.

After Direct Triple Tempering (DTT) (Figure 4.2-5-b), these regions remain identifiable; however, the contrast between them is reduced and the apparent width of the transition zones slightly decreases, indicating partial homogenization of the martensitic structure formed during deposition.

In the Quenched and Triple Tempered (QTT) condition (Figure 4.2-5-c), although the SU and DED regions can still be recognized based on their morphological features and layered texture, the fusion boundary and intermediate transition zones become significantly less distinct. The interface is difficult to clearly define, suggesting that re-austenitization and subsequent quenching have largely diminished the initial thermal gradients created during the DED process.

4.2.2.1 Heat-Affected Zone of As-Built repaired sample

In Figure 4.2-6, the region in Partially Melted Zone (PMZ) of As-Built sample, observed directly beneath the first deposited layer, close to the fusion boundary with the substrate, does not appear to be a classical white band[90]. A typical white band is usually a narrow and continuous layer formed by complete remelting and rapid resolidification of the substrate, showing a relatively uniform and highly refined microstructure parallel to the interface [57]. In the present case, morphology is neither fully continuous nor uniformly refined.

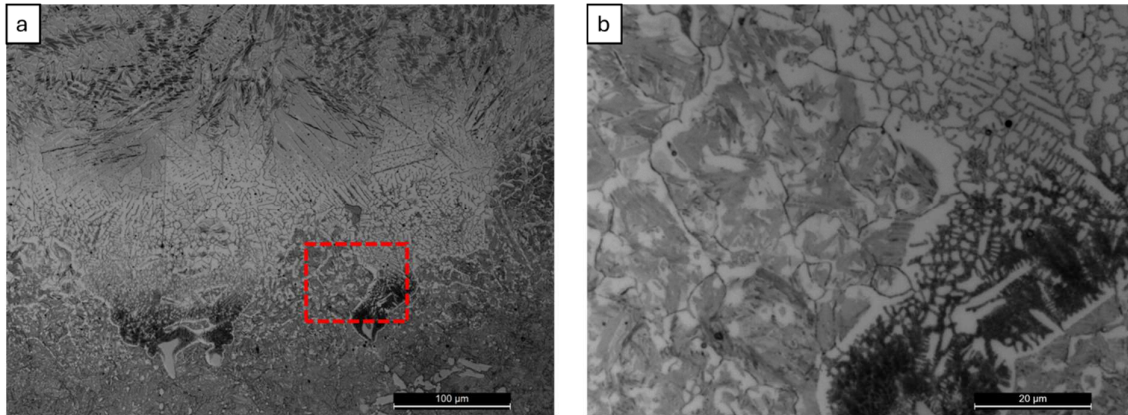


Figure 4.2-6 LOM image of PMZ of As-Built Sample, x200, x1000

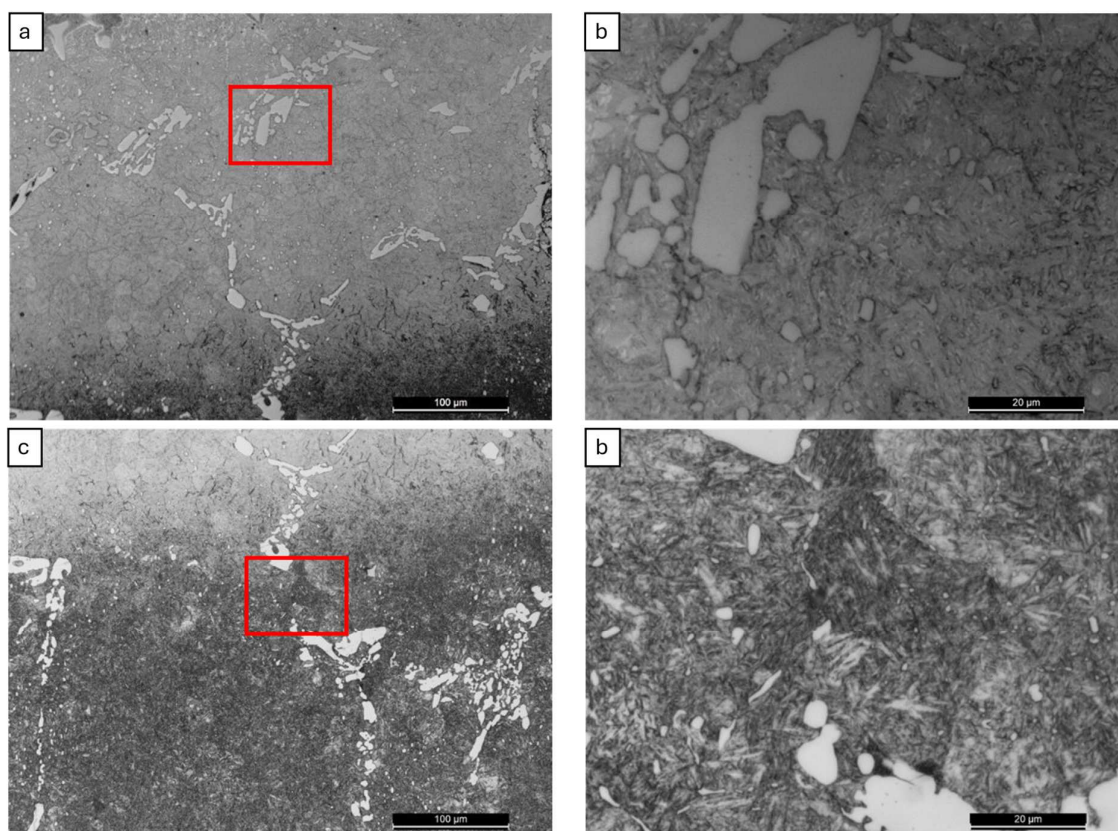
Instead, this area contains locally rounded or globular features with martensite inside, surrounded by a very fine and irregular network-like structure, which appears to result from re-solidification. This morphology suggests that the region is strongly influenced by solidification phenomena within the melt pool rather than being only the result of a simple reheating effect. This combination of features is characteristic of a partially remelted and re-solidified region. It suggests that the substrate was not completely melted; rather, certain areas such as grain boundaries or alloy-enriched regions locally entered the liquid state and then rapidly solidified.

The globular features are most likely remnants of the original substrate grains that grew epitaxially from the melt. The presence of martensite within them indicates that these regions were locally austenitized during the thermal cycle and subsequently transformed during rapid cooling. This behavior is consistent with a partially melted zone, often referred to in the literature as a mushy zone or partially remelted zone. [57]

During the first DED pass, a steep thermal gradient exists between the molten pool and the relatively cold substrate. Solidification progresses directionally from the substrate toward the deposited material. At the same time, alloying elements may segregate into the remaining liquid during the final stages of solidification. As a result, locally enriched liquid regions can form near the fusion boundary. After rapid cooling, these enriched areas may transform into high-alloy martensite, while the surrounding zones solidify as a fine interdendritic network [56], [57].

In addition, dilution between the deposited material and the substrate may contribute to local compositional variations in this region. Therefore, this zone can be interpreted as a complex transition region formed by a combination of melt pool solidification, micro-segregation, and local remelting effects, rather than a classical thermally formed white band [57].

Below the fusion boundary, two distinct sub-regions of the heat-affected zone can be identified: HAZ-F (Fully Austenitized HAZ) and HAZ-P (Partially Austenitized HAZ).



*Figure 4.2-7 LOM of HAZ-F and HAZ-P of As-Built Sample
a) HAZ-F x200, b) HAZ-F x1000, c) HAZ-P x200, d) HAZ-P x1000*

HAZ-F is located closer to the deposited layer. This region experienced the highest peak temperature and most likely exceeded A_{c3} , leading to full austenitization during the DED process. After rapid cooling, a relatively uniform and fine martensitic structure formed. Due to this more homogeneous microstructure, the etched surface appears brighter, and the measured hardness is higher.

Further below, HAZ-P is observed. Although its general morphology may look similar to HAZ-F, it appears darker and shows a significant drop in hardness. This region likely did not fully exceed A_{c3} or remained at high temperature for a shorter time. As a result, austenitization was incomplete, some carbides did not fully dissolve, and the final

microstructure became more heterogeneous. This heterogeneity increases etching contrast, which explains the darker appearance and lower hardness.

The transition from HAZ-F to HAZ-P is not a phase boundary but a thermal gradient effect. Both regions belong to the HAZ, but they differ in peak temperature and degree of austenitization. Below HAZ-P, the unaffected substrate is reached, where the original microstructure remains largely preserved.

Colombo et al. investigated the laser deposition of M4 steel on a D2 substrate, identifying three distinct zones within the thermally modified region: the HAS (HAZ), which undergoes re-austenitization and rapid quenching without melting, resulting in a fresh martensite/bainite matrix with a high volume fraction of spherical secondary carbides that exhibits the highest hardness across the specimen; the Mushy Zone (M), a complex solid-liquid transition area where localized melting around blocky M_7C_3 carbides leads to their functioning as nuclei for a eutectic-like austenite/carbide structure enveloped by a retained austenite ring upon rapid solidification; and the White Zone (WH), which forms strictly from the rapid melting and solidification of the D2 substrate itself (with zero M4 dilution) and consists entirely of equiaxed retained austenite dendrites free of martensite, bounded by intercellular eutectic M_7C_3 carbides [57].

4.2.2.2 Heat Affected Zone of DTT repaired sample

In the Direct Triple tempered (DTT) condition, the overall morphology of the interface region remains largely similar to the as-built state. The PMZ (Figure 4.2-8 a, b) directly beneath the first deposited layer is still identifiable, and its general characteristics resemble those observed in the as-built sample. However, the martensitic structure appears more clearly defined, with distinct lath and block morphology embedded in a matrix containing retained austenite and very fine alloy-enriched streaks. The fine interdendritic network surrounding the globular features is still visible, suggesting that the solidification-induced segregation pattern formed during deposition has not been eliminated by tempering.

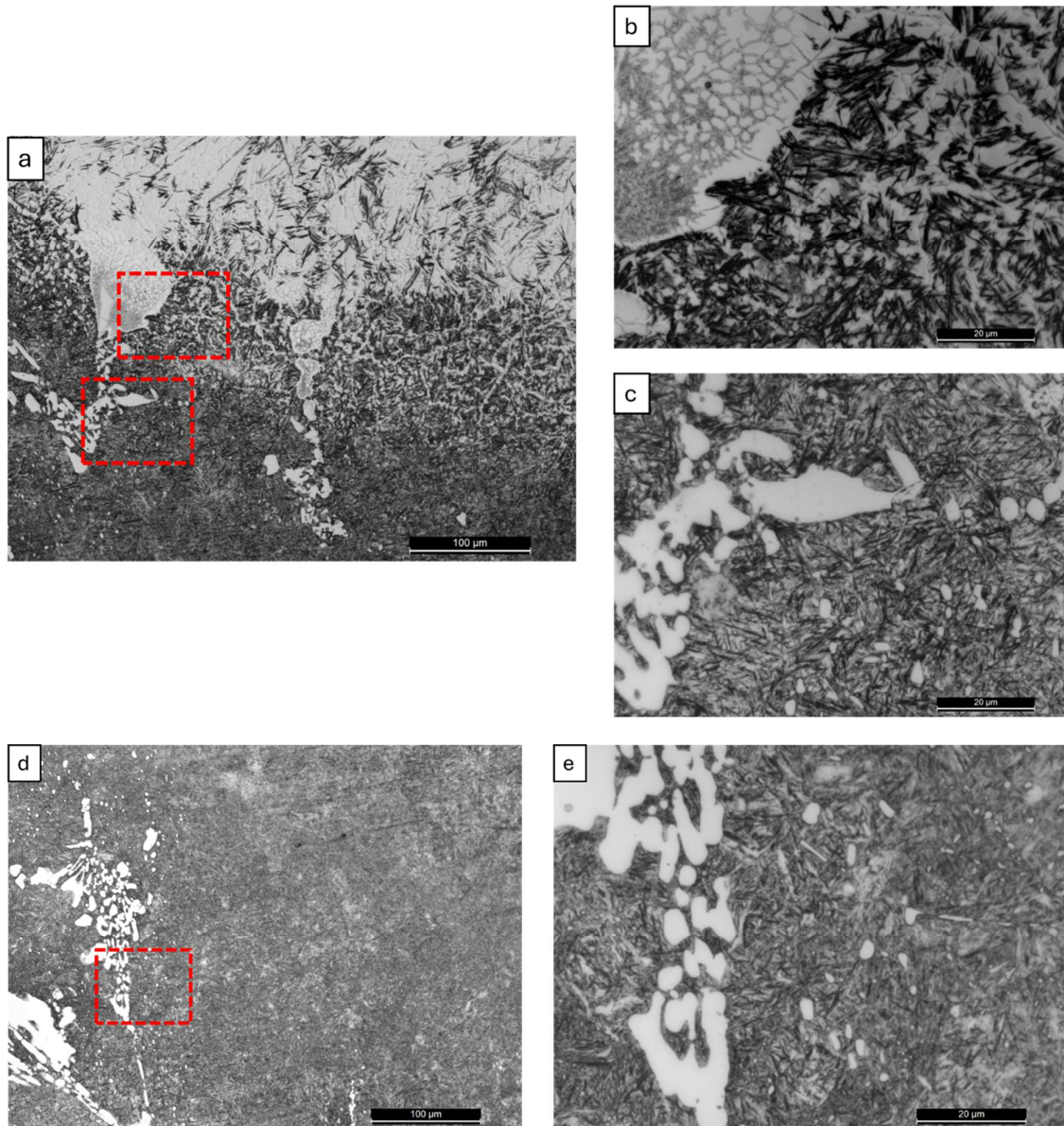


Figure 4.2-8 LOM of HAZ of DTT repaired sample
a) PMZ and HAZ-F region (x200) b) PMZ region (x1000) c) HAZ-F region (x1000), d, e) HAZ-P region (x200, x1000)

Below the PMZ, two sub-regions similar to HAZ-F (Figure 4.2-8 a, c) and HAZ-P (Figure 4.2-8 d, e) can still be distinguished. The upper region shows a relatively bright contrast, comparable to the substrate, while the lower region appears darker, although the contrast difference is less pronounced than in the as-built condition. This indicates that tempering has reduced the microstructural and hardness gradient previously created by rapid cooling.

At higher magnification, the martensite formed in the PMZ exhibits morphology comparable to that observed in the adjacent upper HAZ layers. Large and fine carbides are present in both regions. However, unlike the PMZ, where martensite is embedded within a matrix containing retained austenite and segregation-related network features,

the underlying HAZ region shows a more uniform martensitic surface with less evident matrix contrast and without a pronounced interdendritic network around the carbides. This suggests that while tempering modifies the internal martensitic structure and reduces residual stresses, it does not fundamentally alter the solidification-related features established in the PMZ during the DED process.

4.2.2.3 Heat Affected Zone of QTT repaired sample

As in figure 4.2-9 shown, in the quenched and triple tempered (QTT) condition, the interface is no longer sharply defined as in the as-built or tempered states. While the deposited (DED) region and the substrate (SU) can still be differentiated, classical PMZ or HAZ sub-regions are absent, indicating that re-austenitization and quenching have largely homogenized the prior thermal gradients.

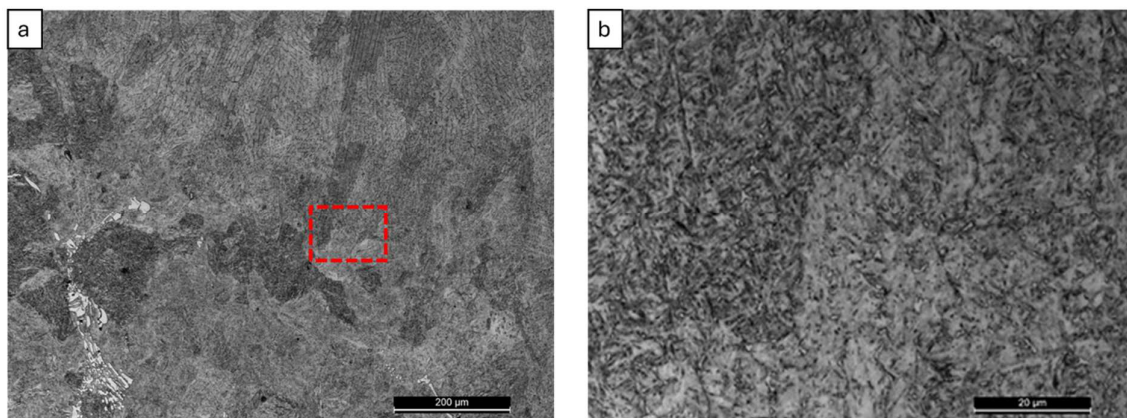


Figure 4.2-9 LOM of interface of QTT sample a) Interface HAZ (x100) b) PMZ (x1000)

At higher magnification, subtle features still allow an approximate identification of the interface. In the substrate, fine carbides are dispersed relatively uniformly, whereas in the DED region, carbides—though also fine—form linear, chain-like arrangements roughly perpendicular to the interface. In addition to this difference in carbide morphology, a contrast difference is apparent: the DED region appears darker than the SU under LOM, highlighting subtle microstructural differences between the two areas. This contrast, together with the carbide alignment, serves as the main visual clue for the former interface location.

The disappearance of classical HAZ or PMZ features confirms the effective homogenization of thermal gradients by the QTT treatment. However, the persistence of contrast differences and the distinctive carbide arrangement suggests that solidification-induced segregation patterns established during the DED process are not entirely erased and still influence the microstructure after quenching and tempering.

4.2.2.4 Phase evolution of HAZ regions in different conditions

EDS point analyses (Table 4.2-2) were performed in the HAZ region for the AB, DTT, and QTT conditions, and the results are summarized in a single table together with the corresponding micrographs. These data, combined with the elemental mapping results, provide insight into the carbide types and microstructural evolution within the PMZ, HAZ-F, and HAZ-P sub-regions. The measured compositions were compared with the carbide composition ranges reported in the two reference tables previously discussed.

Table 4.2-2 EDS of Element distribution in HAZ of Repaired sample in different condition

*AB: As-Built, DTT: Direct Triple Tempered, QTT: Quench and Triple Tempered									
Point	C	AL	Si	V	Cr	Mn	Fe	Nb	Mo
1-HAZ-AB (PMZ)	21.8	0.84	0.79	0.15	5.18	0.35	69.4	nd	1.5
2-HAZ-AB (PMZ)	24.62	0.21	0.22	2.73	32.89	0.64	34.79	0.22	3.68
3-HAZ-AB (HAZ-F)	28.85	0.05	0.13	3.88	33.54	0.55	28.83	nd	4.15
4-HAZ-AB (HAZ-F)	30.46	0.54	0.57	0.89	4.13	0.18	43.56	17.46	2.21
5-HAZ-AB (HAZ-P)	28.73	0.74	0.8	0.18	5.01	0.29	62.87	nd	1.39
6-HAZ-AB (HAZ-P)	29.57	0.07	0.13	3.88	33.14	0.53	28.71	nd	3.98
1-HAZ-DTT	15.31	0.04	0.05	4.44	38.75	0.39	35.44	0.11	5.46
2-HAZ-DTT	26.35	0.06	0.06	2.25	0.075	0.011	3.35	62.34	4.74
1-HAZ-QTT	22.82	0.71	0.81	0.17	5.54	0.32	68.13	nd	1.49
2-HAZ-QTT	24.68	0.18	0.3	3.04	27.58	0.55	39.86	0.06	3.75
3-HAZ-QTT	33.71	0.02	0.16	3.2	0.87	nd	2.05	51.33	8.65

In the As-Built condition, the PMZ point (Figure 4.2-10), Point 1-HAZ-AB, shows compositions dominated by Fe and C with moderate Cr and Mo, indicating a martensitic matrix affected by solidification-induced micro segregation. In some locations such as Point 2-HAZ-AB, the high Cr content (~33 wt%) together with V and Mo suggests the presence of Cr-rich carbides, with respect to round and massive shape consistent with M7C3 carbide reported in reference tables 3.3-1 to 3.3-3.

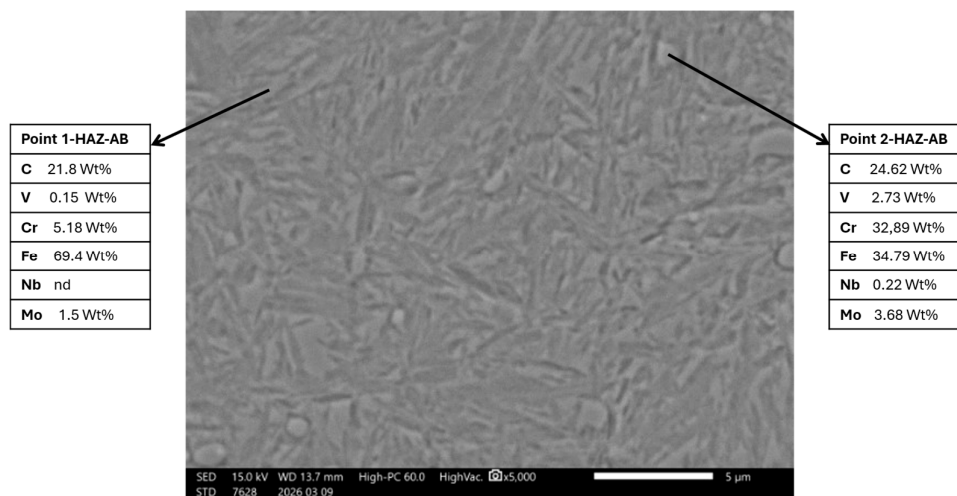


Figure 4.2-10 EDS points of HAZ, PMZ region of As-Built repaired sample

Within the HAZ-F region of the As-Built sample (Figure 4.2-11), the peak temperature exceeded A_{c3} , resulting in full austenitization followed by rapid martensitic transformation. EDS results (e.g., Point 3-HAZ-AB) show a massive shape with high Cr, V, and Mo, which is characteristic of Cr-rich carbides and probably M_7C_3 . In Point 4-HAZ-AB, the relatively high Nb content (~17 wt%) and bright spot indicates the presence of Nb-rich MC-type carbides, which are known for their high thermodynamic stability and limited dissolution during thermal cycles.

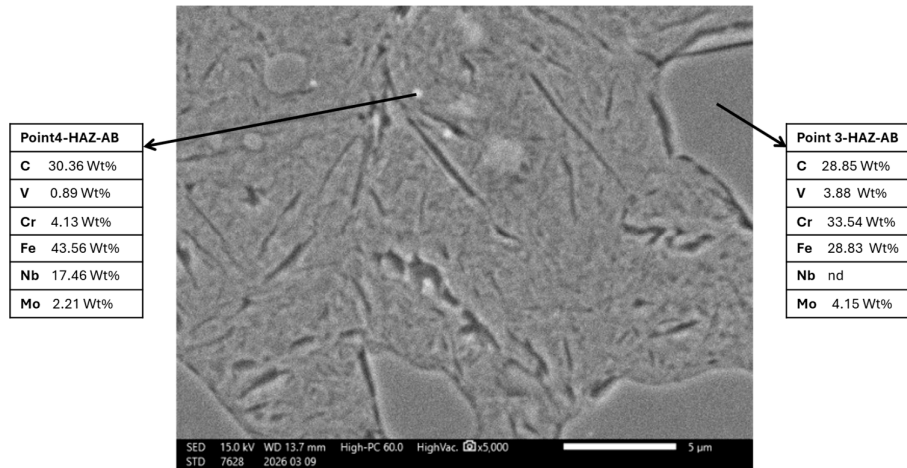


Figure 4.2-11 EDS points of HAZ-F region of As-Built repaired sample

In the HAZ-P region (Figure 4.2-12), the peak temperature likely remained below full austenitization, leading to incomplete carbide dissolution. As a result, the compositions are more heterogeneous. Points such as 5 -HAZ-AB show Fe-rich matrix compositions with local enrichment in Cr and alloying elements, which is consistent with the microstructural observation that some carbides remain undissolved in this region. according to percentage of alloys and massive shape of point 6-HAZ-AB, it is likely to be an M_7C_3 carbide.

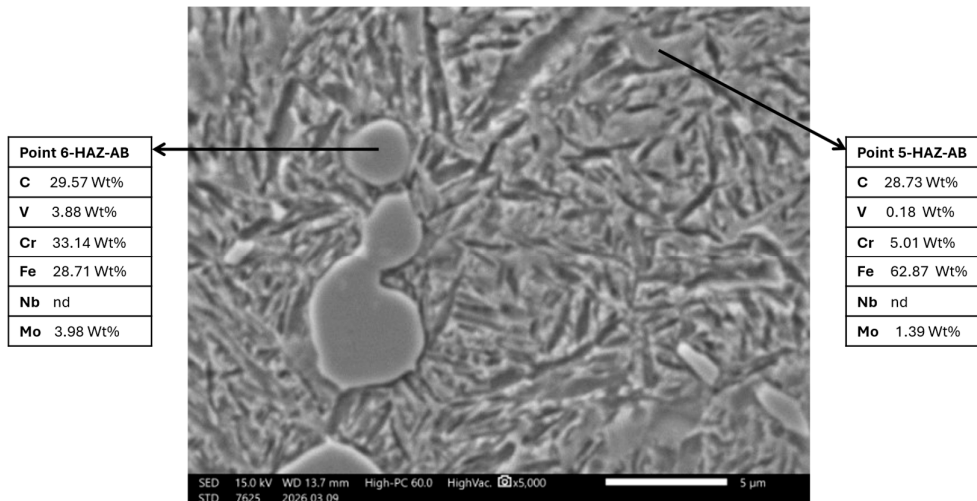


Figure 4.2-12 EDS points of HAZ-P region of As-Built repaired sample

For the DTT condition, elemental mapping of PMZ-DTT reveals a clear cellular segregation network formed during rapid solidification. In this network, Cr and Mo are enriched in the interdendritic regions with narrow, elongated network lines, while Nb and V are present only in minor amounts. This observation supports the previously described microstructure, where martensite with distinct lath and block morphology is embedded within a matrix containing retained austenite and very fine alloy-enriched streaks, indicating that the solidification-induced segregation pattern is preserved after tempering. They are likely to be primary carbide of M_7C_3 . [19], [57]

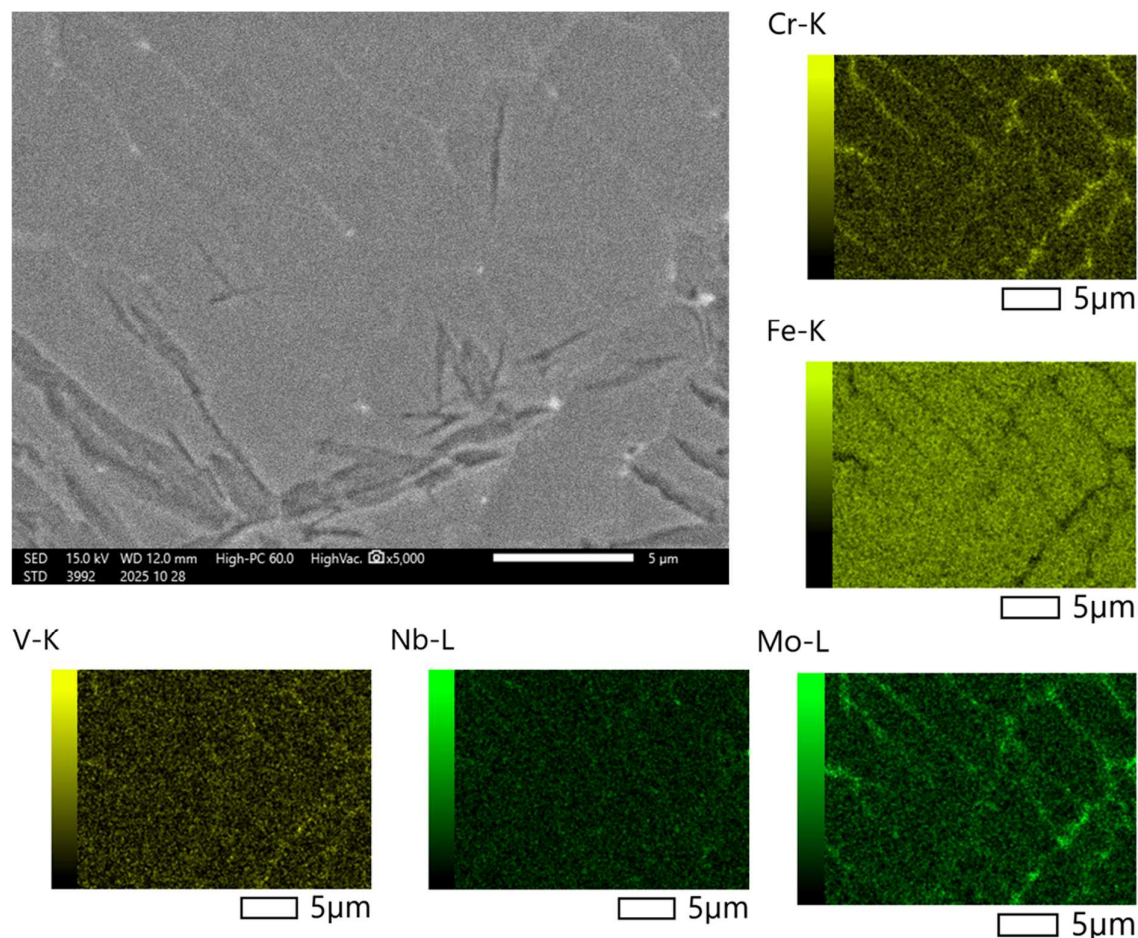


Figure 4.2-13 MAP of PMZ of DTT repaired sample

In the HAZ-F region of DTT condition (Figure 4.2-14), the microstructure contains large primary Cr-rich carbides, M_7C_3 -type, (point 1-HAZ-DTT), while Nb-rich carbides appear as fine, MC -type, (point 2-HAZ-DTT), dispersed secondary particles. The concentrations of V, Mo, and Nb are also higher than in the PMZ. This can be explained by the absence of complete melting in the HAZ-F region, which limits solidification segregation and allows carbide-forming elements to remain within the microstructure and precipitate during tempering.

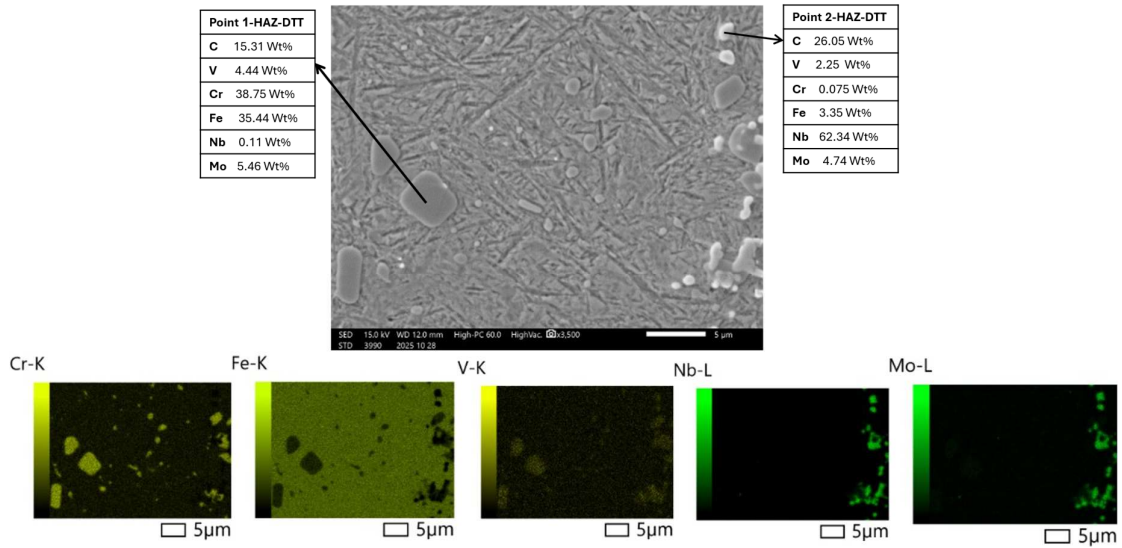


Figure 4.2-14 MAP and EDS point analysis of HAZ-F of DTT repaired sample

In the HAZ-P region of DTT condition (Figure 4.2.15), the composition becomes more similar to that of the substrate. The Nb-rich carbides (likely MC) appear larger, and the overall contents of V, Mo, and Nb are higher. This suggests that less carbide dissolution occurs in this region, and the microstructure remains closer to the original substrate condition. The massive shape with high Cr, V and a few Mo are likely referred to as M_7C_3 carbide.

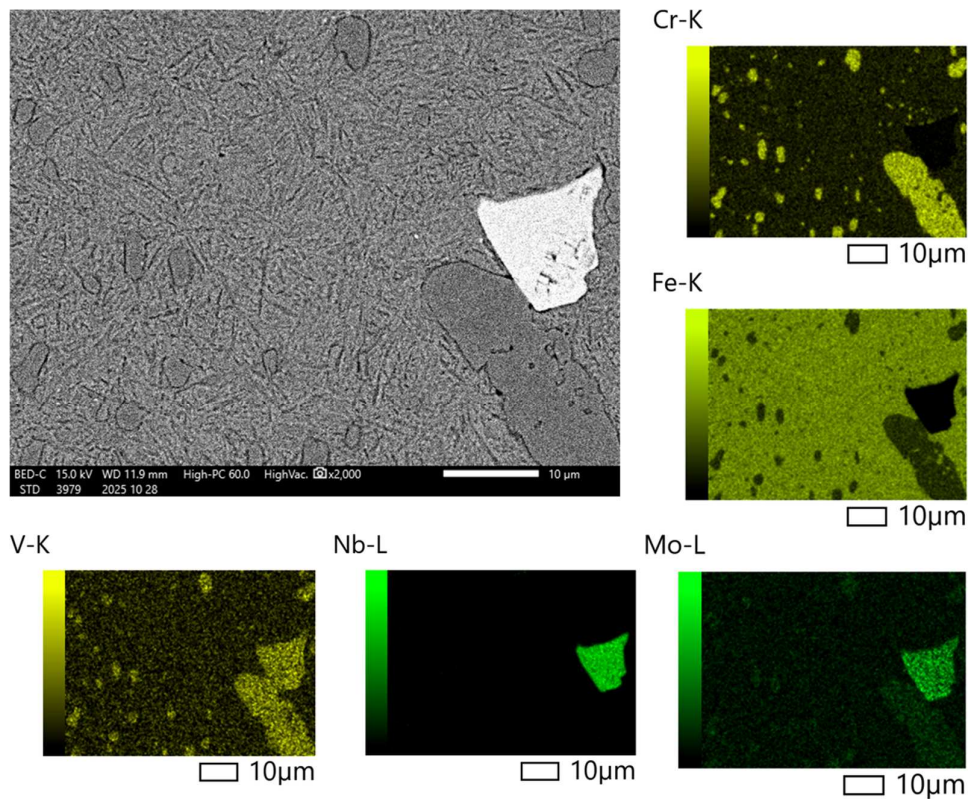


Figure 4.2-15 MAP of HAZ-P region of DTT repaired sample

In the QTT condition, as also observed in the LOM images, the classical PMZ, HAZ-F, and HAZ-P sub-regions are largely absent, and only a very narrow PMZ can be identified. In Point 1-HAZ-QTT, the high Fe content (~68 wt%) suggests that the analysis mainly corresponds to the martensitic matrix. EDS results indicate a complex network of Cr-, V-, and a few Mo-rich carbides (point 2-HAZ-QTT), forming alloy-enriched clusters, while the central regions of these clusters remain relatively Fe-rich, likely associated with solidification-induced segregation. With respect to the narrow, elongated network lines and their percentage of alloys, they are likely to be secondary carbide of $M_{23}C_6$.

Nb-rich carbides (point 3-HAZ-QTT) are mainly associated with the substrate region, rather than the PMZ. These observations suggest that the QTT treatment significantly homogenized the original thermal gradients, although the solidification-related segregation patterns generated during the DED process are still partially preserved.

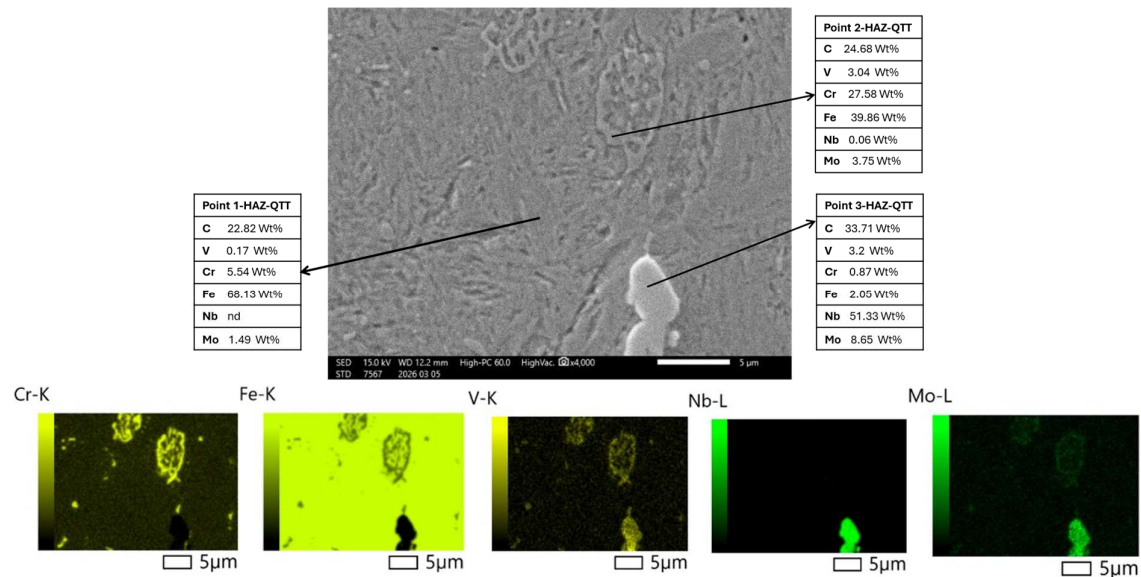


Figure 4.2-16 MAP and EDS point analysis of HAZ region (PMZ) of QTT repaired sample

4.2.3 Deposition Zone

In the repaired samples produced by the DED process under pure Argon shielding, the deposited region shows a microstructural evolution similar to that observed in the bulk specimens fabricated under Argon atmosphere. The solidification behavior is governed by high energy input, steep thermal gradients, rapid cooling rates, and repeated thermal cycling caused by subsequent layer deposition. Although the groove geometry slightly modifies local heat extraction conditions, the overall solidification mechanism remains unchanged.

4.2.3.1 Deposition zone of As-Built repaired sample

Optical micrographs of As-Built repaired sample (Figure 4.2-17 a) reveal a regular layer-by-layer structure with overlapping melt pools and clearly defined arc-shaped fusion boundaries, indicating stable processing conditions during repair. The deposited material is dense and free of significant macroscopic defects. At higher magnification (Figure 4.2-17 b), the microstructure is dominated by columnar prior austenite grains (PAGs) growing epitaxially from the fusion boundary into the melt pool. This epitaxial growth occurs because the molten metal solidifies on the solid substrate and inherits its crystallographic orientation. The grains preferentially grow along the build direction, aligned with the maximum heat flow [4], [11], [56].

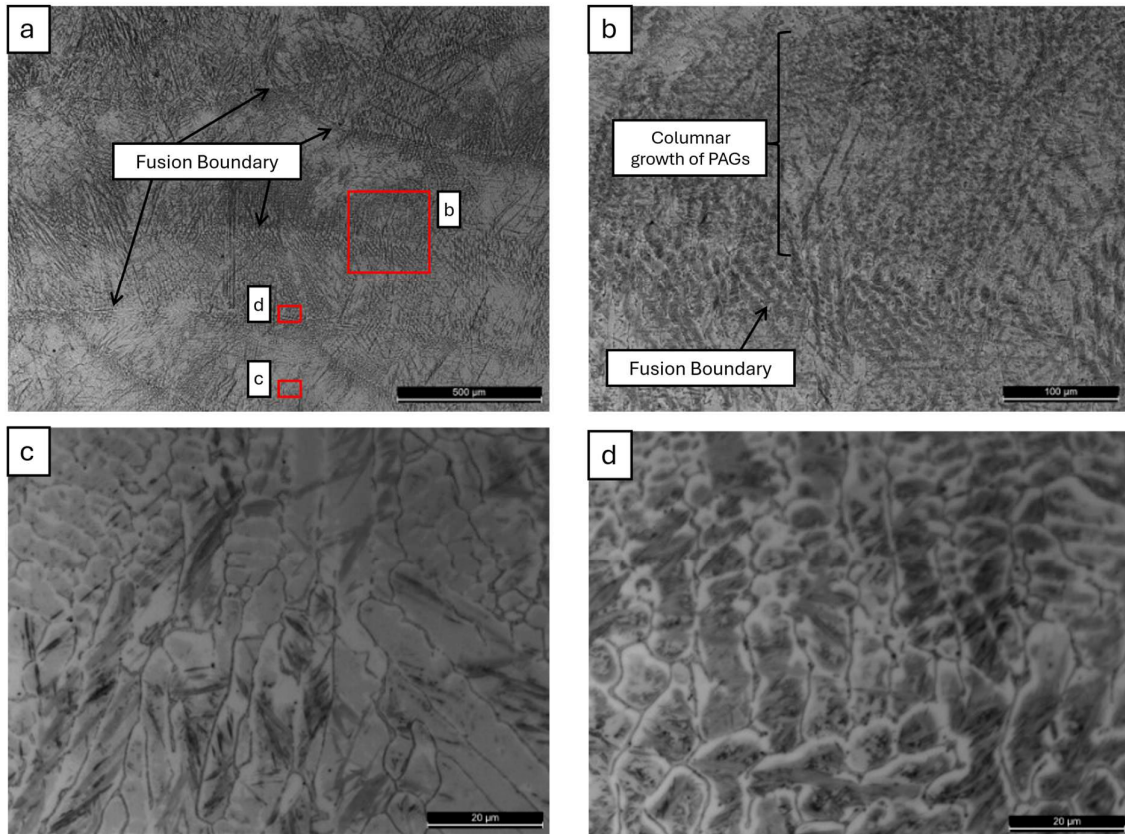


Figure 4.2-17 LOM image of layer of microstructure of As-Built repaired sample
a) overlapping of layers and melt pool (x50) b) Fusion Boundary and Columnar Growth of PAGs (x200),
c) Microstructure near of SU (x1000), d) Microstructure toward center (x1000)

Within prior austenite grains, a refined cellular–dendritic substructure is observed, which is characteristic of rapid directional solidification in laser-based additive manufacturing. Near the fusion boundary (Figure 4.2-17 c), efficient heat extraction into the substrate results in a high thermal gradient and a fine cellular structure. As the distance from the interface increases toward the center of the deposited track (Figure 4.2-17 d), the effective thermal gradient decreases due to reduced heat extraction and thermal accumulation within the melt pool. Consequently, constitutional supercooling becomes

more significant and columnar–dendritic features develop more clearly [56], [84], [85], [87].

A gradual increase in etching contrast is observed when moving from the fusion boundary toward the center of the deposition zone, where the microstructure appears darker. This change is attributed to the reduction in cooling rate and thermal gradient, which enhances micro-segregation of alloying elements such as C, Cr, and Mo in the interdendritic regions during the final stages of solidification. The increased chemical heterogeneity in these regions leads to stronger etching response and higher contrast. In addition, the central areas are more affected by repeated reheating from subsequent layers, which may cause localized tempering of martensite and further increase microstructural heterogeneity [56], [58], [86].

The material shows a hierarchical microstructure. After etching, LOM observations (Figure 4.2-18) at the microscale reveal a martensitic (M) structure embedded within retained austenite (RA). In addition, a cellular–dendritic solidification pattern is observed, along with chemically different interdendritic regions. The prior austenite grains are relatively coarse, mainly due to the solidification conditions in the DED process. During cooling, the martensitic phase transformation occurs within this solidified structure. As a result, martensitic lath blocks appear as dark, zigzag needle-like features, mostly forming inside individual solidification cells [4], [91].

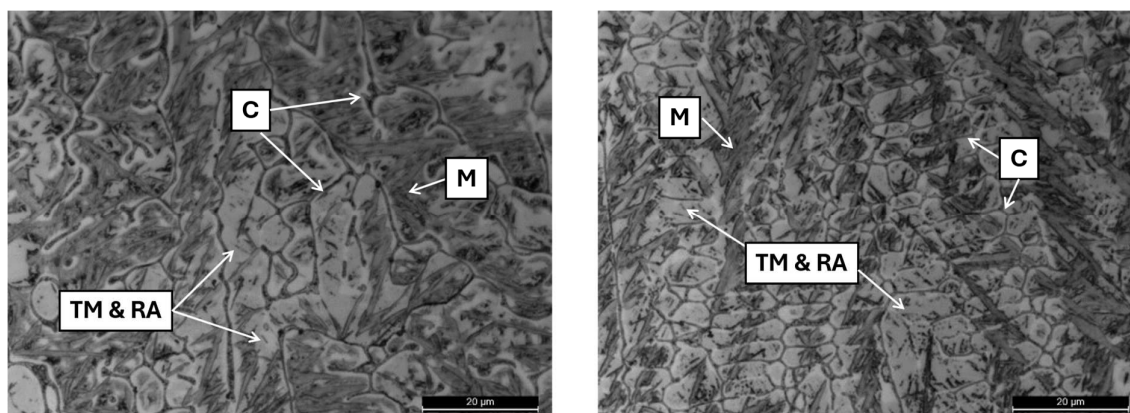


Figure 4.2-18 LOM of Microstructure of As-Built repaired sample,
M: Martensite, TM: Tempered Martensite, RA: Retained Austenite, C: Carbide

In many areas, these martensitic blocks are confined within the cells by retained austenite located at the cell boundaries, which appears bright in etched images. This retained austenite forms due to chemical segregation during rapid solidification, leading to a solute-rich interdendritic network composed of eutectic carbides and retained austenite. This network appears as thin continuous lines along the cell boundaries and helps stabilize the austenite while limiting the growth of martensite across these boundaries [69], [91].

4.2.3.2 Deposition zone of DTT repaired sample

As shown in figure 4.2-19, in the repaired samples after Direct Triple Tempering (DTT), the microstructure of the deposited region is similar to that observed in the bulk specimens. Triple tempering, conducted below A_{c1} , acts as a controlled softening treatment and does not induce full austenitization. Under these conditions, the as-built martensitic matrix undergoes stress relief and partial recovery, with gradual reduction of residual stress, partial decomposition of retained austenite, and carbide precipitation [74], [93], [94].

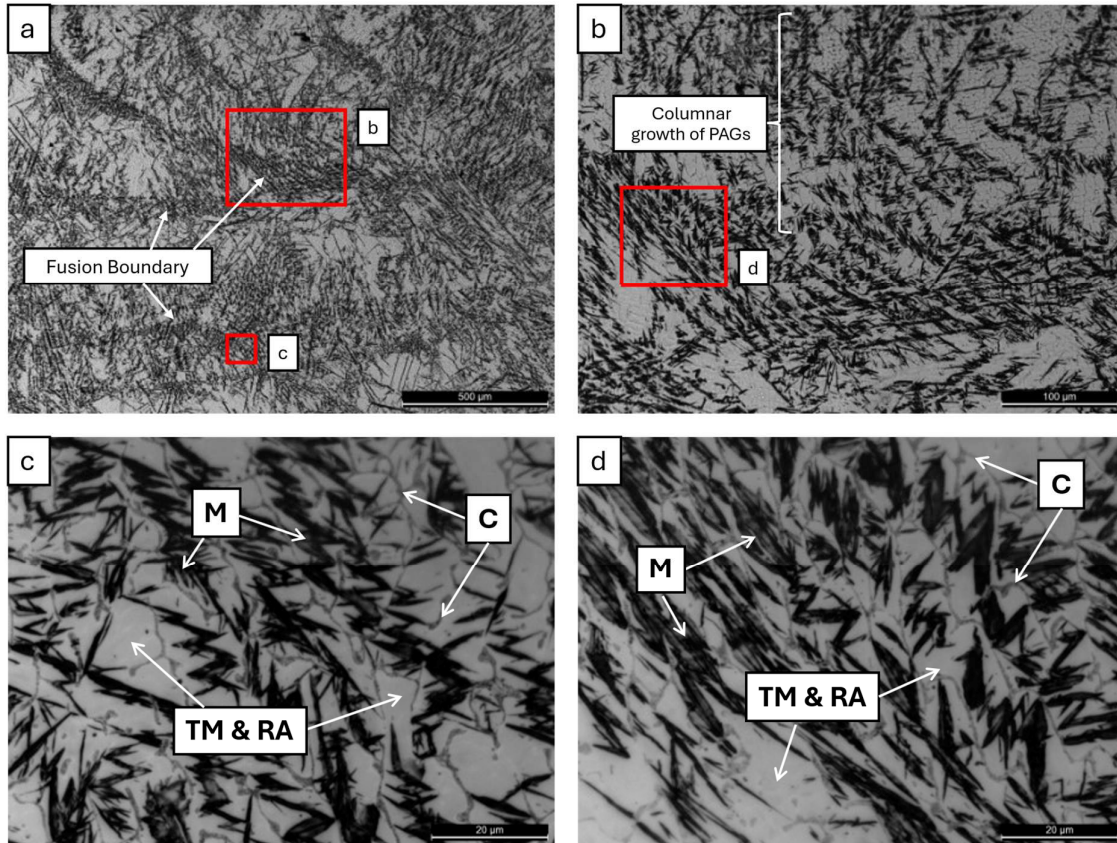


Figure 4.2-19 LOM image of microstructure of DTT repaired sample
a) Overlapping of layers and melt pool (x50) b) Fusion Boundary and Columnar Growth of PAGs (x200),
c) Microstructure near of SU (x1000), d) Microstructure into fusion boundary (x1000)
M: Martensite, TM: Tempered Martensite, RA: Retained Austenite, C: Carbide

Unlike the as-built condition, there is no gradual change in etching contrast or microstructure from the HAZ toward the center of the deposition; the structure is largely uniform. Only near the beginning of the deposition zone, at the interface with the HAZ, the etching appears slightly brighter, and the amount of retained austenite in the matrix is higher, which can be attributed to local thermal gradients during deposition and repeated inter-layer heating [56].

The primary DED microstructural features, including layer boundaries, cellular–dendritic substructure, and columnar epitaxial grains, are preserved. The martensitic lath

structure is more distinct and evenly distributed and previously observed circular or chaotic features at layer interfaces are absent. The bright matrix corresponds to a tempered martensitic–retained austenite mixture, with faint linear traces of segregation or fine tempering-induced carbides rather than solidification boundaries. Overall, the microstructure after triple tempering in the repaired deposition closely resembles the bulk condition while remaining structurally uniform [4], [19], [91].

4.2.3.3 Deposition zone of QTT repaired sample

As shown in Figure 4.2-20, the microstructure of the Quench and Triple Tempered (QTT) samples is very similar to bulk materials. After quenching and tempering, the DED-processed K340 steel mainly exhibits a tempered lath martensitic structure. The martensite is arranged in well-defined blocks and packets within each layer, indicating reduced internal stresses and a more homogeneous microstructure compared to the as-built and triple-tempered conditions. The original cellular–dendritic structure and the interlayer boundaries can still be recognized due to the thermal history of the process. The lighter regions correspond to tempered martensite with lower internal strain, while a small amount of retained austenite is present as thin films between martensitic laths. Fine and dispersed carbides formed during tempering are also observed, contributing to microstructural refinement and affecting the stability of retained austenite [26], [69], [93], [94].

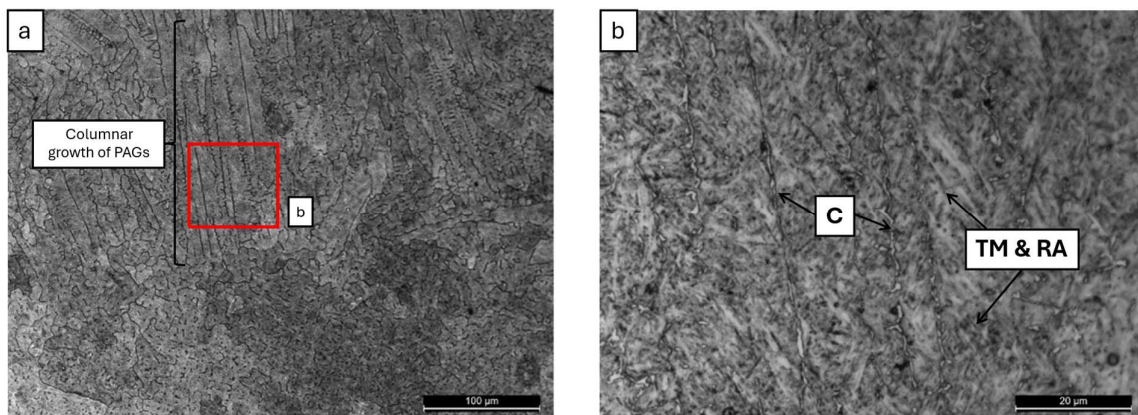


Figure 4.2-20 LOM image of microstructure of QTT repaired sample
a) Overlapping of layers and Columnar Growth of PAGs (x200), b) Microstructure Columnar Growth of PAGs (x1000)
TM: Tempered Martensite, RA: Retained Austenite, C: Carbide

4.2.3.4 Phase evaluation of Deposition zone in different conditions

The EDS point analysis results (Table 4.2-20) in the DED region for the As-Built (AB), Direct Triple Tempered (DTT), and Quenched & Triple Tempered (QTT) conditions reveal a clear evolution in carbide type, distribution, and formation mechanisms, directly governed by the thermal history of each condition. In combination with elemental mapping, and

shape of carbides, these data provide valuable insight into the carbide types and the microstructural evolution within the deposited material. The obtained compositions were interpreted by comparison with the carbide composition ranges reported in the tables 3.3-1 to 3.3-3.

Table 4.2-3 EDS of Element distribution in Deposited of Repaired sample in different condition

*AB: As-Built, DTT: Direct Triple Tempered, QTT: Quench and Triple Tempered									
Point	C	AL	Si	V	Cr	Mn	Fe	Nb	Mo
1-DED-AB	21.39	0.68	0.91	0.35	6.94	0.33	67.71	0.02	1.67
2-DED-AB	26.15	0.13	0.58	1.64	16.4	0.39	44.38	0.59	9.73
3-DED-AB	21.09	0.54	0.89	1.1	7.98	0.34	63.32	1.18	3.57
1-DED-DTT	8.88	0.5	0.5	0.44	8.73	0.25	78.67	0.06	1.97
2-DED-DTT	10.95	0.34	0.4	1.9	14.75	0.93	31.28	0.99	38.47
3-DED-DTT	10.38	0.43	0.4	0.59	9.96	0.39	75.56	nd	2.19
1-DED-QTT	17.18	0.39	0.39	2.75	22.98	0.37	51.37	0.27	4.32
2-DED-QTT	17.41	0.68	0.68	0.69	6.48	0.23	64.98	6.46	2.39

In the as-built condition, the microstructure is fully dominated by melt pool solidification. The composition of Point 2-DED-AB, with high Cr (~16 wt%) and Mo (~9.7 wt%), indicates the formation of Cr-rich carbides, that according to almost their massive shape, it is likely to be M_7C_3 , within interdendritic regions. These carbides result from microsegregation during rapid solidification. In contrast, points such as 1-DED-AB, with high Fe (~68 wt%), correspond to the martensitic matrix. In Point 3-DED-AB, the presence of Nb along with Cr and Mo suggests the formation of mixed carbides or the initiation of Nb-rich MC carbides..

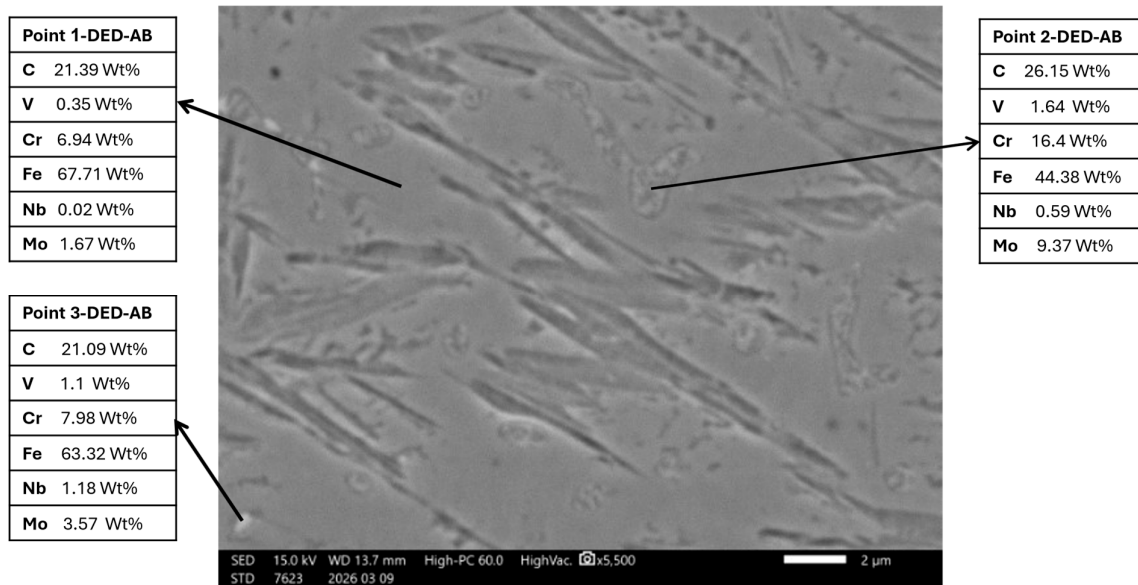


Figure 4.2-21 EDS points analysis of Deposited region of As-Built sample

These observations are fully consistent with the elemental maps showing a clear Cr-Mo lattice. As shown in Figure 4.2-22, the accumulation of Cr and Mo elements, followed by the other elements, is formed in a ring and chain pattern, indicating that micro-segregation of alloying elements such as C, Cr, and Mo in the interdendritic regions during the final stages of solidification. The increased chemical heterogeneity in these regions leads to stronger etching response and higher contrast .

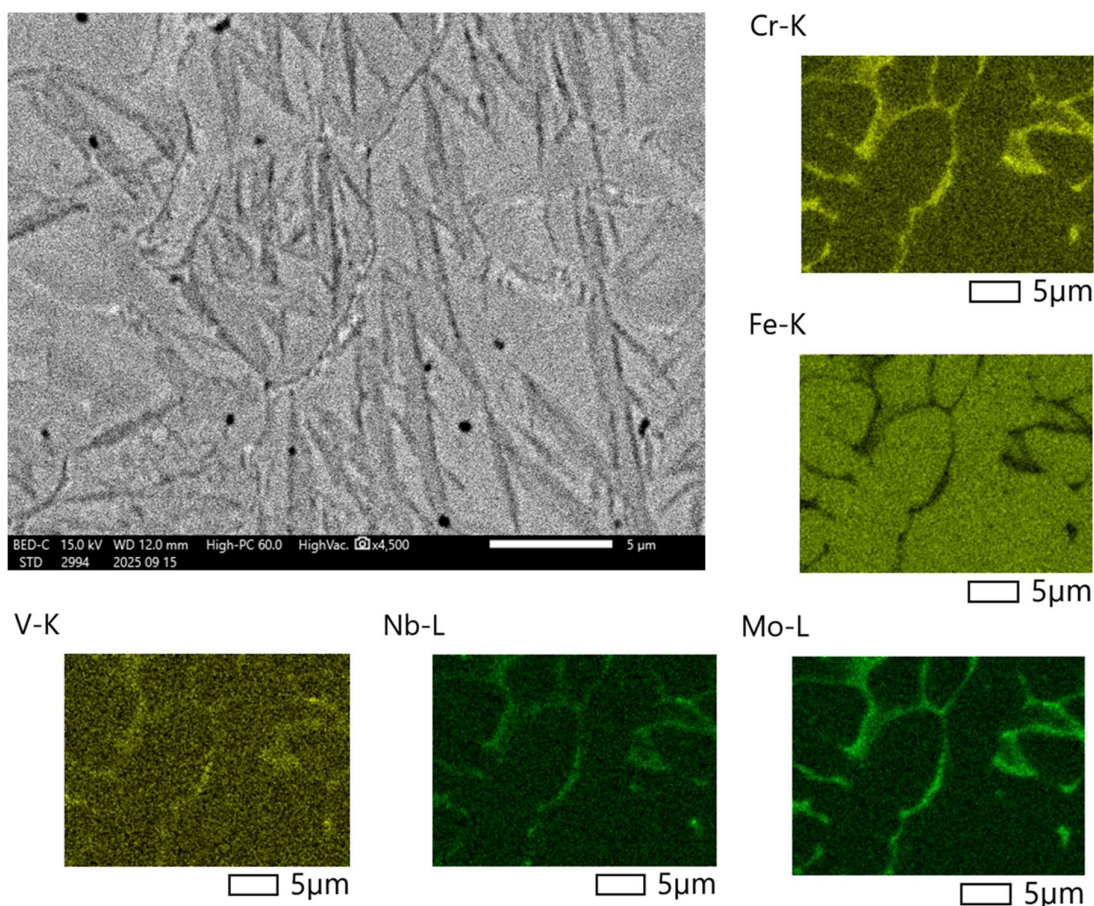


Figure 4.2-22 MAP of Deposited of As-Built repaired sample

In the Direct Triple Tempered (DTT) condition, although tempering reduces residual stresses and modifies the martensitic structure, the microstructure is now influenced by secondary carbide precipitation, the solidification-induced segregation pattern is not eliminated (Figure 4.2-23, 4.2-24). Notably, point 2-DED-DTT exhibits extremely high Mo (~38 wt%), indicating the formation of Cr-Mo-rich carbides (likely $M_{23}C_6$ or complex carbides) as secondary precipitates during tempering, as suggested by the reference tables. Point 1-DED-DTT represents the tempered martensitic matrix. Additionally, point 3-DED-DTT, which is located on martensite, shows a composition close to the matrix (high Fe, low alloying elements), confirming that no distinct carbide is present at this location. The reduced Nb signal in the maps also suggests its limited role in carbide formation in this condition.

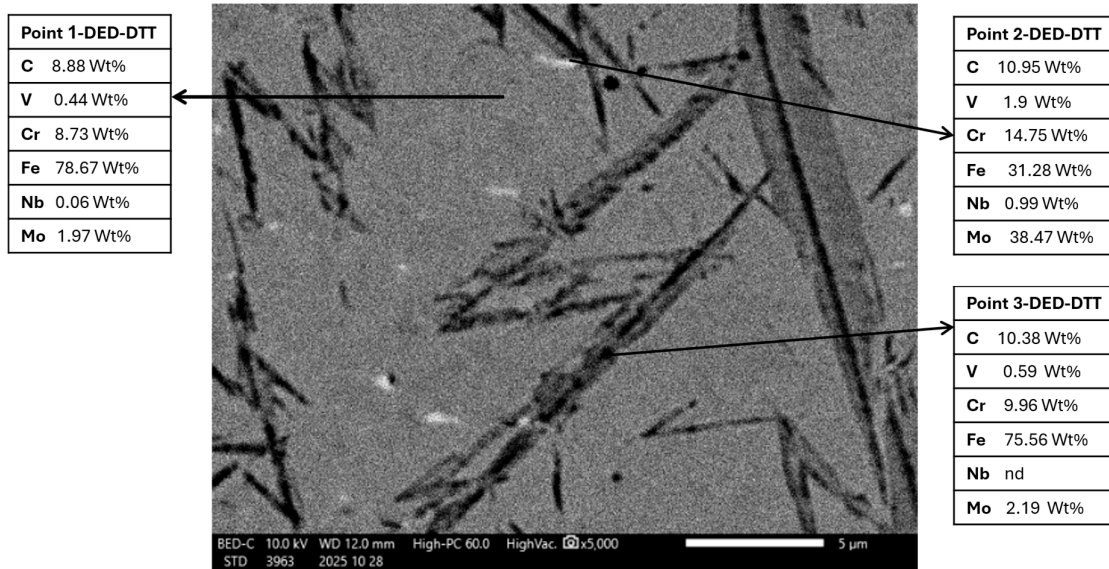


Figure 4.2-23 EDS points analysis of Deposited region of DTT repaired sample

In the DTT map images, the accumulation of Cr and Mo elements is scattered but massive with a small amount of V, which, given their appearance, are probably primary carbide M_7C_3 . In a few bright spots, a combination of all alloys and Nb is observed, which is secondary carbide $M_{23}C_6$.

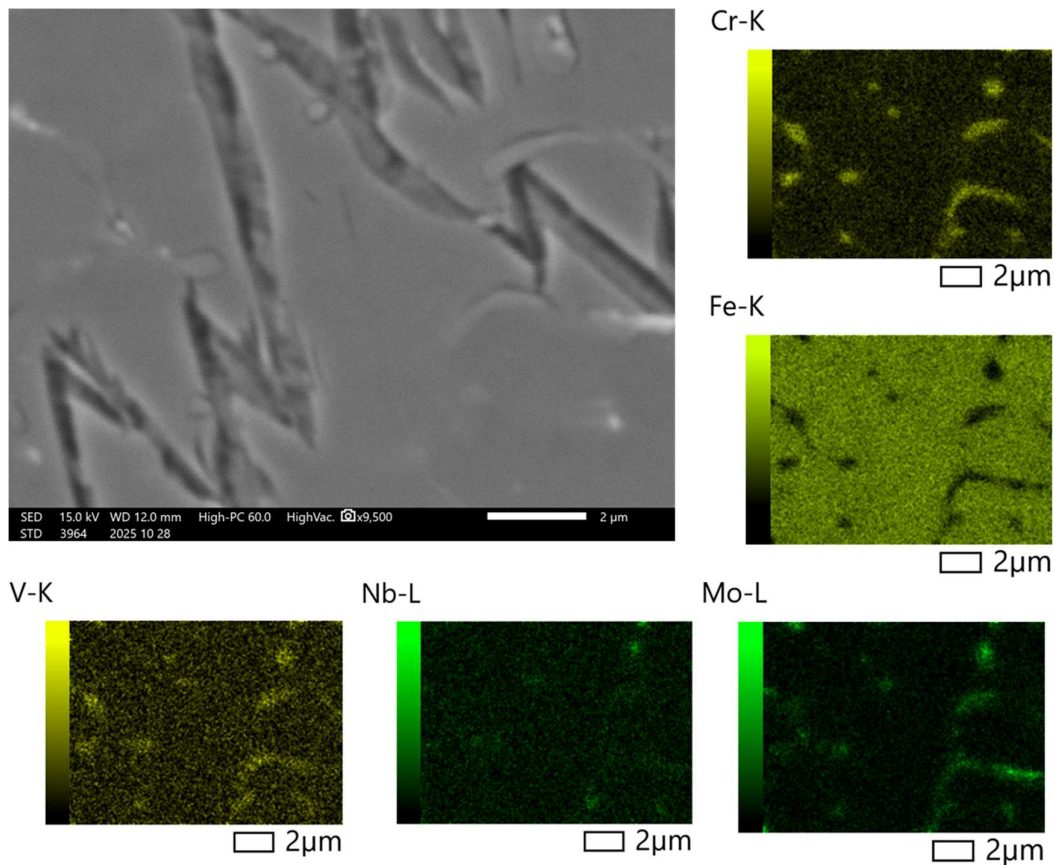


Figure 4.2-24 MAP of Deposited of DTT repaired sample

In the Quenched and Triple Tempered (QTT) condition, re-austenitization followed by quenching leads to a more homogenized microstructure, traces of prior segregation remain. However, EDS results indicate that certain carbides remain stable. In Point 1-DED-QTT, the high Cr (~23 wt%) along with V suggests the presence of Cr-rich carbides with V contribution, and with respect to narrow and film shape of them, which may correspond to $M_{23}C_6$. In Point 2-DED-QTT, the significant Nb content (~6.4 wt.%) and being bright spot, clearly indicates the presence of highly stable Nb-rich MC carbides, which persist even after QTT.

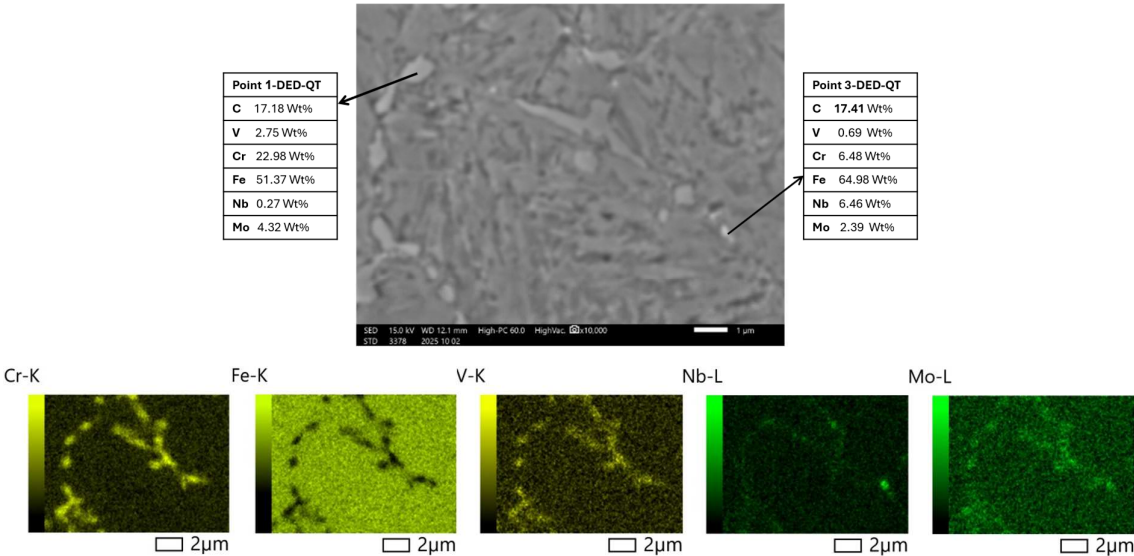


Figure 4.2-25 EDS points analysis and MAP of Deposited region of QTT repaired sample

4.2.4 EBSD Analysis of Interface and Adjacent Regions

Based on the microstructural observations presented in the previous sections, EBSD analysis was performed across the HAZ and DED regions to further clarify the phase distribution, grain structure, and local misorientation.

The analyzed surface is a polished cross-section of a sample produced by the Directed Energy Deposition and examined under a scanning electron microscope. Due to the layer-by-layer nature of this process, the microstructure is formed as a series of stacked layers. In this section, three main directions are defined for crystallographic analysis. The Y-axis represents the build direction (BD), which is perpendicular to the layers. The X-axis corresponds to the transverse or scanning direction within each layer, while the Z-axis lies in the section plane and is perpendicular to X.

From a morphological point of view, the AB, DTT, and QTT conditions show clearly different behaviors. In the as-built (AB) condition (Figure 4.2-26), the DED region consists of very coarse columnar grains elongated along the build direction, which typically span over several deposited layers due to the epitaxial growth from the previously solidified

underlying tracks [77], while the heat-affected zone (HAZ) shows a fine, mosaic-like structure with random orientation. After Direct Triple Tempering (DTT), the overall morphology remains largely unchanged, and the coarse columnar grains are still present, but their internal structure becomes more ordered (Figure 4.2-27). In contrast, the QTT sample (Figure 4.2-28) shows a complete transformation: the coarse columnar grains disappear and are replaced by a fine, uniform martensitic structure with no visible layering or preferred direction, resulting in a fully isotropic behavior.

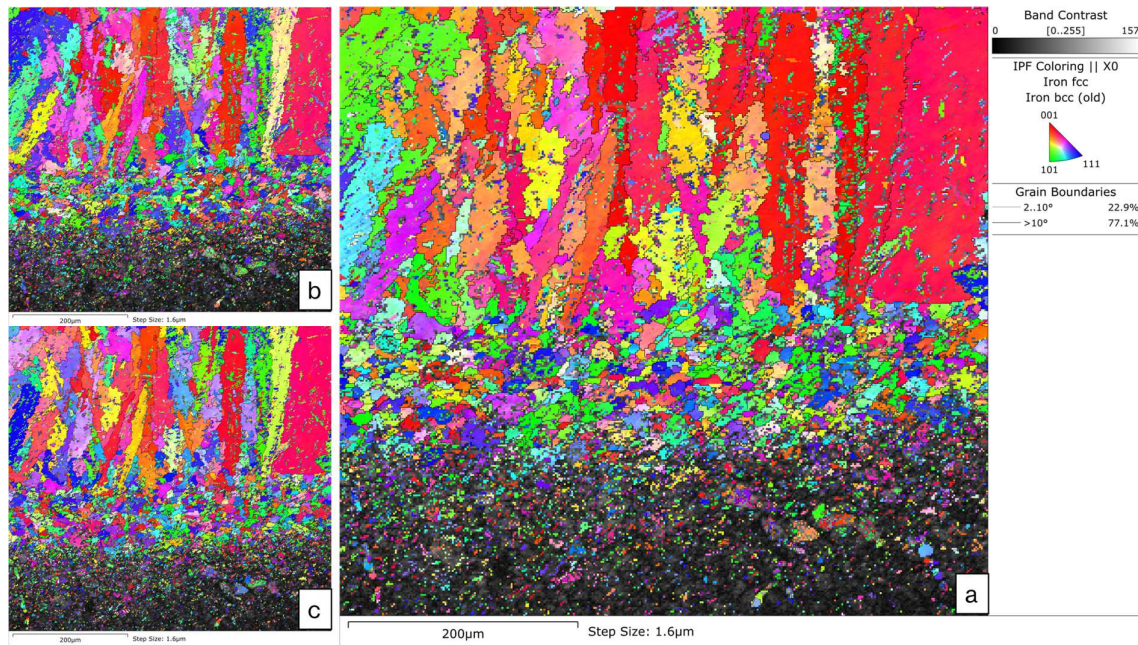


Figure 4.2-26 IPF coloring of As-Built repaired Sample in a) X0, b) Y0, c) Z0 directions

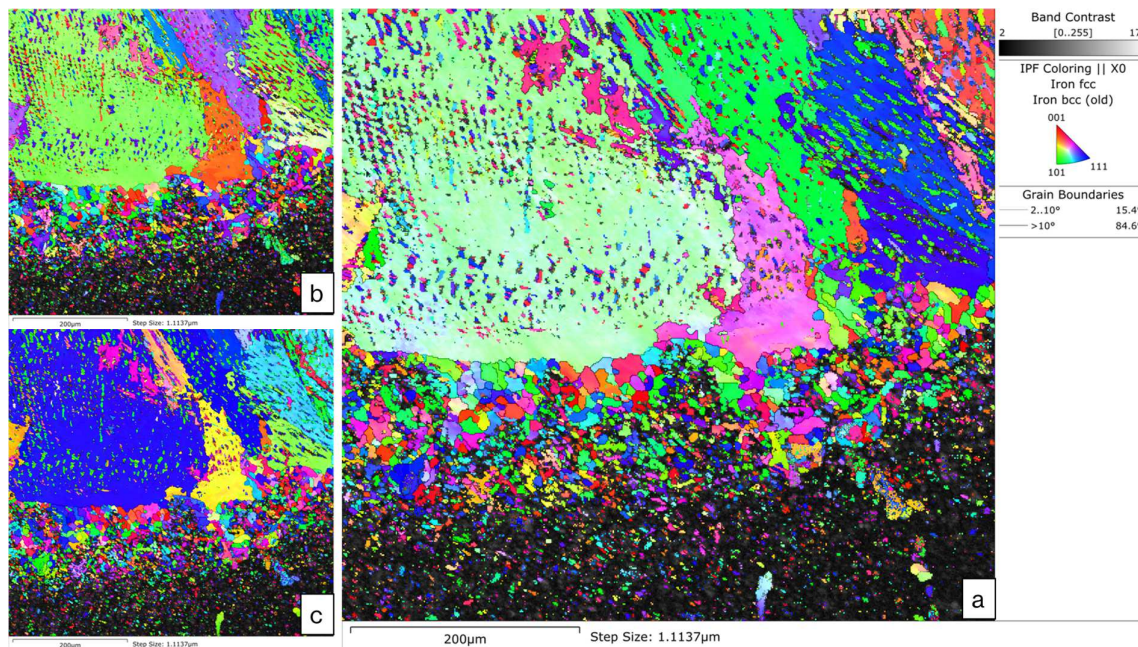


Figure 4.2-27 IPF coloring of DTT repaired Sample in a) X0, b) Y0, c) Z0 directions

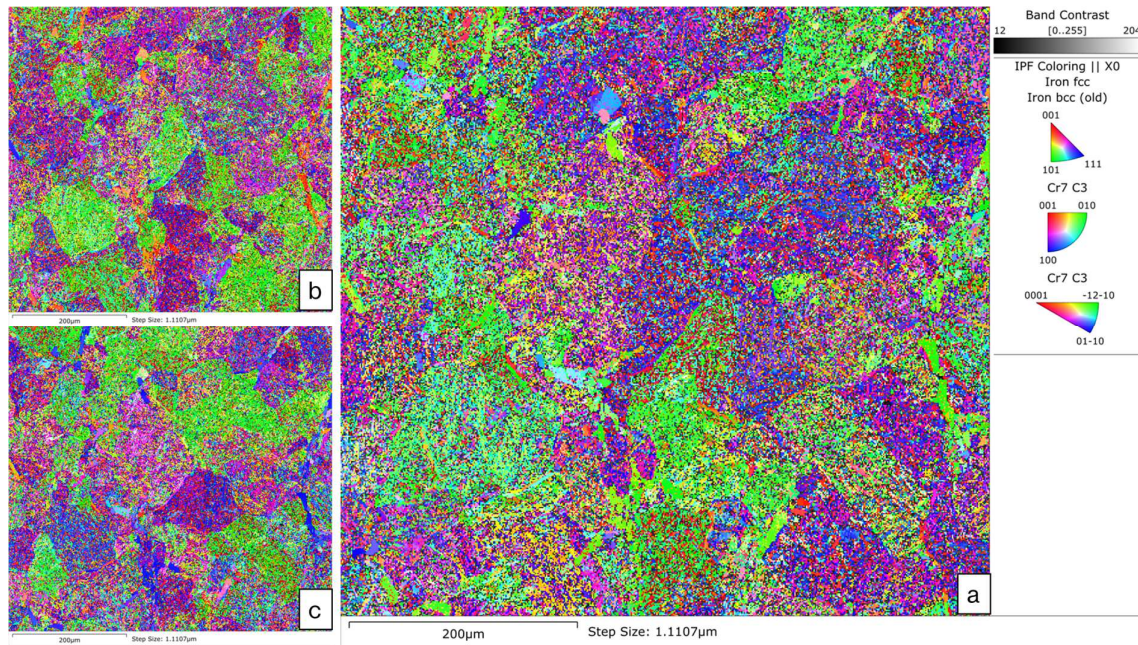


Figure 4.2-28 IPF coloring of QTT repaired Sample in a) X0, b) Y0, c) Z0 directions

From a quantitative perspective, the grain size statistics extracted from the EBSD datasets provide a definitive confirmation of the microstructural evolution across the as-built (AB), DTT, and QTT conditions. The grain size distribution characteristics for each state are summarized in Figures 4.2-29 to 4.2-31.

The As-Built (AS) state (Figure 4.2-29) shows a distinctive bimodal distribution (ASTM 9.6); the high frequency of fine grains (mean: 63.96 pix) represents the mosaic-like intercellular regions/HAZ zones, while the isolated peak at 6521 pix confirms the coarse columnar prior-austenite grains from directional solidification.

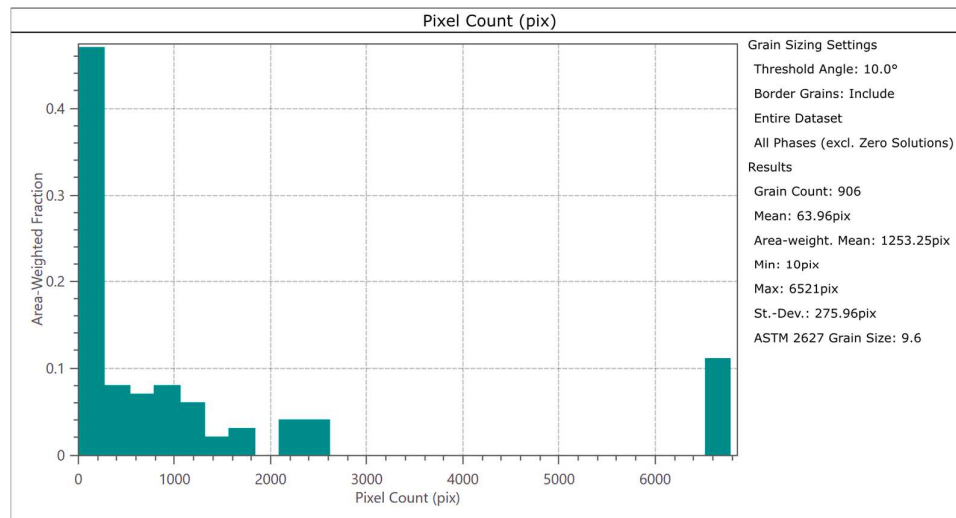


Figure 4.2-29 Grain size distribution charts obtained from EBSD analysis of As-built (AB)

Upon direct triple tempering (DTT) (Figure 4.2-30), the hierarchical solidification morphology is not eliminated, but rather intensified. The grain count increases to 1520 (ASTM 10.1) due to the partial decomposition of retained austenite into fine tempered martensite. However, the maximum grain size expands drastically to 52058 pixels, proving that the coarse columnar solidification texture remains highly stable and active thermal grain boundary migration occurs.

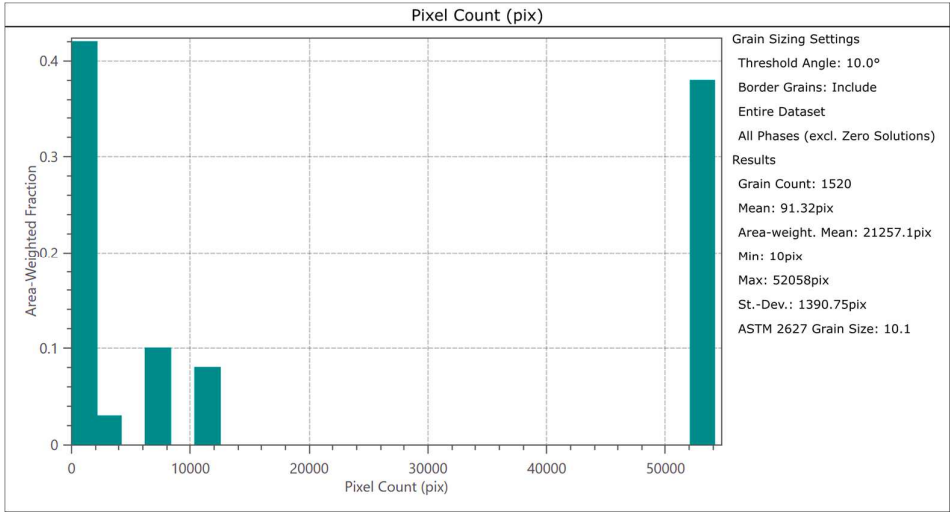


Figure 4.2-30 Grain size distribution charts obtained from EBSD analysis of DTT

In contrast, the QTT condition (Figure 4.2-31) produces a complete microstructural transformation. The previous bimodal distribution disappears and is replaced by uniform grain structure with an ASTM grain size number of about 12.6. The coarse columnar grains are fully eliminated and replaced by refined martensitic blocks and laths, resulting in a homogeneous and isotropic microstructure. These results confirm that austenitization followed by quenching and tempering effectively removes the solidification-induced structure and promotes full recrystallization and phase transformation.

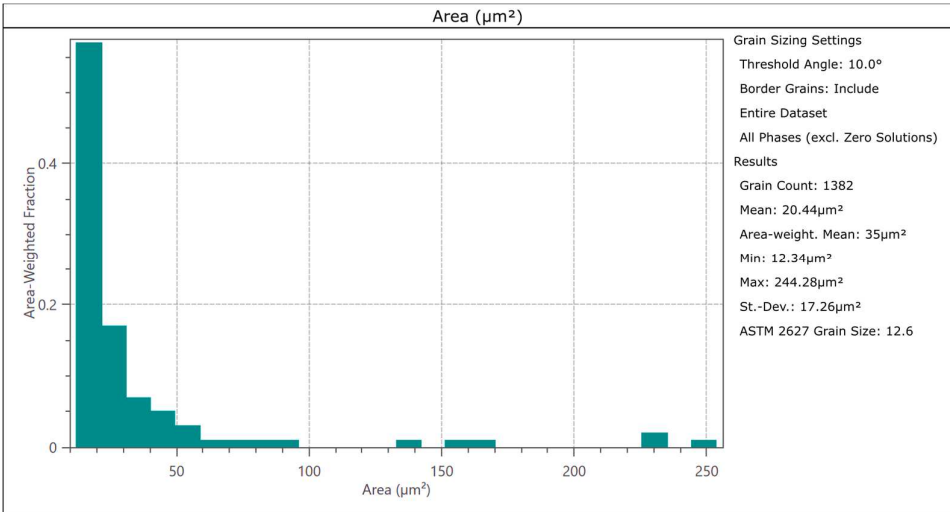


Figure 4.2-31 Grain size distribution charts obtained from EBSD analysis of QTT.

Phase analysis combined with IPF maps gives a clearer picture of the microstructure (Figure 4.2-32). In the AB and DTT conditions, about 55.3% and 53.4% of the DED region consist of retained austenite (FCC), which is stabilized by rapid cooling and alloying effects. More specifically, as demonstrated by Jäggle et al. (2016)[97] and Chou et al. (2021)[77] during laser additive manufacturing of steels, solute elements undergo severe microsegregation into the interdendritic/inter-cellular regions during directional solidification. This localized enrichment of alloying elements significantly lowers the martensite start (M_s) temperature and leads to the mechanical and chemical stabilization of a high fraction of retained austenite within these regions. This shows that tempering alone is not sufficient to fully transform austenite into martensite. In the IPF maps, this region is characterized by columnar grains with preferred orientations close to $\langle 001 \rangle$ and $\langle 101 \rangle$ along the build direction, indicating directional solidification. This anisotropy is still present after DTT [98].

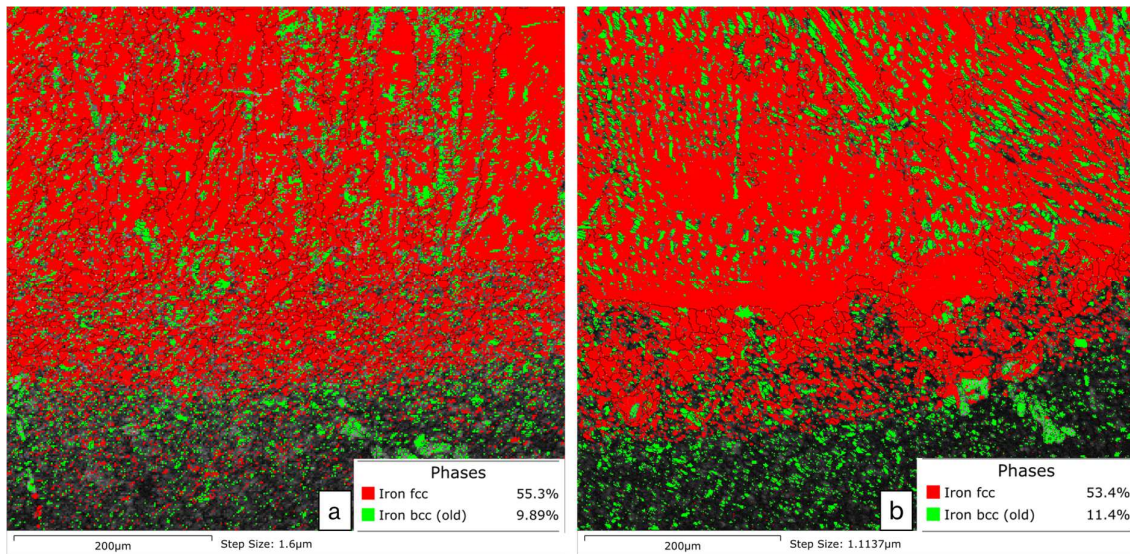


Figure 4.2-32 EBSD of Phases of a) AB, b) DTT repaired Sample

In the QTT sample (Figure 4.2-33), the retained austenite (FCC) is significantly reduced to 0.77% , while the microstructure is mainly composed of BCC martensite (60.3%) , confirming that quenching followed by triple tempering effectively transforms the matrix into a stable martensitic structure. The random color distribution in the IPF maps indicates that the directional solidification texture has been completely removed, resulting in a homogeneous and nearly isotropic microstructure. The phase map also shows the presence of hexagonal Cr_7C_3 (10.9%) and a smaller fraction of orthorhombic Cr_7C_3 (1.47%). These carbides mainly appear as thin, elongated, and continuous particles, suggesting preferential precipitation along prior austenite grain boundaries, martensitic lath boundaries, or Cr-rich segregation paths. Although the QT treatment homogenizes the martensitic matrix, these elongated Cr_7C_3 carbides remain as local heterogeneities that contribute to secondary hardening while locally affecting lattice distortion and crack propagation behavior.

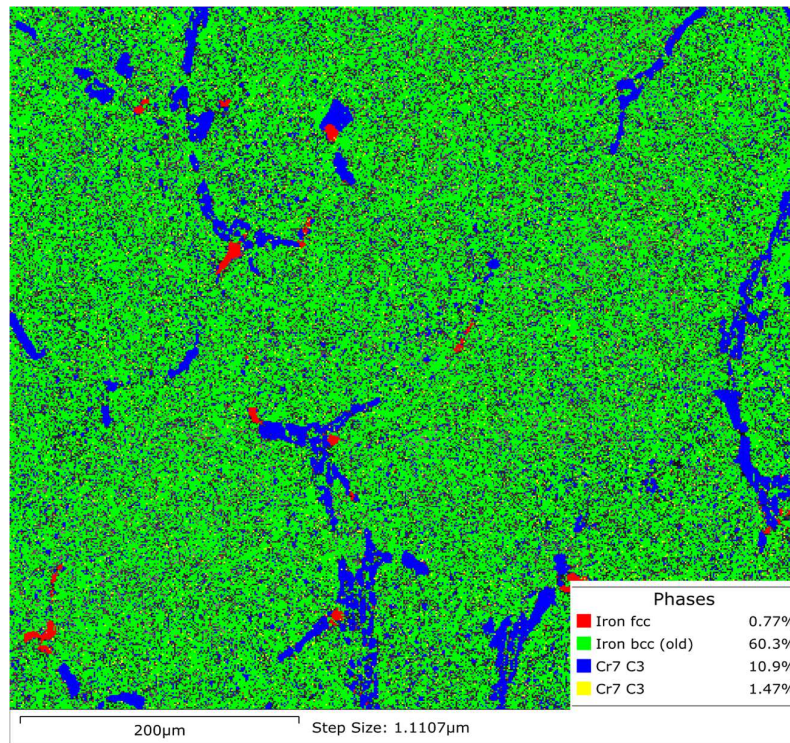


Figure 4.2-33 EBSD of Phases of QTT repaired Sample

According to Grain Orientation Spread, lattice distortion parameters, in the AB condition, GOS is about 6° , and appears as green and yellow regions inside grains, indicating high internal strain and orientation variation. In DTT, it decreases to about $4\text{--}5^\circ$, and the maps become more blue points, showing recovery and reduced internal strain (Figure 4.2-34). In QTT, GOS may increase again to around $6\text{--}7^\circ$, but this appears as fine, scattered variations rather than large patches, reflecting a finer structure (Figure 4.2-35).

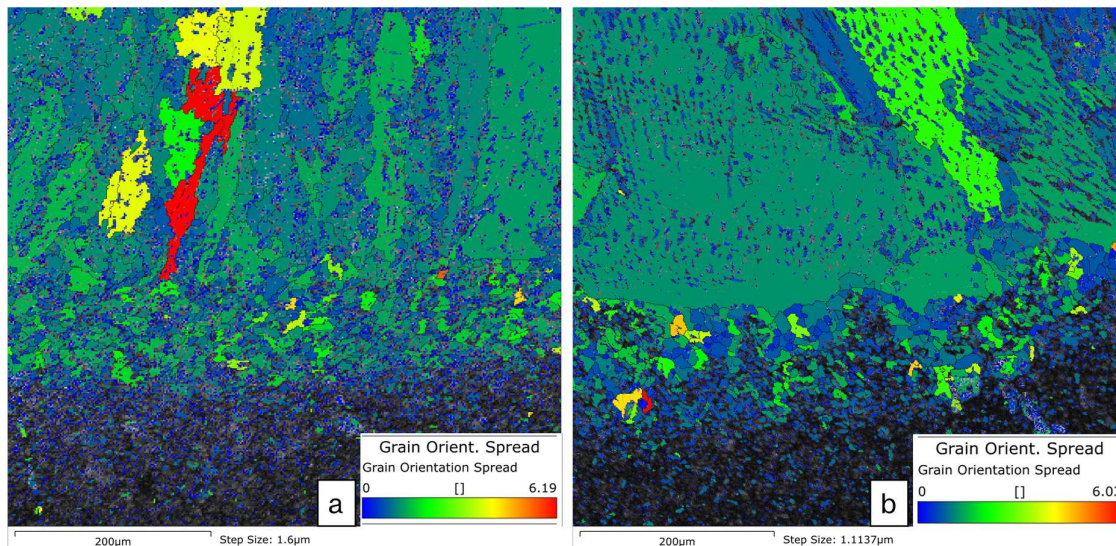


Figure 4.2-34 GOS of a) AB, b) DTT repaired sample

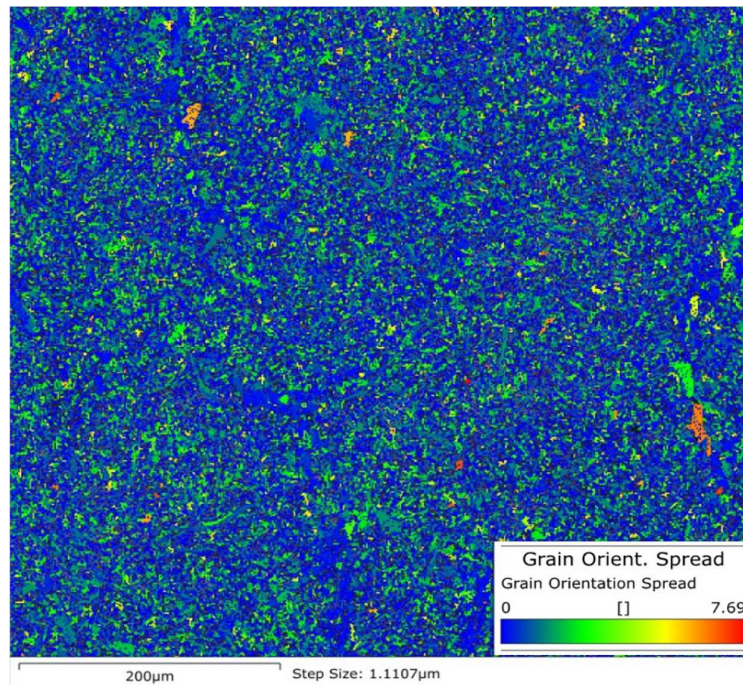


Figure 4.2-35 GOS of QTT repaired sample

A similar trend is observed for GROD (Figure 4.2-36 and 4.2-37). In AB, it is about 13.3° , indicating strong lattice distortion caused by solidification stresses. In DTT, it decreases to about 10.9° , showing stress relief. In QTT, it increases again to around $15\text{--}16^\circ$. Although this higher value may initially appear counterintuitive, the origin of this lattice strain state is fundamentally different. In QTT, the structure consists of very fine lath martensite and dispersed carbides. These features introduce local misorientations in the lattice, which increase the measured values, but this is inherently related to microstructural complexity rather than harmful residual stress provided by manufacturing.

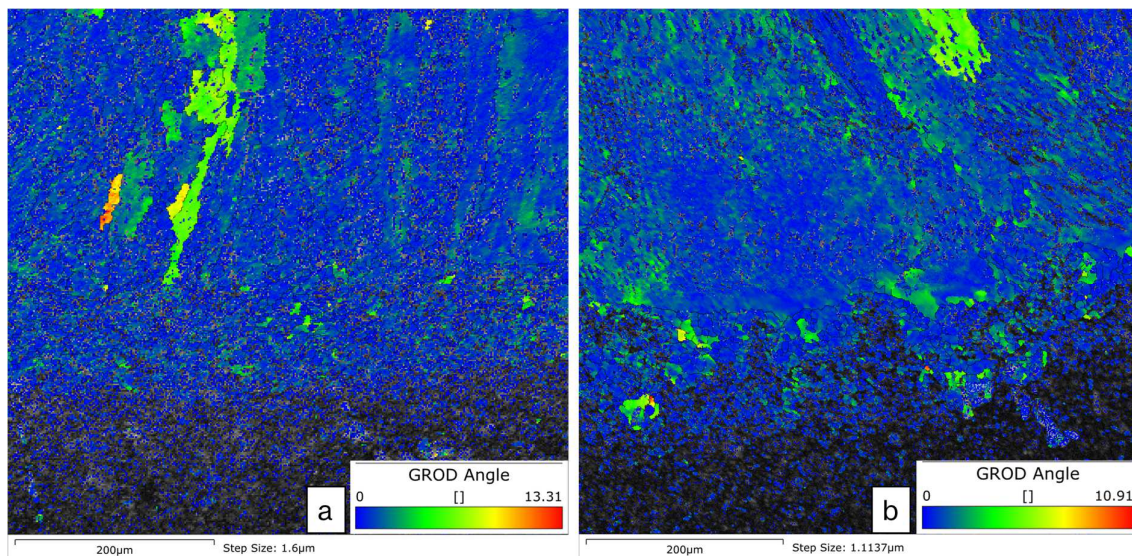


Figure 4.2-36 GROD of a) AB, b) DTT of repaired sample

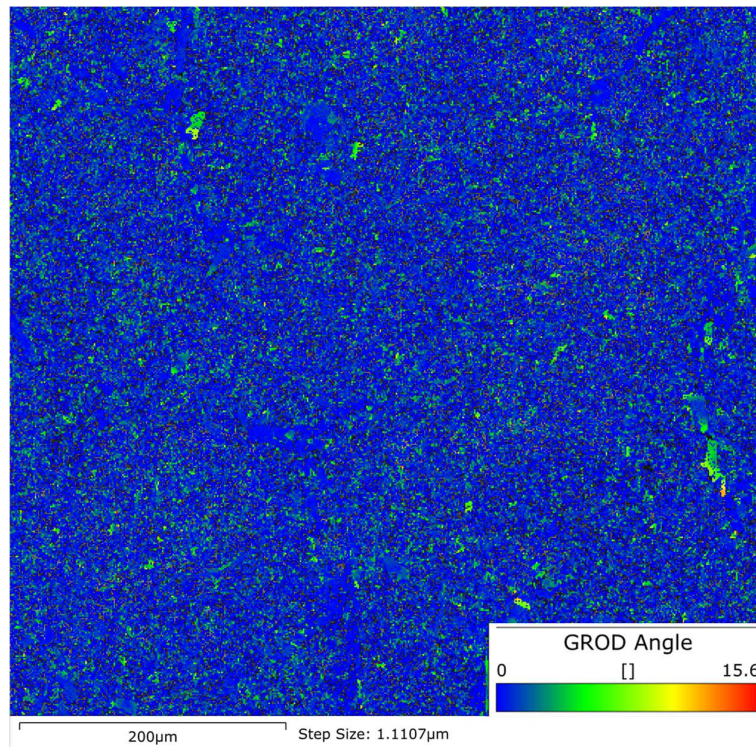


Figure 4.2-37 GROD of QTT of repaired sample

The KAM parameter confirms this interpretation. In AB, KAM is about 4.9° and appears as green and yellow regions along grain boundaries and into grains, indicating high local strain from solidification. In DTT, it remains almost at the same level (4.8°), and the maps become more uniform and bluer, showing stress relaxation. In QTT, KAM remains in the range of $4.5\text{--}5^\circ$, but its distribution changes: instead of being spread across large grains, it becomes localized around martensitic laths and carbides. This indicates transformation-induced strain rather than residual stress from processing (Figure 4.2-38 and 4.2-39).

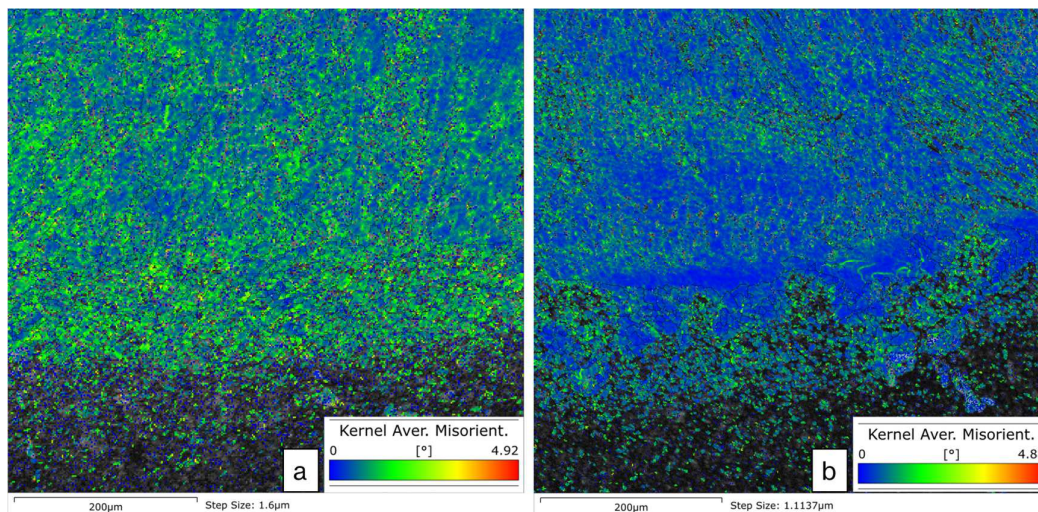


Figure 4.2-38 KAM of AB, DTT of repaired sample

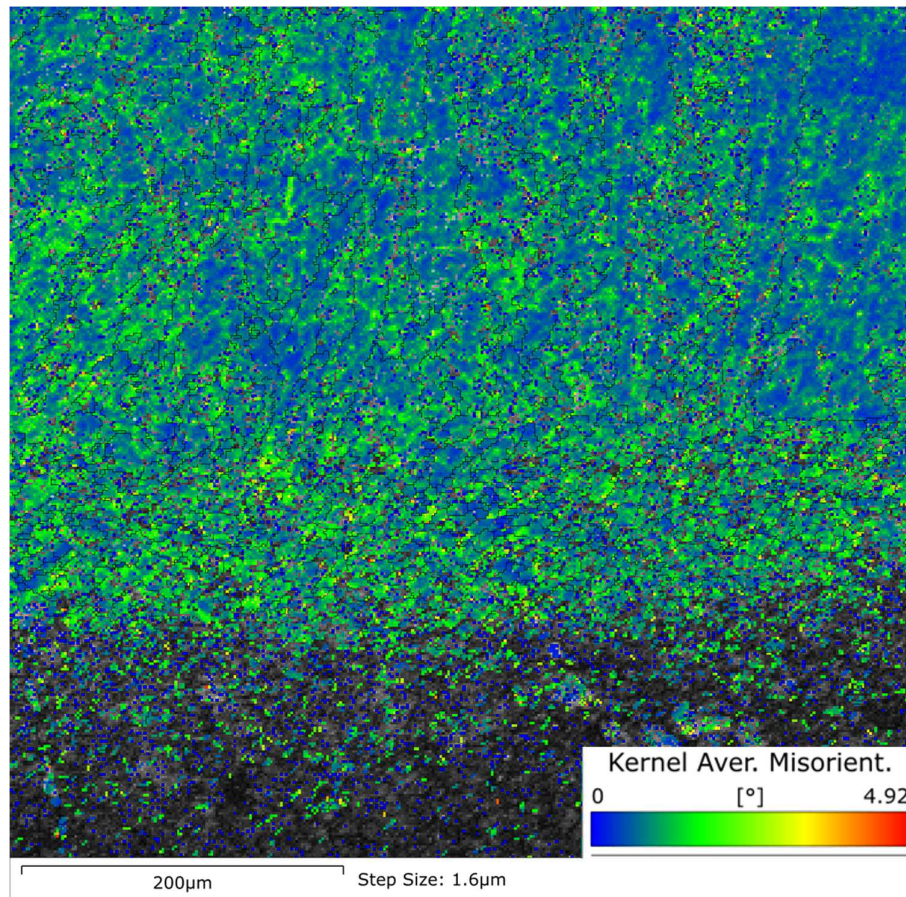


Figure 4.2-39 KAM of QTT of repaired sample

4.2.5 Thermo-Calc Simulation of Solidification and Carbide Evolution

Thermo-Calc simulations were performed to better understand the solidification behavior, elemental segregation, phase stability, and carbide evolution in the DED-processed K340 tool steel. Different simulation modules were used to investigate microstructural evolution during the as-built condition and subsequent heat treatment processes. For all calculations described above, the composition Fe 1.08C, 7.86Cr, 0.41Mn, 1.52Mo, 0.61Si, 0.40V, 0.68Al, 0.12Nb (weight-percent, wt.%) was used. The Thermo-Calc 2026a, TCFE15 (Steels/Fe-Alloys v15.0), MOBFE9 (Steels/Fe-Alloys Mobility v9.0) databases.

4.2.5.1 Solidification and Micro-segregation Modeling in the As-Built State

Due to the extremely high cooling rates during the Directed Energy Deposition (DED) process, the K340 tool steel develops a non-equilibrium dendritic microstructure accompanied by strong microsegregation of alloying elements in the interdendritic regions. To simulate the solidification behavior and elemental redistribution during the as-built condition, Scheil-Gulliver simulations were performed using Thermo-Calc, and the results are presented in Figure 4.2-40.

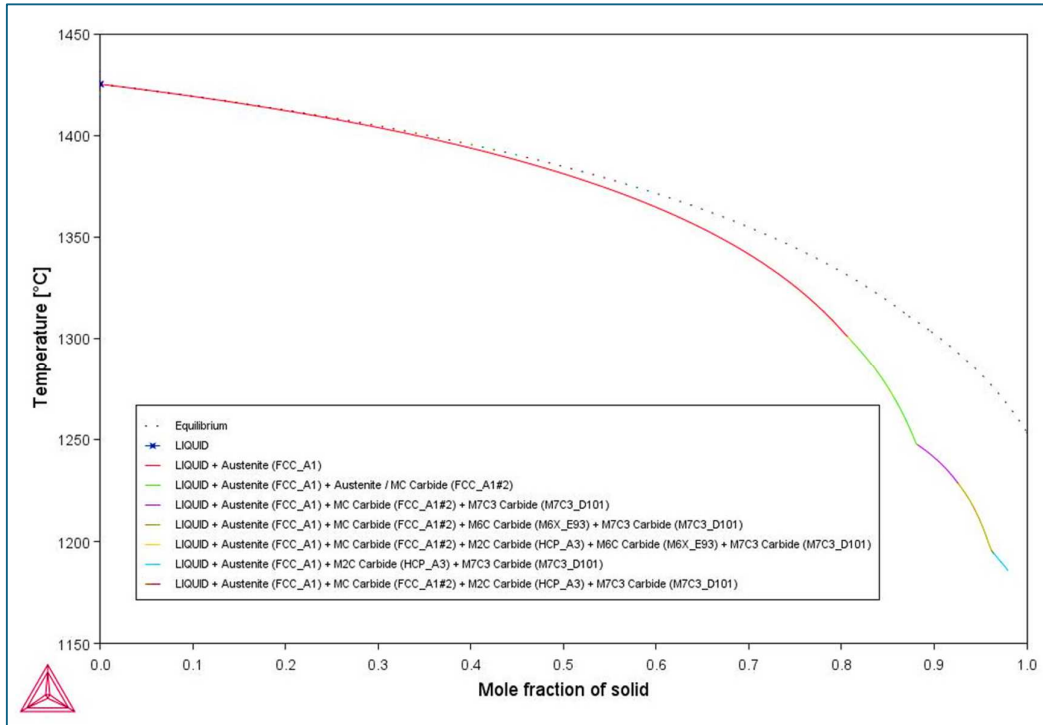


Figure 4.2-40 Scheil-Gulliver simulation of DED process by Thermo-Calc.
Equilibrium condition: point line, Non equilibrium condition (DED) : colored line

Since the conventional Scheil–Gulliver computational model assumes zero interstitial diffusion in the solid phase by default, this simplification introduces significant inaccuracies for light interstitial elements such as carbon and nitrogen, which exhibit exceptionally high diffusivity in steels. Therefore, following the methodology established by Chou et al. (2021)[77], carbon and nitrogen were designated as “Fast Diffusers” during the Scheil–Gulliver solidification path simulation of the K340 tool steel within the Thermo-Calc software. This modification explicitly accounts for the back-diffusion phenomenon of interstitial solutes into the solidifying phase, thereby ensuring a highly accurate prediction of both the constituent phase evolutions and the critical transformation temperatures [99].

The Scheil model assumes negligible diffusion in the solid phase and complete diffusion in the liquid phase, which reasonably represents the rapid solidification conditions occurring during additive manufacturing. The comparison between equilibrium and Scheil solidification paths shows a significant gap in the final solidification temperature. Under equilibrium conditions, solidification ends at a higher temperature, whereas in the Scheil simulation solidification continues down to approximately 1180 °C. This difference confirms the occurrence of severe microsegregation in the as-built microstructure.

According to the simulation, solidification starts with the formation of austenite (FCC_A1) at approximately 1425 °C. As solidification progresses, alloying elements such as Cr, V, Nb, and Mo are gradually rejected into the remaining liquid, leading to enrichment of the interdendritic regions. During the final stage of solidification, when the liquid fraction becomes very small, primary MC and M_7C_3 carbides begin to precipitate from the highly segregated liquid. At even lower temperatures, complex Mo-rich carbides such as M_6C and M_2C may also form [19], [69].

The phase evolution observed in the Scheil diagram demonstrates that the carbide network observed in the as-built microstructure originates directly from the last solidified liquid. Furthermore, the large separation between equilibrium and non-equilibrium solidification curves indicates the strong chemical heterogeneity generated during the DED process.

4.2.5.2 Simulation of Precipitation Kinetics during Direct Triple Tempering (DTT)

A critical challenge in simulating precipitation kinetics is the accurate determination of the supersaturated matrix composition. In this study, instead of relying directly on experimental EDS data for interdendritic regions, the "Peak Concentration" derived from the Scheil-Gulliver model was utilized. This methodology is substantiated by the findings of Chou et al. [77] and Baptista et al. (2020) [99]. Their research indicates that EDS analysis in highly segregated areas is prone to error because the electron beam's interaction volume captures both the matrix and the nano-sized solidification carbides that form at cell junctions during the final stages of solidification. This interaction leads to unrealistic elemental percentages, particularly for carbon.

To elucidate the distribution of alloying elements within the as-built microstructure, the chemical evolution of the liquid phase during solidification was simulated using the Scheil-Gulliver model. Solidification in high-alloy steels is inherently non-equilibrium, where the forming solid phase (austenite) cannot fully accommodate alloying elements within its lattice.

As illustrated in Figure 4.2-41, the horizontal axis represents the mass fraction of the remaining liquid (f_L) while the vertical axis shows the mass percentage of Chromium in the liquid. At the onset of solidification ($(f_s \rightarrow 1)$ or $(f_L \rightarrow 0)$), the Chromium concentration in the liquid equals the nominal alloy composition (~ 7 wt.%). As dendritic arms grow, Chromium is rejected from the solid phase into the liquid due to a partition coefficient of less than unity. This phenomenon leads to the progressive enrichment of the residual liquid. As the liquid fraction decreases, the concentration of alloying elements intensifies, reaching a maximum value (Peak). This critical point corresponds to the chemistry of the final solidified liquid, which serves as the origin of the interdendritic regions. Any subsequent decrease in Chromium concentration after the peak is typically attributed to the onset of secondary phase formation, such as primary carbides, which

begin to consume alloying elements directly from the melt. In agreement with Baptista et al. (2020) [99], the peak concentration at terminal solidification was selected as the PRISMA matrix input, since rapid cooling suppresses long-range diffusion and locks this highly segregated, maximum supersaturation state, into the lattice via solute trapping.

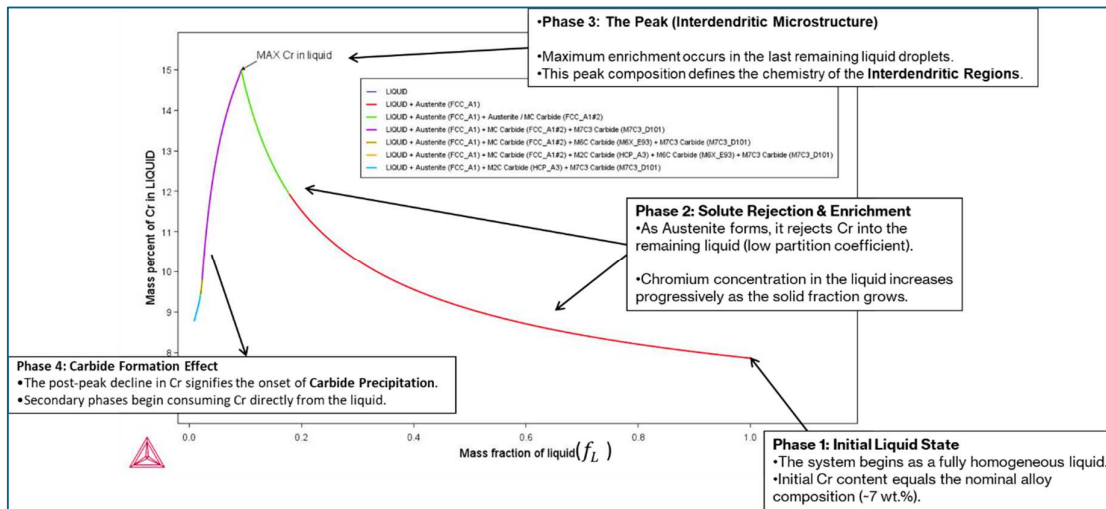


Figure 4.2-41 Peak Concentration" of Cr derived from the Scheil-Gulliver model of during solidification of As-Built

In this study, the "Peak Concentration" extracted from the Scheil model (as detailed in Table 4.2-4) was utilized as the primary input for precipitation kinetics modeling (PRISMA). This methodological choice is justified by the "Solute Trapping" phenomenon characteristic of the DED process. Due to extreme cooling rates, diffusional transformations are suppressed, causing this enriched chemical composition to be "frozen" within the martensitic lattice. Utilizing these peak values provides a more accurate representation of the matrix supersaturation and the secondary hardening potential compared to raw EDS data. This is because EDS interaction volumes often capture an average of both the matrix and nano-sized carbides, failing to isolate the true concentration of atoms trapped within the lattice.

Table 4.2-4 Comparison of Mass Percentage of alloys in liquid during solidification with nominal composition of K340

Mole fraction of liquid	Mass percentage of alloys in LIQUID						
	Cr	Mo	V	Mn	Si	Al	Nb
0.09779	14.98	5.59	2.24	0.63	0.50	0.053	0.29
Nominal Composition	7.86	1.52	0.4	0.41	0.61	0.68	0.12

Using the TC-PRISMA module, the carbide precipitation kinetics were simulated during the tempering cycle. The evolution of the dendritic core was compared against the interdendritic regions.

A comparative analysis of the Mean Radius plots (Figure 4.2-42) reveals the mean radius results indicate that carbide growth accelerates after approximately 15,000 s,

corresponding to enhanced diffusion and precipitate coarsening during tempering. In both regions, $M_{23}C_6$ and M_7C_3 carbides exhibit the largest growth, while MC carbides remain relatively fine and stable.

In addition, Although the mean radius plots exhibit a lower growth slope in the interdendritic regions, precipitation kinetics initiate significantly faster in the interdendritic regions than in the dendritic core. This apparent discrepancy is attributed to the exceptionally high nucleation density in these segregated zones. The elevated thermodynamic driving force leads to the formation of a vast number of carbide nuclei. The extreme enrichment of alloying elements (Cr, Mo, V) in these zones, caused by solidification segregation, drastically increases the thermodynamic driving force (ΔG) for carbide nucleation[100]. This results in a high nucleation rate and the formation of a dense population of nano-sized carbides at very early stages of tempering. Due to the high density of nucleation sites, the available solute is distributed among a larger number of particles, which maintains a finer mean radius in the interdendritic zones for a longer duration compared to the dendritic core.

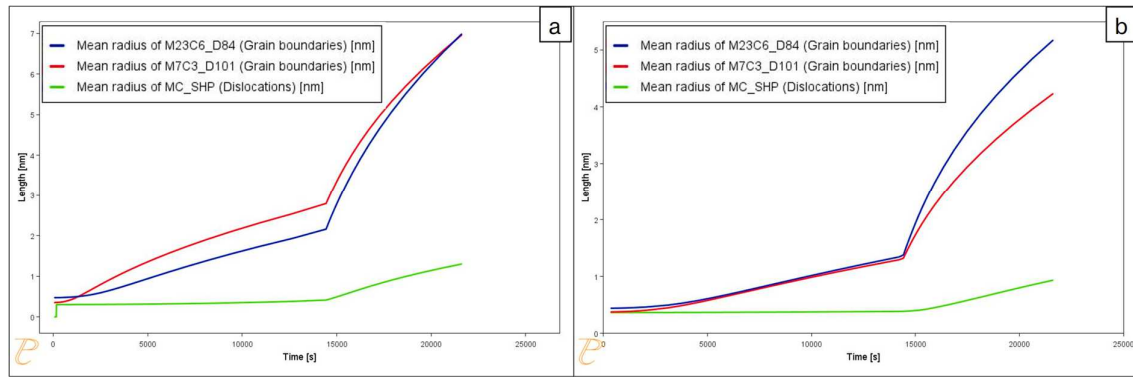


Figure 4.2-42 Mean radius and Time of DTT a) Dendritic (core) b) Interdendritic

The volume fraction results further confirm the stronger precipitation potential of the interdendritic regions (Figure 4.2-43). Among all precipitates, M_7C_3 carbides dominate the tempering response and exhibit a sharp increase in volume fraction after approximately 15,000 s, whereas $M_{23}C_6$ remains limited and MC carbides show only minor precipitation. The final carbide volume fraction in the interdendritic regions is significantly higher than in the dendritic core due to the local enrichment of alloying elements inherited from solidification segregation. This high density of nano-scale secondary carbides is considered the primary mechanism responsible for secondary hardening during tempering. Furthermore, the differences in precipitate size distribution and volume fraction between the dendritic and interdendritic regions provide a fundamental explanation for the microscopic hardness heterogeneity observed in the as-built DED microstructure.

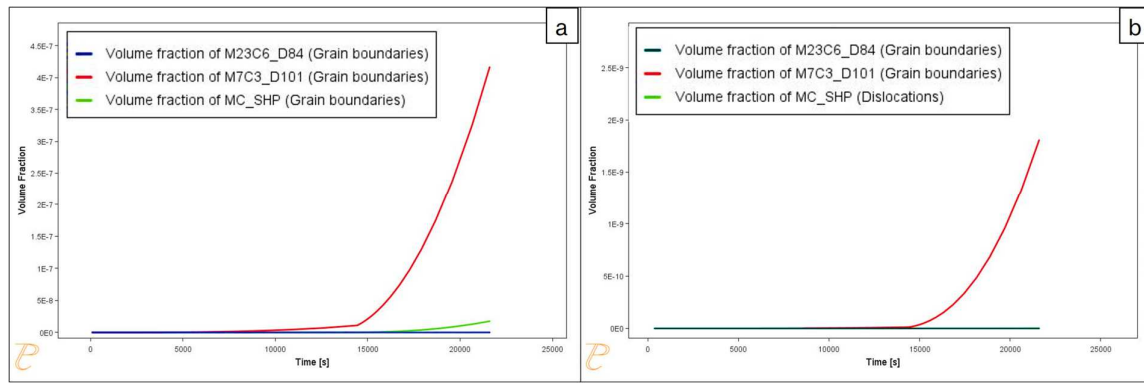


Figure 4.2-43 Volume Fraction and Time of DTT a) Dendritic (core) b) Interdendritic

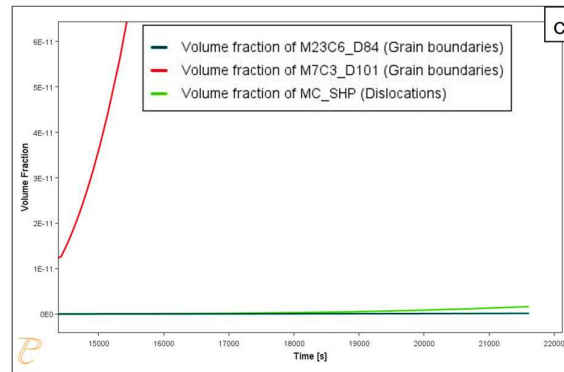


Figure 4.2-44 Volume Fraction and Time of Interdendritic in the last roots of DTT

4.2.5.3 Simulation of Precipitation Kinetics during Quenching Triple Tempering (QTT)

To evaluate the precipitation behavior following standard heat treatment, an austenitizing temperature of 1070°C was selected. According to equilibrium calculations, the steel is fully austenitic at this temperature, with alloying elements dissolved into the matrix to their maximum thermodynamic potential. Extracting the composition of the stable FCC phase at this temperature for tempering simulations is justified because it represents the actual chemical state of the matrix immediately prior to quenching. Any carbide remaining stable at 1070°C is not dissolved into the matrix and therefore do not contribute to the solute available for subsequent secondary precipitation.

The fundamental reason for utilizing the high-temperature austenite composition for the martensitic matrix lies in the nature of the martensitic transformation. The transition from austenite (FCC) to martensite (BCC/BCT) during quenching is a Displacive or Diffusionless transformation[101], [102], [103]. This process occurs without long-range atomic diffusion, meaning alloying atoms does not have sufficient time to be rejected from the matrix. Consequently, Solute Trapping occurs, and the martensitic lattice

perfectly inherits the chemical composition of the parent austenite. The mass fractions extracted for the FCC phase in both dendritic and interdendritic regions are detailed in Table 4.2-5. serving as the chemical foundation for the subsequent tempering simulations [101], [102], [104].

Table 4.2-5 Stable Phase Austenite of K340 after Quench

Stable Phase Austenite									
FCC		Moles					0.84215		
		Mass					46.39006		
Component		Fe	Cr	Mo	V	Mn	Si	Al	Nb
Mass Fraction	Dendritic	0.901	0.061	0.013	0.007	0.007	0.006	0.001	0.00005
	Interdendritic	0.829	0.120	0.027	0.009	0.007	0.005	0.0006	0.00014

The TC-PRISMA simulations after the Quench Triple Tempered (QTT) treatment show that the chemical segregation inherited from solidification strongly affects the precipitation behavior in both the dendritic core and the interdendritic regions. The Mean Radius curves (Figure 4.2-45) indicate that precipitates formed in the dendritic core grow to larger sizes, while the precipitates in the interdendritic regions remain finer throughout the tempering process. This behavior is related to the high concentration of alloying elements trapped in the segregated interdendritic zones after quenching. The high supersaturation increases the nucleation driving force and leads to the formation of a very large number of carbide nuclei. As a result, the available solute atoms are shared between many particles, limiting the growth of individual precipitates and maintaining a smaller average radius in the interdendritic regions [100].

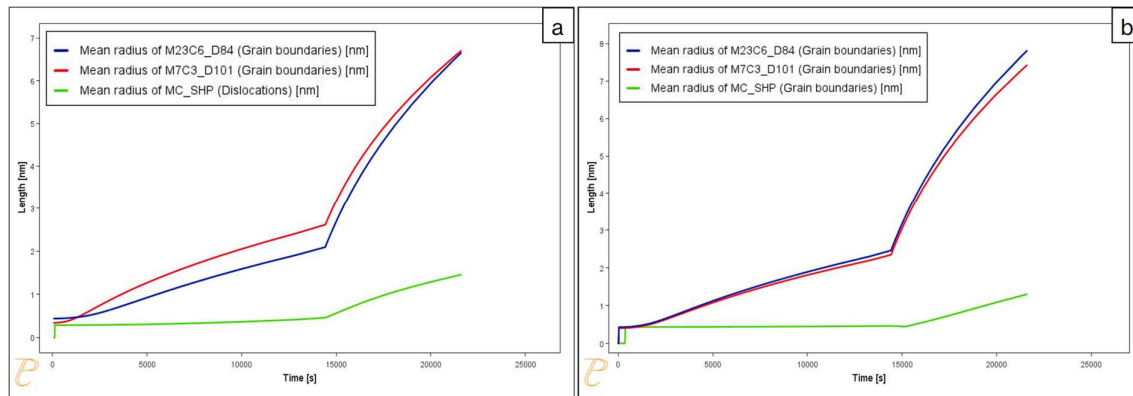


Figure 4.2-45 Mean radius and Time of QTT a) Dendritic (core) b) Interdendritic

The Volume Fraction curves show the opposite trend (Figure 4.2-46). The interdendritic regions exhibit a significantly higher precipitate volume fraction compared to the dendritic core. This is mainly caused by solute trapping during the diffusionless martensitic transformation after quenching from 1070 °C, where carbide-forming

elements remain trapped inside the matrix. Among all precipitates, M7C3 carbides show the highest volume fraction and dominate the precipitation response during tempering, indicating their important role in secondary hardening of the K340 steel. In contrast, MC and M23C6 carbides remain at much lower volume fractions. The combination of high precipitate density and fine particle size in the interdendritic regions explains the local hardness heterogeneity and the stronger strengthening effect observed in these segregated zones.

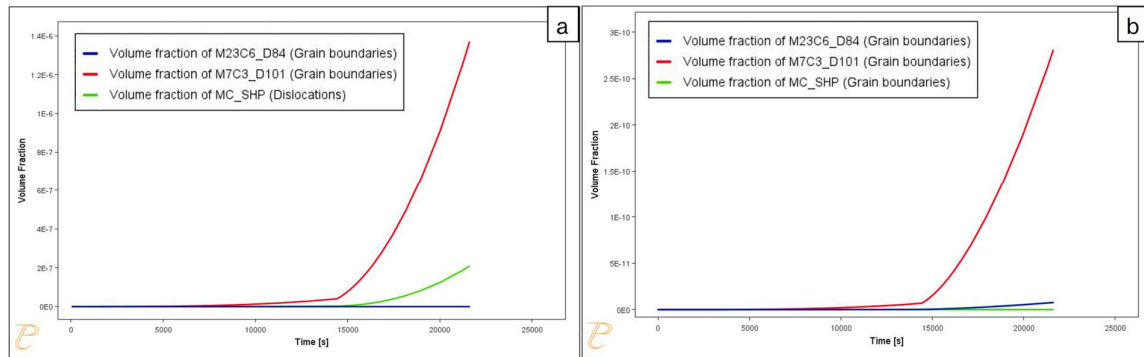


Figure 4.2-46 Volume Fraction and Time of QTT a) Dendritic (core) b) Interdendritic

A comparison between the DTT and QTT simulations reveals that the quenching step significantly modifies the precipitation behavior during tempering. In the DTT condition, carbide precipitation mainly originates from the segregation inherited directly from the as-built microstructure, leading to relatively lower precipitation kinetics and smaller differences between dendritic and interdendritic regions. In contrast, the QTT condition shows a much stronger precipitation response, particularly in the interdendritic regions.

This behavior is related to the austenitization treatment at 1070 °C followed by rapid quenching. During this process, alloying elements remain trapped inside the supersaturated martensitic matrix due to the diffusionless martensitic transformation. As a result, the matrix after quenching retains a higher amount of dissolved Cr, Mo, and V, which greatly increases the thermodynamic driving force for secondary carbide precipitation during tempering.

The Mean Radius plots indicate that carbides formed in QTT conditions generally exhibit faster growth kinetics and larger particle sizes compared to the DTT condition. However, the interdendritic regions still maintain relatively finer precipitates due to the extremely high nucleation density and competitive growth effect. According to the high-temperature in-situ synchrotron X-ray diffraction study by Fonseca et al. (2022)[100], the as-built microstructure inherits an ultra-high dislocation density and dense film-like retained austenite networks at the cellular boundaries due to the rapid cooling rates of the additive process. This intense solute trapping creates a massive thermodynamic driving force for explosive nucleation along the segregated boundaries. Consequently, the precipitate number density escalates drastically, triggering a competitive growth

mechanism where alloying elements are rapidly partitioned among countless nuclei; this stunting effect restricts individual particle growth, maintaining a finer mean radius in DTT (~4–5 nm) compared to QT (~6–8 nm).

The Volume Fraction vs. Time results reveal an opposite kinetic trend between the two heat-treatment routes. In the DTT condition, the severe segregation of carbide-forming alloying elements within the last solidified regions, as predicted by the Scheil model, causes the volume fraction of the dominant M_7C_3 carbide to increase more rapidly and at shorter times compared to the QT condition. This behavior is consistent with the mechanism proposed by Deirmina [100], where direct tempering promotes the decomposition of retained austenite together with the precipitation of secondary nano-carbides, leading to a strong secondary hardening response. On the other hand, in the QT route, although the interdendritic regions still contain a higher precipitate fraction than the dendritic core, the evolution of the curves becomes smoother due to partial redistribution of alloying elements during the austenitization stage. Both the kinetic simulations and previous metallurgical studies indicate that, although QT treatment improves matrix homogenization and reduces the cellular features inherited from the as-built structure, the chemical effects of the original solidification process are not eliminated. The remaining difference between dendritic and interdendritic regions in both heat-treatment conditions explains the persistence of local hardness heterogeneity within the material.

4.3 Hardness Distribution and Structure–Property Relationship

Micro-Vickers hardness measurements were carried out along the build direction, covering the deposited region, heat-affected zone, and base material. Three repaired sample conditions were examined: As-Built, DTT, and QTT. For each condition, 120 indentations were performed in 4 series of 30 measurements, equally distributed between the DED region and the base material for direct comparison. The distance between adjacent indentations was kept above three times the indentation diagonal length (about 100 μm) according to standard recommendations.

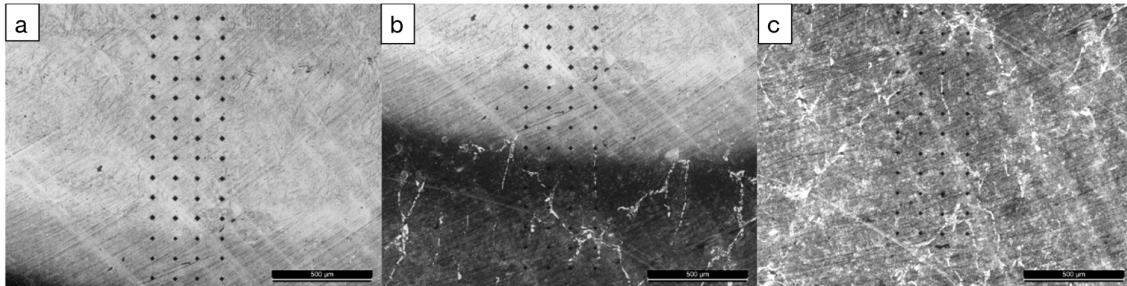


Figure 4.3-1 Micro-Vickers Hardness of As-Built Sample a) Deposited region, b) HAZ region, c) Substrate region

In the As-Built condition (Figure 4.3-2), the deposited region (0–1000 μm) exhibits the lowest hardness (450–550HV), while the substrate (SU) (starting from 1800 μm) shows the highest values (800HV). This significant difference indicates that the deposited material is still dominated by rapid solidification effects and alloy microsegregation. SEM observations and EDS analysis revealed that carbide-forming elements such as Cr and Mo are concentrated within interdendritic regions; however, these elements have not yet precipitated as fine and uniformly distributed strengthening carbides. Consequently, despite the presence of martensite, the lack of effective secondary carbide precipitation results in the relatively low hardness of the DED zone [19], [56], [69], [97].

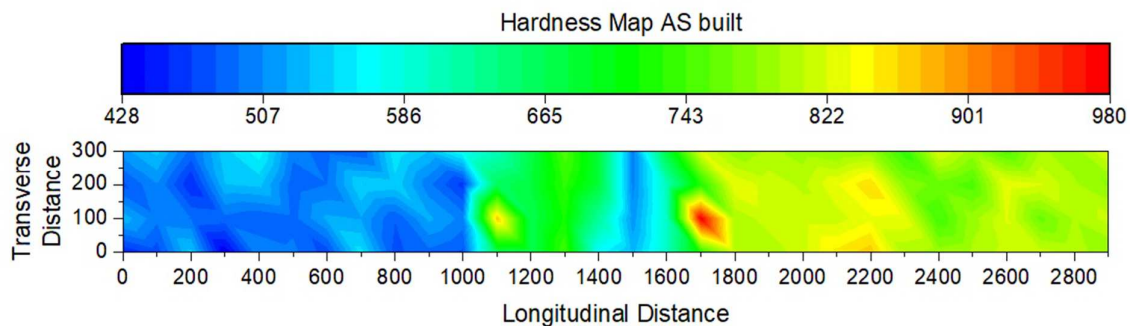


Figure 4.3-2 MAP of distribution of Hardness test of As-Built sample from Deposited to Substrate region

Significant hardness fluctuations are observed across the PMZ and HAZ regions. In the PMZ, localized melting and rapid resolidification partially reduce the segregation level

and promote local redistribution of alloying elements. As a result, the hardness increases compared to the DED region. Nevertheless, the microstructure in this area remains highly heterogeneous, leading to local scatter in hardness values.

In the HAZ-F region (around 1100–1700 μm), where the temperature exceeded A_{c3} , complete austenitization occurred followed by rapid self-quenching, resulting in the formation of fine and relatively homogeneous martensite. This explains the localized hardness peak observed in this region. The hardness increase is mainly associated with martensitic refinement and partial dissolution of pre-existing carbides [56].

In the DTT condition, the hardness of the DED region (0–1600 μm) increases significantly compared to the As-Built state, reaching values of 600–700 HV. This behavior is characteristic of secondary hardening, where tempering promotes redistribution of alloying elements from segregation zones and precipitation of fine Mo- and V-rich secondary carbides, thereby strengthening the matrix (Figure 4.3-3).

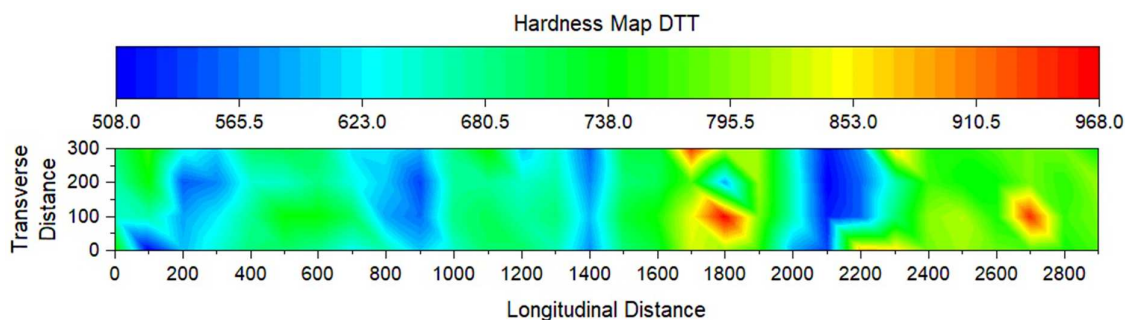


Figure 4.3-3 MAP of distribution of Hardness test of DTT sample from Deposited to Substrate region

However, the HAZ-P region, points 21–23, around 2100–2300 μm , still exhibits a pronounced hardness reduction (524 HV). This region experienced only partial austenitization, leaving some primary carbides undissolved. Additionally, the tempering cycle promoted carbide coarsening and softening of the martensitic matrix. Consequently, the microstructure became more heterogeneous, leading to the formation of a local softening zone. Therefore, the hardness reduction in HAZ-P is primarily associated with over-tempering and carbide coarsening rather than a decrease in carbide content [56], [97].

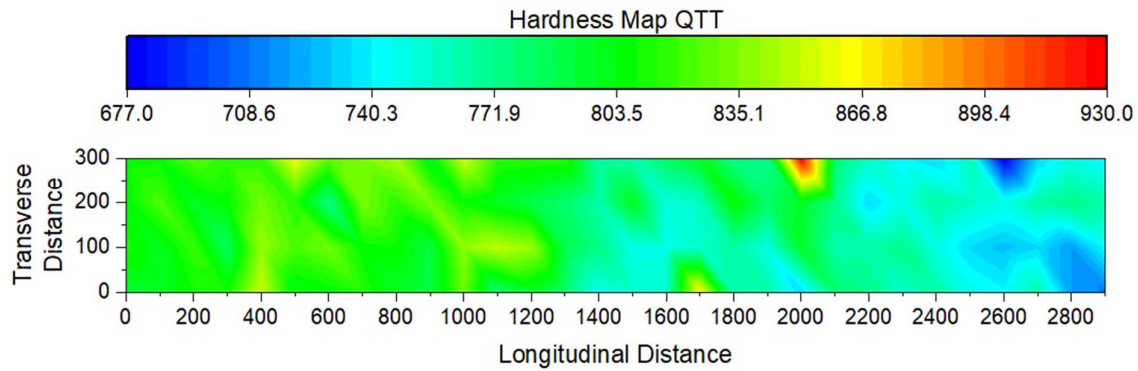


Figure 4.3-4 MAP of distribution of Hardness test of QTT sample from Deposited to Substrate region

The QTT sample exhibits the most homogeneous hardness distribution across the entire cross-section. Re-austenitization and quenching effectively reduce most of the solidification-induced segregation, while the final tempering treatment promotes uniform precipitation of fine carbides within the tempered martensitic matrix. As a result, hardness fluctuations and local softening regions are significantly minimized, leading to a relatively uniform hardness level in the range of approximately 750–830 HV throughout the sample. This behavior demonstrates that QTT treatment is the most effective approach for microstructural homogenization and mitigation of the mechanical gradients induced by the DED process [56], [97].

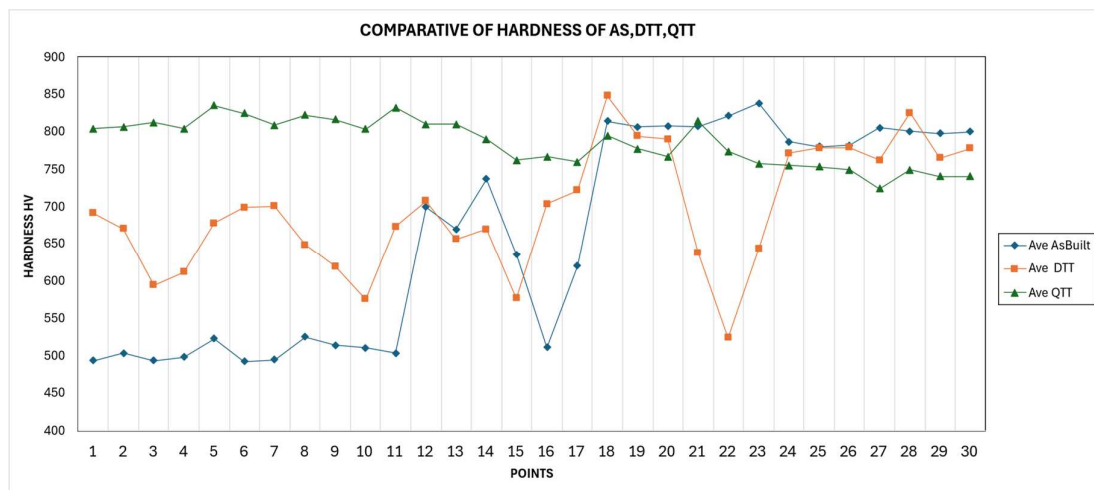


Figure 4.3-5 Comparative of Hardness of AB, DTT, QTT samples

The comparative line graph (Figure 4.3-5) highlights the superior stability of the QTT process. While the AB condition suffers from a soft deposit zone and the DTT condition exhibits a dangerous softening dip in the HAZ-P (point 22), the QTT condition provides a uniform response. The transition from solidification-controlled hardness (AB) to precipitation-controlled hardness (DTT and QTT) is evident. Specifically, the QTT treatment successfully bridges the ~ 300 HV gap between the deposit and the substrate that existed in the as-built state.

5 Conclusion

Based on the systematic experimental investigation, advanced characterization, and thermodynamic simulations of DED-processed and heat-treated Böhler K340 cold-work tool steel, the following main conclusions are drawn:

- **Bulk Deposition Zone Characteristics:** Microstructural evaluation of the bulk specimens established that the feedstock atomization gas deeply dictates the initial structural integrity of the deposition zone. Argon-atomized powder yielded fully dense bulk tracks, whereas Nitrogen-atomized feedstock introduced process-induced microporosity. In the as-built state, the deposition zone is dominated by a highly non-equilibrium, anisotropic cellular-dendritic matrix and epitaxial columnar grain growth along the build direction, where an intrinsic inter-layer heat treatment (IHT) effect causes localized tempering and partial recovery strictly in the upper regions of the underlying layers.
- **Comprehensive Repair Interface Macrostructure:** Multi-zone macroscopic analysis of the repaired specimens demonstrated that defect susceptibility is critically governed by the groove aspect ratio. Successful, defect-free filling and strong metallurgical wall-bonding were achieved in wider grooves (8 mm) and shallow depths (2 mm). Conversely, deep, narrow grooves (4 mm width, 4–5 mm depth) formed side-wall "dead zones" parallel to the laser beam, leading to lack-of-fusion defects and trapped porosity. Severe thermal gradients during the first pass induced localized cracking at the substrate interface, which acted as stress concentrators and propagated during post-process quenching due to transformation-induced volumetric expansion.
- **Interfacial Microstructural Evolution (SU/HAZ/DED):** The repair configuration introduced distinct microstructural zones across the fusion boundary. Directly below the first layer, a complex Partially Melted Zone (PMZ) formed, characterized by a non-continuous, irregular re-solidified network rather than a classical white band. Below the fusion line, the heat-affected zone resolved into two distinct sub-regions: a fully austenitized HAZ-F (exhibiting refined, hard martensite due to peak temperatures exceeding A_{c3}) and a partially austenitized HAZ-P (exhibiting a heterogeneous, over-tempered matrix with incomplete carbide dissolution), which subsequently transitions into the unaffected pre-hardened substrate.
- **EBSD Texture and Retained Austenite Quantification:** Advanced EBSD analysis conducted across the repair interfaces revealed that the as-built and DTT DED zones are highly anisotropic, possessing a strong $\langle 001 \rangle$ preferred orientation and an exceptionally high fraction of retained austenite (55.3% and 53.4%, respectively). This stable FCC phase was mechanically and chemically locked along cellular boundaries due to solidification microsegregation. The QTT

treatment successfully achieved a complete microstructural reset, breaking down the directional solidification texture, completely relaxing processing residual stresses (as verified by GOS, GROD, and KAM parameters), and reducing the retained austenite to a fully isotropic 0.77%.

- **Thermo-Calc Solidification and Kinetics Modeling:** Exclusively applied to the repair configuration, Thermo-Calc Scheil-Gulliver simulations verified that severe microsegregation rejects Cr and Mo into the final liquid during rapid solidification, forming the interdendritic primary M_7C_3 network. TC-PRISMA kinetic modeling demonstrated that the diffusionless martensitic transformation during QTT quenching triggers a massive thermodynamic driving force (ΔG) for nucleation. This promotes an ultra-dense population of fine, nano-sized secondary carbides (M_7C_3 and $M_{23}C_6$) in the interdendritic zones during tempering, whereas DTT kinetics remain limited by the un-resolved as-built segregation state.
- **Hardness Distribution and Structure-Property Optimization:** Micro-Vickers hardness profiling across the SU/HAZ/DED cross-sections confirmed severe mechanical mismatch in the untreated and DTT repair states. The as-built zone suffered from a soft matrix (450–550 HV) due to unprecipitated solutes, while the DTT route introduced a dangerous softening dip (524 HV) directly inside the HAZ-P due to over-tempering. The QTT process completely eliminated these mechanical gradients, bridging the ~300 HV gap across the fusion boundary to deliver a highly uniform and continuous hardness profile (750–830 HV) across all repaired zones, ensuring microstructural continuity and optimal structural durability for industrial tooling restoration.

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