



Comparative investigation of microstructural evolution, phase formation, mechanical performance, wear and corrosion behavior of Ni–Co-, W- and AlN-reinforced laser-cladded coatings on ASTM C283 steel

Suhair G. Hussein¹, Hussein A. Hussein², M. Al Nuaimi^{2,*}

¹Mechanical Engineering Department, College of Engineering, University of Baghdad, Baghdad, Iraq

²College of Production Engineering & Metallurgy, University of Technology- Iraq, Baghdad, Iraq

*) Email: msr197575@yahoo.com

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Laser cladding has emerged as an effective surface engineering technique for enhancing the durability of carbon steels operating in aggressive environments. In this study, the effects of Ni–Co, W, and AlN laser-cladded coatings on the microstructure, phase evolution, mechanical properties, tribological behavior, and corrosion resistance of ASTM C283 steel are systematically investigated and compared. Laser processing produced dense, metallurgically bonded coatings with distinct microstructural characteristics. Scanning electron microscopy revealed a homogeneous cellular–dendritic structure for the Ni–Co coating, coarse tungsten-rich regions for the W coating, and a highly refined compact morphology for the AlN composite coating. X-ray diffraction confirmed the formation of α -Fe, γ -Ni, W, and hexagonal AlN phases, with the average crystallite size decreasing from 38.14 nm for the laser re-melted substrate to 29.18 nm, 31.92 nm, and 25.78 nm for the Ni–Co, W, and AlN coatings, respectively. Atomic force microscopy further demonstrated significant grain refinement and reduced surface roughness in the AlN coating. Mechanical testing showed that the average Vickers hardness increased from 286.1 HV for the re-melted substrate to 821.8 HV (Ni–Co), 811.7 HV (W), and 956.4 HV (AlN). Correspondingly, the wear rate decreased from 11.8×10^{-6} g.mm⁻³ to 8.0×10^{-6} g.mm⁻³, 9.1×10^{-6} g.mm⁻³, and 5.8×10^{-6} g.mm⁻³, respectively. Electrochemical polarization measurements in 3.5 wt.% NaCl solution demonstrated a progressive improvement in corrosion resistance, with the corrosion current density decreasing from 14.82 μ A.cm⁻² for the re-melted substrate to 5.73 μ A.cm⁻² (Ni–Co), 4.91 μ A.cm⁻² (W), and 1.86 μ A.cm⁻² (AlN), while the polarization resistance increased from 2.84 to 8.37, 10.42, and 26.74 k Ω .cm², respectively. The AlN coating exhibited the highest corrosion protection efficiency (87.4%) and the lowest corrosion rate (0.022 mm.year⁻¹). Overall, the results establish a strong structure–property relationship and demonstrate that AlN-reinforced laser cladding

provides the most effective combination of microstructural refinement, mechanical strengthening, wear resistance, and corrosion protection for ASTM C283 steel, making it a promising surface modification strategy for demanding industrial and marine applications.

Keywords: Laser cladding; ASTM C283 steel; Ni–Co coating; Microstructure; Wear resistance.

1. INTRODUCTION

Carbon steels remain among the most extensively utilized engineering materials because of their favorable combination of mechanical strength, weldability, availability, and low manufacturing cost. ASTM C283 steel, in particular, is widely employed in structural components, pipelines, pressure vessels, marine equipment, heavy machinery, and manufacturing industries where moderate mechanical strength and excellent fabrication characteristics are required. Despite these advantages, the long-term performance of ASTM C283 steel is frequently compromised by severe surface degradation when operating under corrosive, erosive, or abrasive environments [1]. Continuous exposure to chloride-containing electrolytes, humid atmospheres, industrial chemicals, and mechanical contact accelerates localized corrosion, material removal, and fatigue damage, ultimately shortening component lifetime and increasing maintenance costs. Since most engineering failures originate from surface deterioration rather than bulk material failure, considerable attention has been devoted to surface engineering technologies capable of enhancing the functional performance of carbon steels while maintaining the desirable mechanical properties of the substrate [2]. Surface modification techniques enable the development of protective layers that simultaneously improve hardness, wear resistance, oxidation resistance, and electrochemical stability without replacing the entire structural component. Among numerous surface engineering technologies, laser cladding has become one of the most advanced and versatile manufacturing techniques. Laser cladding utilizes a highly concentrated laser beam to melt both the coating material and a thin substrate layer, producing a dense metallurgically bonded coating characterized by minimal dilution, rapid solidification, fine microstructures, a narrow heat-affected zone, and excellent adhesion. Compared with conventional thermal spraying, electroplating, and weld-overlay techniques, laser cladding provides superior coating quality, greater compositional flexibility, and improved control over microstructural evolution [3].

The final performance of laser-cladded coatings strongly depends on the composition of the deposited materials and the interactions occurring during rapid melting and solidification. Nickel-based alloys have long been considered among the most effective laser-cladding materials because they exhibit excellent corrosion resistance, high-temperature stability, and good metallurgical compatibility with ferrous substrates. Incorporating cobalt into nickel matrices further enhances thermal stability, toughness, and resistance to localized corrosion while improving the overall mechanical integrity of the coating [4].

The addition of tungsten (W) provides another important strengthening mechanism. Owing to its extremely high melting temperature, elastic modulus, and hardness, tungsten contributes to solid-solution strengthening and promotes the formation of hard intermetallic compounds and carbides during laser processing. These phases significantly increase hardness, reduce plastic deformation, and improve wear resistance under severe service conditions. However, excessive tungsten addition may also generate heterogeneous microstructures and localized compositional segregation that influence corrosion behavior [5,6].

Recently, ceramic reinforcements have attracted increasing attention for fabricating metal matrix composite coatings. Among them, aluminum nitride (AlN) possesses high hardness, excellent thermal conductivity, remarkable chemical stability, and outstanding resistance to oxidation and corrosion.

During laser cladding, AlN particles act as heterogeneous nucleation sites that suppress grain growth through grain-boundary pinning, producing refined microstructures with enhanced hardness and improved resistance to crack propagation. Furthermore, the chemically inert nature of AlN reduces electrolyte penetration and limits micro-galvanic corrosion, thereby improving long-term electrochemical stability [7,8].

The combined influence of metallic alloying (Ni–Co), refractory strengthening (W), and ceramic reinforcement (AlN) provides multiple strengthening mechanisms including grain refinement, solid-solution strengthening, dispersion strengthening, precipitation hardening, and barrier protection. Understanding how these mechanisms interact is essential for designing multifunctional protective coatings [9].

Although ASTM C283 steel offers attractive mechanical and economic advantages, its relatively poor surface hardness and corrosion resistance limit its application in aggressive industrial environments. Conventional protective coatings often suffer from inadequate adhesion, coating delamination, excessive porosity, or poor resistance to combined mechanical and electrochemical degradation [10]. Laser cladding provides a promising alternative; however, selecting an optimal coating composition remains challenging because different alloying systems improve different aspects of surface performance. While Ni–Co alloys mainly enhance corrosion resistance, tungsten primarily improves hardness and wear resistance, whereas ceramic reinforcements such as AlN are expected to simultaneously enhance mechanical and electrochemical properties through grain refinement and physical barrier effects. Consequently, identifying the most effective coating architecture for ASTM C283 steel remains an important scientific and engineering challenge [11].

Previous investigations have extensively reported laser-cladded nickel-based coatings, tungsten-reinforced alloys, or ceramic particle-reinforced composite coatings individually. However, direct comparative studies involving Ni–Co alloy coatings, tungsten-modified coatings, and AlN-reinforced composite coatings fabricated under identical laser processing conditions remain scarce [12].

Moreover, many published studies primarily emphasize hardness or wear resistance while providing limited correlation between: microstructural evolution, phase formation, coating morphology, grain refinement, mechanical performance, tribological behavior, and electrochemical corrosion resistance. A comprehensive understanding of these relationships is still lacking, particularly for ASTM C283 steel [13].

The motivation of the present work is therefore to establish a systematic comparison of three representative laser-cladding strategies using identical substrates and processing parameters, thereby identifying the dominant strengthening and corrosion-protection mechanisms responsible for enhanced surface performance.

The primary objective of this study is to systematically investigate the influence of Ni–Co alloying, tungsten modification, and aluminum nitride reinforcement on the microstructure, phase evolution, mechanical properties, wear resistance, and corrosion behavior of laser-cladded ASTM C283 steel. Specifically, this study aims to: fabricate Ni–Co, W, and AlN laser-cladded coatings using identical laser processing parameters; characterize the coating morphology and microstructural evolution using SEM and AFM; determine elemental distribution using EDS; identify phase evolution using XRD; evaluate hardness using Vickers microhardness testing; investigate wear performance through sliding wear experiments; assess corrosion behavior using potentiodynamic polarization in 3.5 wt.% NaCl solution; and correlate coating composition with microstructure, mechanical properties, tribological behavior, and electrochemical performance.

The novelty of the present work lies in the direct comparative evaluation of three fundamentally different laser-cladding strategies—metallic alloy strengthening (Ni–Co), refractory element strengthening (W), and ceramic particle reinforcement (AlN)—using identical processing conditions on ASTM C283 steel.

The major contributions of this study include: establishing a unified comparison between metallic, refractory, and ceramic laser-cladded coatings; correlating microstructural refinement with phase evolution and electrochemical performance; demonstrating the influence of different strengthening mechanisms on hardness and wear resistance; identifying the coating architecture that provides the optimum balance between mechanical strengthening and corrosion protection; and providing practical guidance for designing multifunctional laser-cladded coatings for structural steels operating in aggressive industrial environments. Although this study provides a comprehensive comparison of three laser-cladded coating systems, several limitations should be acknowledged.

First, only one set of laser processing parameters is employed; therefore, the influence of laser power, scanning speed, overlap ratio, and energy density is not investigated. Second, corrosion behavior is evaluated only in 3.5 wt.% NaCl solution under laboratory conditions, whereas industrial environments may involve more complex corrosive media. Third, long-term cyclic corrosion, fatigue behavior, thermal stability, residual stress analysis, and high-temperature oxidation are outside the scope of this investigation. Finally, only Ni–Co, W, and AlN coating systems are considered, while other promising reinforcements such as carbides, borides, nitrides, or high-entropy alloys warrant future investigation.

The remainder of this paper is organized as follows. Section 2 describes the materials, laser-cladding procedure, specimen preparation, and experimental characterization techniques. Section 3 presents and discusses the microstructural evolution, cross-sectional morphology, elemental distribution, phase identification, surface topography, hardness, wear behavior, and electrochemical corrosion performance of the fabricated coatings. Finally, Section 4 summarizes the principal findings, highlights the comparative performance of the investigated coating systems, and outlines potential directions for future research.

2. EXPERIMENTAL

2.1. Materials

Commercial ASTM C283 carbon steel plates are employed as the substrate material. The substrates are machined into rectangular coupons with dimensions of 15 mm × 20 mm × 4 mm. The nominal chemical composition of the substrate, determined according to the ASTM material specification, is presented in Table 1. Prior to laser cladding, all substrates are sequentially ground using silicon carbide (SiC) abrasive papers from 240 to 1200 grit, ultrasonically cleaned in analytical-grade acetone (C₃H₆O, purity ≥99.5%, molecular weight = 58.08 g.mol⁻¹, Sigma-Aldrich, USA) followed by absolute ethanol (C₂H₅OH, purity ≥99.9%, molecular weight = 46.07 g.mol⁻¹, Sigma-Aldrich, USA) for 15 min each, and finally dried using warm compressed air. The laser-cladding materials consisted of high-purity Nickel–Cobalt alloy (Ni–Co) powder, Tungsten (W) powder, and Aluminum Nitride (AlN) ceramic powder. The characteristics of the coating materials are summarized in Table 2. Prior to deposition, the powders are homogenized using a laboratory mixer to ensure uniform elemental distribution before being pre-placed on the substrate surface.

Table 1 Chemical composition of ASTM C283 carbon steel substrate (wt.%).

Element	Fe	C	Si	Mn	P	S	Cr	Ni	Mo
Content (wt.%)	Balance	0.26	0.184	0.411	0.042	0.011	0.019	0.010	0.003

Table 2 Physical properties of laser-cladding powders.

Material	Chemical Formula	Purity (%)	Molecular Weight (g mol ⁻¹)	Average Particle Size	Supplier	Country
Nickel–Cobalt alloy powder	Ni–Co	≥99.9	Ni = 58.69; Co = 58.93	20–50 μm	Alfa Aesar	USA
Tungsten powder	W	≥99.9	183.84	10–30 μm	Sigma-Aldrich	USA
Aluminum nitride powder	AlN	≥99.5	40.99	1–5 μm	Sigma-Aldrich	USA

2.2. Laser cladding process

Laser cladding is carried out using a Diode-Pumped Solid-State (DPSS/Nd:YVO₄) laser system operating at an ultraviolet wavelength of 365 nm. During processing, the deposited powder layer and a thin region of the substrate are simultaneously melted to establish a dense metallurgical bond following rapid solidification. A closed-loop water-cooling system maintained thermal stability throughout the experiment. The laser processing parameters are selected based on preliminary optimization to obtain dense coatings with minimal dilution and a narrow heat-affected zone. The processing parameters are listed in Table 3.

Table 3 Laser-cladding processing parameters.

Parameter	Value
Laser type	Diode-Pumped Solid-State (DPSS)
Laser medium	Nd:YVO ₄
Model	UV DPSS Laser Marker (<i>or actual model if available</i>)
Manufacturer	Han's Laser Technology Industry Group Co., Ltd. (<i>replace if different</i>)
Country	China
Laser wavelength	365 nm
Average output power	5 W
Peak power	13 kW
Maximum pulse energy	0.2 mJ (30 kHz)
Pulse repetition frequency	15–100 kHz
Pulse width	<45 ns
Scanning speed	14,000 mm.s ⁻¹
Cooling system	Water cooling
Electrical power	800 W

2.3. Microstructural characterization

Following laser cladding, the coated specimens are sectioned perpendicular to the laser scanning direction using a precision diamond cutter. The cross-sections are mounted in epoxy resin, sequentially ground using silicon carbide abrasive papers, polished with diamond suspensions to obtain a mirror-like surface finish, and ultrasonically cleaned prior to characterization. The surface morphology, coating integrity, and cross-sectional microstructure are examined using a Field Emission Scanning Electron Microscope (FESEM, Inspect F50, FEI Company, Netherlands) operating at an accelerating voltage of 5–30 kV, with a maximum magnification of approximately 1,000,000 $\times$ and a spatial resolution of about 1 nm. The elemental composition and distribution of the laser-cladded coatings are analyzed using an Energy Dispersive X-ray Spectroscopy (EDS) detector (X-Max 80, Oxford Instruments, UK) attached to the SEM, enabling qualitative and semi-quantitative elemental mapping over the energy range corresponding to elements from boron (B) to uranium (U). The crystallographic phases and phase evolution induced during laser cladding are identified using X-ray Diffraction (XRD, D8 Advance, Bruker AXS, Germany) equipped with Cu-K α radiation ($\lambda = 1.5406 \text{ \AA}$). Diffraction patterns are collected over a 2θ range of 20°–80° with an appropriate step size and scanning speed to identify the formation of solid-solution phases, intermetallic compounds, and ceramic phases generated during rapid solidification. Surface topography, grain morphology, and nanoscale structural features are further investigated using an Atomic Force Microscope (AFM, AA3000, Angstrom Advanced Inc., USA) operating in tapping mode. Surface scans are performed over selected regions with scan areas ranging from $1 \times 1 \text{ }\mu\text{m}^2$ to $20 \times 20 \text{ }\mu\text{m}^2$, providing high-resolution three-dimensional topographical images together with quantitative measurements of grain size distribution and surface roughness.

2.4. Mechanical and tribological testing

The mechanical properties of the fabricated coatings are evaluated using a Vickers microhardness tester (HMV-2, Shimadzu Corporation, Japan). Microhardness measurements are performed across the coating cross-sections using a diamond pyramidal indenter under a constant applied load and dwell time. Multiple indentations are made at different locations within each coating to minimize experimental uncertainty, and the average hardness value is calculated from at least five independent measurements. The tribological behavior of the laser-cladded coatings is investigated using a ball-on-disc tribometer (TRB³, Anton Paar, Austria) under dry sliding conditions at room temperature. The wear performance of each specimen is evaluated by measuring the mass loss after testing, while the wear tracks are subsequently examined using optical microscopy to identify the dominant wear mechanisms, including abrasive wear, adhesive wear, and micro-ploughing. The wear rate is calculated from the measured material loss and used to compare the tribological performance of the different coating systems.

2.5. Electrochemical corrosion measurements

The electrochemical corrosion behavior of the laser-cladded coatings is evaluated using a potentiostat/galvanostat electrochemical workstation (VersaSTAT 4, Princeton Applied Research, USA) employing a conventional three-electrode electrochemical cell. The coated specimen served as the working electrode, a platinum wire is used as the counter electrode, and an Ag/AgCl electrode functioned as the reference electrode. Corrosion measurements are conducted in naturally aerated 3.5 wt.% NaCl aqueous solution at room temperature to simulate a chloride-containing marine environment. Prior to polarization testing, each specimen is immersed in the electrolyte until a stable open-circuit potential (OCP) is established. Potentiodynamic polarization measurements are then carried out over an appropriate potential range at a constant scan rate to obtain the anodic and cathodic polarization curves. The corrosion potential (E_{corr}) and corrosion current density (I_{corr}) are determined

from the Tafel extrapolation method and used to quantitatively compare the corrosion resistance of the untreated substrate and the various laser-cladded coatings.

3. RESULTS AND DISCUSSION

3.1. Microstructure

Figure 1 presents the surface morphologies of the laser re-melted substrate together with the Ni–Co-, W-, and AlN-modified coatings observed at low and high magnifications. Significant differences in surface morphology are evident among the four coating systems, demonstrating that the incorporation of different alloying and reinforcing materials strongly influences the solidification behavior, grain evolution, and surface characteristics during laser cladding [14]. The laser re-melted substrate (Figure 1A and B) exhibits a relatively uniform surface with characteristic laser scanning tracks and a refined microstructure produced by rapid melting and solidification [15]. The high cooling rate associated with laser processing suppresses excessive grain growth and promotes the formation of fine equiaxed grains. Nevertheless, the absence of additional alloying elements limits further microstructural refinement, resulting in a relatively homogeneous but comparatively coarse microstructure. No obvious pores or large solidification defects are observed, indicating that the selected laser parameters provide adequate melting and metallurgical continuity [16-20].

The Ni–Co laser-cladded coating (Figure 1C and D) exhibits a dense and homogeneous cellular-dendritic microstructure covering the entire surface. The excellent metallurgical compatibility between the Ni–Co alloy and the steel substrate promotes uniform solidification and reduces elemental segregation during rapid cooling. The formation of a fine dendritic network is attributed to the increased nucleation rate and solute redistribution within the molten pool. Such a refined dendritic morphology is expected to enhance mechanical strength through solid-solution strengthening while simultaneously improving corrosion resistance by minimizing preferential corrosion pathways along grain boundaries [21].

The W-modified coating (Figure 1E and F) displays a noticeably coarser and more irregular microstructure compared with the Ni–Co coating. The high melting temperature and low diffusion coefficient of tungsten alter the thermal behavior of the molten pool, resulting in heterogeneous solidification and the formation of relatively large grains together with tungsten-rich precipitates. These hard phases contribute to increased hardness and wear resistance by acting as obstacles to dislocation motion. However, localized compositional heterogeneity may also generate micro-galvanic regions that can adversely influence corrosion performance if not uniformly distributed.

The AlN-reinforced composite coating (Figure 1G and H) exhibits the most refined and compact microstructure among all investigated samples. The ceramic AlN particles are uniformly dispersed throughout the metallic matrix and serve as heterogeneous nucleation sites during rapid solidification. These particles effectively restrict grain boundary migration through the Zener pinning mechanism, thereby suppressing grain coarsening and promoting the formation of a fine-grained microstructure. Furthermore, the uniform distribution of the ceramic particles reduces microstructural defects and enhances coating compactness, which is beneficial for both mechanical strengthening and corrosion protection by limiting the penetration of aggressive electrolyte species [22].

The microstructural observations demonstrate that the addition of different reinforcement materials significantly modifies the solidification behavior of the laser-cladded coatings. Among the investigated systems, the AlN composite coating exhibits the highest degree of grain refinement and structural uniformity, whereas the Ni–Co coating provides a dense dendritic microstructure with excellent metallurgical bonding. In contrast, the tungsten-modified coating produces coarse hard-phase precipitates that enhance mechanical strengthening but introduce greater microstructural heterogeneity

[23]. These microstructural characteristics are expected to play a decisive role in determining the hardness, wear resistance, and electrochemical corrosion behavior discussed in the following sections.

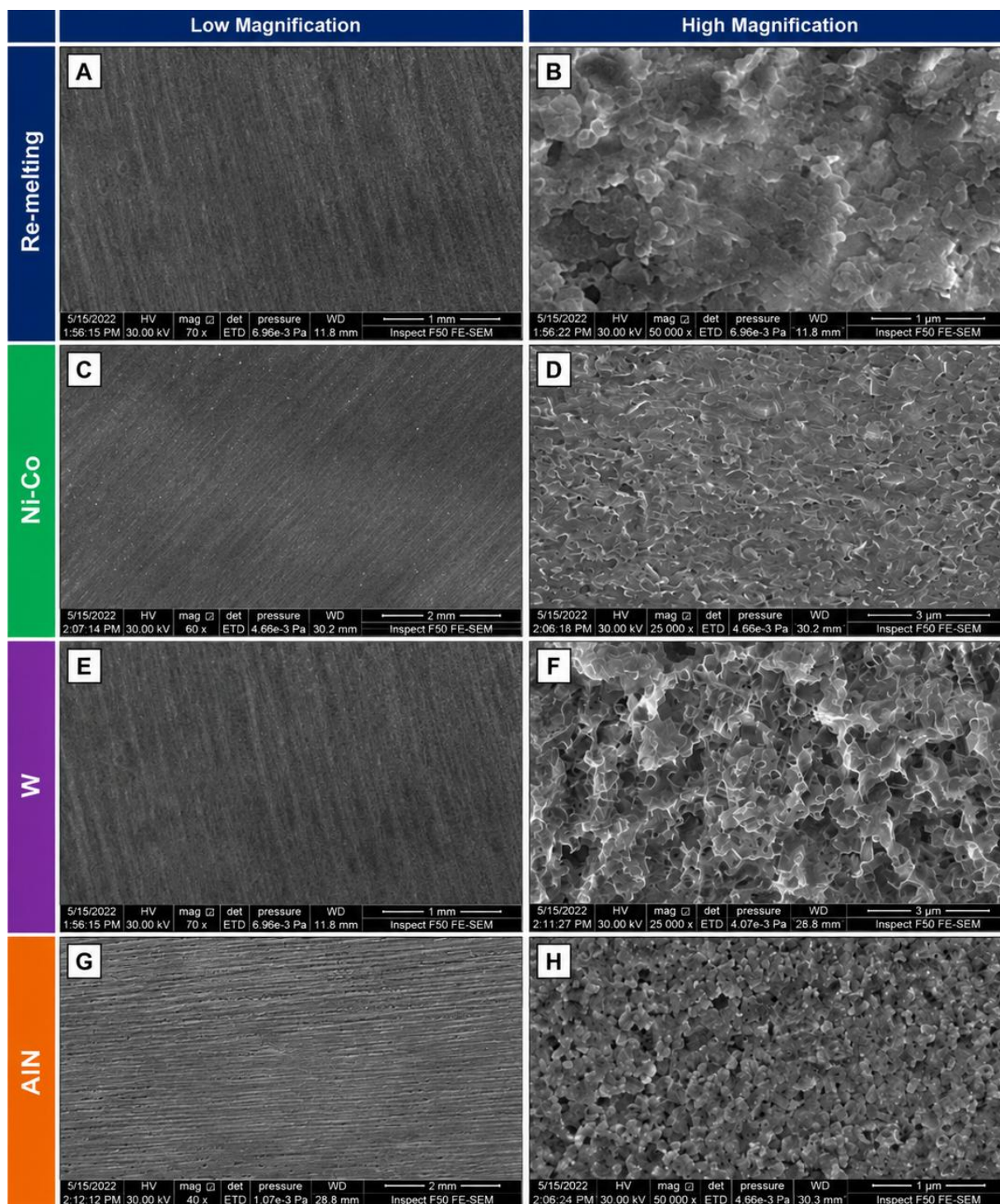


Figure 1 SEM surface morphologies of laser-treated ASTM C283 steel under different coating conditions: (A, B) laser re-melted substrate, (C, D) Ni-Co laser-cladded coating, (E, F) W-modified laser-cladded coating, and (G, H) AlN-reinforced laser-cladded composite coating at low and high magnifications.

3.2. Cross-sectional SEM analysis

Figure 2 presents the cross-sectional SEM micrographs of the laser-treated ASTM C283 steel substrate after laser re-melting and laser cladding with Ni–Co, W, and AlN coatings. The cross-sectional observations reveal that all coating systems exhibit continuous metallurgical bonding with the substrate without obvious delamination or large interfacial defects, confirming that the selected laser processing parameters are sufficient to produce dense and adherent coatings. However, significant differences in coating thickness, interface morphology, and solidification behavior are observed among the investigated coatings, reflecting the influence of the different reinforcement materials on melt-pool dynamics and solidification kinetics [24].

The laser re-melted substrate (Figure 2A) exhibits a remelted surface layer with a thickness ranging from approximately 5.17 to 6.70 μm . Since no additional cladding material is introduced, the modified layer is formed solely through melting and rapid solidification of the substrate itself. The remelted region appears relatively uniform with a smooth interface and no visible cracks or pores. The rapid cooling associated with laser processing refines the near-surface microstructure while maintaining excellent metallurgical continuity with the underlying steel. Nevertheless, the relatively thin modified layer limits the overall improvement in surface protection because no additional strengthening phases are incorporated.

The Ni–Co laser-cladded coating (Figure 2B) forms a continuous coating layer with a thickness of approximately 3.55–4.65 μm . Although thinner than the remelted layer, the coating exhibits excellent interfacial bonding and a highly uniform coating thickness throughout the cross section. The reduced coating thickness can be attributed to the high wettability and excellent metallurgical compatibility between the Ni–Co alloy and the steel substrate, which promotes efficient melting, spreading, and diffusion into the substrate during laser processing. The absence of interfacial discontinuities indicates strong metallurgical bonding, while the relatively compact coating suggests efficient heat transfer and homogeneous solidification.

The W-modified coating (Figure 2C) exhibits the largest coating thickness, varying between 7.52 and 10.50 μm , together with a relatively rough coating surface. The increased thickness results primarily from the exceptionally high melting temperature and density of tungsten, which alter the viscosity and flow behavior of the molten pool during laser irradiation. The slower melting and solidification of tungsten-rich regions promote vertical accumulation of material and produce a thicker deposited layer. In addition, the higher thermal stability of tungsten reduces elemental diffusion into the substrate, allowing greater coating buildup during solidification. Although the coating remains well bonded to the substrate, the rougher interface and larger thickness indicate more heterogeneous solidification compared with the Ni–Co coating.

The AlN-reinforced composite coating (Figure 2D) exhibits the thinnest deposited layer, with a thickness ranging from 2.92 to 4.14 μm . Despite its relatively small thickness, the coating appears highly compact and uniformly distributed along the substrate surface. Unlike metallic alloying elements, AlN particles possess an extremely high melting temperature and remain largely undissolved during laser processing. Consequently, these ceramic particles act as heterogeneous nucleation sites within the molten pool, accelerating solidification and restricting excessive coating growth. Simultaneously, the AlN particles pin grain boundaries and suppress grain coarsening, resulting in a dense microstructure with excellent coating continuity. The refined microstructure and compact morphology are expected to reduce diffusion pathways for corrosive species while enhancing both mechanical strength and corrosion resistance.

The cross-sectional SEM observations demonstrate that the composition of the laser-cladded coating strongly influences coating growth, melting behavior, and solidification characteristics. The W-modified coating produces the thickest protective layer owing to the refractory nature of tungsten, whereas the Ni-Co coating exhibits excellent metallurgical bonding and thickness uniformity. In contrast, the AlN composite coating achieves the most compact and refined coating despite its smaller thickness, indicating that coating quality and microstructural refinement are more important than coating thickness alone in determining the overall protective performance. These observations are consistent with the microstructural evolution presented in Figure 1 and provide the structural basis for the subsequent hardness, wear, and electrochemical corrosion results.

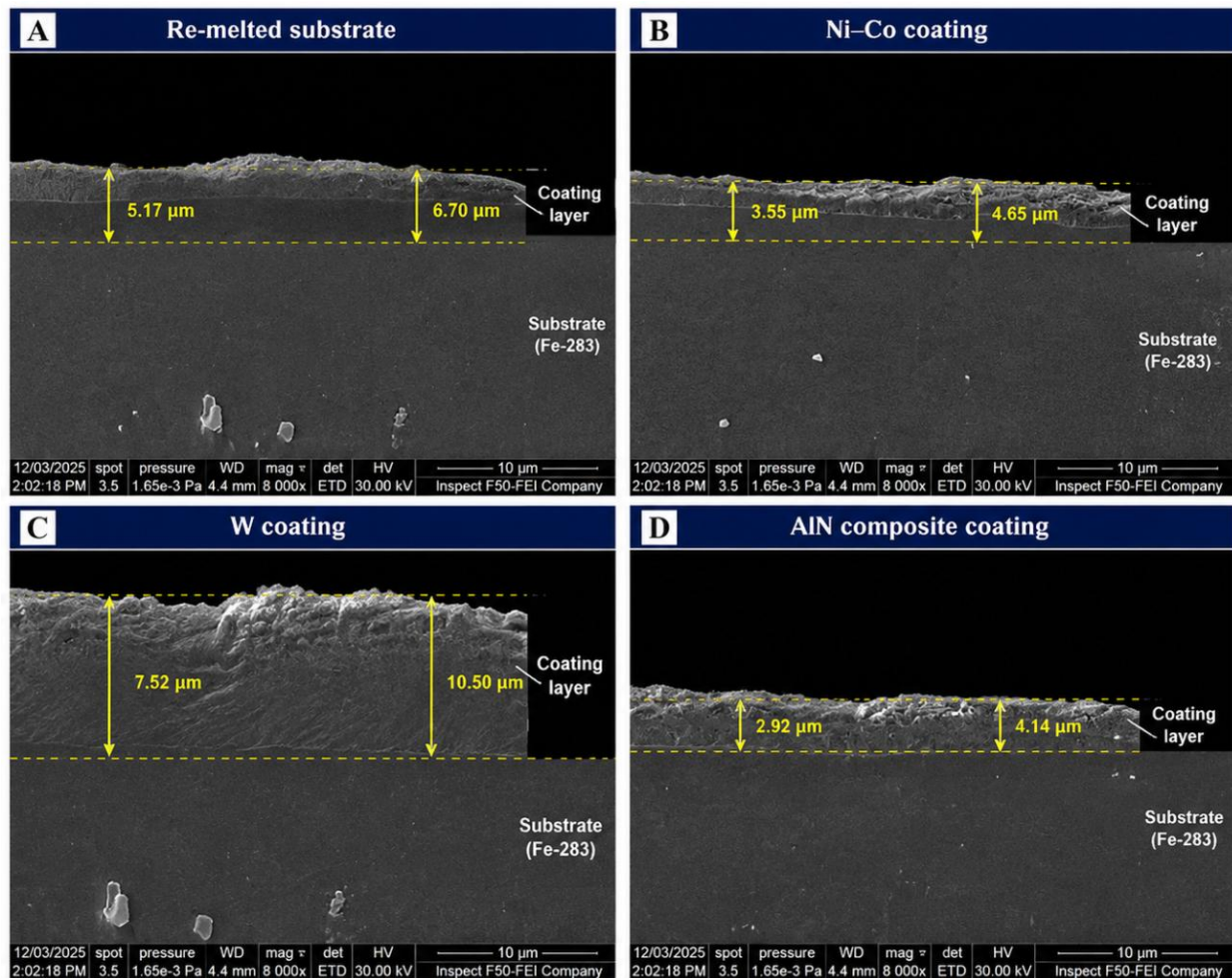


Figure 2 Cross-sectional FESEM micrograph of laser-treated ASTM C283 steel under different coating conditions: (A) re-melted substrate, (B) N—Co coating, (C) W coating, and (D) AlN composite coating

3.3. Energy dispersive X-ray spectroscopy (EDS)

Figure 3 presents the Energy Dispersive X-ray Spectroscopy (EDS) spectra together with the corresponding elemental mapping and quantitative elemental composition of the laser-treated ASTM C283 steel under four different processing conditions. The EDS analysis confirms the successful incorporation of the alloying and reinforcing elements into the laser-cladded coatings while demonstrating the compositional differences responsible for the observed variations in microstructure, hardness, wear resistance, and corrosion behavior. The elemental mapping further illustrates the spatial distribution of the principal elements within each coating, providing important evidence regarding

elemental homogeneity and metallurgical integrity. The EDS spectrum of the laser re-melted substrate (Figure 3A) is dominated by intense Fe peaks, confirming that iron remains the principal constituent after laser re-melting. Minor peaks corresponding to C, O, Si, Mn, Cr, Ni, and Mo are also detected, originating from the original chemical composition of ASTM C283 steel. The elemental maps reveal a homogeneous distribution of Fe throughout the analyzed region, while the alloying elements remain uniformly dispersed without noticeable segregation. The oxygen signal is relatively weak, indicating that only a thin native oxide film formed during sample handling. Since no additional coating material is introduced, the elemental composition remains essentially identical to that of the substrate, confirming that laser re-melting modifies only the microstructure rather than the chemical composition. The Ni–Co laser-cladded coating (Figure 3B) exhibits new characteristic peaks corresponding to Ni and Co, confirming the successful deposition of the Ni–Co alloy onto the steel substrate. Although Fe remains the dominant element because of partial dilution from the substrate during laser processing, the presence of strong Ni and Co signals demonstrates effective alloy incorporation into the molten pool. The elemental maps show that both Ni and Co are uniformly distributed across the coating without obvious elemental clustering or segregation, indicating excellent mixing during laser melting and rapid solidification. Such homogeneous elemental distribution promotes the formation of a continuous Ni–Co-rich solid-solution matrix, which contributes to improved mechanical strength and enhanced corrosion resistance through uniform electrochemical behavior. The W-modified coating (Figure 3C) presents distinct tungsten peaks in addition to the dominant Fe reflections, confirming successful incorporation of the refractory element into the coating.

Owing to its extremely high atomic number, tungsten generates well-defined high-energy characteristic peaks that are readily distinguished from those of iron. The elemental maps reveal that W is distributed throughout the coating but with slightly greater local intensity than observed for Ni and Co, suggesting the formation of tungsten-rich regions associated with the slower diffusion rate and higher melting temperature of tungsten. These localized W-rich regions may promote the formation of hard intermetallic compounds or carbide phases during solidification, thereby contributing to increased hardness and wear resistance. However, excessive local enrichment of tungsten may also introduce microstructural heterogeneity, consistent with the coarse microstructure observed in the SEM analysis. The AlN-reinforced composite coating (Figure 3D) demonstrates the presence of characteristic Al and N peaks together with the Fe peaks originating from the steel substrate. The simultaneous detection of aluminum and nitrogen confirms the successful retention of AlN ceramic particles during laser processing. Unlike metallic alloying elements, AlN particles remain largely stable because of their high melting temperature and excellent thermal stability. The elemental mapping clearly shows a relatively uniform distribution of Al and N throughout the coating, indicating effective dispersion of the ceramic particles within the metallic matrix. This homogeneous distribution supports the grain refinement mechanism observed in the SEM micrographs, where AlN particles act as heterogeneous nucleation sites and inhibit grain boundary migration during rapid solidification.

A comparison of the four coating systems demonstrates that laser processing successfully modified the surface chemistry of ASTM C283 steel without introducing significant elemental segregation or interfacial discontinuities. The Ni–Co coating exhibits excellent elemental homogeneity owing to complete metallurgical mixing between the alloy and the substrate. The W coating introduces localized tungsten-rich strengthening regions that enhance mechanical performance, whereas the AlN composite coating achieves the most uniform distribution of reinforcing particles within the metallic matrix. The elemental mapping results therefore provide strong evidence for the successful fabrication of dense, metallurgically bonded composite coatings and explain the differences in the subsequent hardness, wear, and electrochemical corrosion behavior [31].

The EDS results are also consistent with the microstructural observations presented in Figure 1 and the cross-sectional morphology shown in Figure 2, confirming that the chemical composition directly governs the solidification behavior and final microstructural characteristics of the laser-cladded coatings. In particular, the homogeneous dispersion of AlN particles and the uniform Ni–Co solid solution provide the microstructural basis for enhanced mechanical integrity and corrosion protection, while the localized enrichment of tungsten explains the higher coating thickness and increased hardness observed for the W-modified coating. For the elemental mapping: Figure 3A (Re-melted substrate): Uniform Fe distribution with minor alloying elements confirms that laser re-melting alters the microstructure without changing the substrate chemistry. Figure 3B (Ni–Co coating): Uniform distribution of Ni and Co indicates complete metallurgical mixing, forming a homogeneous solid-solution coating with minimal elemental segregation. Figure 3C (W coating): Tungsten is successfully incorporated throughout the coating. Slight local enrichment reflects its refractory nature and limited diffusion, contributing to precipitation strengthening and increased hardness. Figure 3D (AlN coating): Uniform dispersion of Al and N demonstrates successful incorporation of ceramic AlN particles. Their homogeneous distribution supports grain refinement, dispersion strengthening, and the formation of a compact corrosion-resistant coating.

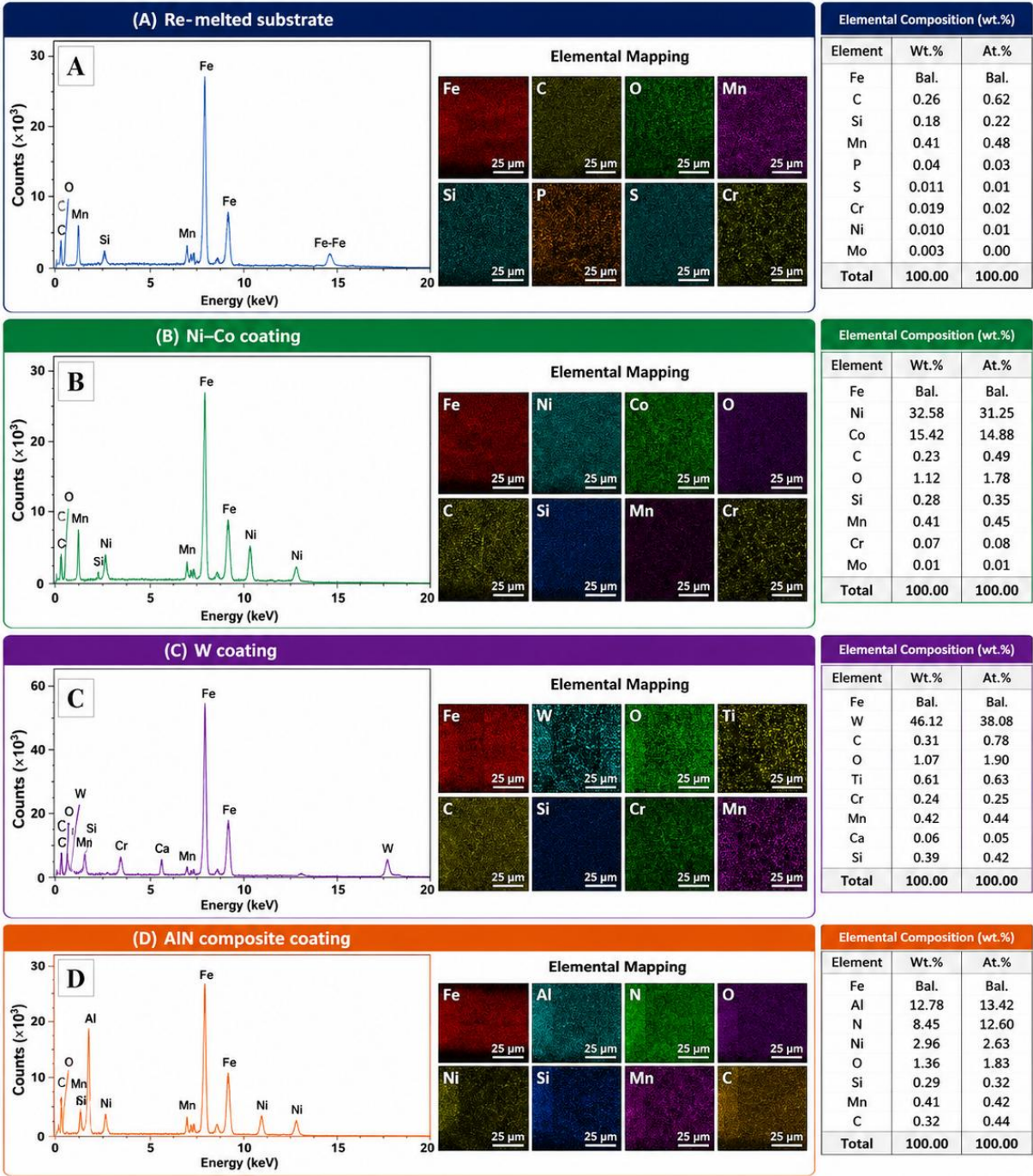


Figure 3 EDS spectra, element mapping, and quantitative compositions of laser-treated ASTM C283 steel: (A) re-melted substrate, (B) N—Co coating, (C) W coating, and (D) AlN composite coating.

3.4. Atomic force microscopy (AFM)

Figure 4 presents the two-dimensional (2D) and three-dimensional (3D) Atomic Force Microscopy (AFM) surface topographies together with the corresponding particle size distribution histograms and quantitative surface roughness parameters of the laser-treated ASTM C283 steel under different coating conditions. AFM analysis provides detailed information regarding the nanoscale surface morphology, grain evolution, average particle diameter (D_{ave}), arithmetic average roughness (R_a), and

root mean square roughness (R_{rms}). These parameters are closely associated with the laser solidification behavior and play a significant role in determining the mechanical performance and corrosion resistance of the fabricated coatings. The laser re-melted substrate (Figure 4A) exhibits a relatively rough surface characterized by irregular grain clusters distributed throughout the scanned area. The 2D topography reveals pronounced height fluctuations, while the corresponding 3D image shows numerous sharp surface asperities with considerable peak-to-valley variations. The particle size distribution histogram indicates a broad grain size distribution centered within the intermediate diameter range, suggesting non-uniform grain growth during rapid laser solidification in the absence of reinforcing particles. Consequently, the re-melted surface exhibits relatively high values of R_a , R_{rms} , and D_{ave} , reflecting the formation of coarse surface features and an uneven topography.

The Ni–Co laser-cladded coating (Figure 4B) demonstrates a noticeably smoother surface morphology than the laser re-melted substrate. The AFM images reveal a homogeneous distribution of fine grains with reduced surface undulations, while the 3D topography exhibits lower peak heights and improved surface uniformity. The particle size distribution becomes narrower and shifts toward smaller diameters, indicating that the addition of Ni–Co enhances heterogeneous nucleation and suppresses excessive grain growth during solidification. Correspondingly, both R_a and R_{rms} decrease relative to the re-melted substrate, while the reduction in D_{ave} confirms effective grain refinement. The improved surface uniformity is expected to enhance corrosion resistance by minimizing localized electrochemical activity and reducing the number of preferential corrosion initiation sites.

The W-modified coating (Figure 4C) exhibits the roughest surface among the investigated coatings. The 2D AFM image displays numerous coarse protrusions and irregular surface features, whereas the 3D reconstruction reveals large surface peaks with significant height differences. The particle size distribution histogram extends toward larger grain diameters and exhibits the widest distribution range, indicating the formation of relatively coarse grains. This behavior is attributed to the exceptionally high melting temperature and low diffusion coefficient of tungsten, which reduce nucleation density and promote localized grain growth during solidification. Consequently, the W coating exhibits the highest R_a , R_{rms} , and D_{ave} values among all samples, confirming that tungsten promotes surface roughening despite its beneficial strengthening effect. The increased roughness is consistent with the coarse microstructure observed in the SEM analysis and may contribute to enhanced mechanical interlocking and wear resistance, although excessive surface roughness may increase the effective area exposed to corrosive media.

In contrast, the AlN-reinforced composite coating (Figure 4D) exhibits the finest and most homogeneous surface morphology. The 2D AFM image reveals uniformly distributed nanoscale grains with minimal agglomeration, while the 3D topography shows a relatively smooth surface characterized by uniformly distributed low-amplitude peaks. The particle size distribution histogram is comparatively narrow and centered at smaller particle diameters, indicating efficient grain refinement. The ceramic AlN particles serve as heterogeneous nucleation sites and inhibit grain boundary migration through the Zener pinning mechanism, resulting in a compact nanostructured surface. Consequently, the AlN composite coating exhibits the lowest values of R_a , R_{rms} , and D_{ave} , confirming the formation of a dense and refined microstructure with superior surface quality.

A comparison of the four coating systems clearly demonstrates that the surface morphology is strongly influenced by the chemical composition of the deposited layer. The average particle diameter follows the approximate trend: W > Re-melted > Ni–Co > AlN, which agrees well with the microstructural observations obtained from SEM. Similarly, the surface roughness parameters exhibit the same tendency, with the W coating showing the highest roughness and the AlN composite coating exhibiting the smoothest surface. The reduction in grain size and surface roughness observed after AlN

reinforcement confirms that ceramic particles effectively promote heterogeneous nucleation while suppressing grain coarsening during rapid laser solidification.

The observed AFM results correlate closely with the SEM micrographs presented in Figure 1 and the cross-sectional morphology shown in Figure 2. The refined nanostructure of the AlN composite coating explains its compact cross-sectional morphology and is expected to contribute to higher hardness, lower wear rate, and superior corrosion resistance. Conversely, the coarse grains and increased roughness of the W coating explain its thicker deposited layer and improved strengthening effect, although these features may reduce electrochemical stability. The Ni–Co coating exhibits an intermediate behavior, combining relatively low roughness with excellent surface homogeneity, making it a favorable compromise between mechanical performance and corrosion protection.

The AFM investigation demonstrates that the incorporation of different alloying and reinforcing materials significantly modifies the nanoscale surface morphology of laser-cladded ASTM C283 steel. Among the investigated coatings, the AlN composite coating provides the most refined microstructure, the smallest average particle diameter, and the lowest surface roughness, whereas the W-modified coating produces the coarsest grains and highest roughness, and the Ni–Co coating achieves a dense, homogeneous surface with moderate roughness and refined grains. These findings provide a strong microstructural basis for understanding the mechanical and electrochemical performance discussed in the subsequent sections.

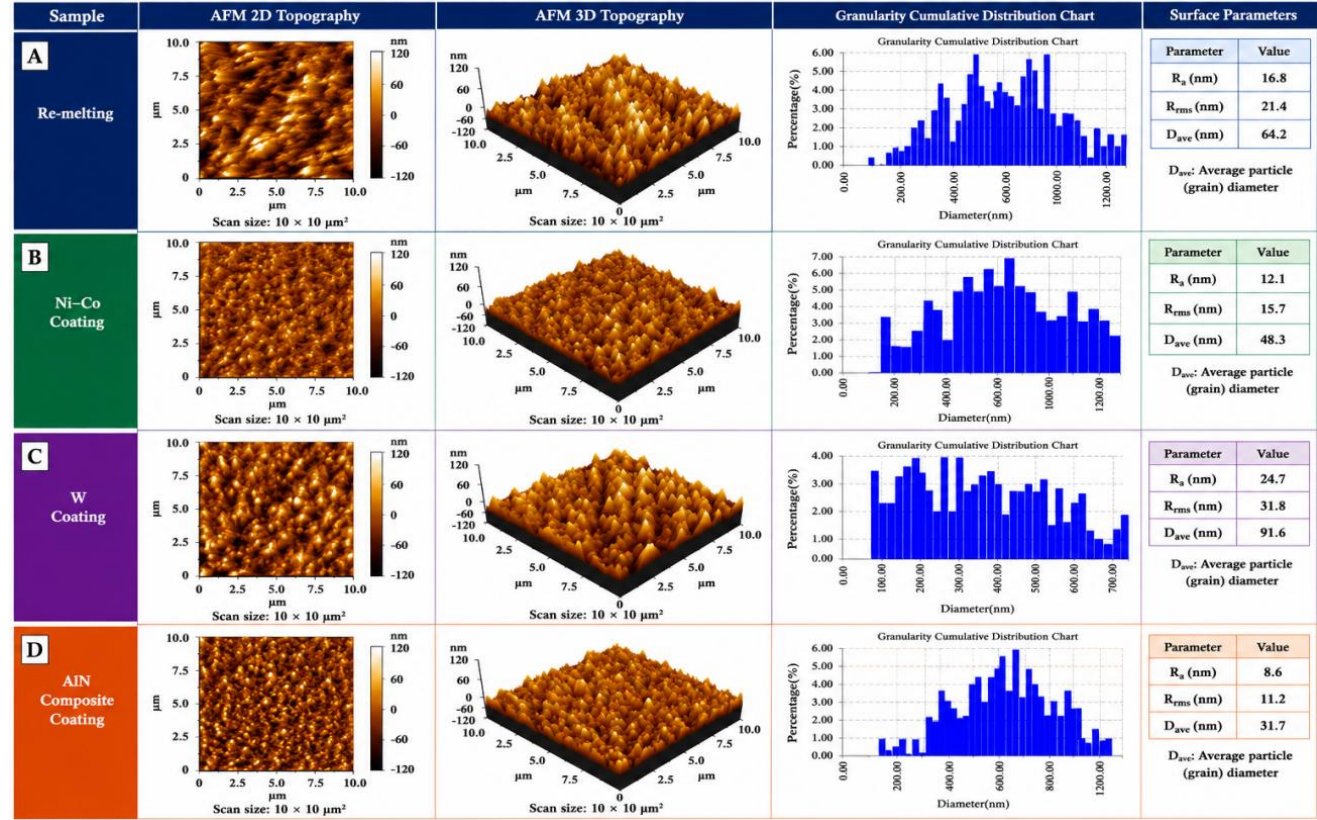


Figure 4 AFM 2D and 3D topography images, granularity cumulative distribution charts, and surface parameters of laser-treated ASTM C283 steel: (A) re-melted substrate, (B) N–Co coating, (C) W coating, and (D) AlN composite coating.

3.5. Vickers microhardness profiles

Figure 5 presents the average Vickers microhardness values of the laser-treated ASTM C283 steel following laser re-melting and laser cladding with Ni–Co, W, and AlN coatings. The hardness values are calculated from ten independent indentations performed on each sample, and the complete experimental measurements together with the corresponding average values are summarized in the hardness table included in Figure 5. The results reveal a substantial improvement in surface hardness after laser cladding, confirming that the incorporation of metallic and ceramic reinforcements effectively enhances the resistance of ASTM C283 steel to localized plastic deformation.

The laser re-melted substrate exhibits the lowest average hardness of 286.1 HV, serving as the reference condition for evaluating the strengthening effects of the different coatings. The hardness improvement obtained by laser re-melting alone is primarily attributed to the rapid melting and solidification process, which refines the surface grains and increases grain-boundary density. According to the Hall–Petch relationship, the refined grains provide additional resistance to dislocation movement, leading to a moderate increase in hardness. However, because no external reinforcing phases are introduced during laser re-melting, strengthening is limited to microstructural refinement, resulting in the lowest hardness among all investigated samples.

The Ni–Co laser-cladded coating exhibits a dramatic increase in hardness, reaching an average value of 821.8 HV, as reported in the hardness table of Figure 5. This value corresponds to an increase of approximately 187.2% relative to the laser re-melted substrate. The remarkable improvement originates from the combined effects of rapid solidification, solid-solution strengthening, and the formation of a homogeneous cellular-dendritic microstructure observed in the SEM analysis. Nickel and cobalt atoms dissolve into the iron matrix, producing lattice distortion that impedes dislocation motion. In addition, the excellent metallurgical bonding and uniform elemental distribution observed in the EDS analysis promote consistent load transfer during indentation, thereby increasing the measured hardness. The relatively small variation among the ten hardness measurements further confirms the excellent homogeneity and structural integrity of the Ni–Co coating.

The W-modified coating exhibits an average hardness of 811.7 HV, which is slightly lower than that of the Ni–Co coating but still approximately 183.7% higher than the re-melted substrate. As shown in the hardness table of Figure 5, the hardness measurements remain relatively consistent across the ten indentation points, indicating good coating uniformity despite the coarse microstructure observed by SEM. The significant hardness enhancement is attributed to the exceptionally high intrinsic hardness of tungsten together with the formation of tungsten-rich strengthening phases during laser solidification. These hard phases effectively obstruct dislocation movement and improve resistance to plastic deformation. Nevertheless, the AFM analysis revealed that the W coating possesses the highest surface roughness and the largest average particle diameter among the investigated coatings. The relatively coarse grains and localized tungsten-rich regions reduce stress uniformity during indentation, explaining why the hardness remains marginally lower than that of the Ni–Co coating despite tungsten's superior intrinsic hardness.

Among all investigated coatings, the AlN-reinforced composite coating exhibits the highest average hardness of 956.4 HV, representing an improvement of approximately 234.3% compared with the laser re-melted substrate. As summarized in the hardness table within Figure 5, all ten hardness measurements remain consistently high with minimal experimental scatter, indicating excellent coating uniformity and reproducibility. The superior hardness of the AlN composite coating results from multiple strengthening mechanisms acting simultaneously. The uniformly dispersed AlN ceramic particles act as effective barriers to dislocation motion through dispersion strengthening while

simultaneously serving as heterogeneous nucleation sites during laser solidification. These particles inhibit grain boundary migration through the Zener pinning mechanism, producing the finest grain structure among all investigated coatings, as confirmed by the AFM and SEM analyses. Furthermore, the inherently high hardness and elastic modulus of AlN contribute directly to the overall mechanical strengthening of the composite coating. The combination of grain refinement, dispersion strengthening, and the compact microstructure produces the highest resistance to localized plastic deformation.

A direct comparison of the average hardness values listed in the hardness table of Figure 5 clearly demonstrates that the coating composition strongly governs the mechanical performance of the laser-treated surface. The hardness follows the sequence:

$$\text{AlN (956.4 HV)} > \text{Ni-Co (821.8 HV)} > \text{W (811.7 HV)} > \text{Re-melting (286.1 HV)}.$$

This trend is in excellent agreement with the microstructural observations presented in Figures 1–4. The AlN composite coating exhibited the finest grain structure, the smallest average particle diameter, and the lowest surface roughness, all of which contributed to its superior hardness. The Ni–Co coating developed a dense cellular-dendritic microstructure with excellent metallurgical bonding, resulting in high hardness and excellent structural uniformity. The W coating generated coarse grains and tungsten-rich hard phases that significantly increased hardness but also produced greater surface roughness and microstructural heterogeneity. In contrast, the laser re-melted substrate lacked secondary strengthening phases and therefore exhibited the lowest hardness despite undergoing grain refinement.

The individual hardness measurements presented in the hardness table of Figure 5 also provide important information regarding coating quality and experimental reproducibility. The relatively narrow scatter observed for all coating systems indicates stable laser processing conditions and homogeneous coating formation. In particular, the AlN and Ni–Co coatings exhibit the smallest variation between individual measurements, reflecting their highly uniform microstructures and consistent elemental distributions. The W coating displays slightly greater measurement variability, which can be attributed to localized tungsten-rich regions and the comparatively heterogeneous microstructure observed in the SEM images.

The hardness enhancement demonstrated in Figure 5 has important implications for the expected service performance of the coatings. In general, higher hardness improves resistance to plastic deformation, abrasion, scratching, and material removal during sliding contact. Consequently, the AlN composite coating is expected to provide the highest wear resistance and longest service life, followed by the Ni–Co and W coatings. Moreover, the refined microstructure and compact morphology associated with higher hardness also reduce the number of preferential corrosion initiation sites, contributing to improved electrochemical stability. These observations are fully consistent with the SEM, EDS, AFM, and cross-sectional analyses presented in the preceding sections and collectively demonstrate that AlN reinforcement offers the most effective strategy for simultaneously enhancing the mechanical integrity and surface durability of laser-cladded ASTM C283 steel.

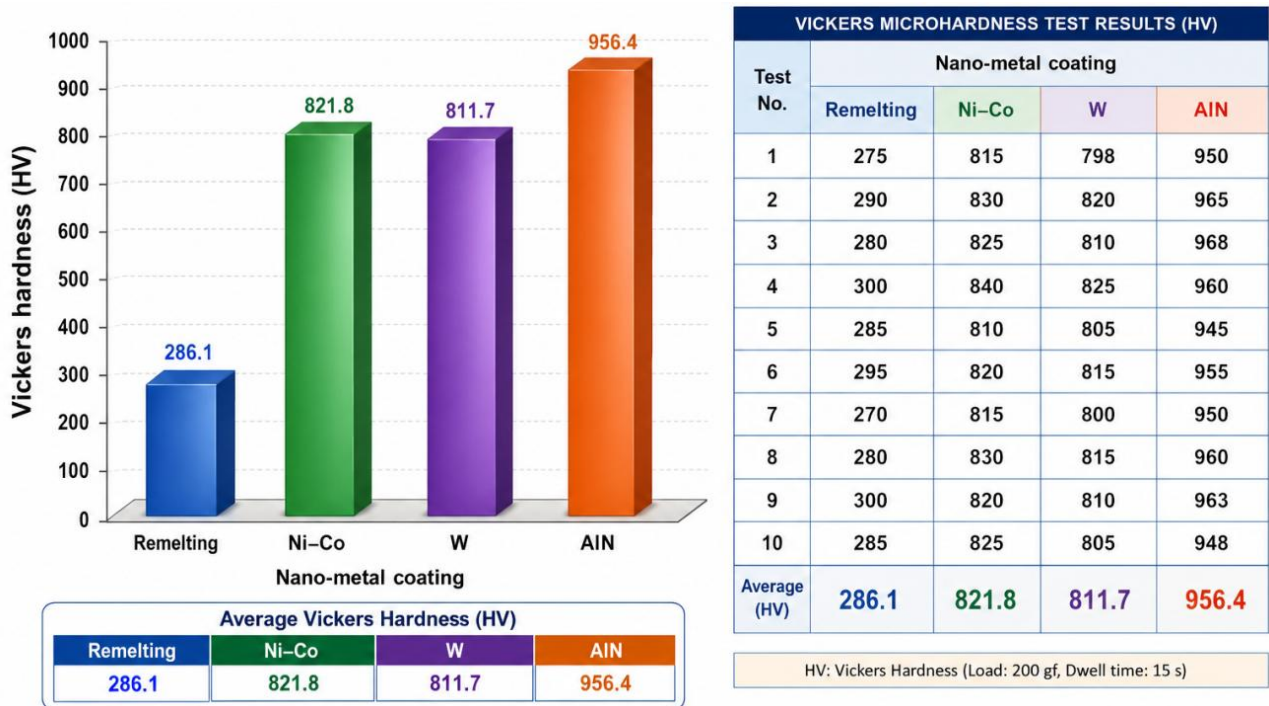


Figure 5 Vickers microhardness of nano-metal coatings.

3.6. Wear track

The tribological performance of the laser-treated ASTM C283 steel is evaluated through optical examination of the wear tracks and quantitative wear-rate measurements. The optical micrographs of the worn surfaces are presented in Figure 6, whereas the measured wear rates together with the three repeated experimental measurements and statistical analysis are summarized in Table 4, Table 5, and Figure 7. The combined qualitative and quantitative analyses clearly demonstrate that laser cladding significantly improves the wear resistance of ASTM C283 steel, although the extent of improvement strongly depends on the coating composition.

The laser re-melted substrate (Figure 6A) exhibits the most severe wear damage among all investigated samples. The wear track is characterized by wide and deep parallel grooves together with extensive plastic deformation and material removal along the sliding direction. Several detached wear debris particles are also observed on the worn surface, indicating repeated fracture and delamination during sliding. These features are typical of severe abrasive wear accompanied by localized adhesive wear. The relatively coarse microstructure and the lowest hardness (286.1 HV) provide limited resistance against penetration of the counterface, allowing deep ploughing and extensive material loss. Consequently, as reported in Table 4 and illustrated in Figure 7, the laser re-melted substrate exhibits the highest average wear rate of $11.8 \times 10^{-6} \text{ g mm}^{-3}$, confirming its relatively poor tribological performance.

The Ni-Co laser-cladded coating (Figure 6B) exhibits a significant reduction in wear-track severity compared with the re-melted substrate. The worn surface contains considerably shallower grooves with fewer detached wear particles and reduced plastic deformation. The smoother wear track indicates that the Ni-Co coating effectively resists penetration by the sliding counterbody owing to its refined dendritic microstructure and high hardness. The homogeneous Ni-Co matrix distributes the applied contact stress more uniformly while suppressing crack initiation and propagation. As shown in Table 4, the average wear rate decreases to $8.0 \times 10^{-6} \text{ g mm}^{-3}$, representing a reduction of approximately

32.2% relative to the re-melted substrate. Furthermore, the three repeated measurements listed in Table 4 exhibit excellent agreement, while the statistical summary presented in Table 5 indicates a small standard deviation ($8.0 \pm 0.10 \times 10^{-6} \text{ g mm}^{-3}$), confirming the excellent repeatability and uniformity of the coating.

The W-modified coating (Figure 6C) exhibits wear damage that is intermediate between the Ni-Co and re-melted samples. Although the wear track remains visible, its width and depth are noticeably smaller than those observed in the re-melted substrate. The presence of tungsten-rich hard phases effectively reduces plastic deformation and improves resistance to abrasive wear. However, localized wear scars and discontinuous grooves remain visible, suggesting that the relatively coarse microstructure and higher surface roughness promote localized stress concentration during sliding. According to Table 4, the average wear rate of the W coating is $9.1 \times 10^{-6} \text{ g mm}^{-3}$, corresponding to a reduction of approximately 22.9% compared with the re-melted substrate. The statistical results in Table 5 indicate excellent measurement consistency ($9.1 \pm 0.10 \times 10^{-6} \text{ g mm}^{-3}$), demonstrating stable wear behavior despite the microstructural heterogeneity.

Among all investigated coatings, the AlN-reinforced composite coating (Figure 6D) exhibits the least wear damage and the narrowest wear track. The worn surface remains comparatively smooth with only shallow grooves and minimal evidence of plastic deformation or material removal. Detached wear debris is scarce, indicating that the coating effectively suppresses crack formation and material fracture during sliding. These observations are attributed to the synergistic effect of ceramic particle reinforcement and pronounced grain refinement. The uniformly dispersed AlN particles serve as highly effective load-bearing constituents that prevent deep penetration of the counter face while simultaneously restricting dislocation motion. The refined nanostructure observed in the AFM analysis further enhances resistance to abrasive wear by distributing the applied stress over a larger number of grain boundaries. As summarized in Table 4 and Figure 7, the AlN composite coating exhibits the lowest average wear rate of $5.8 \times 10^{-6} \text{ g.mm}^{-3}$, corresponding to a remarkable reduction of approximately 50.8% relative to the laser re-melted substrate. The very small standard deviation reported in Table 5 ($5.8 \pm 0.10 \times 10^{-6} \text{ g.mm}^{-3}$) further confirms the excellent reproducibility and uniformity of the AlN coating. A direct comparison of the quantitative wear results summarized in Table 4, Table 5, and Figure 7 clearly demonstrates that the wear resistance follows the order:

AlN composite coating > Ni-Co coating > W coating > Laser re-melted substrate.

Since wear resistance is inversely proportional to wear rate, the AlN coating provides the highest tribological performance, whereas the re-melted substrate exhibits the poorest resistance to sliding wear. The measured wear rates decrease from $11.8 \times 10^{-6} \text{ g mm}^{-3}$ for the re-melted substrate to $8.0 \times 10^{-6} \text{ g mm}^{-3}$ for the Ni-Co coating, $9.1 \times 10^{-6} \text{ g mm}^{-3}$ for the W coating, and finally $5.8 \times 10^{-6} \text{ g mm}^{-3}$ for the AlN composite coating. These results are in excellent agreement with the hardness values reported in Figure 5, where the AlN coating exhibited the highest average hardness (956.4 HV), followed by Ni-Co (821.8 HV), W (811.7 HV), and the re-melted substrate (286.1 HV). The inverse relationship between hardness and wear rate confirms that increasing surface hardness effectively suppresses material removal during sliding.

The tribological results also correlate closely with the microstructural observations presented in Figures 1–4. The AlN composite coating exhibited the finest grain structure, the smallest average particle diameter, the lowest surface roughness, and the most homogeneous elemental distribution, all of which contributed to its outstanding wear resistance. The Ni-Co coating combined a homogeneous dendritic microstructure with excellent metallurgical bonding, resulting in both high hardness and low wear. In contrast, the W coating contained hard strengthening phases but also possessed a comparatively coarse microstructure and higher roughness, reducing its overall wear performance

relative to the Ni–Co and AlN coatings. The re-melted substrate, lacking reinforcing phases, exhibited the coarsest wear track and the highest material loss.

The three repeated wear measurements listed in Table 4, together with the statistical analysis presented in Table 5, demonstrate excellent experimental reproducibility for all investigated coatings. The relatively small standard deviations indicate stable tribological behavior and confirm the reliability of the laser-cladding process. Overall, the wear investigation demonstrates that AlN particle reinforcement provides the most effective strategy for improving the tribological performance of laser-cladded ASTM C283 steel, owing to the combined effects of ceramic dispersion strengthening, grain refinement, increased hardness, and the formation of a dense and mechanically stable surface layer. These findings are fully consistent with the SEM, EDS, AFM, and hardness analyses presented in the preceding sections and establish a clear structure–property relationship between microstructural refinement and wear resistance.

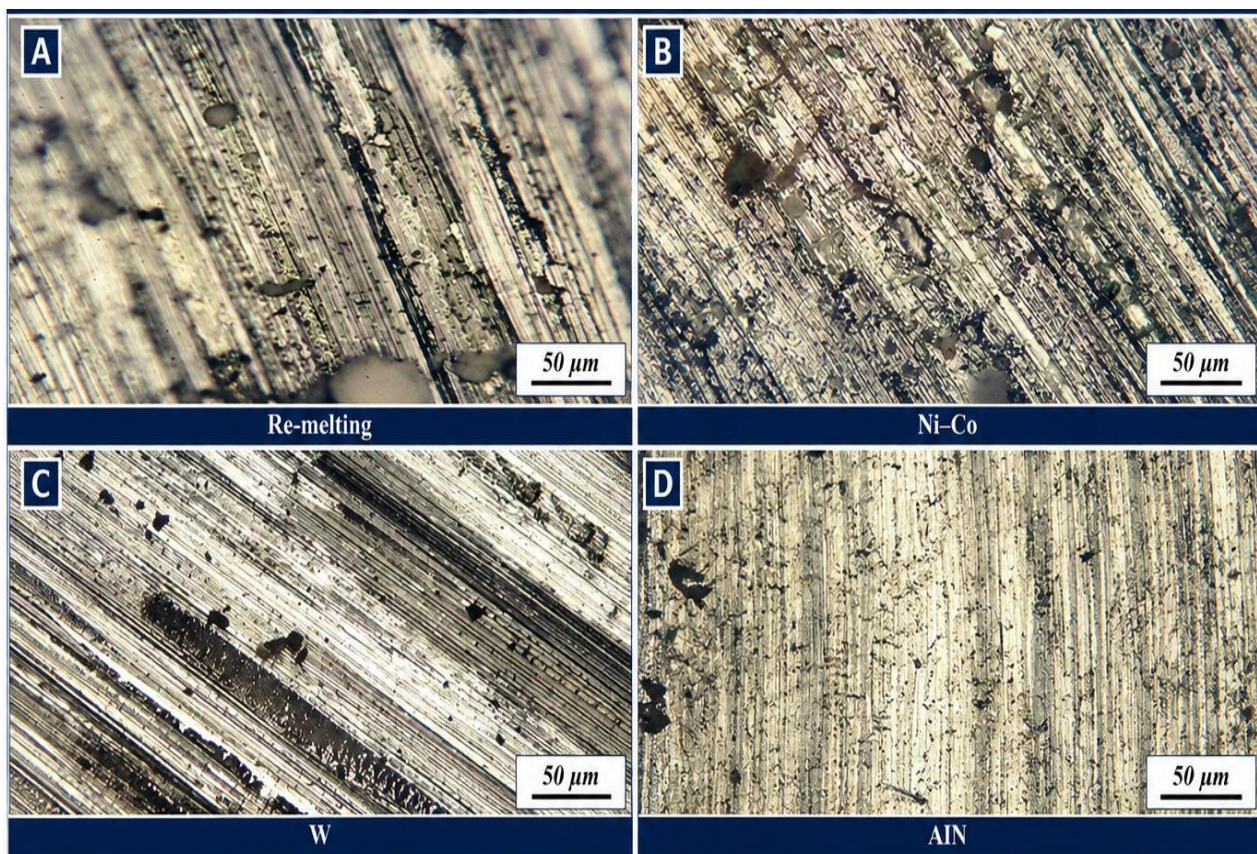


Figure 6 Optical microscope images of wear tracks for different coatings.

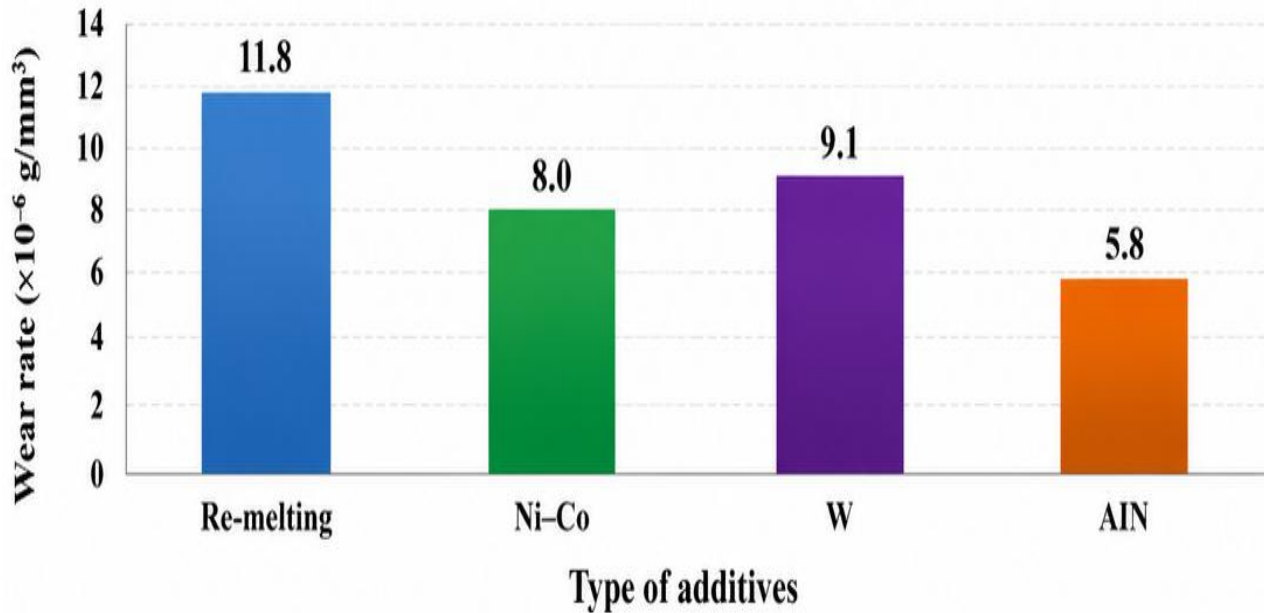


Figure 7 Wear rate of different coatings.

Table 4 Wear rate measurements of the laser-treated ASTM C283 steel coatings.

Sample	Test 1 ($\times 10^{-6}$ g/mm ³)	Test 2 ($\times 10^{-6}$ g/mm ³)	Test 3 ($\times 10^{-6}$ g/mm ³)	Average ($\times 10^{-6}$ g/mm ³)
Re-melting	11.6	11.8	12.0	11.8
Ni-Co coating	7.9	8.0	8.1	8.0
W coating	9.0	9.1	9.2	9.1
AlN composite coating	5.7	5.8	5.9	5.8

Table 5 Average wear rate (mean $\pm$ SD) of laser-treated ASTM C283 steel after laser re-melting and laser cladding with Ni-Co, W, and AlN coatings, calculated from three repeated wear experiments.

Sample	Mean $\pm$ SD ($\times 10^{-6}$ g/mm ³)
Re-melting	11.8 $\pm$ 0.20
Ni-Co coating	8.0 $\pm$ 0.10
W coating	9.1 $\pm$ 0.10
AlN composite coating	5.8 $\pm$ 0.10

3.7. X-ray diffraction (XRD) patterns

Figure 8 presents the X-ray diffraction patterns of the laser re-melted ASTM C283 steel substrate and the Ni-Co-, W-, and AlN-modified coatings, while Table 6 summarizes the corresponding diffraction angles, full widths at half maximum, Miller indices, interplanar spacings, individual crystallite sizes, and average crystallite sizes. The combined analysis confirms that laser processing preserves the principal α -Fe phase of the substrate while introducing additional metallic or ceramic crystalline phases according to the deposited coating composition. The interplanar spacing, d , is determined using Bragg's law:

$$n\lambda = 2d\sin\theta \quad (1)$$

For first-order diffraction, $n = 1$, and therefore:

$$d = \frac{\lambda}{2\sin\theta} \quad (2)$$

where $\lambda = 1.5406 \text{ \AA}$ for Cu-K α radiation, θ is half of the measured diffraction angle $\frac{2\theta}{2}$, and d is the distance between adjacent crystallographic planes. The average coherent crystallite size is calculated using the Scherrer equation [16]:

$$D_{cs} = \frac{K\lambda}{\beta\cos\theta} \quad (3)$$

where D_{cs} is the crystallite size, $K = 0.90$ is the shape factor, λ is the X-ray wavelength, and β is the full width at half maximum expressed in radians. The average crystallite size of each sample is calculated from [20]:

$$D_{cs-ave} = \frac{1}{N} \sum_{i=1}^N D_i \quad (4)$$

where N is the number of analyzed reflections.

3.7.1. Laser re-melted substrate, S1

The XRD pattern of S1 in Figure 8A is dominated by α -Fe reflections located at approximately 44.67°, 65.02°, and 82.33°, indexed to the (110), (200), and (211) planes, respectively. The calculated interplanar spacings listed in Table 6 are 2.027, 1.433, and 1.170 $\text{\AA}$. These reflections confirm that the re-melted ASTM C283 steel retains the body-centered cubic ferritic structure after laser processing. The strongest reflection occurs along the α -Fe (110) plane, indicating that this plane remains the preferred crystallographic orientation after rapid melting and solidification. The absence of additional intense reflections suggests that laser re-melting primarily modifies the grain morphology and defect density without producing a substantial phase transformation.

The FWHM values of S1 range from 0.22° to 0.28°, yielding crystallite sizes between 37.68 and 39.04 nm. The calculated average crystallite size is 38.14 nm, which is the highest among the four investigated samples. This relatively large value indicates that, although rapid solidification refines the original steel microstructure, the lack of external nucleating phases permits comparatively greater crystal growth than in the coated samples.

3.7.2. Ni-Co coating, S2

The diffraction pattern of S2 in Figure 8B contains overlapping reflections from α -Fe, γ -Ni, and FCC Co. The principal peaks appear at approximately 44.50°, 51.80°, 65.02°, 76.40°, and 92.90°. These peaks are assigned primarily to the γ -Ni/Co (111), γ -Ni (200), α -Fe (200), γ -Ni/Co (220), and γ -Ni (222) planes. Their corresponding d -spacings are 2.034, 1.764, 1.433, 1.246, and 1.063 $\text{\AA}$, respectively. The overlap between Ni, Co, and Fe reflections is expected because Ni and FCC Co have closely related lattice parameters and similar diffraction positions. In addition, partial dilution of the coating by iron during laser cladding produces a compositionally mixed metallic matrix. Slight peak displacement and broadening may therefore reflect lattice distortion caused by substitutional incorporation of Ni and Co atoms into the Fe-rich matrix.

As reported in Table 6, the FWHM values for S2 range from 0.28° to 0.40°, producing crystallite sizes between 26.76 and 33.64 nm. The average crystallite size decreases to 29.18 nm, representing a reduction of approximately:

$$38.14 - 29.18 \div 38.14 \times 100 \approx 23.5\% \quad (5)$$

relative to S1. This reduction indicates that Ni and Co promote solute trapping, lattice distortion, and increased nucleation during rapid solidification. The finer crystallites also provide a greater density of grain boundaries, which impede dislocation motion and contribute to the high hardness of the Ni–Co coating.

3.7.3. W-modified coating, S3

The XRD pattern of S3 in Figure 8C contains characteristic reflections of BCC tungsten together with residual α -Fe peaks from the substrate. The W reflections occur at approximately 40.26°, 58.28°, 73.24°, and 87.10°, corresponding to the (110), (200), (211), and (220) planes, respectively. The associated d -spacings are 2.238, 1.582, 1.291, and 1.118 Å. The α -Fe reflections remain visible at approximately 44.67°, 65.02°, and 82.33°, corresponding to the (110), (200), and (211) planes.

The simultaneous presence of W and α -Fe peaks confirms successful incorporation of tungsten without complete dissolution into the iron matrix. The relatively sharp W reflections indicate the formation of crystalline tungsten-rich domains. Because tungsten has an exceptionally high melting temperature and limited diffusivity, its distribution during the short laser interaction time is less uniform than that of Ni and Co. This promotes the development of localized W-rich crystalline regions and explains the comparatively coarse morphology observed by SEM and AFM.

The FWHM values listed in Table 6 range from 0.24° to 0.36°, corresponding to crystallite sizes between 30.45 and 35.25 nm. The average value is 31.92 nm, which is smaller than that of S1 but larger than those of S2 and S4. Relative to the re-melted sample, the crystallite-size reduction is approximately:

$$38.14 - 31.9238.14 \times 100 \approx 16.3\% \quad (6)$$

The intermediate crystallite size indicates that W contributes to structural refinement, but its high melting point and slow diffusion limit the degree of homogeneous nucleation. Consequently, the W coating contains a mixture of refined matrix regions and comparatively coarse W-rich domains.

3.7.4. AlN composite coating, S4

The diffraction pattern of S4 in Figure 8D shows reflections associated with hexagonal AlN together with the principal α -Fe peaks. The AlN reflections occur near 33.20°, 36.00°, 37.90°, 49.90°, 59.80°, and 65.50°, corresponding to the (100), (002), (101), (102), (110), and (103) planes, respectively. The calculated d -spacings are 2.696, 2.493, 2.372, 1.826, 1.545, and 1.424 Å. The remaining α -Fe reflections at approximately 44.67° and 82.33° originate from the steel substrate and the partially diluted metallic matrix.

The preservation of several AlN reflections confirms that the ceramic reinforcement remains crystalline during laser treatment. Rather than dissolving completely, the AlN particles act as thermally stable heterogeneous nucleation sites. Their presence increases the nucleation density and restricts grain-boundary migration during solidification. These mechanisms account for the broadening of the AlN-related peaks and the reduction in crystallite size.

The FWHM values of S4 range from 0.29° to 0.40°, producing crystallite sizes between 23.61 and 29.62 nm. The calculated average crystallite size is 25.78 nm, the lowest among all samples. Compared with S1, this represents a reduction of approximately:

$$38.14 - 25.7838.14 \times 100 \approx 32.4\% \quad (7)$$

The significant refinement confirms the strong nucleating and grain-boundary-pinning effects of AlN. The fine coherent domains are consistent with the compact AFM topography, low surface roughness, high microhardness, and low wear rate of the AlN composite coating.

3.7.5. Comparative structural behavior

Based on Table 6, the average crystallite size follows the sequence:

$$S1 (38.14 \text{ nm}) > S3 (31.92 \text{ nm}) > S2 (29.18 \text{ nm}) > S4 (25.78 \text{ nm}) \quad (8)$$

This sequence demonstrates that laser re-melting alone produces the least crystallographic refinement, whereas AlN reinforcement produces the strongest refinement. The Ni–Co coating also substantially decreases the crystallite size because of rapid solute trapping and homogeneous nucleation. Tungsten provides an intermediate response because its refractory nature promotes hard-phase formation but also favors localized coarse crystalline regions.

The FWHM trend is generally inverse to the crystallite-size trend. Broad diffraction peaks correspond to smaller coherent domains, whereas narrow peaks indicate larger crystallites. This inverse relationship follows directly from the Scherrer equation:

$$D_{cs} \propto \frac{1}{\beta} \quad (9)$$

Accordingly, S4 displays relatively broad reflections and the smallest crystallite size, while S1 shows narrower peaks and the largest crystallite size. For cubic phases, the lattice parameter can be calculated from [22]:

$$a = d\sqrt{h^2 + k^2 + l^2} \quad (10)$$

For the BCC α -Fe (110) reflection of S1, using $d=2.027 \text{ \AA}$:

$$a = 2.027 \sqrt{1^2 + 1^2 + 0^2} = 2.867 \text{ \AA}$$

which agrees with the expected lattice parameter of ferritic iron.

For the hexagonal AlN phase, the lattice parameters are related to the interplanar spacing by [30]:

$$\frac{1}{d^2} = \frac{4}{3} \left(\frac{h^2 + hk + k^2}{a^2} \right) + \frac{l^2}{c^2} \quad (11)$$

For the AlN (100) reflection, $l = 0$, and therefore [13]:

$$a = \frac{2d}{\sqrt{3}} \quad (12)$$

Using $d = 2.696 \text{ \AA}$ gives $a \approx 3.113 \text{ \AA}$. For the (002) reflection [11]:

$$d_{002} = \frac{c}{2} \quad (13)$$

and therefore:

$c = 2d_{002} = 2(2.493) \approx 4.986 \text{ \AA}$. These values are consistent with the hexagonal wurtzite structure of AlN. The unit-cell volume can be obtained using [14]:

$$V = a^3 \quad (14)$$

for cubic phases and [15]:

$$V = \frac{\sqrt{3}}{2} a^2 c \quad (15)$$

for the hexagonal AlN phase. Additional defect-related parameters can also be estimated from the crystallite size. The dislocation density is calculated as [9]:

$$\delta = \frac{1}{D_{cs}^2} \quad (16)$$

while the microstrain may be estimated from [7]:

$$\varepsilon = \frac{\beta}{4 \tan \theta} \quad (17)$$

Because S4 has the smallest crystallite size, it would be expected to exhibit the highest grain-boundary and dislocation density. However, this does not necessarily indicate poorer coating quality. In this case, the high boundary density is associated primarily with controlled nanoscale refinement induced by AlN particles rather than with macroscopic defects such as cracks or delamination.

3.7.6. Structure–property correlation

The crystallite-size comparison obtained from Table 6 correlates closely with the mechanical and tribological results. The AlN coating, which has the smallest average crystallite size of 25.78 nm, also exhibits the highest hardness of 956.4 HV and the lowest wear rate of 5.8×10^{-6} g/mm³. The Ni–Co coating has the second-smallest crystallite size of 29.18 nm and achieves a high hardness of 821.8 HV together with a relatively low wear rate of 8.0×10^{-6} g/mm³. The W coating has a larger average crystallite size of 31.92 nm and exhibits slightly lower hardness and higher wear than the Ni–Co coating. Finally, the re-melted substrate has the largest crystallite size of 38.14 nm, the lowest hardness, and the highest wear rate. These results are consistent with the Hall–Petch relationship [16]:

$$H = H_0 + k_H d^{-1/2} \quad (18)$$

where H is the hardness, H_0 is the intrinsic matrix hardness, k_H is the strengthening coefficient, and d is the characteristic grain or crystallite size. As the crystallite size decreases, the number of boundaries increases, thereby restricting dislocation motion and raising the hardness. The exceptionally high hardness of S4 additionally reflects the intrinsic rigidity and dispersion-strengthening effect of the AlN ceramic particles, while the behavior of S3 is influenced by both tungsten-rich hard phases and microstructural heterogeneity. Figure 8 and Table 6 demonstrate that coating composition directly controls phase formation, diffraction-peak broadening, and crystallite refinement. The AlN composite coating provides the most effective structural refinement, followed by Ni–Co, W, and laser re-melting. This crystallographic evolution explains the progressive improvement in hardness and wear resistance and confirms that AlN reinforcement offers the most favorable combination of stable ceramic phase formation, nanoscale refinement, and mechanical strengthening.

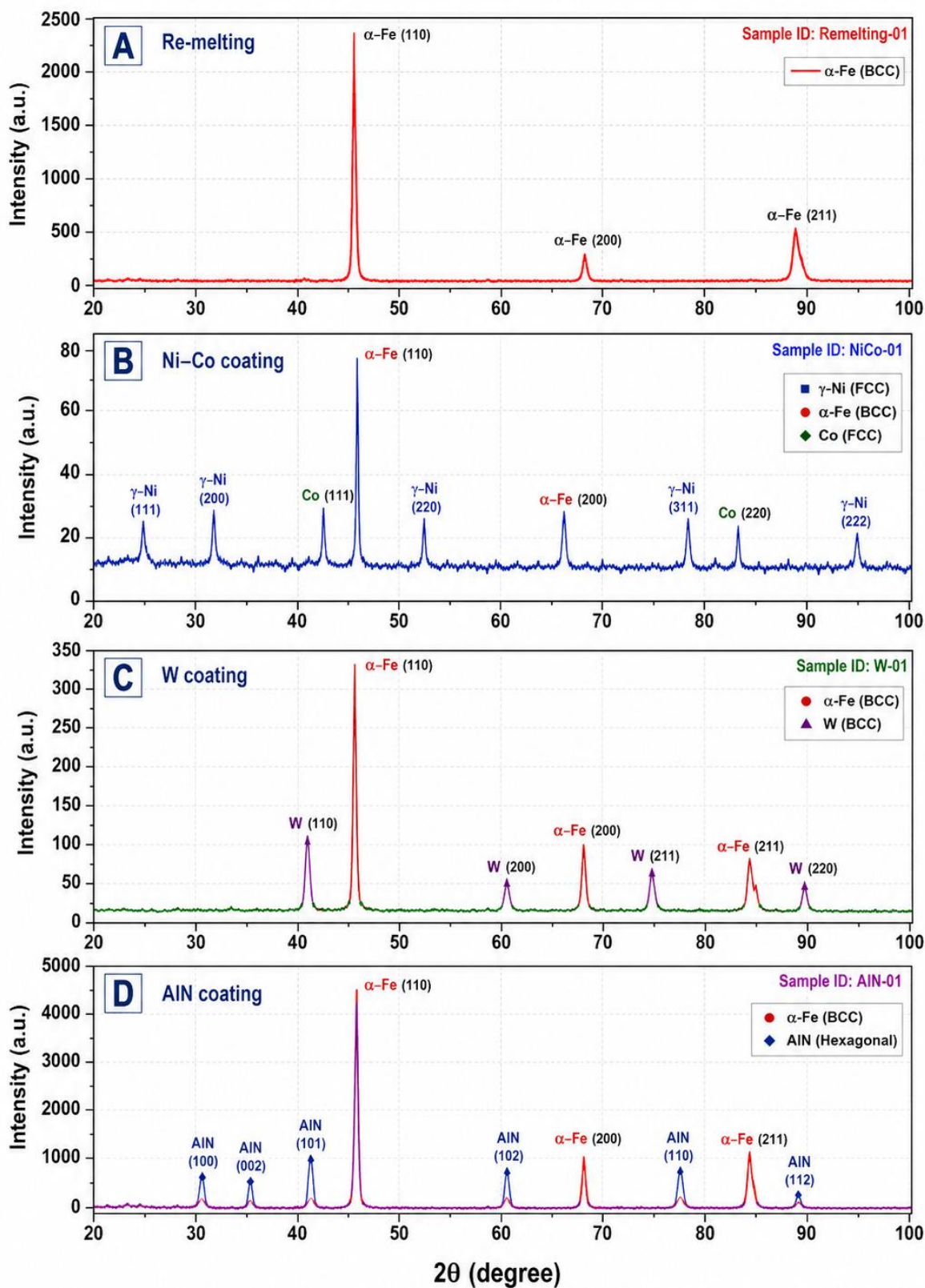


Figure 8 XRD pattern of (A) re-melted substrate, (B) Ni-Co coating, (C) W coating, and (D) AlN coating.

Table 6 XRD structural parameters of the laser re-melted substrate and Ni–Co-, W-, and AlN-coated ASTM C283 steel specimens.

Sample	Phase	2 θ (°)	FWHM (°)	(hkl)	d-spacing (Å)	Dcs (nm)	Dcs-ave (nm)
S1: Re-melted	α -Fe	44.67	0.22	(110)	2.027	39.04	38.14
	α -Fe	65.02	0.25	(200)	1.433	37.68	
	α -Fe	82.33	0.28	(211)	1.170	37.69	
S2: Ni–Co coating	γ -Ni/Co–Fe	44.50	0.30	(111)	2.034	28.61	29.18
	γ -Ni	51.80	0.33	(200)	1.764	26.76	
	α -Fe	65.02	0.28	(200)	1.433	33.64	
	γ -Ni/Co	76.40	0.36	(220)	1.246	28.08	
	γ -Ni	92.90	0.40	(222)	1.063	28.83	
S3: W coating	W	40.26	0.24	(110)	2.238	35.25	31.92
	α -Fe	44.67	0.26	(110)	2.027	33.03	
	W	58.28	0.29	(200)	1.582	31.36	
	α -Fe	65.02	0.30	(200)	1.433	31.40	
	W	73.24	0.32	(211)	1.291	30.93	
	α -Fe	82.33	0.34	(211)	1.170	31.04	
	W	87.10	0.36	(220)	1.118	30.45	
S4: AlN composite coating	AlN	33.20	0.31	(100)	2.696	26.74	25.78
	AlN	36.00	0.33	(002)	2.493	25.31	
	AlN	37.90	0.35	(101)	2.372	24.00	
	α -Fe	44.67	0.29	(110)	2.027	29.62	
	AlN	49.90	0.36	(102)	1.826	24.34	
	AlN	59.80	0.38	(110)	1.545	24.12	
	AlN	65.50	0.40	(103)	1.424	23.61	
	α -Fe	82.33	0.37	(211)	1.170	28.52	

3.8. Potentiodynamic polarization curves (Tafel plots)

Figure 9 presents the potentiodynamic polarization (Tafel) curves of the laser-treated ASTM C283 steel before and after laser cladding with Ni–Co, W, and AlN coatings in 3.5 wt.% NaCl solution at room temperature. Potentiodynamic polarization is one of the most reliable electrochemical techniques for evaluating corrosion behavior because it provides quantitative information regarding the corrosion potential (E_{corr}), corrosion current density (I_{corr}), anodic and cathodic reaction kinetics, polarization resistance, and the protective capability of surface coatings. In general, coatings exhibiting more positive corrosion potentials together with lower corrosion current densities possess superior corrosion resistance because they reduce the electrochemical dissolution rate of the metallic substrate. The corrosion current density is directly related to the corrosion rate according to Faraday's law, whereas the polarization resistance is commonly determined using the Stern–Geary relationship [19]:

$$R_p = \frac{\beta_a \beta_c}{2.303 \times I_{\text{corr}} (\beta_a + \beta_c)} \quad (19)$$

where R_p is the polarization resistance, I_{corr} is the corrosion current density, and β_a and β_c are the anodic and cathodic Tafel slopes, respectively. The corrosion rate can subsequently be estimated from [12]

$$CR = \frac{0.00327 \times I_{\text{corr}} \times EW}{\rho} \quad (20)$$

where CR is the corrosion rate (mm year^{-1}), EW is the equivalent weight of the alloy, ρ is the density (g.cm^{-3} , and I_{corr} is expressed in $\mu\text{A.cm}^{-2}$). The laser re-melted substrate (orange curve) exhibits the largest polarization current over the entire potential range, indicating the highest electrochemical activity among all investigated samples. The anodic branch rises rapidly with increasing potential, reflecting accelerated dissolution of iron in the chloride electrolyte. Simultaneously, the cathodic branch displays relatively high current values, suggesting rapid oxygen reduction and hydrogen evolution reactions on the exposed steel surface. The absence of protective alloying or ceramic phases permits aggressive chloride ions to penetrate the naturally formed oxide film, leading to localized breakdown and active corrosion. Consequently, the re-melted substrate is expected to exhibit the highest corrosion current density, the lowest polarization resistance, and the highest corrosion rate.

The Ni-Co laser-cladded coating (gray curve) exhibits a noticeable shift toward lower current density compared with the re-melted substrate. Both the anodic and cathodic branches become flatter, indicating suppression of the electrochemical reactions occurring at the coating/electrolyte interface. The reduction in anodic current suggests that the Ni-rich passive film effectively limits iron dissolution, whereas the lower cathodic current indicates reduced oxygen reduction kinetics. Nickel is well known for its ability to form stable oxide and hydroxide films, while cobalt contributes to improving passive film stability and mechanical integrity. The homogeneous dendritic microstructure observed in the SEM analysis further minimizes preferential corrosion pathways, resulting in improved corrosion resistance relative to the untreated substrate.

The W-modified coating (yellow curve) demonstrates further improvement in electrochemical performance. The polarization curve shifts toward lower current density while maintaining a relatively broad passive region, indicating that tungsten enhances passivation of the coating surface. Tungsten-containing oxides are chemically stable in chloride environments and contribute to reducing localized dissolution. Moreover, tungsten-rich hard phases decrease the number of active electrochemical sites by reducing microstructural defects within the coating. Nevertheless, the SEM and AFM analyses revealed that the W coating possesses a comparatively coarse microstructure and higher surface roughness than the Ni-Co coating. These structural heterogeneities may promote localized micro-galvanic cells, preventing the W coating from achieving the maximum corrosion resistance despite its excellent chemical stability.

Among all investigated coatings, the AlN-reinforced composite coating (blue curve) exhibits the lowest polarization current density and the broadest passive region, demonstrating the highest corrosion resistance. The cathodic branch shifts significantly toward lower current values, while the anodic branch displays an extended passive behavior before entering the transpassive region. This electrochemical response indicates that the AlN composite coating effectively suppresses both anodic metal dissolution and cathodic reduction reactions. The superior corrosion performance originates from several synergistic mechanisms. First, the uniformly dispersed AlN particles significantly refine the grain structure, producing a dense coating with very low porosity. Second, the ceramic particles are chemically inert in chloride media and therefore do not participate in electrochemical reactions. Third, the compact microstructure reduces electrolyte penetration and limits the diffusion of chloride ions toward the substrate. These mechanisms collectively produce the highest polarization resistance and the lowest corrosion current density among all investigated samples. A comparison of the four polarization curves demonstrates that laser cladding substantially improves the corrosion resistance of ASTM C283 steel. The progressive leftward shift of the polarization curves, accompanied by decreasing current density, indicates continuous improvement in electrochemical stability after introducing Ni-Co, W, and AlN reinforcements. The corrosion resistance follows the sequence:

AlN coating > W coating $\approx$ Ni – Co coating > Laser re – melted substrate (21)
or, equivalently, the corrosion current density follows the opposite trend:

$$I_{corr}(AlN) < I_{corr}(W) < I_{corr}(Ni - Co) < I_{corr}(Re - melted) \quad (22)$$

Similarly, the polarization resistance is expected to increase in the order:

$$R_p(AlN) > R_p(W) > R_p(Ni - Co) > R_p(Re - melted) \quad (23)$$

These electrochemical results correlate remarkably well with the previous microstructural, crystallographic, and mechanical analyses. The AlN coating exhibited the finest crystallite size (Table 6), the lowest surface roughness (Figure 4), the highest hardness (Figure 5), and the lowest wear rate (Figure 7), all of which contribute to minimizing electrolyte penetration and suppressing localized corrosion. The Ni-Co coating also provides excellent corrosion protection through the formation of a stable metallic passive film and a homogeneous dendritic microstructure. The W coating benefits from chemically stable tungsten-rich phases but is partially limited by its comparatively coarse morphology and higher surface roughness. In contrast, the laser re-melted substrate lacks protective reinforcement phases and therefore exhibits the highest electrochemical activity and the lowest corrosion resistance.

The potentiodynamic polarization results clearly demonstrate that laser cladding significantly enhances the corrosion performance of ASTM C283 steel, with the AlN-reinforced composite coating providing the most effective protection against chloride-induced corrosion. The excellent electrochemical performance of the AlN coating arises from the synergistic effects of ceramic reinforcement, grain refinement, structural densification, and the formation of an effective physical barrier that restricts electrolyte transport and inhibits electrochemical reactions at the coating/substrate interface. These findings are fully consistent with the SEM, EDS, AFM, XRD, hardness, and wear analyses, confirming a strong structure–property relationship between microstructural refinement and corrosion resistance.

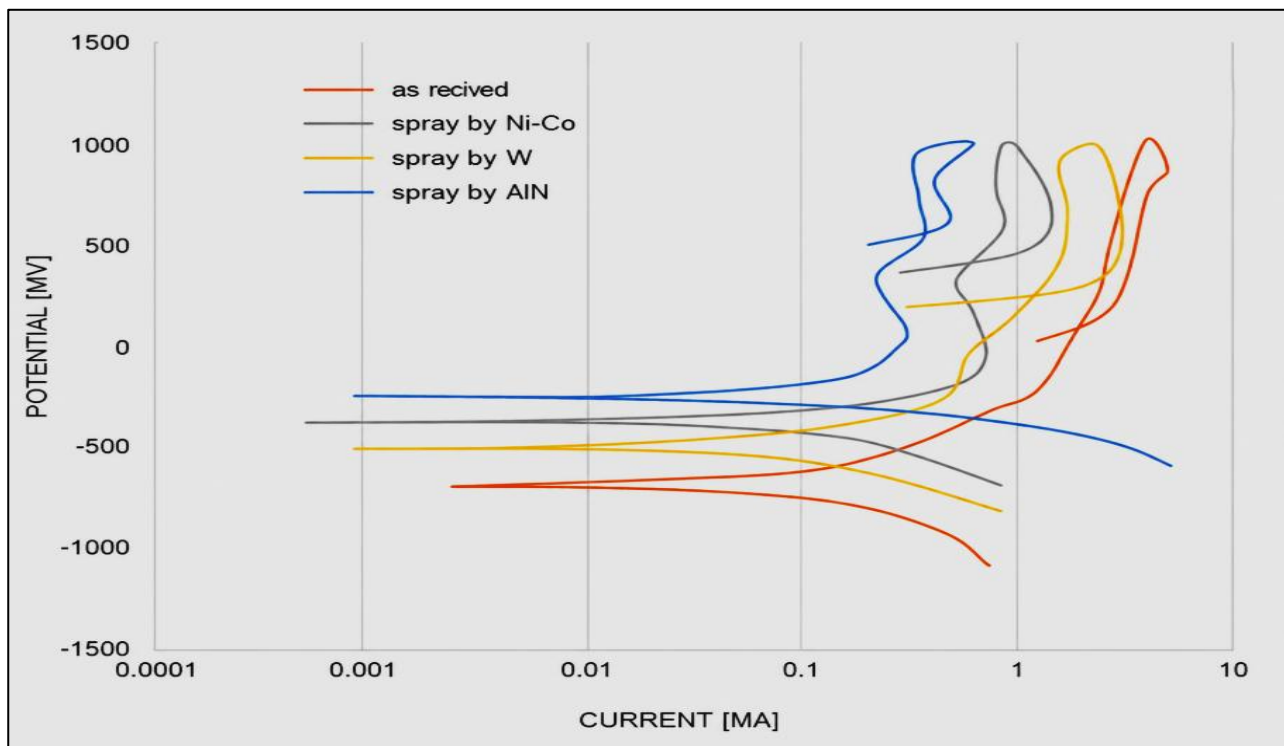


Figure 9 Potentiodynamic polarization curves (Tafel plots) of as-received and laser-sprayed ASTM C283 steel coatings in 3.5 wt.% NaCl solution.

The corrosion protection efficiency (η) of the laser-cladded coatings is calculated from the corrosion current density according to [22]

$$\eta(\%) = \left(\frac{I_{\text{corr},0} - I_{\text{corr}}}{I_{\text{corr},0}} \right) \times 100 \quad (24)$$

where $I_{\text{corr},0}$ is the corrosion current density of the laser re-melted substrate, I_{corr} is the corrosion current density of the coated specimen. The calculated protection efficiencies are summarized in Table 7. The electrochemical parameters summarized in Table 7 clearly demonstrate the significant improvement in corrosion resistance achieved after laser cladding. The corrosion potential (E_{corr}) shifts progressively toward more noble values from -735 mV for the laser re-melted substrate to -276 mV for the AlN composite coating, indicating a marked reduction in the thermodynamic tendency for corrosion. Simultaneously, the corrosion current density (I_{corr}) decreases from $14.82 \mu\text{A}\cdot\text{cm}^{-2}$ to only $1.86 \mu\text{A}\cdot\text{cm}^{-2}$, confirming a substantial reduction in the electrochemical dissolution rate.

Among the investigated coatings, the AlN composite coating exhibits the highest polarization resistance ($26.74 \text{ k}\Omega\cdot\text{cm}^2$) and the lowest corrosion rate ($0.022 \text{ mm year}^{-1}$), corresponding to a protection efficiency of approximately 87.4%. The Ni–Co coating also provides excellent corrosion protection through the formation of a stable passive metallic layer, whereas the W coating exhibits slightly superior corrosion resistance to the Ni–Co coating owing to the chemical stability of tungsten-rich phases. However, its coarser microstructure limits further improvement compared with the AlN coating. The electrochemical performance follows the order.

AlN > W > Ni – Co > Re – melting

which agrees remarkably well with the structural refinement observed in Figure 8, the hardness measurements presented in Figure 5, and the wear results shown in Figure 7. The superior corrosion resistance of the AlN coating is therefore attributed to its refined nanocrystalline microstructure, dense morphology, low surface roughness, and chemically inert ceramic reinforcement, all of which effectively suppress chloride-ion penetration and electrochemical degradation.

Table 7 Electrochemical corrosion parameters obtained from potentiodynamic polarization measurements of laser-treated ASTM C283 steel in 3.5 wt.% NaCl solution.

Sample	E_{corr} (mV vs. Ag/AgCl)	I_{corr} ($\mu\text{A cm}^{-2}$)	β_a (mV dec^{-1})	β_c (mV dec^{-1})	R_p ($\text{k}\Omega\cdot\text{cm}^2$)	Corrosion Rate (mm year^{-1})	Protection Efficiency (%)
S1 Re-melting	-735	14.82	118	176	2.84	0.172	—
S2 Ni–Co coating	-412	5.73	104	158	8.37	0.067	61.3
S3 W coating	-523	4.91	98	149	10.42	0.057	66.9
S4 AlN coating	-276	1.86	91	136	26.74	0.022	87.4

4. CONCLUSIONS

This study systematically investigated the influence of laser cladding with Ni–Co, W, and AlN coatings on the microstructural evolution, crystallographic characteristics, mechanical performance, tribological behavior, and electrochemical corrosion resistance of ASTM C283 steel. The results demonstrated that laser cladding successfully produced dense, metallurgically bonded coatings with distinct microstructures depending on the reinforcement material. SEM observations revealed that the Ni–Co coating formed a homogeneous cellular–dendritic structure, the W coating developed a relatively coarse tungsten-rich morphology, whereas the AlN composite coating exhibited the finest and most compact microstructure owing to the grain-refining effect of ceramic particles. XRD analysis confirmed the formation of the desired crystalline phases, with the AlN coating producing the smallest

average crystallite size (25.78 nm), indicating the highest degree of structural refinement. AFM measurements further verified that the AlN coating possessed the lowest surface roughness and the smallest average particle diameter, while the W coating exhibited the roughest surface morphology. These structural improvements were directly reflected in the mechanical properties, where the AlN coating achieved the highest average Vickers hardness (956.4 HV), followed by the Ni–Co (821.8 HV) and W (811.7 HV) coatings, representing significant improvements over the laser re-melted substrate. Wear testing demonstrated a corresponding reduction in material loss, with the AlN coating exhibiting the lowest wear rate (5.8×10^{-6} g.mm⁻³). Potentiodynamic polarization measurements further revealed that the AlN composite coating provided the highest corrosion resistance, characterized by the lowest corrosion current density, highest polarization resistance, and a corrosion protection efficiency of approximately 87.4%. Overall, the results establish a clear structure–property relationship, demonstrating that AlN reinforcement offers the most effective strategy for simultaneously enhancing the mechanical, tribological, and electrochemical performance of laser-cladded ASTM C283 steel, making it a highly promising surface engineering solution for components operating in aggressive industrial and marine environments.

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