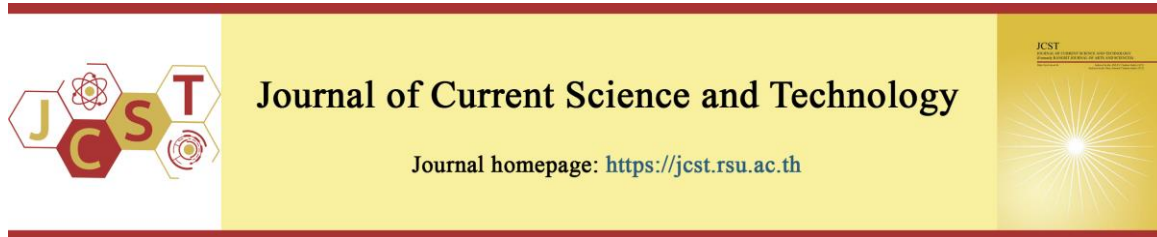


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Enhancing Wear Resistance and Interfacial Properties of Hardfaced AISI 1050 Steel Utilizing an Austenitic Buffer Layer

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Abstract

The performance of hardfaced carbon steels is strongly influenced by interfacial integrity and microstructural evolution during welding. This study investigated the effects of an austenitic E309L-16 buffer layer on the microstructure, hardness distribution, and abrasive wear resistance of SMAW-hardfaced AISI 1050 steel. A functionally graded hardfacing architecture incorporating the buffer layer was developed to improve metallurgical compatibility and interfacial stability. Two conditions were examined: direct hardfacing (S1) and buffer-assisted hardfacing (B1). Optical microscopy (OM), scanning electron microscopy coupled with energy-dispersive X-ray spectroscopy (SEM-EDS), microhardness measurements, and ASTM G65 abrasion testing were conducted. The results showed that direct hardfacing promoted carbon redistribution toward the interface, resulting in a brittle interfacial region and an abrupt hardness transition. In contrast, the E309L-16 buffer layer moderated carbon redistribution, eliminated the interfacial white band, and promoted a more homogeneous microstructure with a gradual hardness profile. The B1 specimen predominantly exhibited lath martensite and demonstrated superior wear resistance despite its slightly lower peak hardness (586 HV compared with 647 HV for S1). The specific wear rate decreased from $(10.08 \pm 0.43) \times 10^{-4} \text{ mm}^3/\text{N}\cdot\text{m}$ for the base metal to $(4.30 \pm 0.10) \times 10^{-4} \text{ mm}^3/\text{N}\cdot\text{m}$ for S1 and further to $(3.62 \pm 0.11) \times 10^{-4} \text{ mm}^3/\text{N}\cdot\text{m}$ for B1. These findings demonstrate that interfacial engineering using an austenitic buffer layer can improve wear resistance and provide practical guidance for designing durable multilayer hardfacing systems.

Keywords: abrasive wear, aisi 1050, austenitic buffer, carbon diffusion, hardfacing, microstructure, smaw

1. Introduction

Medium-carbon steel, specifically AISI 1050 (S50C), is extensively used in the manufacture of machinery components such as shafts, gears, and structural parts that require high load-bearing capacity and toughness. AISI 1050 contains a relatively high carbon content of approximately 0.47–0.55%, which provides a typical ultimate tensile strength (UTS) of 630–690 MPa and a yield strength of 355–400 MPa (Equbal et al., 2016; Hasanah et al., 2025). Despite its

excellent bulk mechanical properties and high hardenability, the surface hardness of AISI 1050 in the conventional normalized state is relatively limited, typically ranging from 190 to 220 HV (ASM Handbook Committee, 1990; Hasanah et al., 2025). This limitation renders the material susceptible to abrasive and adhesive wear. Consequently, during operation, these components are frequently subjected to continuous friction and impact, leading to severe surface deterioration and reduced service life. Surface

modification to enhance wear resistance is therefore essential for prolonging service life and minimizing maintenance costs (Ning et al., 2025).

Various advanced surface-coating technologies are currently available to address wear-related failures. High-Velocity Oxygen Fuel (HVOF) thermal spraying is widely recognized for producing dense coatings with excellent wear resistance and exceptionally low heat input to the substrate. In this process, semi-molten powder particles are accelerated to supersonic velocities and impinge on the substrate, thereby minimizing thermal distortion and metallurgical degradation of the base metal (Anusha et al., 2023). However, the primary limitation of HVOF lies in its bonding mechanism. The coating relies predominantly on mechanical interlocking rather than fusion, resulting in relatively low interfacial bond strength. Consequently, the coating layer is highly susceptible to macrocracking, spalling, or catastrophic delamination when subjected to heavy-duty loads, severe impact forces, or localized high contact stresses (Testa et al., 2020; Ning et al., 2025). Conversely, laser cladding uses a high-energy-density laser beam to simultaneously melt the feedstock material and a thin surface layer of the substrate. This technique offers significant metallurgical advantages, including the formation of a true metallurgical bond, an extremely narrow heat-affected zone (HAZ), fine microstructures resulting from rapid solidification, and a low dilution rate (Chaichawalit et al., 2025). Despite these superior microstructural characteristics, the industrial application of laser cladding is severely constrained by several practical and economic factors. The process requires substantial initial capital investment in high-power laser equipment and necessitates complex, rigid, automated multi-axis robotic systems. Furthermore, its relatively low deposition rate, extreme sensitivity to substrate surface preparation, and lack of portability make it highly inflexible and economically unviable for the large-scale or in situ repair of heavy industrial components (Dubourg et al., 2005).

Consequently, conventional shielded metal arc welding (SMAW) hardfacing remains one of the most widely adopted surface-engineering approaches in industrial maintenance because of its cost-effectiveness, high deposition efficiency, operational flexibility, and ability to produce a robust metallurgical bond capable of withstanding severe impact and abrasive conditions. In addition to these advantages, SMAW offers practical benefits, including simple equipment requirements, excellent portability, ease of deposition under field conditions, and suitability for the in situ

repair of large-scale or immovable industrial components. These characteristics make SMAW hardfacing particularly attractive for maintenance and remanufacturing applications in which minimizing production downtime and operational costs is essential. Nevertheless, the major metallurgical challenge associated with direct hardfacing remains the control of heat input and dilution. When high-hardness electrodes conforming to DIN 8555: E 3-UM-600 are directly deposited onto AISI 1050 steel, the relatively high carbon content of the substrate can promote carbon redistribution, severe thermal gradients, brittle martensitic transformation within the heat-affected zone, and increased susceptibility to interfacial cracking (Srisuwan et al., 2019; Tippayasam et al., 2023).

To overcome these limitations, the use of an austenitic buffer layer has been proposed as an effective metallurgical strategy. Austenitic stainless steel E309L-16 is widely recognized for its high ductility and ability to accommodate residual stresses while limiting elemental redistribution across dissimilar interfaces. However, to the best of the authors' knowledge, few studies have investigated the integration of an E309L-16 buffer layer with DIN 8555: E 3-UM-600 hardfacing specifically on AISI 1050 steel, particularly with respect to interfacial microstructural evolution, compositional transition, and wear performance. Therefore, the novelty of this study lies in the strategic implementation of an E309L-16 austenitic buffer layer to engineer a functionally graded hardfacing architecture through the SMAW process. Unlike conventional direct hardfacing, the buffer layer was intentionally employed as a metallurgical and compositional transition layer to moderate carbon redistribution, reduce the tendency for brittle interfacial region formation, and facilitate the development of a lath martensitic microstructure in the top hardfacing layer. This engineered approach was expected to improve interfacial compatibility, establish a more gradual hardness transition, and enhance abrasive wear resistance.

2. Objectives

Following the identification of the existing research gap, the primary objective of this study was to systematically investigate the influence of an E309L-16 austenitic stainless steel buffer layer on the microstructural and mechanical behavior of hardfaced AISI 1050 steel. Specifically, this study aimed to:

- 1) Evaluate the effectiveness of the E309L-16 buffer layer in mitigating macroscopic surface defects

and controlling carbon migration across the fusion boundaries.

2) Quantitatively compare the interfacial microstructural evolution and elemental diffusion profiles of directly hardfaced specimens (S1) and buffer-assisted specimens (B1).

3) Assess the mechanical response by correlating the microhardness distribution gradients with the resulting resistance to three-body abrasive wear.

3. Materials and Methods

To provide a comprehensive overview of the experimental program, Figure 1 presents a schematic flowchart illustrating the complete workflow, from input materials to performance evaluation. The subsequent sections describe the materials, welding procedures, and characterization techniques employed in this study.

3.1 Materials and Equipment

The base material used in this study was AISI 1050 medium-carbon steel purchased from N.C. Rungrueng Machine Trade Co., Ltd. (Thailand). The material was machined into rectangular specimens measuring $38 \times 100 \times 6$ mm. Prior to welding, the substrate surfaces were mechanically ground to remove oxide scale and subsequently cleaned with acetone to eliminate surface contaminants.

Two types of commercial welding electrodes, both 3.2 mm in diameter, were selected for the coating process. A lime–titania-type austenitic stainless steel electrode, classified according to AWS A5.4 as E309L-16 and DIN 8556 as E 23 12 L 26, was used for the buffer layer. For the top hardfacing layer, a martensitic hardfacing electrode conforming to DIN 8555: E 3-UM-600 was employed to provide exceptional abrasion resistance under medium-impact conditions. The nominal chemical composition of the base metal, together with the typical chemical compositions and mechanical properties of the deposited weld metals, is summarized in Table 1.

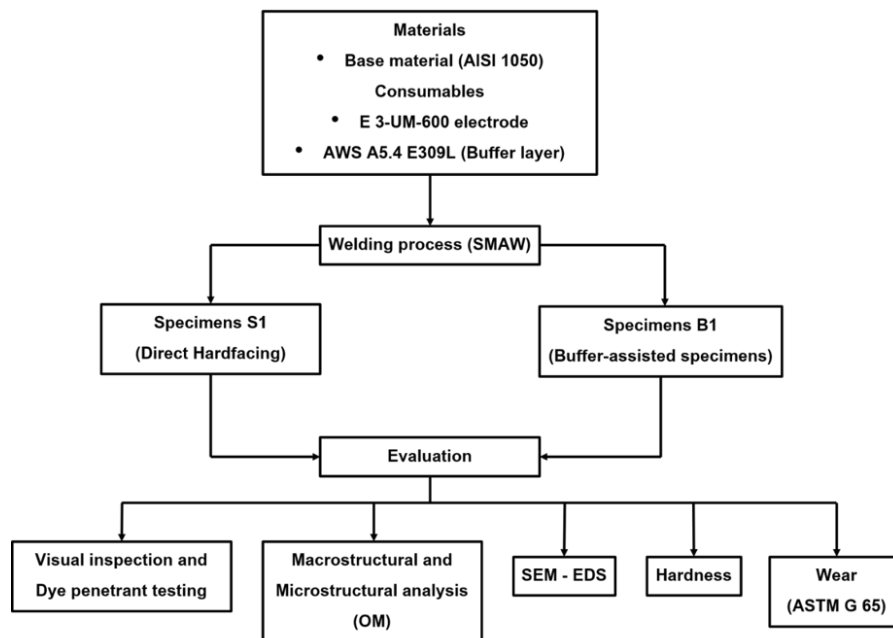


Figure 1 Schematic overview of the experimental workflow from material preparation to characterization and performance evaluation

Table 1 Chemical compositions (wt.%) and standard hardness of the base metal and deposited weld metals

Material	Classification standard	C	Si	Mn	Cr	Ni	Mo	Hardness (HV)
Base metal	AISI 1050	0.5	0.2	0.7	-	-	-	210
Buffer layer	AWS A5.4 E309L-16	0.03	0.85	1.3	24	12.4	0.12	200
Hardfacing layer	DIN 8555: E 3-UM-600	0.4	0.37	1.15	4.65	-	0.53	624

For metallographic preparation, analytical-grade nitric acid and ethanol were used to prepare the Nital etching solution. For the low-stress abrasive wear tests, AFS 50/70 rounded quartz silica sand was used as the abrasive medium. The hardfacing process was performed using a SUMO 200 A inverter welding machine. During the multipass welding process, the welding current and arc voltage were continuously monitored using a UNI-T UT200 digital clamp meter to ensure stable welding parameters and accurate heat-input estimation. For subsequent characterization of the welded specimens, macrostructural and microstructural features were examined using optical microscopy (OM), while elemental compositions and distributions were analyzed using a scanning electron microscope equipped with energy-dispersive X-ray spectroscopy (SEM-EDS). In addition, mechanical integrity was evaluated using a Vickers microhardness tester, and abrasive wear resistance was assessed using a dry sand/rubber wheel abrasion tester in strict accordance with ASTM G65.

3.2 Welding Procedure

Prior to hardfacing, the welding consumables were redried in an electrode oven according to the manufacturer's recommendations to minimize the risk of hydrogen-induced cold cracking and porosity. The E309L-16 electrodes were baked at 250°C for 60 minutes, whereas the E 3-UM-600 hardfacing electrodes were baked at 350°C for 60 minutes. Deposition was performed using a SUMO 200 A inverter welding machine through the manual shielded metal arc welding (SMAW) process with direct current electrode negative (DCEN) polarity. Welding was initiated at ambient temperature without preheating the base metal.

For the buffer-assisted specimens (B1), a buffer layer consisting of five overlapping parallel beads deposited using E309L-16 electrodes was first applied to cover the substrate surface at a nominal welding current of 85 A. Subsequently, a top hardfacing layer, also consisting of five overlapping parallel beads deposited using the E 3-UM-600 electrode, was applied directly over the entire buffer layer at a nominal welding current of 100 A. For the directly hardfaced specimens (S1), an identical

hardfacing layer comprising five overlapping parallel beads deposited using the E 3-UM-600 electrode was applied directly to the AISI 1050 substrate under the same conditions.

During the multipass welding operations, the real-time welding current and arc voltage were continuously monitored and recorded using a UNI-T UT200 digital clamp multimeter to ensure the stability and consistency of the applied heat input. The corresponding welding heat input (Q) for each pass was calculated using the standard empirical equation (Kou, 2021).

$$Q = \frac{\eta \times V \times I}{S \times 1000} \quad (1)$$

Where η is the arc thermal efficiency (0.8 for the SMAW process), V is the arc voltage (V), I is the arc current (A), and S is the travel speed (mm/s).

Because both the buffer and hardfacing layers were deposited using a multibead technique, the interpass temperature between consecutive beads in all layers was strictly monitored and maintained at approximately 150°C. This precise thermal control was implemented to mitigate the high hardenability of the medium-carbon steel substrate and prevent cold cracking. Furthermore, the slag was thoroughly removed using a chipping hammer and wire brush between consecutive passes to prevent slag inclusions. The primary welding parameters are summarized in Table 2.

3.3 Characterizations

To preserve the true metallurgical characteristics, transverse cross-sectional specimens were precisely extracted from the welded plates using a water-cooled abrasive cut-off machine. The extracted specimens were subsequently mounted in epoxy resin, sequentially ground using silicon carbide (SiC) abrasive papers ranging from 240 to 2000 grit, and finally polished with a 1 μ m diamond suspension to achieve a scratch-free, mirror-like finish. To reveal the microstructural features, the polished surfaces were chemically etched at room temperature using a freshly prepared 2% Nital solution.

Table 2 Welding parameters

Specimen	Deposited weld layer	Interpass temperature (°C)	Current (A)	Voltage (V)	Travel speed (mm/s)	Heat input (kJ/mm)
S1	Hardfacing	150	100	25	1.8	1.11
B1	Buffer	150	85	23	2.5	0.63
	Hardfacing	150	100	25	1.8	1.11

The macrostructure and microstructure of the weld cross-sections were initially examined using an optical microscope (OM) to assess the weld profile, microstructural evolution, and integrity of the fusion boundaries among the AISI 1050 base metal, heat-affected zone (HAZ), buffer layer, and hardfacing layer. To further investigate high-resolution morphological features, a scanning electron microscope (SEM) was used. In addition, energy-dispersive X-ray spectroscopy (SEM-EDS) was employed to determine the elemental compositions and analyze the distributions of key alloying elements, such as Cr, Ni, and Mn, across the distinct weld interfaces.

3.4 Hardness and Wear Resistance

To evaluate the mechanical integrity and hardening effect of the multipass welding process, the microhardness distribution was continuously measured from the unaffected base metal, across the HAZ and fusion lines, to the top surface of the hardfacing layer. The measurements were conducted using a Vickers microhardness tester in strict accordance with ASTM E384 (ASTM International, 2022). A constant test load of 200 gf (HV0.2) and a dwell time of 15 seconds were applied for each indentation. To ensure statistical reliability, hardness measurements were performed along three parallel lines across the cross-sections of three independent specimens for each condition. The average values and standard deviations were then calculated to construct the microhardness profiles.

Furthermore, to simulate severe industrial operating conditions, the low-stress three-body abrasive wear resistance of the directly hardfaced (S1) and buffer-assisted (B1) specimens was evaluated using a dry sand/rubber wheel abrasion tester. Testing was conducted in accordance with a modified ASTM G65 Procedure B (ASTM International, 2017) to systematically assess wear progression. Prior to testing, the top hardfaced surfaces were ground to ensure uniform contact. Rounded quartz silica sand (AFS 50/70) was used as the abrasive medium and continuously fed into the contact zone between the specimen and a rotating chlorobutyl rubber wheel at a controlled flow rate of approximately 300–400 g/min. The tests were performed under a constant applied normal load of 130 N and a rubber wheel rotational speed of 200 rpm.

To accurately monitor the wear rate, the total test duration of 600 s was divided into three distinct intervals of 200 s each. The specimens were periodically removed, thoroughly cleaned, and weighed before and

after each 200-second interval using an analytical balance with a resolution of 0.1 mg to construct cumulative mass-loss curves. Finally, to ensure statistical reliability and reproducibility, three independent specimens were tested for each experimental condition, with data variations quantified and represented using standard deviations. To facilitate standardized evaluation and direct comparison with contemporary literature, the cumulative mass loss (Δm) obtained from the abrasive wear tests was subsequently converted into the specific wear rate (K) using the standard Archard wear equation (Holmberg & Matthews, 2009).

$$K = \frac{V}{WS} = \frac{\Delta m}{\rho WS} \quad (2)$$

Where V is the total volume loss (mm^3), calculated by dividing the mass loss (Δm) by the nominal density of the steel matrix ($\rho = 7.85 \times 10^{-3} \text{ g/mm}^3$); w is the applied normal load (130 N); and S is the total sliding distance (m), determined from the wheel diameter, rotational speed, and test duration. Furthermore, following completion of the wear tests, the worn surfaces of representative specimens were systematically examined using scanning electron microscopy (SEM) equipped with energy-dispersive X-ray spectroscopy (EDS) to elucidate the dominant abrasive wear mechanisms, localized deformation behavior, and elemental integrity of the hardfacing layers under the applied loads.

4. Results

4.1 Hardfacing and Surface Inspection

Figure 2 illustrates the macroscopic surface morphology of the multipass hardfaced specimens deposited using the shielded metal arc welding (SMAW) process. Figure 2(a) shows the directly hardfaced specimen without a buffer layer (S1), whereas Figure 2(b) shows the buffer-assisted specimen (B1). Visual inspection of both specimens revealed relatively uniform bead profiles with consistent overlap and regular ripple patterns. No macroscopic surface defects, such as severe porosity, macrocracks, slag inclusions, or undercut, were observed on the top surface of either specimen. This indicated that the selected welding parameters, including the welding current, arc voltage, and strictly controlled interpass temperature of 150°C , were appropriate for achieving a stable arc and sound multibead deposition.

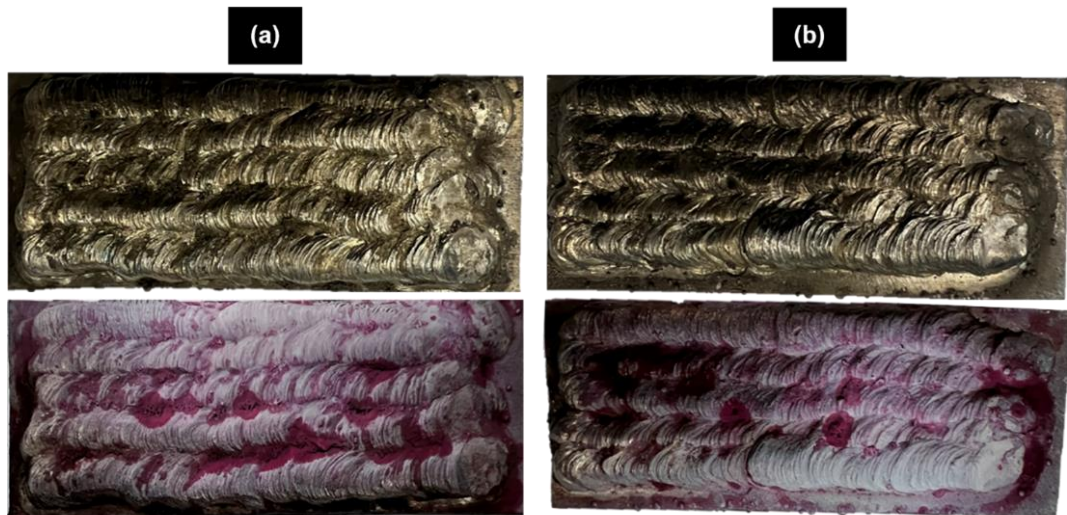


Figure 2 Visual inspection and penetrant testing: (a) S1 (without a buffer layer) and (b) B1 (with a buffer layer)

To further assess surface integrity, dye penetrant testing (PT) was performed. As shown in the lower panels of Figure 2, localized red penetrant indications were observed, particularly on specimen S1. However, detailed examination suggested that these indications were predominantly aligned along the longitudinal valleys and weld toes between overlapping beads. They were therefore interpreted as nonrelevant or spurious indications resulting from the mechanical entrapment of penetrant within the inherent surface irregularities produced by the manual multipass SMAW process, rather than as actual surface-breaking metallurgical defects such as cracks or porosity (Hellier, 2013). Similarly, no clear defect indications were observed on specimen B1.

Although no genuine linear crack indications were detected on the examined surfaces, PT is inherently limited to detecting surface-opening discontinuities and cannot conclusively exclude the presence of subsurface or microscale cracking. Therefore, these observations should be interpreted as evidence of satisfactory surface integrity rather than definitive proof of crack-free conditions. For further verification, subsequent macrostructural and microstructural cross-sectional examinations, described in Sections 4.2–4.4, were conducted and revealed no visible crack propagation across the examined interfaces.

Overall, these results suggest that, although no preheating was applied, precise thermal control through the maintenance of an interpass temperature of 150°C, combined with stringent low-hydrogen practices such as prior electrode baking, was

sufficient to prevent hydrogen-induced cold cracking and maintain acceptable weld quality in both specimens under the evaluated conditions (Kou, 2021; Lippold, 2014).

4.2 Macrostructural Characterization

To further investigate the weld profiles, transverse cross-sections of the specimens were extracted. Figure 3 presents the macrostructures and corresponding quantitative geometric analyses of the deposited layers. Figures 3(a) and 3(b) show the macroscopic cross-sections of the directly hardfaced specimen (S1) and the buffer-assisted specimen (B1), respectively. At this magnification, neither specimen exhibited evidence of major internal defects, such as lack of fusion or severe macrocracking.

The cross-section of specimen B1 revealed four distinct metallurgical zones: the AISI 1050 base metal, the heat-affected zone (HAZ), the E309L-16 buffer layer, and the top E 3-UM-600 hardfacing layer. All interfaces exhibited continuous and coherent fusion boundaries. The austenitic stainless steel buffer layer effectively covered the entire base metal surface, serving as a transitional barrier to accommodate thermal stresses and mitigate carbon diffusion from the substrate (Lippold, 2014; Mvola et al., 2014).

Quantitative geometric analysis of the multipass hardfacing profiles was conducted using image analysis, and the results are summarized in Figures 3(c) and 3(d). As shown in Figure 3(c), the average total width of the overlapping passes for B1 (35.48 mm) was slightly greater than that for S1

(32.14 mm). This broader profile was attributed to the superior wetting behavior of the austenitic E309L-16 buffer layer on the AISI 1050 substrate during initial deposition. Furthermore, the average reinforcement height (convexity) of B1 (4.14 mm) was notably greater than that of S1 (3.06 mm), directly resulting from the intermediate buffer layer, which contributed to a greater total deposited volume.

The average penetration depths into the AISI 1050 substrate were shallow and comparable for both specimens, measuring 1.14 mm for S1 and 1.30 mm for B1. Such shallow penetration is desirable in hardfacing applications because it minimizes excessive melting of the base metal and limits dilution. To quantitatively assess this effect, geometric dilution was evaluated, as shown in Figure 3(d). The calculated dilution for S1 was 25.01%, indicating a

controlled level that ensured adequate metallurgical bonding while minimizing incorporation of the base metal. In contrast, B1 exhibited a higher dilution of 31.31%. This increase was attributed not only to the initial buffer pass, which established the primary metallurgical bond with the substrate, but also to the cumulative thermal effects associated with multipass deposition. The additional buffer layer introduced extra heat input, followed by reheating during deposition of the top hardfacing layer. Together, these effects enlarged the effective fusion zone and promoted greater interfacial mixing. However, because this dilution was confined to the intermediate E309L-16 buffer layer rather than the top hardfacing layer, it enabled metallurgical bonding while preserving the high-alloy integrity and wear-resistant properties of the final working surface.

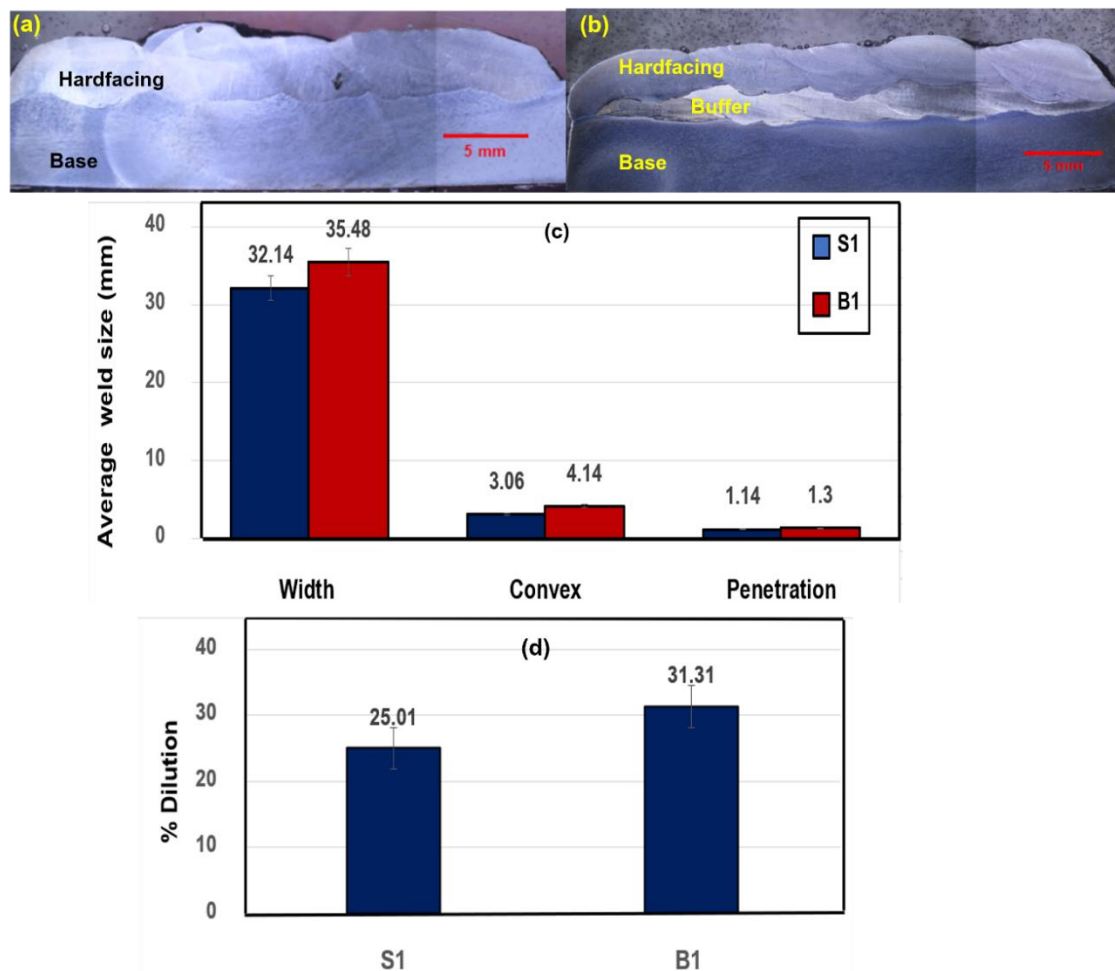


Figure 3 Macroscopic results: (a) S1 (without a buffer layer), (b) B1 (with a buffer layer), (c) weld dimensions, and (d) specimen dilution

4.3 Microstructural Characterization

The microstructural characteristics of the deposited layers, heat-affected zones (HAZs), and fusion interfaces of both specimens were systematically examined using optical microscopy (OM). Figure 4 presents the cross-sectional microstructure of the directly hardfaced specimen (S1), whereas Figure 5 presents that of the buffer-assisted specimen (B1).

As shown in Figure 4(a) at 400 μm , the low-magnification cross-section of S1 revealed three distinct metallurgical zones: the AISI 1050 base metal, the HAZ, and the E 3-UM-600 hardfacing layer, separated by a curved fusion boundary. Direct deposition of the highly alloyed hardfacing electrode onto the AISI 1050 substrate produced a steep thermal gradient at the interface, despite maintaining a controlled interpass temperature of 150°C. This severe thermal condition led to the formation of a distinct, unetched white band along the fusion boundary. Although OM does not provide compositional information, such etch-resistant features are commonly associated with either an unmixed zone or a carbon-enriched interfacial region resulting from localized elemental redistribution between the substrate and the weld pool. This region likely corresponded to an untempered high-carbon martensitic layer or iron carbide precipitates, constituting a structurally weak zone susceptible to residual stress concentration and crack initiation (Lippold, 2014; Gupta et al., 2024). Detailed compositional verification using SEM-EDS is presented in Section 4.4.

Immediately beneath the fusion boundary, the HAZ exhibited a coarser microstructure than the base metal. This was attributed to the elevated peak temperatures near the weld pool, where rapid austenite grain growth occurred at temperatures exceeding the A_{c3} transformation line. Subsequent rapid cooling imposed by the surrounding material promoted partial martensitic transformation within the coarse-grained HAZ (Makhdoom et al., 2022). Although this transformation increased local hardness, it significantly reduced interfacial toughness.

At intermediate magnification, as shown in Figure 4(b) at 200 μm , the region immediately above the fusion boundary exhibited an equiaxed dendritic morphology. This structure formed under conditions of constitutional supercooling ahead of the solid-liquid interface. A locally low thermal-gradient-to-growth-rate ratio (G/R) promoted simultaneous nucleation at multiple sites and suppressed directional columnar growth. This feature was identified as a columnar-to-equiaxed transition (CET), driven by solute redistribution resulting from the steep compositional gradient between the substrate and the weld pool (Kou, 2021).

Further into the bulk of the hardfacing layer, the solidification structure transitioned into a fully martensitic matrix. At higher magnification (50 μm), the microstructure predominantly consisted of fine acicular martensite. This morphology resulted from a diffusionless shear transformation under conditions of high undercooling. The elevated carbon content (≈ 0.40 wt.%) and the presence of alloying elements such as Cr and Mo suppressed diffusion-controlled transformations and lowered the martensite start temperature (M_s) (Krauss, 2015), thereby promoting the formation of a high-dislocation-density martensitic structure responsible for the high hardness and abrasion resistance of the hardfacing layer.

Figure 4(c) further examines the fusion-boundary region of S1. The intermediate-magnification image (200 μm) showed columnar solidification structures growing perpendicular to the fusion boundary and aligned with the dominant direction of heat extraction toward the substrate. This abrupt transition into the underlying coarsened HAZ highlighted the absence of an intermediate layer. At higher magnification (50 μm), the HAZ predominantly exhibited untempered martensite within coarse prior-austenite grains. The coexistence of this brittle martensitic HAZ and the carbon-enriched interfacial white band created a mechanically vulnerable region. The combined effects of residual stresses, thermal expansion mismatch, and microstructural brittleness increased susceptibility to cold cracking and interfacial failure.

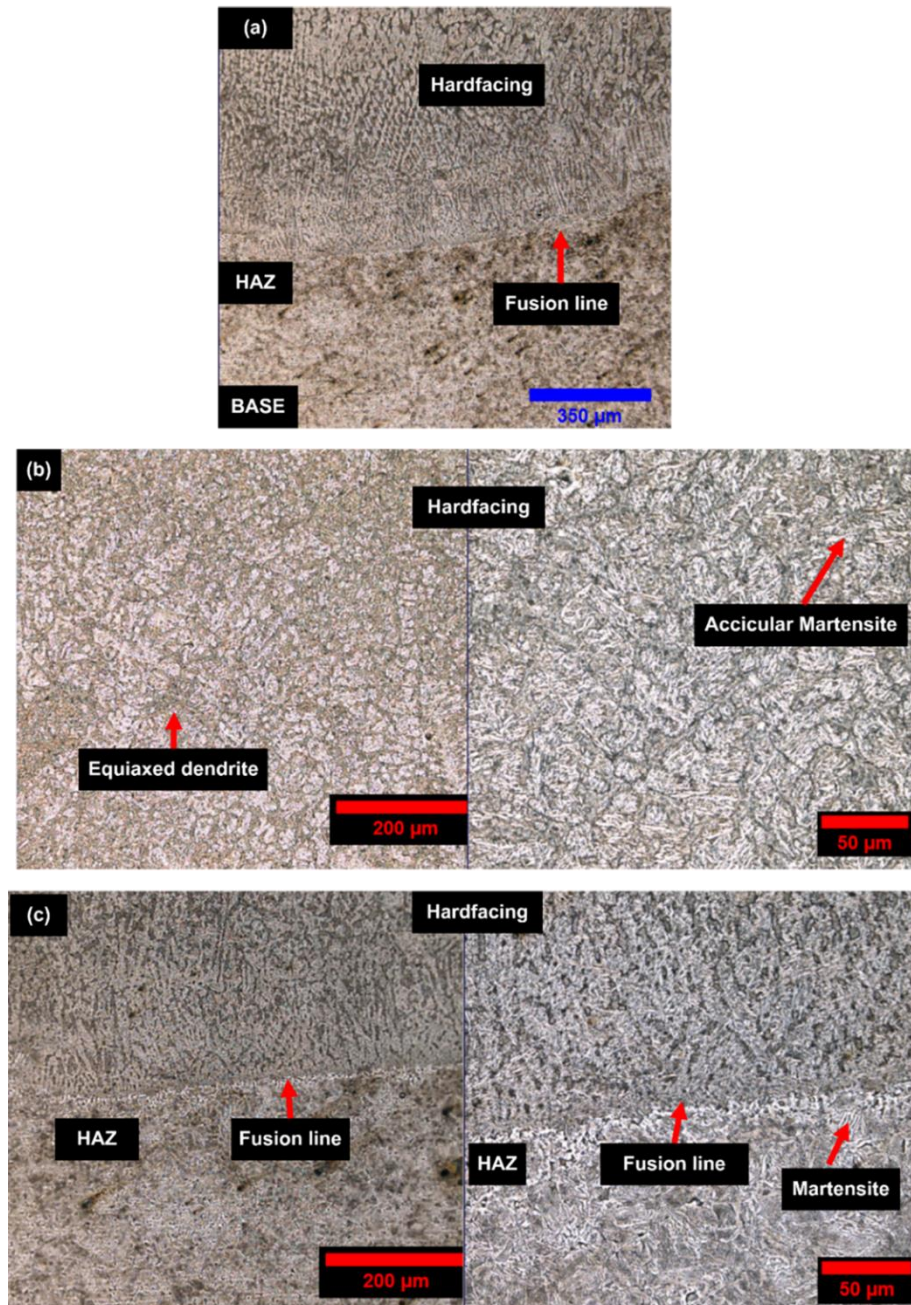


Figure 4 Optical micrographs of the directly hardfaced specimen (S1). (a) Overall view (400 μm) showing base metal, HAZ, and hardfacing layer with a white band at the fusion line. (b) Higher magnifications (200 μm and 50 μm) showing equiaxed dendrites near the fusion line and acicular martensite in the hardfacing layer. (c) Detailed images (200 μm and 50 μm) of the fusion boundary and coarsened HAZ

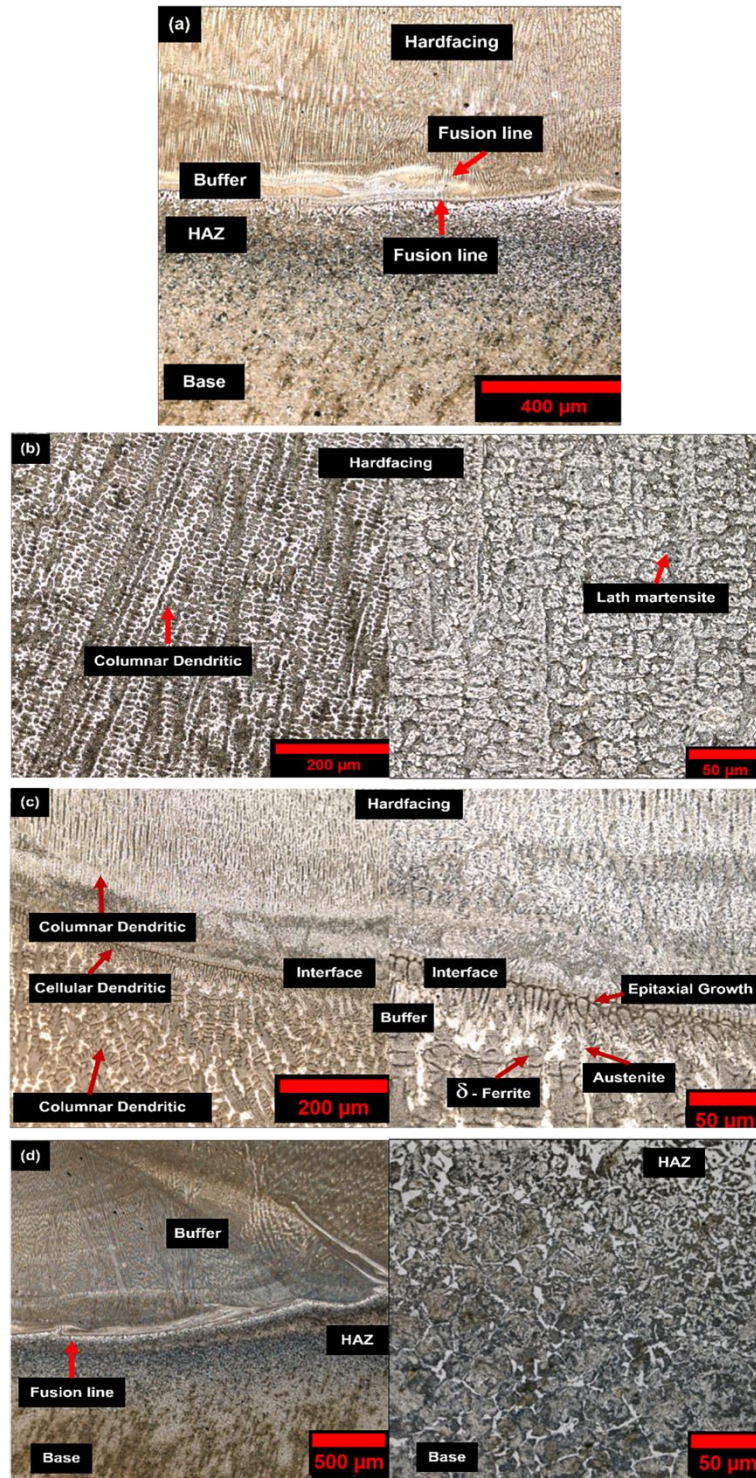


Figure 5 Optical micrographs of the buffer-assisted specimen (B1): (a) overall view (350 μm) showing the base metal, HAZ, E309L-16 buffer layer, and hardfacing layer, with no white band observed at the fusion boundaries; (b) higher-magnification images (200 μm and 50 μm) showing columnar dendritic austenite in the buffer layer and lath martensite in the hardfacing layer; (c) detailed images (200 μm and 50 μm) of the secondary interface showing epitaxial solidification; and (d) primary fusion boundary at 500 μm and 50 μm

In contrast, Figure 5(a) at 350 μm revealed a markedly different microstructural architecture in B1. Five distinct zones were identified: the AISI 1050 base metal, the HAZ, the E309L-16 austenitic buffer layer, a secondary fusion interface, and the E 3-UM-600 hardfacing layer. Two continuous fusion boundaries were clearly observed: one at the base metal/buffer interface and the other at the buffer/hardfacing interface. This configuration transformed the system from an abrupt binary interface into a functionally graded multilayer structure. At low magnification, the HAZ in B1 appeared less extensive and less severely coarsened than that in S1. This observation was consistent with the lower thermal impact of the initial buffer pass ($Q = 0.63 \text{ kJ/mm}$) compared with direct hardfacing ($Q = 1.11 \text{ kJ/mm}$). Furthermore, the primary fusion boundary between the substrate and the buffer layer appeared fine and continuous, without the pronounced white band observed in S1, indicating a more gradual compositional transition.

At intermediate magnification, as shown in Figure 5(b) at 200 μm , the E309L-16 buffer layer exhibited a columnar dendritic morphology. The dendrites grew perpendicular to the fusion boundary, consistent with epitaxial solidification under a high G/R ratio. This condition suppressed constitutional supercooling and promoted directional solidification along preferred crystallographic orientations. At higher magnification (50 μm), the hardfacing layer in B1 exhibited a lath martensitic morphology, in contrast to the acicular martensite observed in S1. The lath martensite was organized into packets within prior-austenite grains and contained a high density of mobile dislocations along the lath boundaries rather than the internal twins typical of plate martensite. This morphology provided improved toughness and resistance to crack propagation (Krauss, 2015; Balos et al., 2025). The transition from acicular to lath martensite was consistent with the compositional modification introduced by the buffer layer. As confirmed by EDS analysis in Section 4.4, the E309L-16 layer suppressed carbon migration from the substrate into the hardfacing layer. This reduced carbon enrichment, increased the effective M_s temperature, and promoted the formation of the tougher lath martensitic structure.

Figure 5(c) provides a detailed view of the buffer layer and the secondary interface. The buffer layer exhibited a microstructural gradient, with a finer cellular-dendritic structure near the secondary interface owing to the localized reheating effect of the subsequent hardfacing pass. This structure transitioned

into robust columnar dendrites deeper within the bulk of the buffer layer. The gradient reflected variations in the local thermal gradient and solidification conditions across the buffer layer. Two important metallurgical features were observed. First, δ -ferrite was present within the austenitic matrix as vermicular precipitates aligned with the solidification direction. The presence of δ -ferrite, governed by the Cr/Ni equivalent balance, disrupted the continuity of low-melting-point liquid films and reduced susceptibility to hot cracking (Lippold, 2014). Second, clear evidence of epitaxial solidification was observed at the secondary interface. The hardfacing layer nucleated directly from the partially remelted grains of the buffer layer without the formation of an unmixed zone. This resulted in a continuous crystallographic transition and a high-integrity metallurgical bond at the interface.

Figure 5(d) shows the primary fusion boundary and the adjacent HAZ. The buffer layer fully covered the substrate and formed a continuous interface. The base metal retained its ferritic-pearlitic structure, whereas the adjacent HAZ exhibited a refined and partially transformed microstructure. Unlike S1, severe grain coarsening was not observed. This was attributed to the lower peak temperatures and shorter thermal exposure associated with buffer deposition, which limited austenite grain growth and improved microstructural stability.

Overall, the microstructural observations revealed a clear distinction between S1 and B1. Specimen S1 exhibited an abrupt interface, a carbon-enriched white band, and a brittle martensitic HAZ, resulting in a mechanically vulnerable structure. In contrast, B1 developed a functionally graded architecture comprising a refined HAZ, a compositionally graded buffer layer, a continuous epitaxial interface, and a lath martensitic hardfacing layer. This engineered microstructural gradient reduced stress concentration, improved interfacial toughness, and enhanced resistance to crack initiation and propagation, thereby providing the metallurgical basis for the improved wear performance discussed in subsequent sections.

It is important to note that the phase identification presented in this section was primarily inferred from microstructural morphologies and supported by semi-quantitative EDS analysis. Although these microstructural observations are strongly aligned with established welding metallurgy principles for the selected consumables, definitive crystallographic confirmation and quantification of specific phases, such as the exact fractions of complex

carbides or retained austenite, would require advanced characterization techniques such as X-ray diffraction (XRD) or electron backscatter diffraction (EBSD). The application of these advanced techniques falls outside the scope of the present study but is recommended for future investigations.

4.4 Compositional and Morphological Characterization

To substantiate the microstructural phenomena observed by optical microscopy and evaluate elemental redistribution semi-quantitatively across the distinct metallurgical zones, scanning electron microscopy (SEM) coupled with energy-dispersive X-ray spectroscopy (EDS) was systematically performed. In the directly hardfaced specimen (S1), the high-magnification SEM micrograph of the bulk hardfacing layer shown in Figure 6 revealed a dense, finely dispersed acicular martensitic matrix. This morphological evolution was a direct consequence of the severe thermal gradients and rapid cooling rates inherent to the manual shielded metal arc welding (SMAW) process. The suppressed martensite start (Ms) temperature, driven by the combined effects of localized carbon enrichment and alloy content, promoted a diffusionless transformation into fine plate-like martensite.

However, the critical metallurgical limitation of direct deposition was particularly pronounced at the fusion boundary. Consistent with the unetched white band previously identified by OM, the SEM-EDS results indicated the presence of a carbon-enriched transition zone. Specifically, point analysis at the S1 interface revealed a sharp localized carbon peak of approximately 3.96 wt.%. Because EDS has inherent limitations in accurately quantifying light elements, these carbon signals were interpreted strictly on a semi-quantitative basis to establish relative compositional trends rather than absolute concentrations. The relatively high uncertainty associated with EDS measurements of absolute carbon concentration is primarily attributable to surface contamination by adventitious carbon, such as environmental hydrocarbons or residual vacuum-pump oil within the SEM chamber, together with the limited sensitivity of standard ZAF matrix-correction algorithms for low-energy carbon $K\alpha$ X-rays in Fe–Cr–Ni-rich matrices. Therefore, the reported carbon values should not be interpreted as absolute concentrations but rather as indicators of localized compositional redistribution.

Despite this limitation, the apparent carbon signal at the S1 interface remained substantially higher than

the nominal carbon contents of both the AISI 1050 substrate (0.50 wt.%) and the hardfacing electrode (0.40 wt.%), supporting the presence of localized interfacial enrichment. From a thermodynamic perspective, this apparent carbon enrichment may be associated with the chemical potential gradient established between the medium-carbon substrate and the molten weld pool during the transient high-temperature stage of welding (Lippold, 2014; Gupta et al., 2024). Concurrently, the chromium content at this interface was measured at only 4.56 wt.%, which directly correlated with the 25.01% geometric dilution from the iron-rich base metal calculated in Section 4.2.

The localized combination of apparent carbon enrichment and an insufficient chromium reservoir restricted the stabilization of complex alloy phases. Instead, this microstructural condition favored the formation of carbon-enriched hard phases, potentially associated with brittle iron-carbide precipitation along the fusion line, together with microstructural regions consistent with untempered high-carbon martensitic transformation. This abrupt compositional discontinuity and microstructural heterogeneity rendered the S1 interface highly susceptible to residual stress concentration, cold cracking, and catastrophic spalling under impact loading (Allou et al., 2021).

In contrast, the strategic incorporation of the E309L-16 austenitic buffer layer in specimen B1 fundamentally reconfigured the compositional architecture and mitigated adverse metallurgical reactions at the interfaces. SEM analysis of the primary fusion boundary between the AISI 1050 base metal and the buffer layer shown in Figure 7 demonstrated a continuous, defect-free, and coherent metallurgical bond. EDS mapping and point analysis in this region clarified the pivotal role of the buffer layer as an effective carbon-accommodation zone. Although the absolute carbon values obtained by EDS carry considerable uncertainty, the relative distribution trends remained consistent across the analyzed regions and indicated localized carbon redistribution from the base metal toward the interfacial region.

Elevated semi-quantitative carbon signals detected in the near-interface HAZ (approximately 4.40 wt.%) and on the adjacent buffer side (approximately 4.35 wt.%) suggested that the migrated carbon was preferentially accommodated within the highly alloyed austenitic matrix. As established in Section 4.2, the initial buffer pass produced a higher geometric dilution of 31.31% because of the cumulative heat input and greater fusion area. Despite this increased mixing with the substrate, the primary

interface retained substantial concentrations of chromium (13.17 wt.%) and nickel (5.95 wt.%).

Within the bulk of the buffer layer shown in Figure 8, the microstructure exhibited a robust cellular-dendritic morphology characteristic of high-alloy austenitic stainless steels. The retained chromium and nickel contents increased to 18.42 wt.% and 8.22 wt.%, respectively, together with an apparent entrapped semi-quantitative carbon signal of 4.12 wt.%. The inherent face-centered cubic (FCC)

crystal structure of this austenitic matrix provides exceptional solid solubility for carbon and greater ductility than body-centered tetragonal (BCT) martensite. Consequently, the buffer layer likely served as a carbon-accommodation region while simultaneously accommodating stresses arising from thermal expansion mismatch, thereby significantly suppressing the formation of the deleterious brittle white band observed in S1 (Mvola et al., 2014).

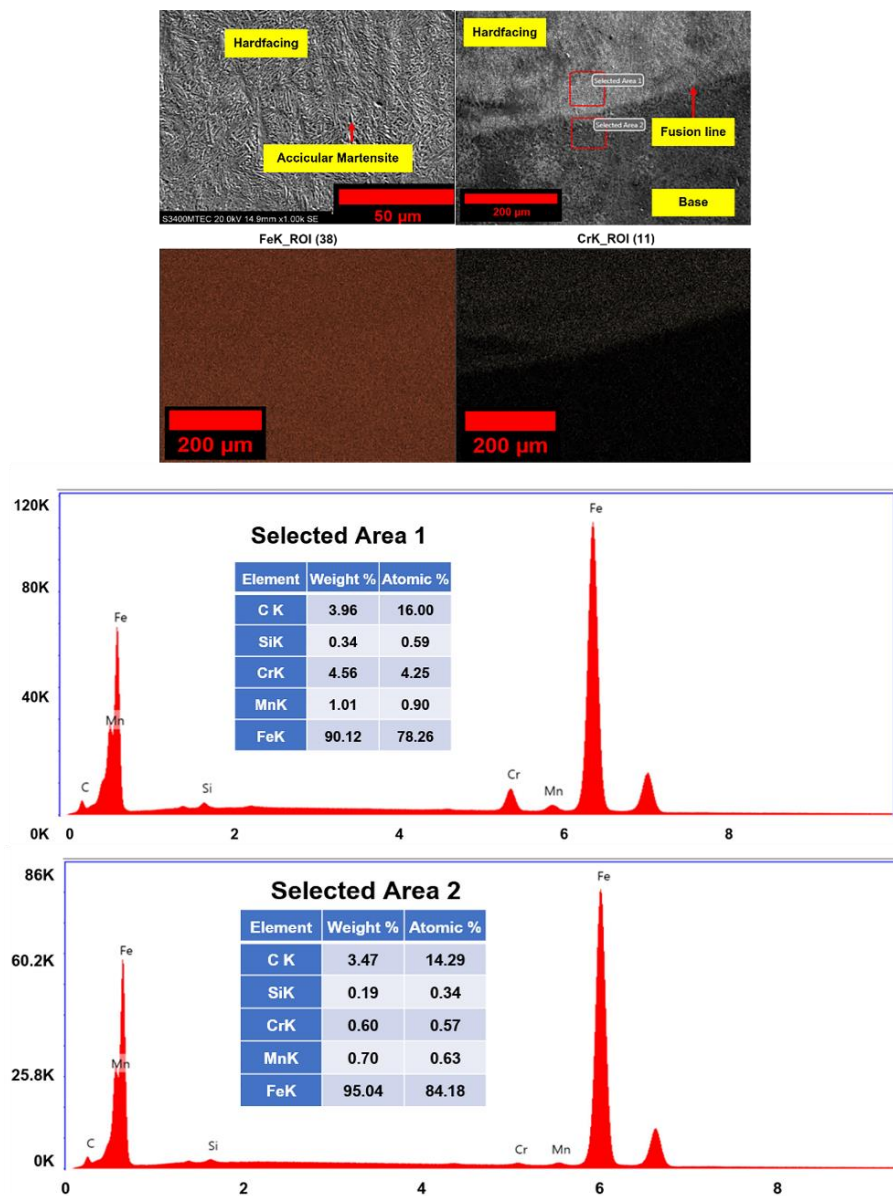


Figure 6 SEM – EDS of S1

The progressive stabilization of the multilayer system was further demonstrated at the secondary interface between the buffer layer and the top hardfacing layer (Figure 9). SEM observations revealed an epitaxial growth mechanism in which the columnar dendrites of the hardfacing layer nucleated directly from the crystallographic planes of the underlying austenitic buffer grains, indicating improved interfacial integrity without microfissures. Because the E309L-16 buffer layer effectively acted as a diffusion barrier that reduced the upward migration of carbon from the substrate (Jeong et al., 2021), EDS analysis of the bulk hardfacing layer in B1 showed a significantly lower semi-quantitative carbon signal (approximately 2.62 wt.%) than the highly enriched level observed at the S1 interface.

Furthermore, as indicated by the macrostructural dilution analysis in Section 4.2, the buffer layer prevented direct dilution of the top layer by the iron-rich substrate, thereby confining base metal incorporation to the intermediate zone. This resulted in significantly improved retention of essential alloying elements in the final working surface. The

top hardfacing layer of B1 contained 10.70 wt.% chromium and 3.21 wt.% nickel, effectively establishing a functionally graded compositional transition that gradually bridged the pronounced metallurgical differences between the base metal and the wear-resistant surface (Tandon et al., 2023).

This optimized elemental redistribution likely promoted a more stable carbon distribution within the top hardfacing layer. These conditions suggested the possible presence of carbon-enriched hard regions potentially associated with chromium-rich precipitate formation rather than deleterious brittle phases. However, definitive crystallographic phase identification is beyond the capability of OM and SEM-EDS alone and would require complementary techniques such as X-ray diffraction (XRD) or electron backscatter diffraction (EBSD). Ultimately, the deliberate grading of the microstructure and chemical composition introduced by the austenitic buffer layer is considered to have contributed to reduced stress concentration, improved interfacial toughness, and enhanced abrasive wear performance in this study.

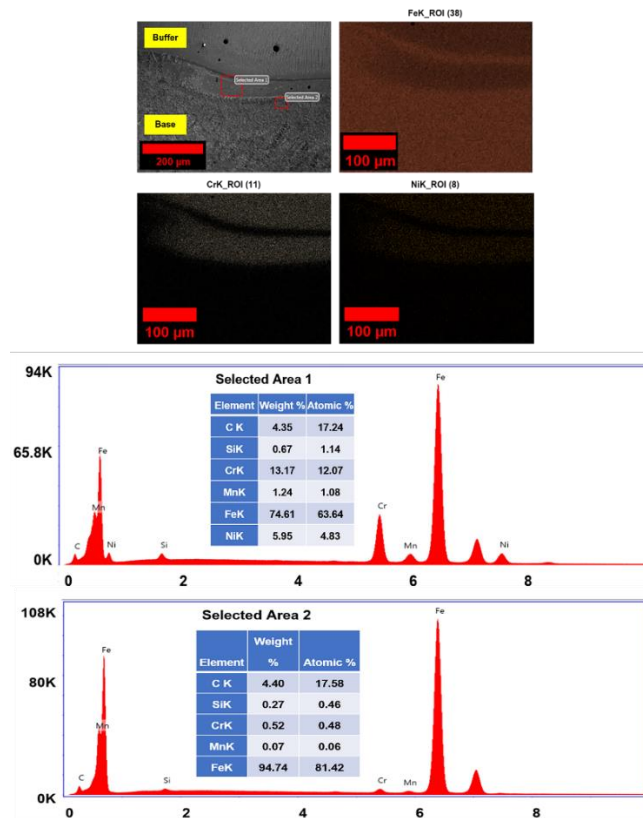
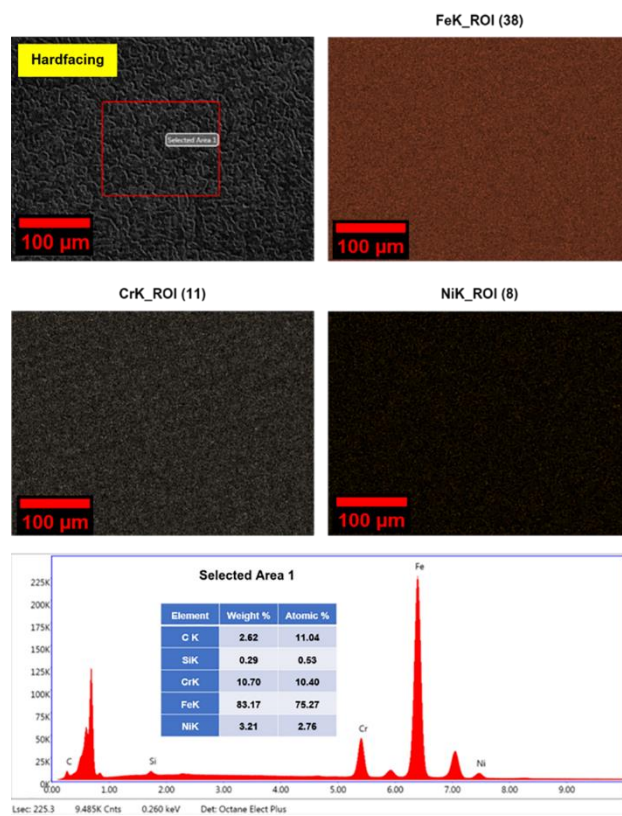
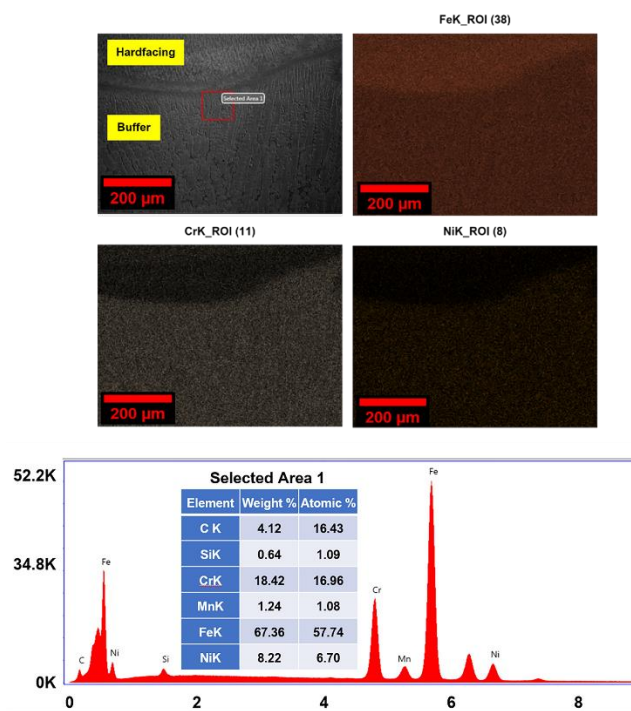


Figure 7 SEM-EDS of B1 between the AISI 1050 base metal and the buffer layer



4.5 Hardness

To evaluate the mechanical response of the hardfaced systems and establish a direct correlation with the macrostructural, microstructural, and compositional results discussed in Sections 4.2–4.4, Vickers microhardness profiles were systematically measured across the different metallurgical zones of both specimens. The corresponding hardness distributions are presented in Figure 10. Each reported value represents the average of three measurements, with the error bars indicating the standard deviation. The hardness of the unaffected AISI 1050 base metal remained relatively constant in both specimens, ranging from 205 to 206 HV. This consistency indicated that the applied welding thermal cycles did not significantly alter the bulk mechanical properties of the substrate, in agreement with the limited penetration depths reported in Section 4.2.

A clear difference in hardness behavior was observed between S1 and B1 in the top hardfacing region. The directly hardfaced specimen (S1) exhibited a higher peak hardness, reaching 647 ± 20 HV at a depth of 0.5 mm and gradually decreasing to 572 ± 15 HV at 2.3 mm. In contrast, the buffer-assisted specimen (B1) exhibited comparatively lower hardness values of 586 ± 25 HV at 0.5 mm and 536 ± 18 HV at 2.1 mm. The lower hardness of B1 was consistent with the compositional and microstructural modifications introduced by the buffer layer. As demonstrated in Section 4.4, the E309L-16 buffer layer restricted carbon migration from the substrate and promoted a more controlled distribution of alloying elements in the top layer. Consequently, the lower apparent carbon enrichment, together with the moderated dilution effect described in Section 4.2, reduced the martensitic hardness. This finding was consistent with the microstructural transition identified in Section 4.3, where the hardfacing layer in S1 was dominated by acicular martensite, whereas B1 exhibited a comparatively tougher lath martensitic structure. Although S1 exhibited a higher peak hardness, this condition was not necessarily advantageous. The higher hardness of S1 was associated with carbon-enriched martensite and potential carbide formation at the interface, as discussed in Section 4.4, which promoted brittleness and increased susceptibility to cracking.

A more critical distinction between the two specimens was observed in the hardness gradient across the interfacial regions. In S1, a sharp and abrupt decrease in hardness occurred, from approximately 572 ± 15 HV in the hardfacing layer to approximately 230 ± 15 HV in the HAZ at a depth of 4.8 mm. This steep gradient reflected a severe mechanical mismatch,

consistent with the abrupt microstructural transition and compositional discontinuity identified previously. Such a mismatch could act as a stress concentration site, significantly increasing the likelihood of interfacial cracking and delamination under service conditions.

In contrast, specimen B1 exhibited a more gradual and continuous hardness transition. A distinct hardness minimum of 175 ± 30 HV was observed at approximately 4.1 mm, corresponding to the austenitic buffer layer. This relatively soft and ductile face-centered cubic (FCC) structure played a critical mechanical role by acting as a strain-accommodating layer. As demonstrated in Sections 4.3 and 4.4, the buffer layer provided both microstructural and compositional gradients, effectively reducing the abruptness of the hardness transition. Furthermore, the buffer layer confined dilution to the intermediate region, as described in Section 4.2, and prevented excessive carbon enrichment in the top layer, as discussed in Section 4.4, resulting in a more balanced hardness distribution. This transformed the system from the abrupt mechanical interface observed in S1 into a functionally graded structure in B1.

Overall, the hardness profiles clearly demonstrated that, although S1 achieved a higher peak hardness, it exhibited severe mechanical discontinuity and brittleness. In contrast, B1 exhibited slightly lower hardness but a substantially improved hardness gradient and greater mechanical compatibility across the interface. This optimized hardness distribution reduced stress concentration, enhanced interfacial toughness, and established the mechanical foundation for improved resistance to cracking, spalling, and coating delamination under the abrasive service conditions evaluated in the subsequent section.

4.6 Wear

The abrasive wear behavior of the specimens was quantitatively evaluated using a dry sand/rubber wheel abrasion test in accordance with the modified ASTM G65 procedure, as shown in Figure 11. Figure 12 presents the macroscopic wear performance of the tested conditions. As shown in Figure 12(a), all specimens exhibited a progressive and approximately linear increase in cumulative mass loss with increasing test duration, indicating the establishment of a relatively steady abrasive wear regime. The uncoated AISI 1050 base metal consistently exhibited the greatest material loss, reaching 1.478 ± 0.063 g after 600 s. In contrast, both hardfaced specimens demonstrated significantly improved wear resistance, with the mass loss reduced to 0.6307 ± 0.0146 g for S1 and 0.5299 ± 0.0154 g for B1 at the end of the test.

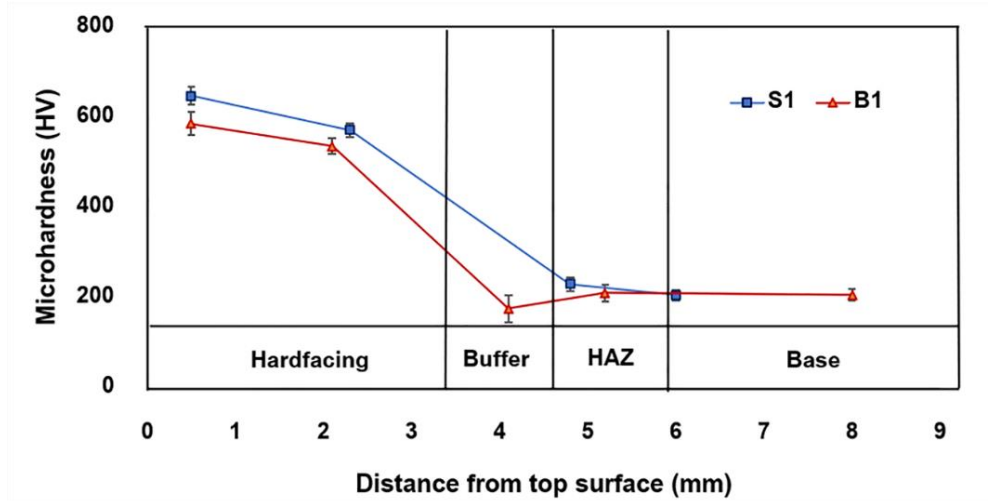


Figure 10 Microhardness profiles across the different metallurgical zones of the S1 and B1 specimens

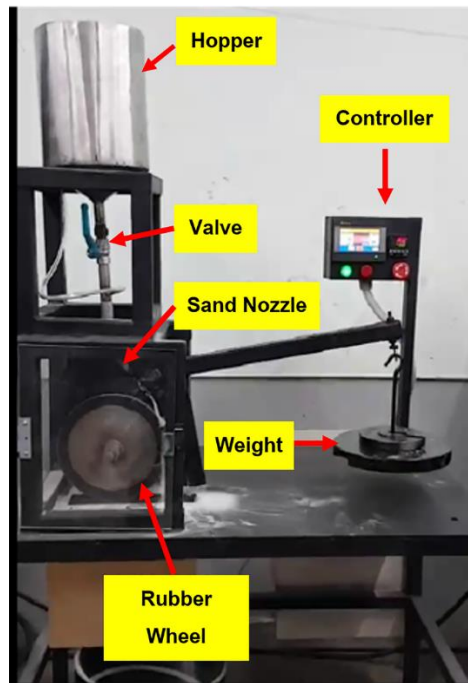


Figure 11 Wear test according to ASTM G65

To facilitate a standardized tribological evaluation, the cumulative mass loss was converted into the specific wear rate (K), as presented in Figure 12(b). The narrow standard deviation error bars confirmed the high reproducibility and statistical reliability of the tests. The base metal exhibited the highest specific wear rate of $(10.08 \pm 0.43) \times 10^{-4} \text{ mm}^3/\text{N}\cdot\text{m}$. Direct hardfacing (S1) reduced the wear rate to $(4.30 \pm 0.10) \times 10^{-4} \text{ mm}^3/\text{N}\cdot\text{m}$, whereas the buffer-assisted specimen (B1) exhibited the greatest wear resistance, achieving the lowest specific wear

rate of $(3.62 \pm 0.11) \times 10^{-4} \text{ mm}^3/\text{N}\cdot\text{m}$. This represented a substantial improvement over the direct hardfacing approach, despite B1 exhibiting a slightly lower peak hardness, as discussed in Section 4.5.

The fundamental mechanisms governing the observed wear behavior were further elucidated through SEM-EDS examination of the worn surfaces, as shown in Figure 13. The wear morphology of specimen S1 (Figure 13(a)) was characterized by deep abrasive grooves and irregular microfractured groove edges. These features indicated a relatively severe

abrasive wear mechanism accompanied by localized brittle-assisted material removal and microspalling during repeated interactions with abrasive particles. Furthermore, EDS mapping of S1 revealed comparatively lower chromium retention at the worn surface, consistent with the greater dilution effect and reduced alloy stability discussed in Section 4.4. The insufficient retention of carbide-forming alloying elements likely contributed to the reduced resistance to localized abrasive damage.

In contrast, the worn surface of specimen B1 (Figure 13(b)) exhibited a more uniform morphology, with shallower microploUGHing grooves and distinct plastic deformation ridges along the groove margins. These features were characteristic of a more stable abrasive wear mechanism dominated primarily by controlled plastic deformation rather than unstable brittle fragmentation. Moreover, EDS elemental mapping confirmed the retention of chromium- and nickel-containing phases across the worn surface,

supporting the effectiveness of the E309L-16 austenitic buffer layer in reducing excessive dilution and moderating interfacial compositional instability. As discussed in Sections 4.3 and 4.4, the buffer layer likely contributed to suppressing excessive carbon redistribution and minimizing the formation of brittle interfacial regions while simultaneously improving stress accommodation across the multilayer structure.

Overall, these findings demonstrated that the abrasive wear resistance of multilayer hardfacing systems was governed not solely by peak hardness but also by the combined effects of microstructural integrity, interfacial toughness, compositional stability, and resistance to localized brittle damage. The incorporation of the austenitic buffer layer effectively produced a functionally graded architecture capable of improving stress distribution and enhancing wear stability under severe three-body abrasive conditions.

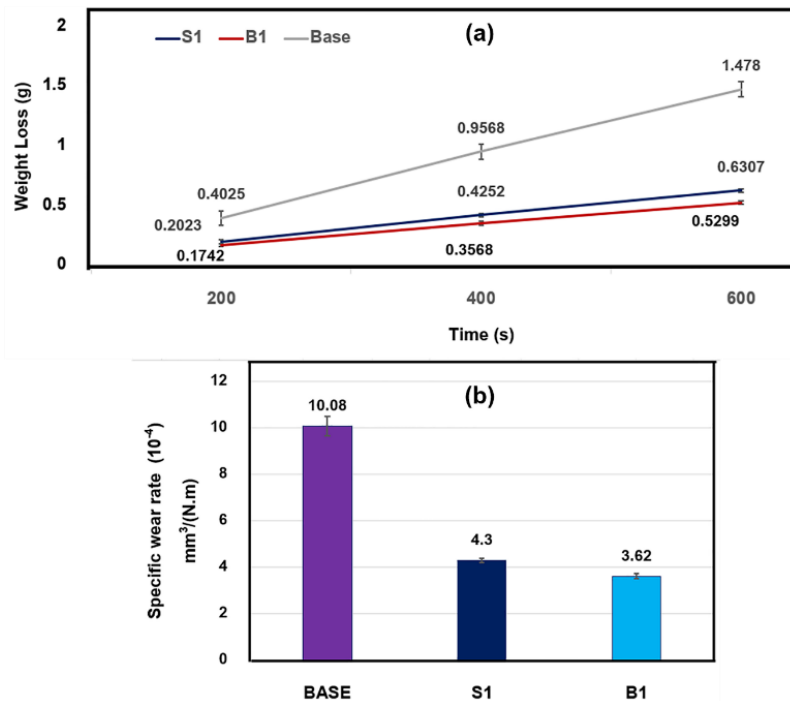


Figure 12 Abrasive wear test results: (a) cumulative weight loss as a function of testing time, and (b) specific wear rates of the base metal, S1, and B1 specimens

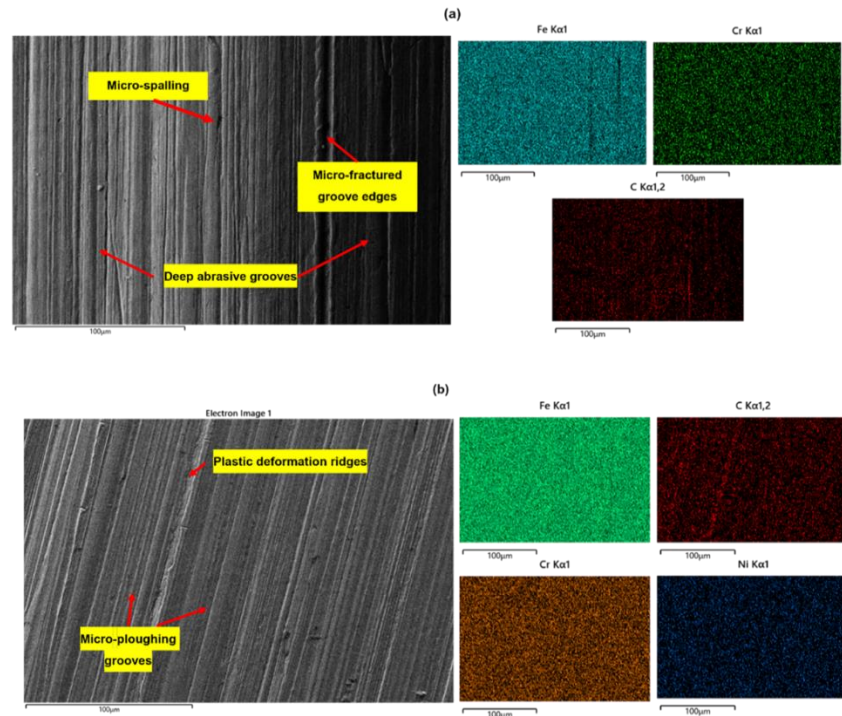


Figure 13 SEM micrographs and corresponding EDS elemental mappings of the worn surfaces: (a) direct hardfacing (S1) exhibiting deep abrasive grooves, micro-fractured edges, and localized micro-spalling indicative of brittle-assisted failure, and (b) buffer-assisted hardfacing (B1) showing shallower micro-ploughing grooves with distinct plastic deformation ridges indicative of improved structural toughness

5. Discussion

The present study demonstrates that incorporating an austenitic E309L-16 buffer layer plays a critical role in modifying the microstructural evolution, interfacial characteristics, and overall performance of hardfaced AISI 1050 steel. The observed improvement in wear resistance, despite a slight reduction in peak hardness, highlights the complex interplay among hardness, microstructure, and interfacial integrity.

5.1 Role of the Austenitic Buffer Layer in Carbon Redistribution

The results clearly demonstrate that the primary metallurgical limitation of direct hardfacing (S1) arises from uncontrolled carbon migration across the fusion boundary. Owing to the high carbon content of AISI 1050 and the steep thermal gradient during welding, carbon rapidly diffuses into the molten weld pool, leading to localized enrichment at the interface. This phenomenon promotes the formation of brittle microstructures, including untempered martensite and possible carbide networks, which are commonly

associated with interfacial embrittlement and susceptibility to cracking.

The introduction of the E309L-16 austenitic buffer layer fundamentally alters this behavior. Owing to its face-centered cubic (FCC) structure and high solid solubility for carbon, the buffer layer acts as an effective carbon sink by redistributing and accommodating diffusing carbon within its matrix. Consequently, the abrupt compositional discontinuity observed in S1 is replaced by a more gradual transition in B1. This graded compositional profile suppresses the formation of the interfacial white band and significantly improves metallurgical compatibility between the substrate and the hardfacing layer. This behavior is consistent with established principles of dissimilar-metal welding, in which austenitic interlayers are frequently employed to mitigate carbon migration and reduce the formation of brittle phases at fusion boundaries.

5.2 Microstructural Evolution and Mechanical Implications

The modification of carbon distribution directly influences the resulting microstructure and,

consequently, the mechanical response of the hardfaced system. In S1, the combined effects of carbon enrichment and rapid cooling lead to the formation of predominantly acicular martensite. Although this microstructure provides high hardness, it is inherently brittle because of its high internal stress state and limited capacity for plastic deformation.

In contrast, the presence of the buffer layer in B1 results in a more controlled chemical composition within the top hardfacing layer. This promotes the formation of lath martensite, which is characterized by a packet structure with improved dislocation mobility and greater toughness than acicular martensite. In addition, the austenitic buffer layer plays a crucial mechanical role as a ductile interlayer. Its FCC structure provides a substantial capacity for plastic deformation, enabling it to absorb thermal and mechanical stresses generated during solidification and subsequent loading. This combination of a tougher martensitic top layer and a ductile intermediate layer improves resistance to crack initiation and propagation.

5.3 Hardness Gradient and Stress Distribution

Beyond absolute hardness values, the distribution of hardness across the interface is a critical factor governing structural reliability. Specimen S1 exhibited a sharp hardness gradient, decreasing abruptly from 572 ± 15 HV in the hardfacing layer to 230 ± 15 HV in the heat-affected zone, indicating a severe mechanical mismatch. Such abrupt transitions are well known to act as stress concentration sites, particularly under cyclic or impact loading conditions.

In contrast, B1 exhibited a more gradual hardness profile because of the presence of the relatively soft buffer layer, which recorded a distinct minimum hardness of 175 ± 30 HV. Quantitative comparisons with the existing literature highlight the effectiveness of the E309L-16 SMAW parameters employed in this study in achieving this profile. For example, a recent study using flux-cored arc welding (FCAW) to apply an austenitic buffer layer to structural steel reported an elevated buffer hardness of approximately 392–408 HV owing to extensive elemental dilution (Tippayasam et al., 2023). Similarly, a comparable SMAW study using a Ni-Cr austenitic buffer layer reported hardness values ranging from 235 to 274 HV (Srisuwan et al., 2019).

In the present study, the E309L-16 buffer layer maintained a significantly lower hardness of 175 ± 30 HV, demonstrating an exceptional ability to minimize excessive dilution and thereby preserve the inherent

ductility of the austenitic phase. This produced a functionally graded structure that distributed stresses more uniformly across the interface. The reduction in mechanical mismatch minimized localized strain accumulation and significantly reduced the likelihood of interfacial cracking or delamination. This finding highlights that quantitatively optimizing the hardness distribution, rather than maximizing peak hardness, is essential for improving the overall performance of multilayer hardfacing systems.

5.4 Correlation Between Microstructure and Wear Mechanism

The superior wear performance of B1, despite its lower peak hardness, underscores the importance of microstructural integrity and toughness in abrasive environments. Under ASTM G65 three-body abrasion conditions, material removal is governed not only by hardness but also by the material's ability to resist crack initiation and propagation under repeated particle impacts. In S1, the brittle acicular martensitic structure and carbon-enriched interfacial region facilitated microcrack initiation. These cracks readily propagated under cyclic abrasive loading, leading to localized microspalling and the formation of microfractured groove edges.

In contrast, B1 benefited from a more balanced microstructure. The lath martensite provided sufficient hardness to resist microcutting, whereas the underlying ductile buffer layer absorbed impact energy and inhibited crack propagation. Furthermore, the absence of a brittle interfacial zone reduced the likelihood of subsurface crack nucleation. Consequently, B1 exhibited a more stable wear mechanism dominated by controlled microploughing and distinct plastic deformation ridges rather than brittle fracture, resulting in a lower wear rate.

Furthermore, the reduction in wear rate achieved by the buffer-assisted B1 specimen, which exhibited a specific wear rate of $3.62 \times 10^{-4} \text{ mm}^3/\text{N}\cdot\text{m}$ and a 64% reduction relative to the base metal, compared favorably with quantitative findings reported in recent hardfacing studies. Although comparable studies using alternative austenitic buffer layers reported apparently greater reductions in wear rate, such as 76–80% for a Ni-Cr buffer layer (Srisuwan et al., 2019) and an improvement in service life of up to 80% for FCAW systems (Tippayasam et al., 2023), this numerical variation was primarily attributable to differences in testing severity and material systems.

Specifically, the wear tests in the cited studies were conducted over substantially shorter durations, such as 30–300 s, whereas the present study employed a more severe and prolonged 600 s testing cycle. In addition, this study used a martensitic E 3-UM-600 electrode on a challenging high-carbon AISI 1050 substrate to balance moderate impact toughness with abrasion resistance, unlike the extreme abrasion-oriented chromium-carbide E10-UM-65 configurations. Therefore, achieving a substantial 64% reduction under these prolonged conditions confirms the high effectiveness and structural reliability of the engineered E309L-16 multilayer architecture.

5.5 Limitations and Future Work

Although the present study provides strong microstructural and mechanical evidence, several limitations should be acknowledged. The compositional analysis of carbon using EDS is inherently semi-quantitative because of the low atomic number of carbon; therefore, the reported values should be interpreted as relative trends rather than absolute concentrations. Furthermore, phase identification was inferred from microstructural observations and compositional data. Additional characterization techniques, such as X-ray diffraction (XRD), would provide more definitive confirmation of the phase constituents, particularly the carbide phases. Future work should also investigate the effects of buffer-layer thickness, alternative buffer compositions, and post-weld heat treatment on the performance of hardfaced systems to further optimize their interfacial properties.

6. Conclusion

This study systematically investigated the influence of an austenitic E309L-16 buffer layer on the microstructural evolution, interfacial characteristics, hardness distribution, and abrasive wear performance of SMAW-hardfaced AISI 1050 medium-carbon steel. The results clearly demonstrate that incorporating the buffer layer fundamentally alters the metallurgical behavior of the system. Under the directly hardfaced condition (S1), pronounced carbon diffusion from the substrate led to the formation of a brittle, carbon-enriched interfacial region accompanied by a steep hardness gradient and an acicular martensitic microstructure. These features contributed to severe mechanical incompatibility and increased susceptibility to crack initiation under abrasive conditions.

In contrast, the buffer-assisted condition (B1) effectively suppressed excessive carbon diffusion and eliminated the formation of the brittle white band at the fusion boundary. This resulted in a more homogeneous interfacial microstructure characterized by improved metallurgical bonding and a gradual compositional transition. The hardfacing layer in B1 consisted predominantly of lath martensite with enhanced toughness, whereas the intermediate austenitic buffer layer provided essential ductility and stress accommodation.

Despite exhibiting a slightly lower peak hardness, specimen B1 demonstrated superior wear resistance. Specifically, the wear mechanism transitioned from severe microspalling in S1 to stable microploughing with plastic deformation ridges in B1. Consequently, B1 achieved the lowest specific wear rate of $(3.62 \pm 0.11) \times 10^{-4} \text{ mm}^3/\text{N}\cdot\text{m}$, representing a reduction of approximately 64% compared with the base material, which exhibited a specific wear rate of $(10.08 \pm 0.43) \times 10^{-4} \text{ mm}^3/\text{N}\cdot\text{m}$, and a distinct improvement over the directly hardfaced S1 specimen, which exhibited a specific wear rate of $(4.30 \pm 0.10) \times 10^{-4} \text{ mm}^3/\text{N}\cdot\text{m}$. This finding firmly establishes that abrasive wear resistance is governed not solely by maximum surface hardness but by the synergistic effects of microstructural integrity, interfacial toughness, and property gradation.

Overall, introducing an austenitic buffer layer effectively transformed the hardfacing system into a functionally graded structure, reducing stress concentration, suppressing crack formation, and significantly enhancing resistance to material removal under severe abrasive conditions. The findings of this study provide valuable insights into the design of advanced multilayer hardfacing systems and emphasize the critical role of interfacial engineering in maximizing the service life of industrial components. Further research should include definitive phase identification using X-ray diffraction (XRD) and evaluation of the mechanical integrity of these functionally graded hardfacing systems under high-impact loading conditions.

7. Acknowledgements

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8. Abbreviations

Abbreviation	Full Term
AISI	American Iron and Steel Institute
ASTM	American Society for Testing and Materials
AWS	American Welding Society
BCT	Body-Centered Tetragonal
CET	Columnar-to-Equiaxed Transition
DCEN	Direct Current Electrode Negative
DIN	Deutsches Institut für Normung (German Institute for Standardization)
EBS	Electron Backscatter Diffraction
EDS	Energy-Dispersive X-ray Spectroscopy
FCAW	Flux-Cored Arc Welding
FCC	Face-Centered Cubic
G/R	Thermal Gradient-to-Growth Rate Ratio
HAZ	Heat-Affected Zone
HV	Vickers Hardness
HVOF	High-Velocity Oxygen Fuel
Ms	Martensite Start Temperature
OM	Optical Microscopy
PT	(Dye) Penetrant Testing
SEM	Scanning Electron Microscopy
SiC	Silicon Carbide
SMAW	Shielded Metal Arc Welding
UTS	Ultimate Tensile Strength
XRD	X-ray Diffraction

9. CRediT Statement

Yodnapha Ketmuang: Conceptualization, Methodology, Resources, Supervision, Validation, Writing – Original Draft, Project Administration, Funding Acquisition

Apichart Chaichawalit: Conceptualization, Methodology, Data Curation, Formal Analysis, Visualization, Writing – Original Draft, Writing – Review & Editing

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