



Influence of ceramic retention on inter-splat bonding and mechanical behavior of cold-sprayed Cu-ZrC composite coatings

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ARTICLE INFO

Keywords:

Cold spray
Metal matrix composites
Inter-splat bonding
Adhesion
Scratch
Wear

ABSTRACT

Cold-sprayed metal-ceramic composite coatings offer a promising route for high-performance surface engineering; however, the influence of ceramic retention on inter-splat bonding and mechanical performance remains insufficiently understood. In this study, Cu-ZrC composite coatings containing 30–85 wt% ZrC were deposited on Al-6061 substrates to investigate deposition behavior, interfacial integrity, and tribological response. Coating build-up was achieved for all compositions, although deposition efficiency decreased at higher ceramic loading due to increased particle rebound and reduced matrix continuity. Hardness increased continuously with ZrC content, while adhesion strength and scratch resistance showed non-linear behavior, with optimum cohesion observed at intermediate compositions (50–70 wt% ZrC). Scratch testing revealed reduced penetration depth and improved resistance to localized deformation with increasing ceramic content, indicating enhanced inter-splat bonding. Ball-on-disk wear testing demonstrated a progressive reduction in wear volume, reaching ~ 93% lower wear for the 85 wt% ZrC coating compared to pure Cu, accompanied by a transition from ductile adhesive wear to brittle fragmentation-dominated mechanisms. Overall, intermediate ceramic contents provided the best balance between deposition efficiency, mechanical integrity, and tribological performance.

1. Introduction

Copper (Cu) and its alloys are widely utilized in energy, power generation, rail transport, aerospace, defence, and electronics due to their outstanding electrical and thermal conductivity combined with high ductility [1–3]. However, their relatively low hardness, poor wear resistance, and susceptibility to mechanical and environmental degradation limit their performance under extreme service conditions [4]. To overcome these limitations, high-performance ceramics or high-strength alloys have historically been incorporated into Cu to improve the mechanical performance of Cu-based coatings and components [5,6]. Recent advancements in ceramic-reinforced coatings, including Al₂O₃-TiO₂ [7] and high-entropy alloy-based composite coatings [8], have demonstrated significant improvements in hardness, adhesion, and wear resistance through mechanisms such as grain refinement and interfacial strengthening. These studies highlight the critical role of ceramic

content in governing coating performance, although they are primarily based on high-temperature processes.

Cold spray (CS) has emerged as a promising alternative for depositing low melting materials and incorporating hard phases into soft metallic matrices. As a solid-state process, CS enables deposition without melting or significant oxidation of feedstock powders [9–11]. By accelerating particles to supersonic velocities, CS can produce dense coatings with low porosity, strong adhesion, and minimal thermal damage to the feedstock and the substrate. This makes CS attractive, especially for Cu-based composites, where retaining thermal and electrical conductivity while improving mechanical durability is essential [10]. Reinforcing Cu with ceramic phases has proven effective in enhancing hardness, wear resistance, and high-temperature performance. In this context, zirconium carbide (ZrC) stands out as a valuable reinforcement due to its ultra-high melting point (~3540 °C), high hardness, low density, and excellent resistance to oxidation, radiation

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<https://doi.org/10.1016/j.matdes.2026.116527>

Received 17 March 2026; Received in revised form 13 May 2026; Accepted 29 June 2026

Available online 29 June 2026

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and ablation [12,13]. The synergy of thermal robustness of ZrC with ductility of Cu suggests strong potential for advanced Cu-based composite coatings in extreme service environments.

Considerable effort has been devoted to synthesizing ZrC and ZrC–Cu-based composites via high-temperature processes [14–18]. Despite these advances, the majority of ZrC–Cu composites reported in the literature rely on high temperature processing routes. These methods often suffer from drawbacks such as poor wettability between Cu and ZrC, residual porosity, and limited scalability. In contrast, cold spray offers a cost-effective and low-temperature technique to fabricate ZrC–Cu coatings while preserving feedstock properties and avoiding interfacial degradation. While recent work [19] has demonstrated the feasibility of depositing ZrC–Cu coatings by cold spray, a systematic understanding of how ceramic content influences coating build-up, inter-splat bonding, and performance remains limited. In particular, the role of ceramic content in governing coating build-up during multi-pass deposition, and its influence on mechanical and tribological behaviour across a wide compositional range, has not been comprehensively investigated. Furthermore, mechanical and tribological properties are often treated independently in cold-sprayed composite coatings. Scratch testing reflects localized deformation under monotonic loading, whereas sliding wear involves cyclic loading and progressive damage evolution. However, the transition in damage mechanisms between these loading conditions from plastic deformation to crack initiation, propagation, and material removal has not been clearly established. To address these gaps, the present study investigates cold-sprayed ZrC–Cu composite coatings across a wide ceramic range (30–85 wt%) and multiple spray passes. The study first examines coating build-up behaviour through thickness evolution and hardness development, and subsequently evaluates adhesion, scratch response, and sliding wear performance of the fully developed coatings. The most significant contribution of this work is the integrated interpretation of scratch testing and ball-on-disk wear to establish a mechanistic understanding of damage evolution under different loading conditions. By linking deposition behaviour with deformation and wear mechanisms, this work establishes a comprehensive processing–structure–property–tribology framework and identifies an optimal ceramic content for balanced performance.

2. Materials and methodology

2.1. Powder and substrate

For cold spray deposition, gas-atomized copper powder of high purity (> 99%) and sintered–crushed ZrC powder (> 99%) were used as the feedstock materials. The copper particles were in the size range of 30–60 µm, whereas the zirconium carbide particles were slightly finer, ranging between 20 and 45 µm. The powders were mechanically blended in different ratios, as presented in Table 1. The compositions are denoted according to their blend designations. Aerospace-grade Al-6061 alloy plates were used as substrate of dimensions 100 mm × 100 mm × 5 mm, which were ultrasonically cleaned with ethanol prior to deposition.

2.2. Coating deposition

The coatings were produced using a commercial high-pressure cold spray unit (PCS-100, Plasma Giken, Japan). The spray gun was mounted

on a six-axis KUKA robotic arm for precise control during deposition. A De-Laval type convergent–divergent nozzle with a circular exit was used to accelerate the powder to supersonic velocity. Nitrogen was employed as the carrier gas to propel the particles and achieve the desired kinetic energy. The spraying parameters applied for deposition of the Cu–ZrC blends are summarized in Table 2.

2.3. Microstructural and mechanical characterisation

The deposited coatings were sectioned using a wire-cut Electrical Discharge Machine (EDM, Electronica Sprintcut, India) following the guidelines for thermal spray coatings [20]. The cross-sections were mounted in bakelite and subsequently polished for microstructural evaluation. Polishing was carried out by sequential grinding with SiC papers of 400, 800, 1000, 1500, and 2000 grit, followed by final polishing using 1 µm alumina suspension. Microstructural and cross-sectional observations of the powders and coatings were performed on a Field Emission SEM (JEOL JSM-7610F Plus, Japan) equipped with Energy-Dispersive X-ray Spectroscopy (EDS). An optical microscope (Olympus GX53) was utilized to quantify ZrC content in the coatings. The area fraction of ZrC was determined using ImageJ software (NIH, USA), where at least ten images per sample were analysed to obtain average values (reported in vol.%). Porosity levels were also calculated using the same image analysis method. To enable comparison with feedstock composition and deposition efficiency, the measured volume fractions were converted to weight percent (wt.%) using the rule of mixtures as per Equation (1):

$$W_i = \frac{V_i \rho_i}{\sum (V_i \rho_i)} \quad (1)$$

where V_i is the volume fraction and ρ_i is the density of the respective phase (Cu: 8.94 g/cm³ [19], ZrC: 6.73 g/cm³ [19]). The contribution of porosity was considered negligible due to very low porosity levels observed in the coatings. Microhardness was evaluated using a Vickers indenter (Wilson 402MVD, China) at a load of 300 gf and a dwell period of 10 s. Fifteen indentations were made for each coating, ensuring that the spacing between indents was at least four times the diagonal length of the previous impression, as per ASTM C1327 [21]. The mean values with standard deviations are reported. Phase identification of the coatings was conducted by X-ray diffraction (XRD, Analytical X'Pert Pro, Netherlands) using Cu-K α radiation. The scans were performed in continuous mode across a 2θ range of 20°–100° at a rate of 4°/min. Adhesion strength was evaluated using a pull-off tester (BGD 500/S Buiged) with 10 mm dollies and a two-part adhesive (80 MPa nominal strength), with at least three measurements per coating. The deposited coatings were evaluated through scratch testing using a Micro Scratch Tester (DUCOM, TR101, India) fitted with a spherical diamond indenter of 200 µm diameter. Experiments were performed in both constant and progressive loading modes. In the constant load case, a force of 40 N was applied across a 4 mm track. For the progressive mode, the applied load varied between 20 and 60 N at a rate of 10 N/mm. In all tests, the indenter traversed the surface at a velocity of 0.25 mm/s. Post-scratch profiles were analyzed to assess penetration depth, scratch width, and to capture high-resolution 3D surface profiles with advanced 3D surface

Table 1
Composition of feedstock blends for cold spray coatings.

Blend Designation	ZrC (wt. %)	Cu (wt. %)
ZC30	30	70
ZC50	50	50
ZC70	70	30
ZC85	85	15

Table 2
Operating parameters for cold spraying ZrC–Cu blends on Al-6061 substrate.

Parameter	Value
Process gas	Nitrogen
Pressure	30 bar
Temperature	700 °C
Scan speed	50 mm/s
Stand-off distance	40 mm
Passes	5

measurement capabilities (Infinite Focus G6, Bruker Alicona). The wear tests were conducted using a ball-on-disk configuration on a Ducom POD-4.0 tribometer. An Al_2O_3 ball with a diameter of 6 mm was used as the counterface. Prior to testing, the coatings were mirror-polished. All experiments were performed under a normal load of 20 N at a rotational speed of 50 rpm, corresponding to a sliding velocity of 26 mm/s. The CoF was continuously recorded for all the samples, and the wear-tested specimens were subsequently evaluated for wear volume loss. Wear volume analysis was carried out using the optical profilometer (Infinite Focus G6, Bruker Alicona). The wear volume was calculated by integrating the acquired 3D surface profiles using dedicated analysis software (MetMaX 5.0.1), ensuring high measurement accuracy.

3. Results and discussions

3.1. Microstructural characterization

The SEM image in Fig. 1 depicts the morphology and size of Cu and ZrC feedstock powders used for developing composite coatings via cold spray on Al-6061 alloy. The micrographs reveal that Cu particles possess a spherical morphology, while ZrC exhibits an irregular geometry. The ZrC particle size was intentionally selected to be irregular in shape while Cu as spherical. This is important because deposition of ceramics in cold spray primarily occurs through mechanical interlocking with the plastically deforming matrix (Cu in our case) [22]. Hence, irregular particles have higher chances of getting entrapped because of more contact points thereby improving the deposition efficiency [23]. Moreover, irregularly shaped particles tend to attain greater impact velocities owing to their higher drag coefficient [24], which enhances their adherence to both the substrate and the surrounding metallic binder [25].

The cross-sectional micrographs of the cold-sprayed ZrC–Cu coatings reveal a clear dependence of coating growth on both the number of spray-passes and the corresponding ratios of ZrC–Cu. The comparison of all the coatings with 1, 2, and 5 passes for each blend (ZC30, ZC50, ZC70, ZC85) is shown in Fig. 2. The variation of coating thickness with respect to ceramic content and number of passes is also summarized in Fig. 3. In every composition, increasing the number of passes yields progressively thicker deposits. From Figs. 2 and 3, it can be observed that the ZC30 coating thickness increases from $\approx 238 \mu\text{m}$ after one pass to $\approx 470 \mu\text{m}$ after two passes, eventually reaching $\approx 1045 \mu\text{m}$ after five passes. A similar doubling trend is seen for ZC50, ZC70, and ZC85, consistent with the cumulative nature of material build-up in cold spray. Across all the combinations, the thickest coatings are observed at 5 passes. Quantitatively, coating thickness scales almost linearly with number of passes for all the compositions, indicating that each additional spray pass contributes a relatively constant increment in material deposition. At a fixed number of passes, the feedstock composition strongly influences the coating thickness and thus the deposition efficiency. As reported in earlier studies [23], higher ceramic content

generally leads to thinner coatings. For instance, after a single pass, the low-ceramic ZC30 blend yielded $\sim 238 \mu\text{m}$ of coating, whereas ZC85 produced only $\sim 72.4 \mu\text{m}$. Even at 5 passes the effect persists, i.e., ZC30 reached $\sim 1045 \mu\text{m}$, ZC50 $\sim 1341 \mu\text{m}$, ZC70 $\sim 936 \mu\text{m}$, and ZC85 only $\sim 568 \mu\text{m}$. Interestingly, the maximum thickness at 5 passes occurs for ZC50 rather than the lowest-ceramic ZC30, this suggests an optimum reinforcement balance around 50 wt% ZrC. This non-monotonic behaviour in thickness with ceramic loading arises from the evolving role of ZrC during deposition. Although ZrC particles are brittle and do not plastically deform, they contribute positively at moderate concentrations by enhancing the deformation of the ductile Cu matrix. Their high stiffness promotes additional compaction of previously deposited material and produces a tamping or peening effect, which increases splat densification and coating build-up. However, the same ceramic particles can fracture upon impact, particularly when they strike previously deposited ZrC rather than the Cu binder. The probability of such ceramic-on-ceramic impacts increases with ceramic content. At moderate ceramic levels, rigid ZrC particles assist deposition by intensifying the deformation of the Cu matrix and promoting mechanical interlocking, thereby enhancing both bonding and coating density. Beyond a threshold of $\sim 70 \text{ wt\%}$ ZrC, however, the deposition process becomes dominated by ceramic-ceramic interactions. Incoming ZrC particles tend to rebound or shatter due to their brittle nature, leading to reduced deposition, increased fragmentation, and disruption of local packing. These effects collectively increase porosity and reduce overall deposition efficiency. Thus, excessive ceramic content limits coating growth by decreasing the amount of ductile Cu available to accommodate and anchor ZrC particles. These observations are consistent with earlier reports on cold-sprayed metal–ceramic composites [26–28]. The schematic representation of the coating deposition behaviour across low, moderate and high ceramic loading conditions is shown in Fig. 4, further illustrating how the evolving role of ZrC particles governs the build-up, retention, and overall deposition efficiency.

The cross-sectional SEM images along with the corresponding elemental distribution maps of various ZrC–Cu coatings on Al-6061 substrates for 5 passes are presented in Fig. 5. Overall, the deposited coatings appear dense, without visible cracks or voids, and exhibit continuous adherence to the substrate. The EDS mapping further confirms the presence of two distinct phases throughout the coatings, with ZrC particles evenly dispersed across the cross-section. The elemental maps also reveal that the amount of ZrC observed within the coatings increases with higher ZrC content in the feedstock blend.

3.2. Deposition efficiency, phase-retention and porosity

Cold spray is well regarded for achieving exceptionally high deposition efficiencies often approaching 100% in ductile metallic systems [29,30]. However, when depositing metal–ceramic composites, the DE typically decreases due to the non-deformable nature of ceramic

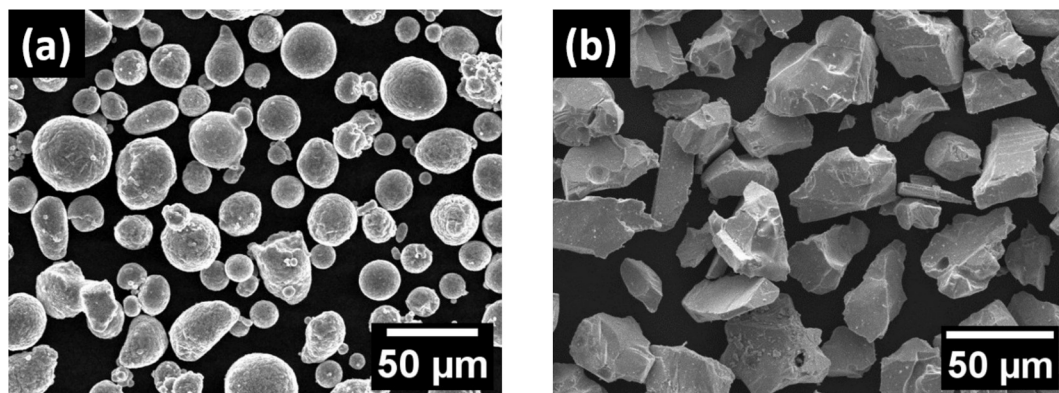


Fig. 1. SEM images showing morphology of (a) Cu and (b) ZrC feedstock used for deposition in cold spray.

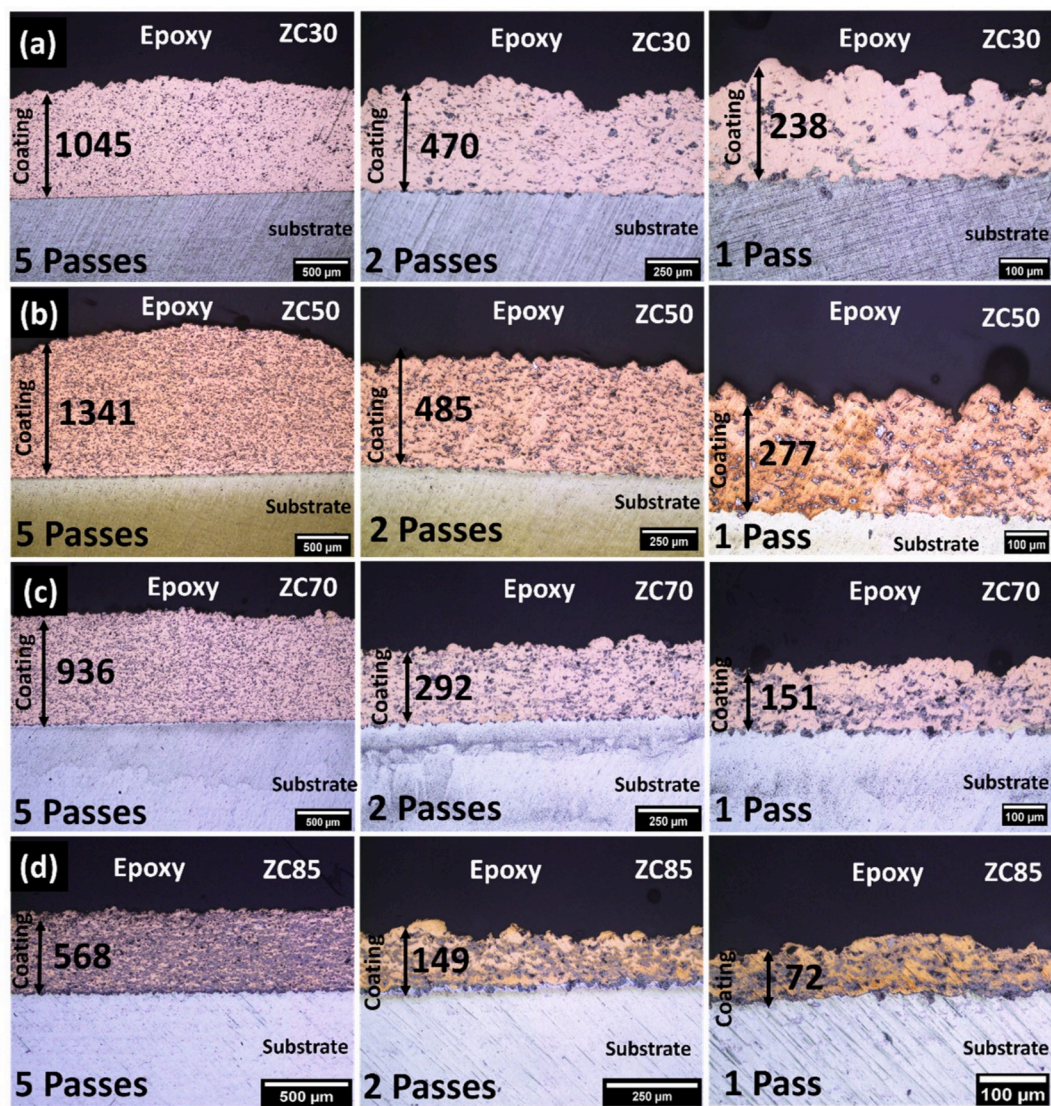


Fig. 2. Optical cross-sections of (a) Cu-30%ZrC, (b) Cu-50%ZrC, (c) Cu-70%ZrC and (d) Cu-85%ZrC cold spray coatings on Al-6061 alloy substrate after 1, 2, and 5 spray passes. ZrC appears as dark particles; Cu matrix is bright-yellow. Coating thickness (in μm) is also indicated. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

particles. Achieving high DE while simultaneously retaining a significant fraction of the ceramic phase therefore becomes a key challenge. The deposition efficiency for different compositions is depicted in Fig. 6. The overall deposition efficiency (DE) of composite coatings is calculated by taking the ratio of the difference in weight of the substrate before and after coating with the total powder added in the feeder. From the deposition efficiency curve shown in Fig. 6, it is visible that with increasing ZrC content in the blend result in overall decrease in DE. This is due to the fact that co-deposition of metal-ceramic is a complex phenomenon driven by various factors such as, oxide layer-break-up of the substrate, plastic deformation of the metal (Cu-binder), interlocking of the ceramic (ZrC-reinforcement), ceramic-metal interaction and tamping effect [19]. Many studies have confirmed that increasing ceramic content in the blend decreases the deposition efficiency in general due to more rebounding of non-deforming ceramic particles when they hit with the previously deposited (interlocked) ceramic particles and possibly due to ceramic-metal interactions [23]. This limitation arises because ceramic particles such as ZrC cannot readily bond with one another ZrC particle during cold spraying. Ceramic particles can adhere only when impacted against a ductile Cu surface. As a result, less thickness is obtained when ZrC percentage was increased to

85% due to absence of sufficient metallic binder, for same number of passes. Therefore, as the ZrC content in the feedstock blend increases, the likelihood of ZrC particles striking and rebounding off each other also increases. These rebounded particles may undergo fragmentation, which explains the higher concentration of fractured ZrC observed in the micrographs at elevated ceramic percentages and resulting in slightly higher porosity. High porosity thus, is due to brittle fracture, fragmentation, and particle rebound during impact causing ceramic fracture. However, there is an increase in DE when the ZrC is increased to 50% after which the DE decreased drastically at 70% of ZrC which is even lower than DE for 30% ZrC coating. Similar increment in DE has been observed in other studies as well with a maximum in the DE with increase in ceramic content and then a decreasing trend. This increase is often facilitated by the tamping effect of the non-deformable ZrC particles helping in deformation of metal binder and more intact coating [28]. Similar trends in the coating are observed in the thickness achieved in the corresponding coatings shown in Fig. 2.

The porosity analysis and ZrC-retention in the coatings was performed by the cross-sectional optical images of the coatings shown in Fig. 7. Ten images per coatings were taken to calculate the porosity and ZrC-retention and mean along with standard deviation is reported. In all

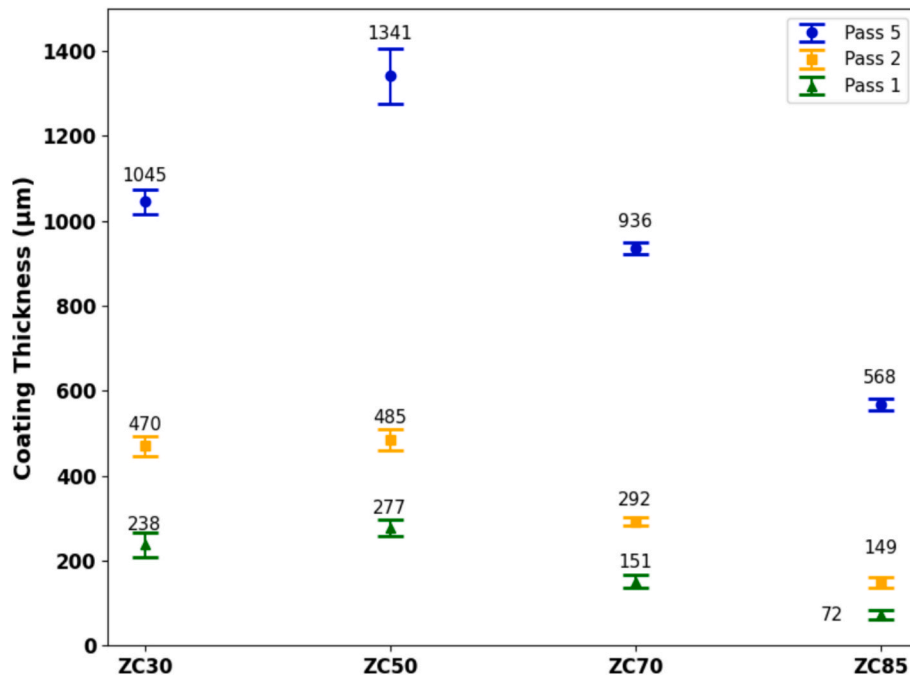


Fig. 3. Coating thickness versus number of passes with varying ZrC composition of coating depicting how coating thickness changes with change in ZrC content in the cold spray coatings.

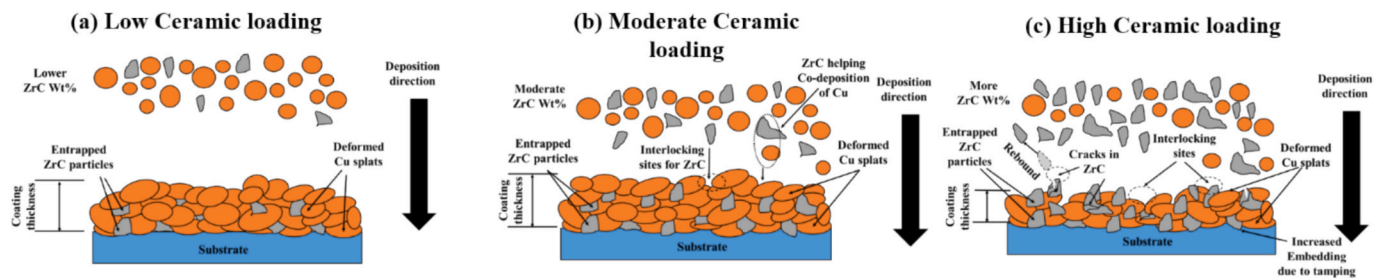


Fig. 4. Schematic illustration of cold-spray deposition behaviour in Cu-ZrC composite coatings under three different ceramic loading conditions: (a) Low ceramic loading: The coating is dominated by ductile Cu particles, which deform readily upon impact and efficiently trap the limited amount of ZrC reinforcement. This results in stable build-up and higher coating thickness. (b) Moderate (optimal) ceramic loading: Increased ZrC content enhances the tamping/peening effect on Cu, promoting stronger particle deformation, improved inter-splat bonding, and effective retention of ceramic particles within the Cu matrix. (c) High ceramic loading: Excess ZrC leads to frequent ceramic-ceramic impacts, causing particle fracture, rebound, and limited consolidation. The reduced availability of ductile Cu hinders bonding, resulting in lower deposition efficiency, higher porosity, and reduced coating thickness.

cases, two distinct phases are observed namely dark grey ZrC particles dispersed within the continuous yellow Cu matrix. Additionally, small black regions indicate the presence of pores. For the ZC30 coating (Fig. 7 (a)), the ZrC particles are relatively sparsely distributed, well embedded in the Cu binder, with only limited porosity ($\sim 0.9\%$). With an increase in ZrC to 50 wt% (Fig. 7 (b)), the number of ceramic particles increases, and the matrix remains sufficiently continuous to accommodate them. However, porosity begins to rise ($\sim 1.1\%$) due to partial ceramic fracture and incomplete filling of interstices. At 70 wt% ZrC (Fig. 7 (c)), clustering of ZrC particles becomes more evident, with localized ceramic-rich regions and a notable increase in porosity to nearly $\sim 1.5\%$. Finally, in ZC85 (Fig. 7 (d)), the coating appears highly ceramic-rich, with a dense network of ZrC particles and reduced Cu continuity. The excessive ceramic content leads to significant porosity ($\sim 1.9\%$) as brittle fracture and particle rebound dominate the deposition process. These observations are in line with earlier findings in cold-sprayed metal-ceramic systems, where higher ceramic fractions improve reinforcement retention but compromise coating density due to ceramic particle fragmentation [31,32]. The results also correlate with the deposition efficiency trend, showing that compositions between 30–50 wt% ZrC

achieve better balance between reinforcement incorporation and coating compactness.

3.3. Mechanical characterization

3.3.1. Microhardness

Vickers micro-hardness test was conducted on the cross-section of the deposited coatings to assess the effect of reinforcement in composite coatings. The micro-hardness values for different coatings along with the different number of passes is shown in Fig. 8. The results reveal a systematic increase in hardness with higher ZrC content. For instance, ZC30 coatings exhibit values in the range of 139–147 HV, while ZC85 coatings achieved values lying around 200 HV in all passes. This enhancement arises from two primary factors: (i) increased incorporation of the hard ZrC phase, and (ii) strain hardening of the Cu matrix due to severe plastic deformation induced by the tamping effect of the ZrC particles [33]. The influence of the number of passes is also depicted in Fig. 8. With increasing number of passes, the material undergoes further densification due to repeated particle impingement, localized plastic deformation, and inter-splat compaction [34]. For instance, ZC85

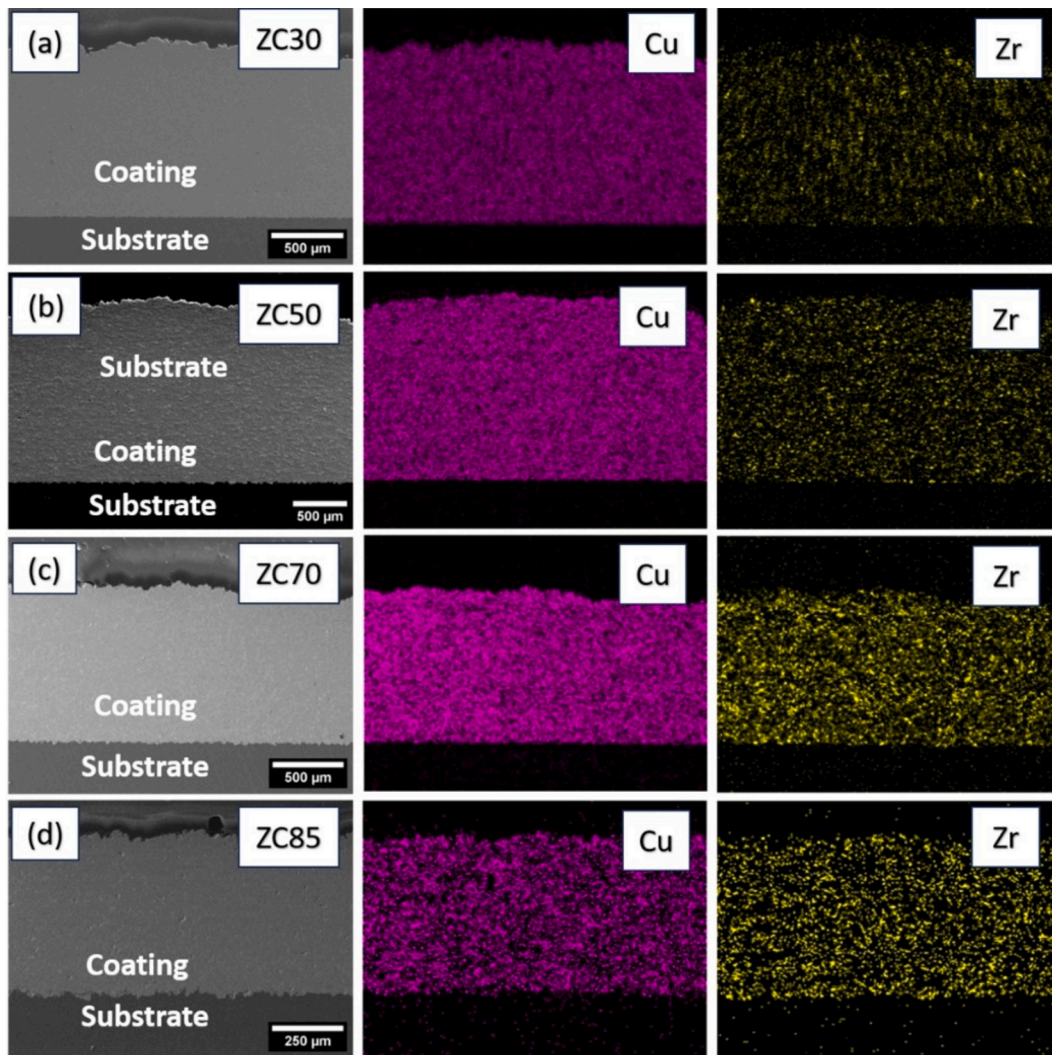


Fig. 5. The cross-sectional SEM images along with the corresponding elemental distribution maps of various (a) Cu-30%ZrC, (b) Cu-50%ZrC, (c) Cu-70%ZrC and (d) Cu-85%ZrC cold spray coatings on Al-6061 substrates for 5 passes showing dense microstructure, good adherence to the substrate, and homogeneous distribution of ZrC throughout the coating.

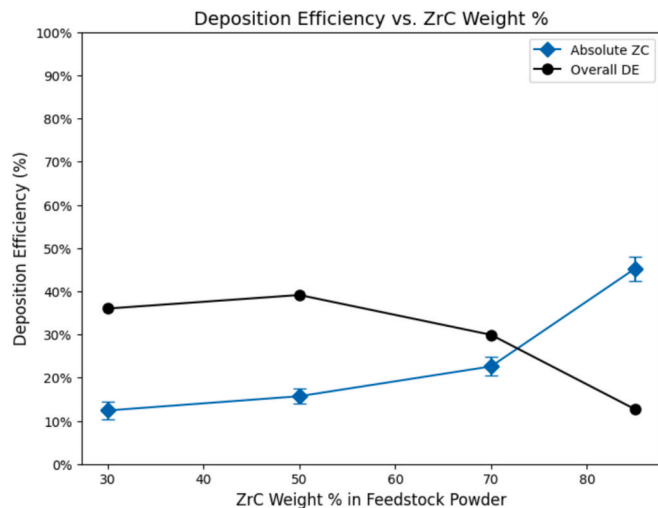


Fig. 6. Deposition efficiency with respect to ceramic retention in the cold spray coatings showing decrease in deposition efficiency with increasing ceramic in the powder blend with highest deposition efficiency at 50% of ZrC in the feedstock blend.

coatings show an increase in hardness from 175.8 HV at 1 pass to 205.3 HV at 2 passes. However, a slight reduction (~ 198.6 HV) is observed at 5 passes, likely due to indents falling on regions with localized porosity, ductile copper or micro-crack formation in the ceramic-rich system. Also, in these kinds of composite coatings the trend is not always ideal as the indent can fall on different areas which are locally rich in ZrC, at interface of the two phases or at the porous rich region. These findings are in agreement with earlier studies in cold spray, where ceramic reinforcement has been shown to substantially increase hardness, but with the trade-off of increased porosity and brittleness when the ductile metallic binder fraction becomes insufficient for plastic accommodation [35]. Thus, the present findings demonstrates that intermediate compositions such as ZC70 offer a balance between high hardness and structural integrity, whereas ZC85 yields the maximum hardness but at the cost of slightly higher porosity. Overall, the micro-hardness results not only highlight the strengthening role of ZrC but also provide insights into optimizing the ceramic-metal ratio and deposition strategy for achieving superior erosion and wear resistance in service applications.

3.3.2. Adhesion strength

The adhesion strength of the coatings was evaluated through pull-off testing for the five deposition passes coatings, and the results are summarized in Fig. 9. For the ZC30 coating, the adhesion strength is ~ 38.5

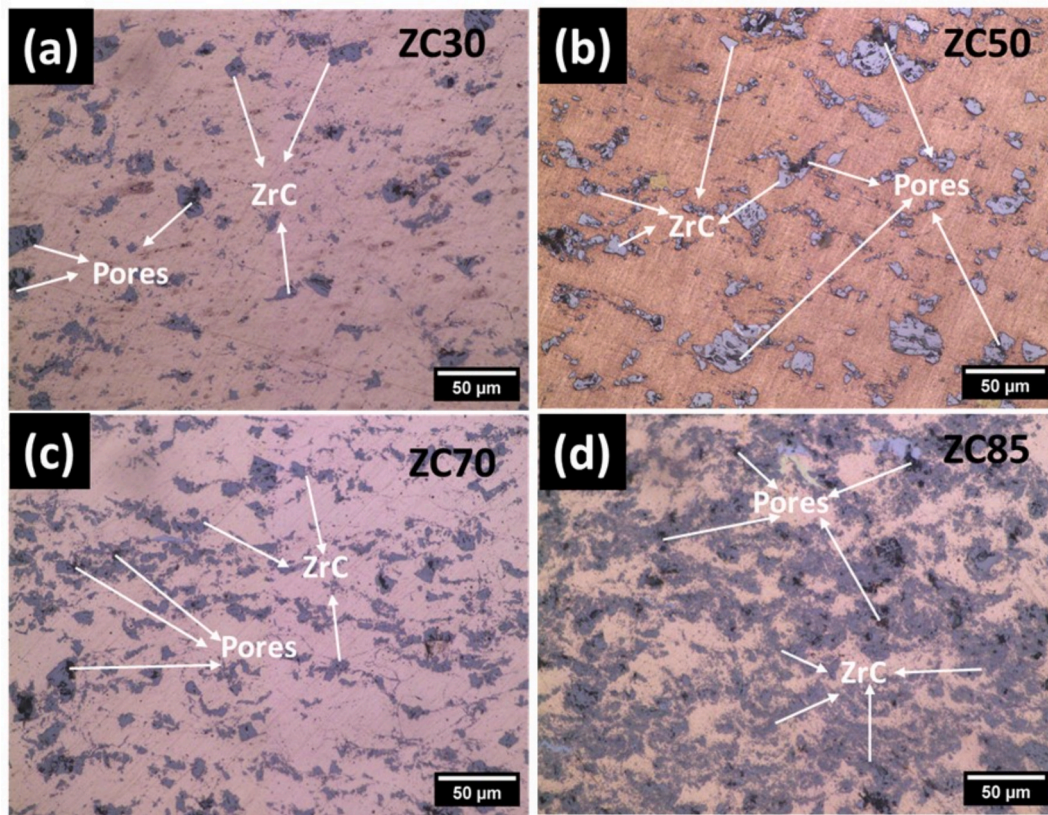


Fig. 7. Cross-sectional optical images of cold sprayed (a) Cu-30%ZrC, (b) Cu-50%ZrC, (c) Cu-70%ZrC and (d) Cu-85%ZrC coatings on Al-6061 alloy substrate, showing increase in ZrC retention in the coatings with increase in ZrC content in the feedstock blend; grey coloured particles are marked as ZrC distributed in the yellow copper matrix along with black regions corresponding to micro-pores. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

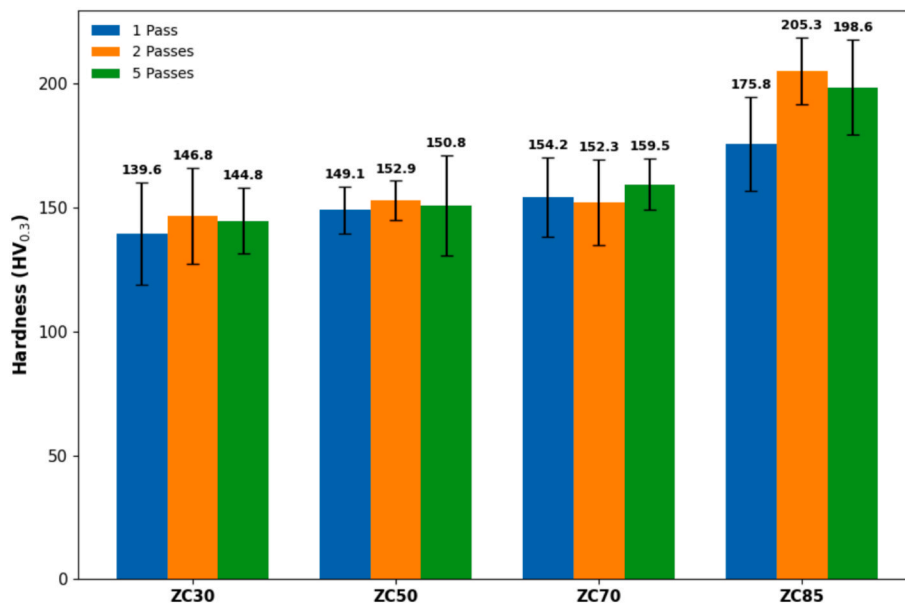


Fig. 8. Microhardness versus feedstock blend composition with different number of passes, showing increase in microhardness with increase in ZrC in the feedstock blend due to increased retention of ZrC in the coatings.

± 6 MPa, and it increases with ceramic content up to 70%, reaching $\sim 43.2 \pm 5$ MPa for ZC50 and $\sim 47.3 \pm 5$ MPa for ZC70, before showing a slight decline to $\sim 44.4 \pm 7$ MPa for ZC85. For comparison, pure copper coatings exhibit an adhesion strength of $\sim 34 \pm 8$ MPa, indicating that

the incorporation of ZrC progressively improves interfacial bonding up to an optimum reinforcement level. The failure modes observed during pull-off testing further clarify the role of ceramic content in interfacial bonding. Pure Cu and the low-ceramic ZC30 coating exhibited primarily adhesive failure at the coating-substrate interface, whereas all other

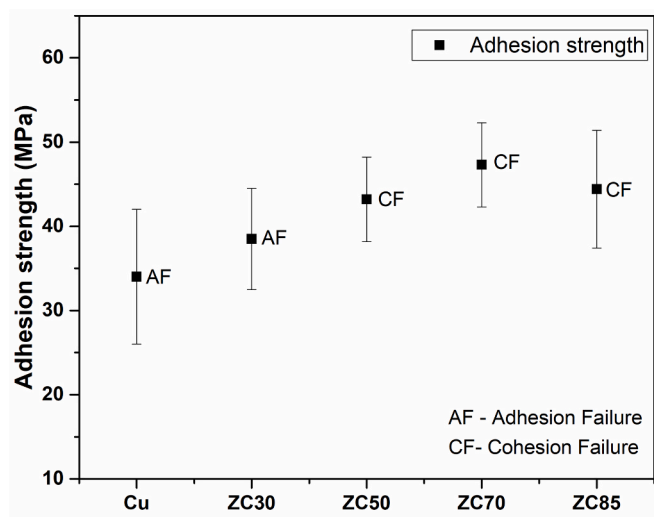


Fig. 9. Adhesion strength obtained from pull-off adhesion test on all the coated samples including the pure copper and all the ZrC blends for five deposition passes. Type of failure i.e., adhesion failure/cohesion failure is also indicated.

compositions (ZC50, ZC70, ZC85) showed cohesive failure, as indicated in Fig. 9. This transition in failure mode indicates that the addition of ZrC particles has a strong activation effect on the substrate surface, thereby enhancing adhesion strength in the ZrC-Cu coatings. This improvement can be attributed to two primary mechanisms: (i) surface roughening of the substrate caused by the high-velocity impact of hard ZrC particles, and (ii) breakup and removal of surface oxide films, which facilitates intimate metallic contact [36,37].

Both these mechanisms increase the likelihood of metallurgical bonding and mechanical interlocking between the substrate and the deforming Cu binder. Additionally, the high-velocity ceramic impacts can disrupt oxide films not only on the substrate but also on the Cu feedstock particles themselves. This enhances chemical and

metallurgical activity at inter-particle interfaces, promoting stronger inter-splat bonding and improved coating cohesion. The interfacial optical images shown in Fig. 10, also support these findings. In the Cu-rich coatings such as ZC30, the coating-substrate interface appears relatively smooth with limited anchoring features, although local mechanical interlocking is visible in certain regions. With higher ZrC additions particularly in ZC70 and ZC85, pronounced interfacial waviness and embedded ceramic particles were observed, which enhance the anchoring effect and contribute to the maximum adhesion strength measured. However, for ZC85, although the interface shows deep anchoring zones (red-circled regions), the overall adhesion strength decreased. This drop can be attributed to excessive ceramic in the coating, which reduces the metallic binder fraction, thereby limiting the ability of copper to bridge up and promote strong metallurgical and mechanical bonding [28]. In addition, brittle fracture in ceramics and weak particle-particle bonding in ceramic-rich coatings can act as stress concentrators, reducing effective adhesion. These results are in good agreement with previous studies on cold-sprayed metal-ceramic systems, where an optimal ceramic content improves adhesion due to enhanced surface roughness and interlocking, but excessive ceramic addition deteriorates bonding by reducing ductile bridging and particle consolidation [28,36].

Further evidence for the transition in failure behavior was obtained from post-fractography SEM analysis of the fractured surfaces after pull-off testing, as shown in Fig. 11 for Cu, ZC30, and ZC85 coatings. The low-magnification SEM images of Cu and ZC30 coatings shown in Fig. 11(a) and (b) exhibit relatively uniform ductile tearing features, indicating predominant adhesive failure at the coating-substrate interface. In contrast, ZC85 shown in Fig. 11 (c), exhibits an adhesive failure with a partial coating pull out and localized fracture within the coating. This can be seen more clearly in high magnification depicted in Fig. 11 (d), (e) and (f), the Cu and ZC30 show clear features of detachment with slight Al signals in EDS, however in higher ZrC coatings (ZC85), rough fractured regions along with fragmented ceramic-rich zones were observed, confirming cohesive fracture within the brittle ceramic-rich coating. These observations validate the measured adhesion trends

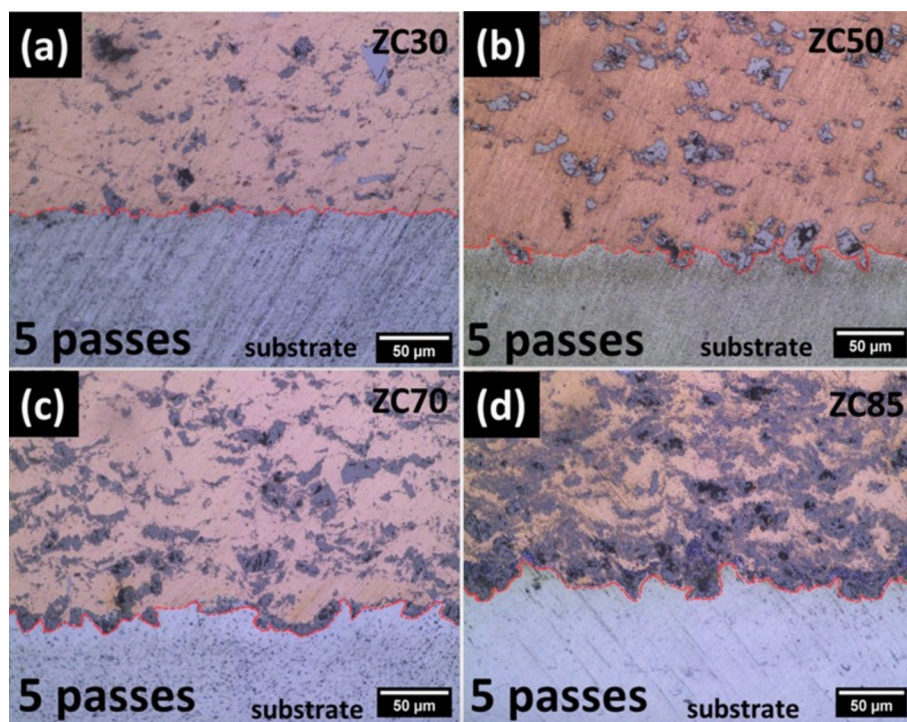


Fig. 10. Cross-sectional optical images of coating-substrate interface for cold sprayed (a) Cu-30%ZrC, (b) Cu-50%ZrC, (c) Cu-70%ZrC and (d) Cu-85%ZrC coatings on Al-6061 alloy substrate showing change in particle embedment in the substrate with increase in ceramic content.

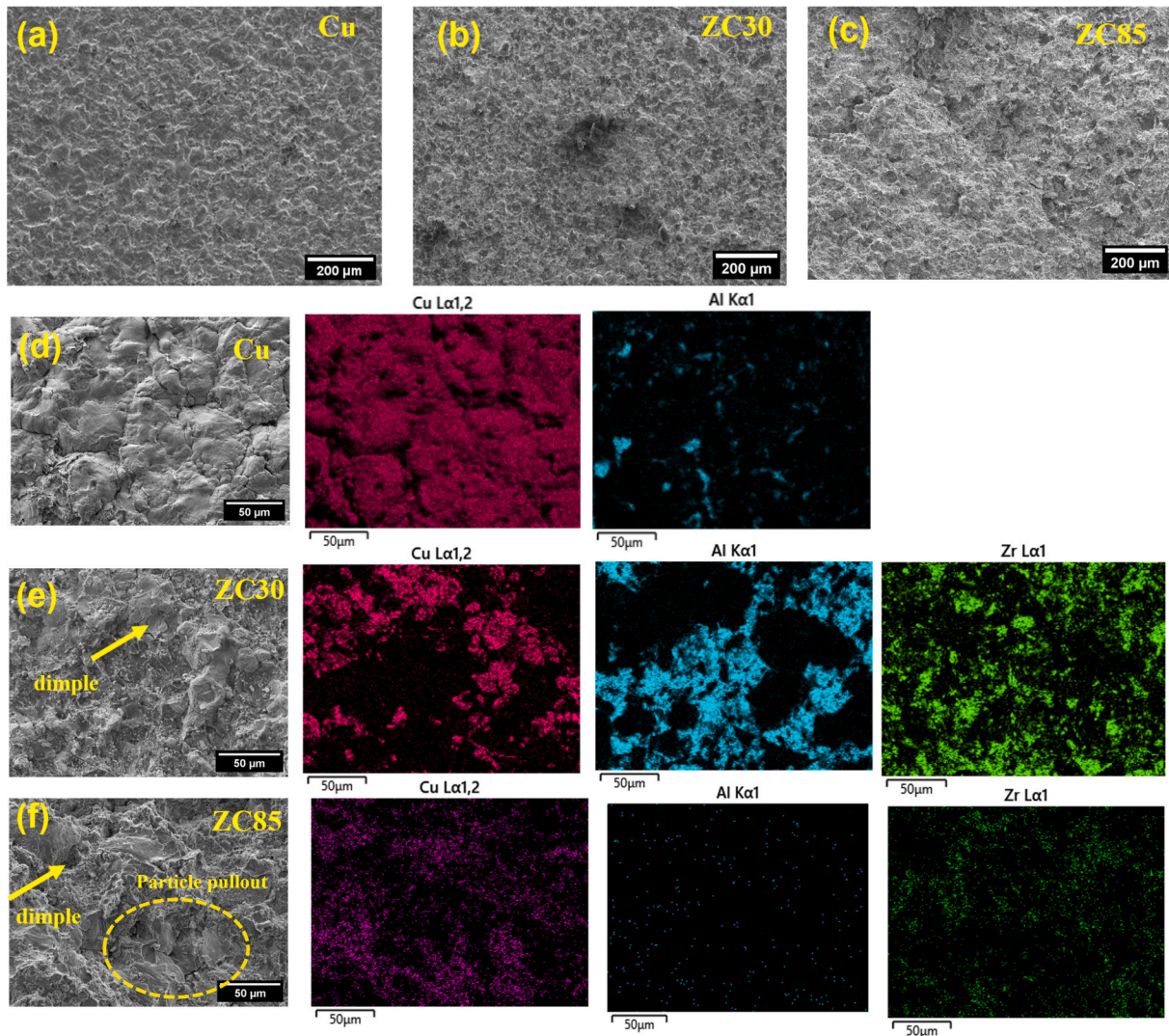


Fig. 11. Post-test fractography, low magnification SEM images of (a) pure Cu, (b) Cu-30%ZrC, (c) Cu-85%ZrC; high magnification SEM images with EDS of (d) pure Cu, (e) Cu-30%ZrC, (f) Cu-85%ZrC cold spray coatings on Al-6061 alloy substrate, showing adhesion failure in pure Cu and coating with 30 wt% ZrC while cohesive failure in the coatings with high content of ZrC in the coatings.

and confirm that the failure mode transitions from predominantly interfacial in Cu-rich coatings to cohesive in high-ZrC coatings. In summary, the adhesion strength of the coatings demonstrates a non-linear trend with ZrC content, where a moderate ceramic content (70 wt%) provides the most effective balance between the ductility of copper and the reinforcing role of ZrC, ensuring both strong bonding and effective mechanical interlocking at the interface.

3.3.3. Scratch analysis

The scratch test is among the most reliable technique for assessing the cohesion and adhesion strength of coatings. Its repeatability and accuracy make it a dependable tool for evaluating interfacial integrity [38]. In scratch test, resistance to deformation of a material is measured against a moving indenter under an applied load. When the indenter traverses the coating with a constant or progressive load, the resistance of the material to the applied normal load is expressed as the coefficient of friction (CoF). In cold-sprayed coatings, the onset of failure is typically sensed by a sudden increase in frictional force and acoustic emission, detectable through friction and acoustic monitoring and post scratch image analysis via SEM or optical microscopy [38]. The fluctuations observed in CoF are associated with variations in the true contact area and the generation of wear debris [39]. By examining scratch worn

tracks through SEM, the wear width, wear mechanisms and overall wear resistance of the coating can be correlated with the frictional response, providing indirect evidence of coating performance under tribological conditions [40]. Pitchuka et al. [41] analysed cold-sprayed aluminium coatings and reported that CoF increased with the true or instantaneous penetration depth of the indenter. This behaviour was attributed to inter-splat failures that alter the contact area and generate wear debris. Ng et al. [39] similarly observed a rise in CoF from 0.2 to 0.6 towards the end of the scratch test, linked to splat loosening, removal, and detachment. The combined effects of increased penetration depth and inter-splat failure increase the traction forces, leading to a higher value of CoF. Since, CoF correlates with scratch penetration depth and wear debris formation, SEM examination of the worn tracks was performed to identify the dominant wear mechanisms. These observations collectively reveal the dynamic behaviour of coatings under varying mechanical loads and provide insight into their adhesion strength and wear resistance.

Fig. 12, depicts the evolution of CoF for various cold-sprayed composite coatings under (a) constant load at 40 N and (b) progressive 20–60 N load conditions. Under constant load, pure Cu exhibited the highest CoF (~0.55–0.6) and substantial fluctuation, indicative of severe plastic deformation and adhesive wear due to its soft matrix. With

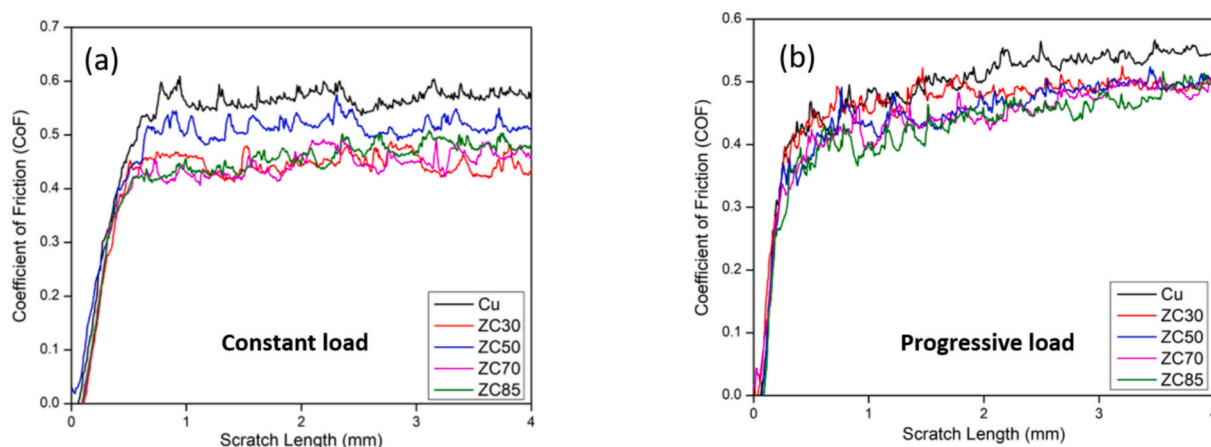


Fig. 12. Scratch test showing the variation in the coefficient of friction (CoF) for the cold-sprayed coatings under (a) a constant load of 40 N over a scratch length of 4 mm, and (b) a progressive load ranging from 20 N to 60 N at a loading rate of 10 N/mm over the same scratch length of 4 mm.

increasing ZrC content, the CoF consistently decreased with ZC70 and ZC85 maintaining CoF values below 0.45. This trend reflects enhanced load-bearing capacity and surface stability imparted by hard ceramic particles, which inhibit plastic flow and ploughing. The fluctuations in the coatings with ZrC can be attributed to the indenter experiencing different regions inside the coatings during scratch including copper ductile phase, ZrC hard phase and inter-splat failures.

The progressive load test further emphasized this effect. As the load increased from 20 N to 60 N, CoF for pure Cu rose steadily and remained the highest among all samples, while ZrC-reinforced coatings exhibited more gradual increases and slightly stable profiles. In particular, ZC85 sustained a low CoF throughout the scratch, showcasing its superior resistance to deformation under constant load. These observations are supported by prior studies, which demonstrate that the incorporation of ceramic reinforcements such as ZrB_2 significantly improves tribological behaviour in metal matrix composites by enhancing surface hardness and reducing the real area of contact [42]. The scratch width and depth of all the coatings taken by optical 3D profilometer is shown Fig. 13 and summarised in Table 3. The corresponding scratch width and depth measurements also supports these observations. Under a constant 40 N load, Cu exhibited the widest ($\sim 315 \mu\text{m}$) and deepest ($\sim 22 \mu\text{m}$) scratches, confirming lowest resistance to deformation among all coating compositions. ZrC-reinforced coatings, especially ZC70 and ZC85, showed significantly narrower widths ($\sim 227\text{--}223 \mu\text{m}$) and shallower depths ($\sim 18\text{--}19 \mu\text{m}$), indicative of higher hardness and structural integrity. This trend persisted under progressive loading. Cu showed the largest scratch width ($\sim 383 \mu\text{m}$) and depth ($\sim 34 \mu\text{m}$), whereas ZC85 maintained the smallest width ($\sim 276 \mu\text{m}$) and depth ($\sim 28 \mu\text{m}$). These results highlight the superior wear resistance and load-bearing capability of ZrC-rich coatings, which effectively hinder material removal even under rising normal loads. Overall, the improved performance of ZrC–Cu composites under both constant and progressive loading arises from the synergistic effects of the hard ZrC phase, which enhances hardness, distributes stress more uniformly, and reduces frictional instability. These findings are in agreement with previous research showing that ZrC additions to copper matrices significantly reduce wear rates and friction coefficients due to constrained plasticity and formation of protective debris layers.

3.3.4. Wear behaviour and wear mechanisms

The tribological behaviour of pure Cu and ZrC–Cu composite coatings was systematically evaluated using ball-on-disk testing. The wear behaviour is analysed by correlating the CoF, 3D wear-track profilometry, wear volume and wear-track morphology examination by SEM examination. The combined analysis provides a comprehensive understanding of the deformation-controlled wear mechanisms governing

material removal in these coatings. The coefficient of friction curves for different coatings are presented in Fig. 14. The CoF curves depicts two phases: an initial run-in phase and comparatively stable phase therein. In the initial run-in stage, as the ball makes contact and gets embedded into the coating and increased area of contact with coating leading to a gradual rise in the CoF. Pure Cu exhibited a sharp rise in the CoF and then decreased to a stable value. This can be attributed to the ductile nature of Cu where the deformation is dominated by ploughing induced by plastic flow of material and ball penetration into the coating. Therefore, initially during the first few seconds in the beginning of wear test the material might have piled up and resulted in increase in CoF and then gradually a stable groove with surface smearing. Also, the CoF curve of Cu is relatively smooth as compared to the Cu–ZrC composite coatings owing to a single material interaction with the applied load, while in the ZrC-based coatings the fluctuation in the traction force arises due to the interaction of ball with material of different hardness. The ZrC-based coatings showed a mixed response of CoF variation. From the Fig. 14, it can be clearly seen that CoF increased in general with the incorporation of hard ZrC ceramic reinforcement. The CoF for ZC30, ZC50 and ZC 70 is higher than pure Cu. It is expected that low CoF corresponds to lower wear rate and high CoF corresponds to high wear rate. However, there is no general rule and several reported results of contradictory behaviour [43–46]. This trend has been observed in similar studies done by Zhou et al. for Cu– Al_2O_3 composites [43]. The addition of hard ceramic results in asperities exposed to the coating surface during the wear test which may increase the friction force during the test. The exposed protrusions of ZrC can therefore hold the wear process and the CoF of the composite coatings become more than pure copper [46]. The ZC85 coating showed the least CoF which might have resulted from the higher ZrC content leading to suppression of ductile behaviour of pure Cu.

To further investigate the wear mechanism of the coatings, the wear track morphologies were measured with a 3D profilometer as shown in Fig. 15. The track width, depth and wear volume are also summarised in Table 4 and wear volume versus the composition of coatings is shown in Fig. 16. With the incorporation of ZrC, a progressive improvement in wear resistance was observed. The pure Cu showed a wear depth of $\sim 12 \mu\text{m}$, width of $\sim 842 \mu\text{m}$ while the ZC85 coating showed a depth of $\sim 1.7 \mu\text{m}$, width of $\sim 572 \mu\text{m}$. Compared to pure Cu ($\sim 0.096 \text{ mm}^3$), the wear volume decreased to $\sim 0.068 \text{ mm}^3$ for ZC30, $\sim 0.050 \text{ mm}^3$ for ZC50, $\sim 0.035 \text{ mm}^3$ for ZC70, and $\sim 0.007 \text{ mm}^3$ for ZC85. Notably, the ZC85 coating exhibited an approximately 92.7% reduction in wear volume relative to pure copper, highlighting the strong influence of ceramic reinforcement on the wear resistance of the coatings. This phenomenon is directly correlates with the microhardness of the coatings. The Archard's law also states that there is an inverse relationship

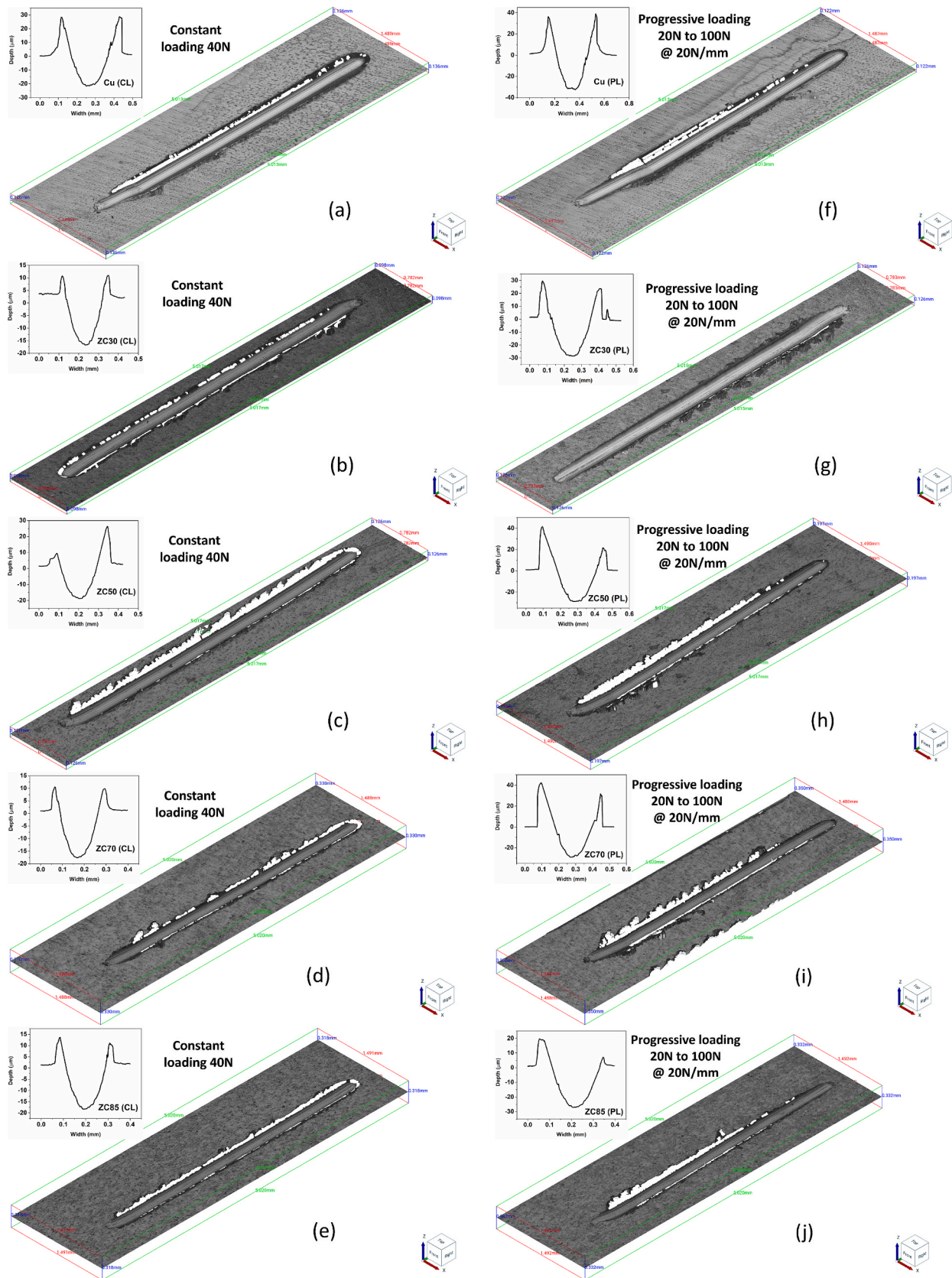


Fig. 13. 3D optical images showing maximum scratch width and depth for different coatings at (a) pure Cu (b) Cu-30%ZrC, (c) Cu-50%ZrC, (d) Cu-70%ZrC and (e) Cu-85%ZrC constant loading (f) pure Cu, (g) Cu-30%ZrC, (h) Cu-50%ZrC, (i) Cu-70%ZrC and (j) Cu-85%ZrC progressive loading measured from 3D profilometer.

Table 3

Maximum scratch width and depth for different coatings at constant and progressive load.

Constant load 40 N		
Coating	Width (μm)	Depth (μm)
Cu	315 $\pm$ 14	22 $\pm$ 5
ZC30	230 $\pm$ 18	20 $\pm$ 3
ZC50	265 $\pm$ 20	20 $\pm$ 5
ZC70	227 $\pm$ 19	18 $\pm$ 4
ZC 85	223 $\pm$ 14	19 $\pm$ 5
Progressive load		
Coating	Width (μm)	Depth (μm)
Cu	383 $\pm$ 20	34 $\pm$ 6
ZC30	331 $\pm$ 18	30 $\pm$ 5
ZC50	352 $\pm$ 24	30 $\pm$ 6
ZC70	350 $\pm$ 21	29 $\pm$ 4
ZC 85	276 $\pm$ 17	28 $\pm$ 4

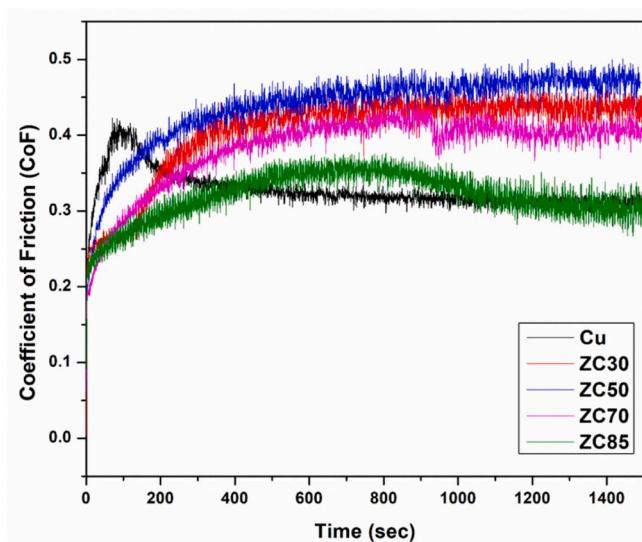


Fig. 14. Evolution of coefficient of friction versus time during ball-on-disk wear test showing the variation of CoF with different Cu/Cu-ZrC composition in cold spray coatings on Al-6061 substrates at 20 N load for 25 min.

between wear rate and surface hardness [47]. Consequently, the wear resistance of the cold sprayed coatings depends on the volume of hard phase particles, the coating hardness as well as the coating density and microstructure [48].

The SEM analysis of the wear tracks were done for different samples as shown in Fig. 17. This provides microstructural evidence of wear trends inferred from CoF, wear volume, and profilometer results to identify the existing wear mechanism in each coating composition. The SEM images of wear scars for ZC30, ZC50, ZC70, and ZC85 coatings shows systematic decreasing width of wear tracks as the ZrC content in the coating increased. Pure Cu exhibited the widest wear tracks ($\sim 842 \mu\text{m}$), followed by ZC30 ($\sim 836 \mu\text{m}$), ZC50 ($\sim 703 \mu\text{m}$), ZC70 ($\sim 670 \mu\text{m}$) and ZC85 ($\sim 572 \mu\text{m}$). The same trend is observed in the wear depth and volume examination. The SEM image for ZC30 of the wear scar revealed extensive plastic smearing, surface tearing, and adhered material patches [49]. It appears that due to more ductile Cu in the coating the material got piled up due to continuous ploughing and then smeared patches got re-adhered again to the wear track. This shows the characteristic of severe adhesive wear dominated by plastic deformation of the Cu matrix [45]. The smeared wear debris which looks like loosely bonded and accumulated along the track shown in Fig. 17a, further indicates the characteristic of severe adhesive wear dominated by plastic deformation of the Cu matrix [38]. These observations are consistent with the low hardness, large scratch penetration depth, and poor

resistance to plastic deformation. The Z50 coating, shown in Fig. 17b, exhibits a well-defined track with a reduced width and relatively smoother wear-track morphology. SEM image also show a localized spallation within the wear track, appearing as an isolated flake-like delamination or ploughing and extrusion of material suggesting an adhesive wear mode [45]. Though, the spallation remains confined and does not propagate across the wear track which indicates that damage is not dominated much by brittle fracture. The ZC70 coatings, exhibit a more narrower wear track width with a smooth interior feature than ZC50. The SEM image shown in Fig. 17c, reveals presence of micro-void like feature or small pull-out pits within the wear track (marked by red circles). The track edges also show the brittle fragmentation of pile-up debris. This indicates that increasing ZrC content promoted discontinuous material accumulation and a gradual transition from ductile plastic deformation to brittle deformation at the track boundaries [22]. The ZC85 coatings show narrowest wear track ($\sim 572 \mu\text{m}$) and the wear morphology is also different from other coating compositions. The SEM image of the wear track shown in Fig. 17d, reveals that there is micro-crack like discontinuities (marked in red arrows), attributed to the brittle fracture in the coatings. In addition, the yellow arrows shown in Fig. 17d, show uplifted ridges aligned in the sliding direction within the wear tracks. This suggests the accumulation of sub surface stress accumulation and limited plastic deformation with repeated sliding contact with the composite systems where hard reinforcements induce stress in the ductile matrix in the high ZrC composition due to limited ductile Cu matrix. The wear track edges further display fragmented and discontinuous pile-up debris rather than ploughing. These features collectively highlight the transition to brittle fragmented wear regime, where the material removal is associated with micro-cracks, uplifted ridges and brittle fragmentation [22].

Overall, the combined COF, wear volume and wear track SEM analysis demonstrate that wear behaviour in ZrC-Cu composite coatings is governed by a balance between plasticity suppression and damage tolerance. While the increase in ceramic content (ZrC) effectively reduces wear volume by restricting plastic deformation. However, excessive ceramic loading promotes brittle fragmentation and localized damage. Therefore, although ZC85 coating exhibits substantially lowest wear volume but the dominant brittle damage mechanism highlights a reduction in damage tolerance when compared to intermediate compositions (ZC50/70). Thus, suggesting that minimal wear volume does not correspond to optimal tribological performance. Although ZrC particles contribute to strengthening, their limited fraction is insufficient to fully suppress matrix plasticity, resulting in a mixed adhesive–abrasive wear regime.

4. Distinct roles of scratch and sliding wear in assessing inter-splat bonding and damage evolution in cold-sprayed coatings.

Scratch testing and ball-on-disk wear testing are both widely used tribological techniques; however, they impose fundamentally different mechanical loading conditions on coatings and therefore probe different aspects of mechanical integrity and damage evolution. A combined interpretation of these tests is essential to understand how cold-sprayed ZrC-Cu composite coatings respond under localized monotonic loading and long-term cyclic sliding conditions. The schematic representation of the fundamental loading conditions and resulting deformation mechanisms observed during scratch and sliding wear testing is illustrated in Fig. 18.

In scratch testing, a hard indenter traverses the coating surface in a single pass under either constant or progressively increasing normal load [38], as schematically shown in Fig. 18a. The response is dominated by highly localized contact stresses beneath the indenter, where deformation occurs through splat displacement, localized pile-up formation, interfacial shearing, and crack initiation at weak inter-splat regions [40]. Consequently, the measured scratch width, penetration depth, and coefficient of friction (CoF) primarily reflect the resistance of

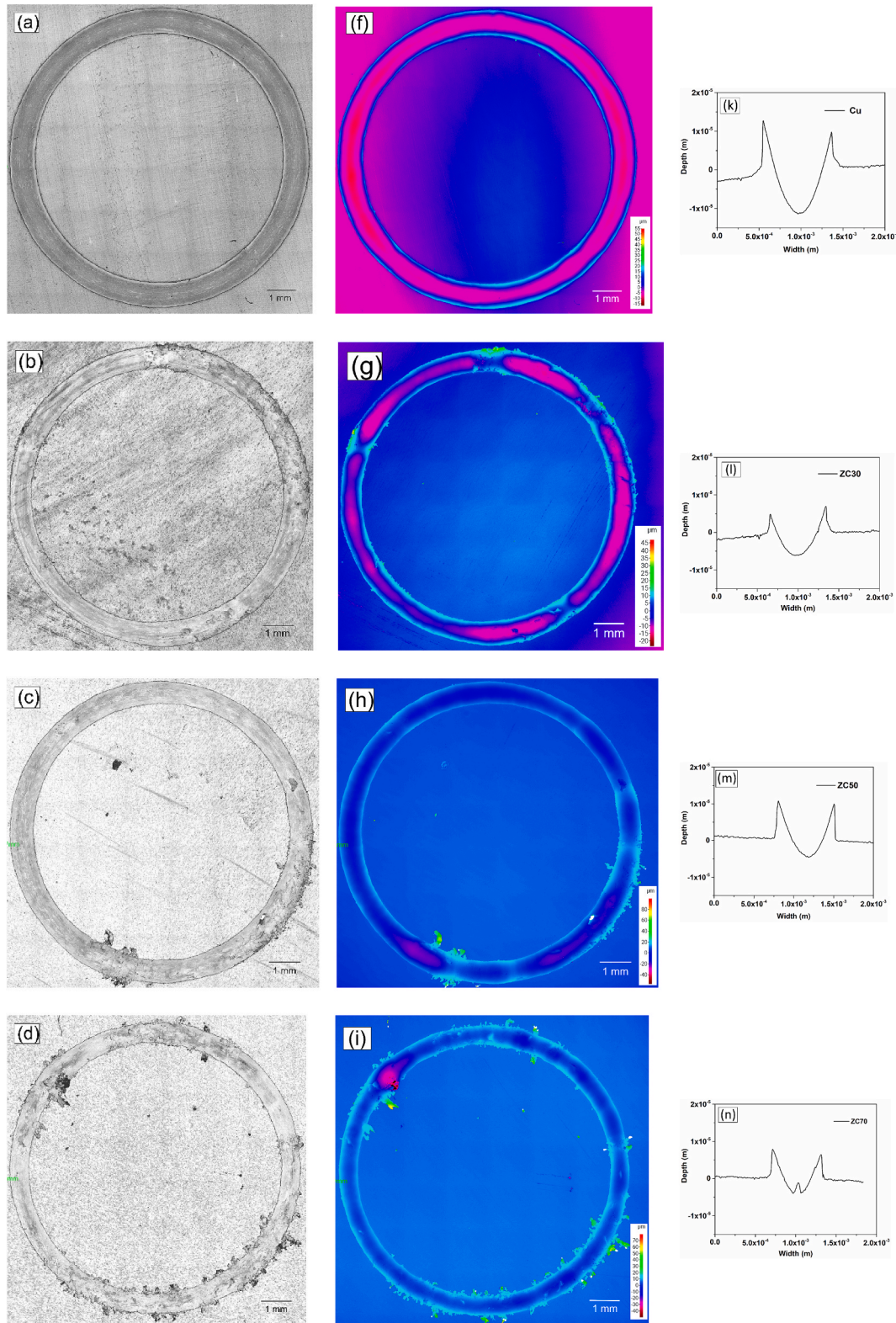


Fig. 15. Wear track of cold spray coatings on Al-6061 substrates (a) pure Cu (b) Cu-30%ZrC, (c) Cu-50%ZrC, (d) Cu-70%ZrC and (e) Cu-85%ZrC; and their respective profilometer images in (f) to (j) and depth profile in (k) to (o) showing the 3D profiles of the wear tracks and corresponding wear depths for different coating compositions.

the coating to localized deformation and material displacement. In cold-sprayed coatings, these responses are strongly governed by inter-splat cohesion, matrix continuity, and coating-substrate interfacial integrity.

Sudden fluctuations in CoF or abrupt changes in scratch morphology are often associated with localized splat pull-out, interfacial cracking, or fragmentation of weakly bonded regions [38]. Therefore, scratch testing

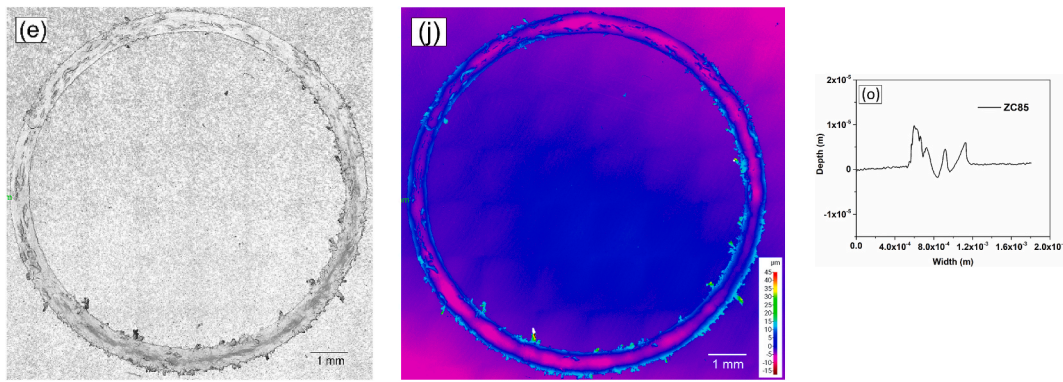


Fig. 15. (continued).

Table 4

Wear width, depth and volume for different Cu/Cu-ZrC coatings at 20 N load for 25 mins for ball-on-disk wear test.

Ball-on-disk wear test at 20 N load and 25 mins			
Coating	Width (μm)	Depth (μm)	Volume (mm^3)
Cu	842 ± 14	12 ± 2	0.096
ZC30	836 ± 12	6.1 ± 3	0.068
ZC50	703 ± 13	4.5 ± 1.7	0.050
ZC70	670 ± 21	3.6 ± 2	0.035
ZC 85	572 ± 19	1.7 ± 0.7	0.007

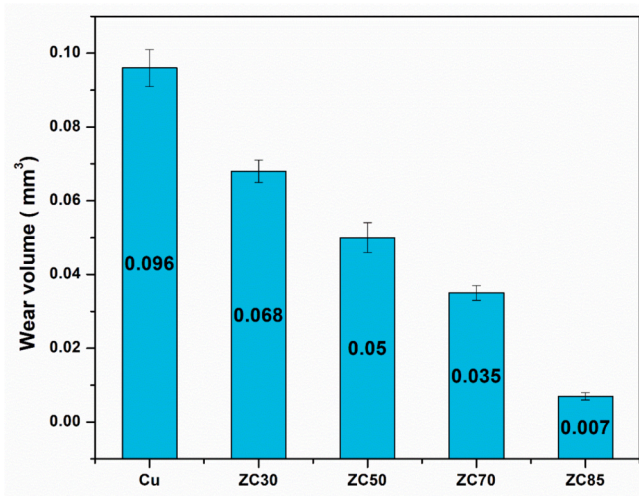


Fig. 16. Comparison of wear volume for different Cu/Cu-ZrC cold spray coatings on Al-6061 substrates composite coatings deposited by cold spray via ball-on-disk wear test. The wear volume is significantly decreased with increase in ZrC % in the coatings.

acts as a sensitive indicator of coating cohesion and localized deformation resistance rather than direct long-term wear performance. In contrast, ball-on-disk wear testing subjects the coating to repeated sliding contact over extended durations, generating cyclic shear stresses, progressive subsurface deformation, debris accumulation, and continuous material removal, as illustrated in Fig. 18b. Unlike scratch testing, where deformation occurs over a single loading event, sliding wear involves repeated asperity interaction and gradual stress accumulation within the coating. The resulting wear behaviour is therefore governed not only by hardness and instantaneous deformation resistance, but also by the ability of the coating to accommodate cyclic stresses without crack propagation or material detachment. Consequently, wear-track width, wear depth, wear volume, and steady-state friction behaviour

provide a more realistic assessment of long-term tribological performance under service-like conditions.

The transition between scratch-induced deformation and sliding wear damage evolution can be understood in terms of how the composite microstructure accommodates applied stresses. During scratch testing, Cu-rich coatings deform predominantly through plastic flow of the ductile matrix, resulting in wider and deeper scratches accompanied by significant pile-up formation, as schematically represented in Fig. 18c. This is evident from the reduction in scratch penetration depth and scratch width from $\sim 22 \mu\text{m}$ and $\sim 315 \mu\text{m}$ in pure Cu to $\sim 19 \mu\text{m}$ and $\sim 223 \mu\text{m}$, respectively, in ZrC-rich coatings under constant loading conditions. The substantial reduction in scratch width indicates effective suppression of lateral material displacement and localized plastic deformation with increasing ceramic content. Under cyclic sliding wear, however, the repeated contact progressively accumulates subsurface deformation and promotes adhesive transfer, material smearing, and debris generation, eventually leading to severe material displacement and larger wear-track volumes in Cu-rich coatings, as shown schematically in Fig. 18e. This behaviour is reflected in the large wear volume of pure Cu ($\sim 0.096 \text{ mm}^3$), where deformation is dominated by extensive plastic flow and adhesive wear. With increasing ZrC content, the hard ceramic phase progressively suppresses localized plastic deformation by constraining matrix flow and improving load-bearing capability. As a result, ZrC-rich coatings exhibit reduced scratch penetration depth, narrower wear tracks, and significantly lower wear volumes. The reduction in wear volume from $\sim 0.096 \text{ mm}^3$ for pure Cu to $\sim 0.007 \text{ mm}^3$ for ZC85 indicates that suppression of localized deformation strongly influences long-term wear resistance. However, the deformation mechanisms evolve significantly with increasing ceramic content. At intermediate compositions (ZC50–ZC70), the combination of improved inter-splat cohesion, sufficient metallic matrix continuity, and effective load sharing produces the most balanced tribological response. These coatings show relatively stable scratch behaviour, reduced material displacement, and lower wear damage while still maintaining sufficient damage tolerance under cyclic loading conditions. The schematic shown in Fig. 18d illustrates the reduced scratch width and limited pile-up formation in ceramic-rich coatings due to improved resistance against localized penetration. At very high ceramic loading (ZC85), although the coating exhibits the lowest wear volume and smallest scratch dimensions, the dominant damage mechanism changes from ductile deformation to brittle micro-fragmentation. The reduced continuity of the Cu matrix limits plastic accommodation during repeated sliding contact, causing localized stress concentration within ceramic-rich regions. Consequently, cyclic loading promotes microcrack initiation, fragmented pile-up debris, and localized brittle fracture features within the wear track, schematically illustrated in Fig. 18f. Importantly, the observed brittle fragmentation does not correspond to catastrophic coating delamination or severe material removal. Instead, the material removal occurs primarily through fine-scale micro-

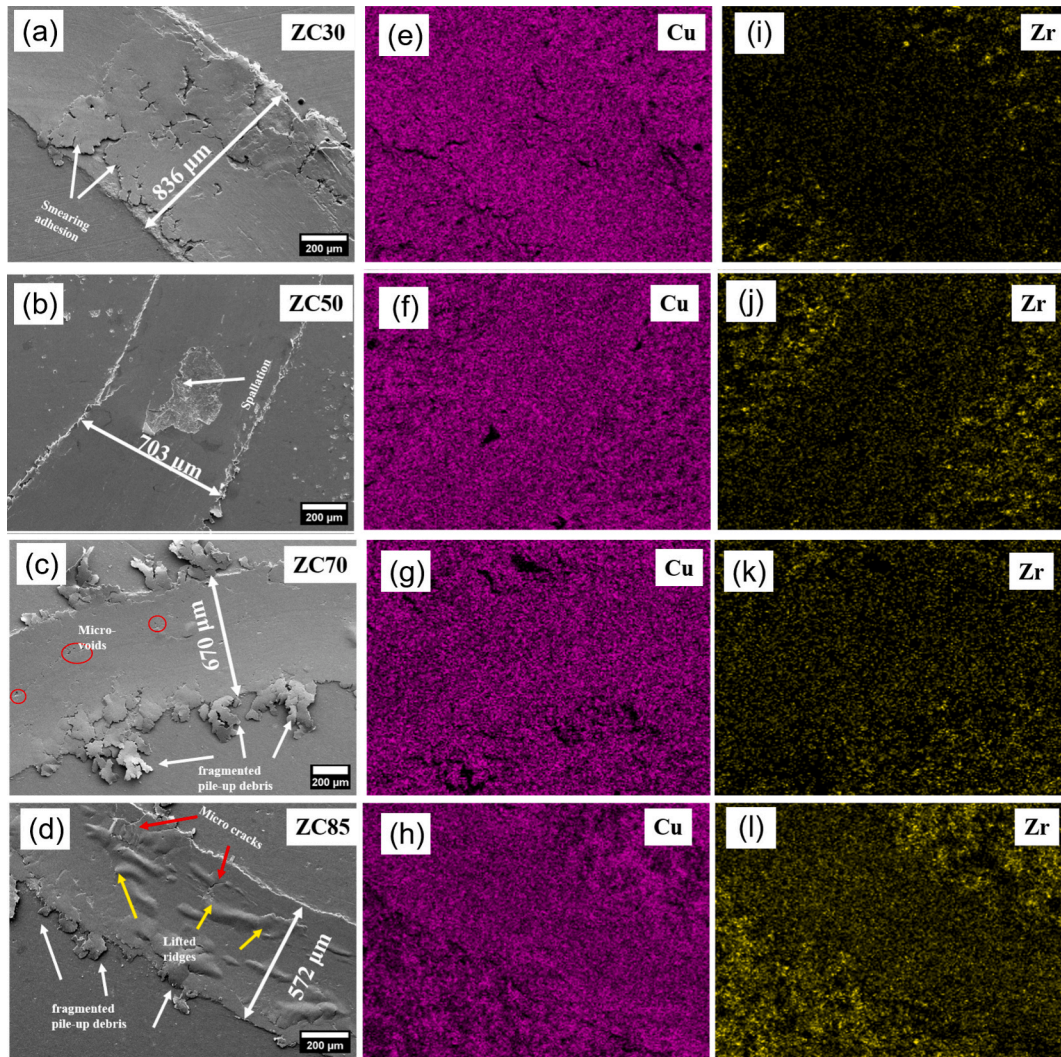


Fig. 17. SEM images of wear tracks of cold spray coatings on Al-6061 substrates (a) pure Cu (b) Cu-30%ZrC, (c) Cu-50%ZrC, (d) Cu-70%ZrC and (e) Cu-85%ZrC; The corresponding EDS elemental maps of Cu are shown in (f–i), and those of Zr are shown in (j–m), respectively, for the coatings in (a–d). The images show the different features in the wear tracks such as smearing, spallation and fragmented wear debris.

fragmentation of hard ceramic-rich regions, while large-scale plastic ploughing and adhesive smearing are significantly suppressed. This explains why the ZC85 coating simultaneously exhibits brittle wear features and the lowest wear volume among all compositions. These observations highlight that minimum wear volume does not necessarily correspond to maximum damage tolerance, particularly in ceramic-rich cold-sprayed composite systems.

The differing nature of scratch and sliding wear testing also explains the contrasting frictional behaviour observed in the present study. In scratch testing, ZC70 and ZC85 coatings exhibited lower and comparatively stable CoF values than pure Cu, indicating enhanced resistance to localized penetration and improved inter-splat integrity. In ball-on-disk wear testing, however, pure Cu exhibited a comparatively lower steady-state CoF than some intermediate composite coatings despite showing the highest wear volume. This behaviour is associated with the formation of a stable plastically smeared sliding layer in Cu-rich coatings, whereas the exposure of hard ZrC asperities during sliding increases frictional interaction in the composite coatings. Therefore, a low CoF under sliding conditions does not necessarily indicate superior wear resistance or reduced material loss.

Overall, the combined interpretation of scratch and sliding wear responses demonstrates that these tests provide complementary but fundamentally non-equivalent information regarding coating

behaviour. Scratch testing primarily probes localized deformation resistance, cohesion, and interfacial stability under monotonic loading conditions, whereas ball-on-disk wear testing captures progressive damage accumulation and material removal under cyclic sliding contact. The present results demonstrate that the most favourable tribological performance is achieved when hardness-induced suppression of plastic deformation is balanced with sufficient metallic matrix continuity to maintain damage tolerance and prevent brittle fragmentation. From this integrated perspective, intermediate ceramic contents (50–70 vol% ZrC) provide the most effective compromise between coating build-up, inter-splat bonding, adhesion strength, deformation resistance, and long-term wear stability. This integrated scratch-wear framework therefore provides a useful approach for understanding damage evolution and designing cold-sprayed metal-matrix composite coatings for demanding tribological applications.

5. Conclusions

- Cold spray was used to deposit dense ZrC–Cu composite coatings while preserving the purity of both the Cu and ZrC. The study demonstrates that ceramic fraction plays a decisive role in governing deposition behavior, retention, interfacial integrity, and tribological performance.

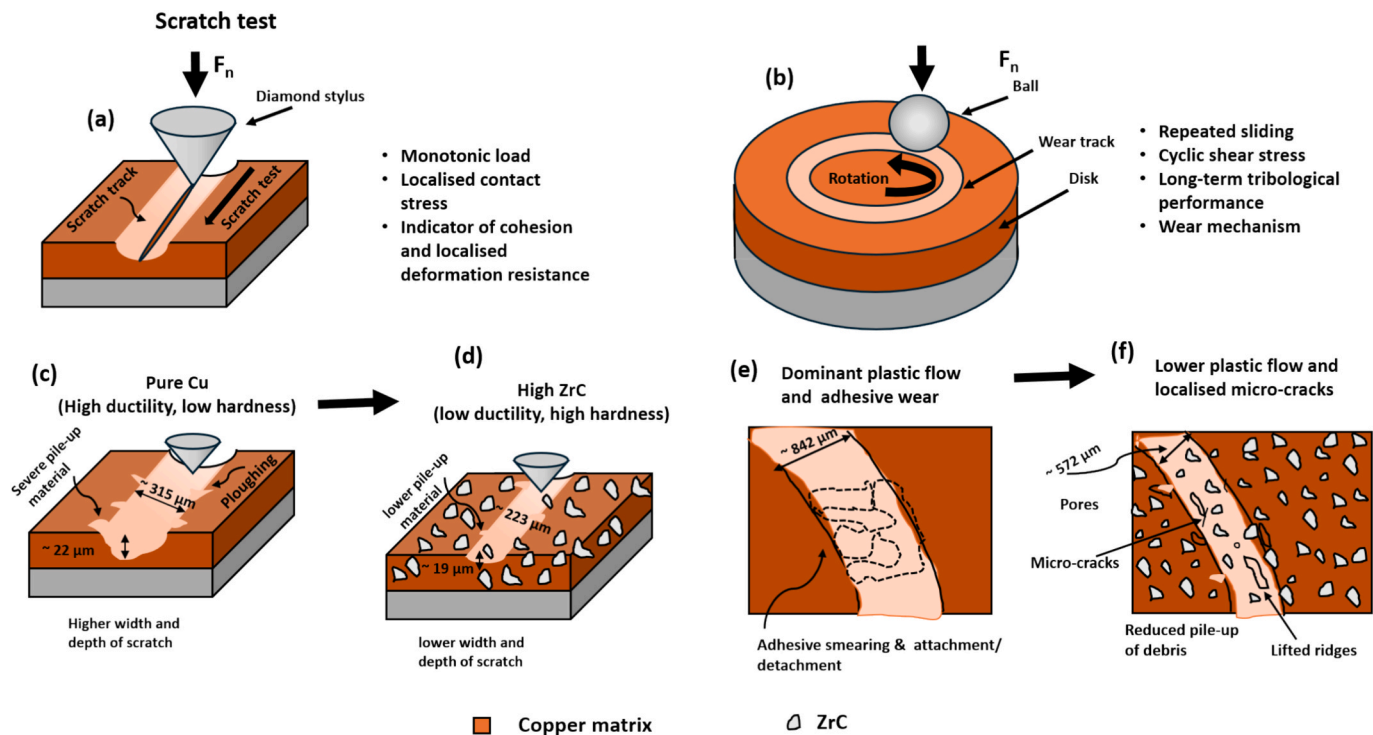


Fig. 18. Schematic illustration showing the integrated interpretation of scratch and sliding wear behaviour in cold-sprayed Cu–ZrC composite coatings. (a) Scratch testing under localized monotonic loading and (b) ball-on-disk wear testing under cyclic sliding conditions. The corresponding deformation and damage mechanisms are illustrated for Cu-rich and ZrC-rich coatings: (c) severe plastic deformation and ploughing in Cu-rich coatings during scratch testing, (d) reduced penetration depth and suppressed material displacement in ceramic-rich coatings, (e) adhesive wear and material smearing with large wear-track formation in Cu-rich coatings during sliding wear, and (f) brittle micro fragmentation and localized crack formation in ZrC-rich coatings under cyclic sliding. The schematic highlights the transition from ductile deformation-dominated wear to brittle fragmentation-controlled wear with increasing ceramic content.

- The results show strong dependence on coating build-up, deposition efficiency, and ceramic retention. Moderate ceramic loading (50–70 wt% ZrC) promoted effective particle tamping and interlocking, resulting in enhanced densification, improved ceramic retention, and maximum coating thickness. Higher ceramic content (85 wt% ZrC) led to increased particle rebound, fragmentation, and porosity due to insufficient ductile binder availability.
- Mechanical characterization revealed a systematic increase in hardness with increasing ZrC content. However, adhesion strength and scratch resistance followed non-linear trends. Maximum adhesion strength and most stable scratch response were achieved for intermediate compositions (ZC50–ZC70), indicating an optimal balance between reinforcement content and coating cohesion. Ceramic-rich coatings exhibited reduced scratch penetration depths but showed reduced damage tolerance due to limited plastic accommodation.
- Ball-on-disk wear testing demonstrated a progressive reduction in wear volume with increasing ceramic content, achieving up to ~93% reduction for ZC85 compared to pure Cu. Wear-track morphology revealed a transition from ductile adhesive wear in Cu-rich coatings to brittle fragmentation-dominated mechanisms at high ceramic loading. These findings highlight that minimal wear volume or low friction coefficient alone does not necessarily correspond to optimal tribological performance.
- An integrated interpretation of scratch and sliding wear responses confirms that scratch testing is primarily sensitive to coating cohesion and adhesion under localized loading, whereas ball-on-disk wear testing captures damage evolution under cyclic, service-like sliding conditions.
- Overall, this work establishes a comprehensive processing–structure–property–tribology framework for designing cold-sprayed ZrC–Cu composite coatings and offers transferable insights

for the development of advanced metal–matrix composite coatings for demanding wear applications.

CRediT authorship contribution statement

Saiful Wali Khan: Writing – review & editing, Writing – original draft, Methodology, Investigation, Formal analysis, Conceptualization. **Gidla Vinay:** Writing – review & editing, Methodology, Investigation, Formal analysis. **Kotturu Leela Satya Sai Vasu:** Writing – review & editing, Investigation. **Ashish Ganvir:** Writing – review & editing, Supervision, Resources, Methodology. **Ekta Singla:** Writing – review & editing, Supervision, Resources, Methodology, Conceptualization. **Harpreet Singh:** Writing – review & editing, Supervision, Resources, Methodology, Conceptualization.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Acknowledgements

The authors like to acknowledge Dr. Ayan Bhowmik (IIT Delhi) for providing access to the adhesion test and Ms Sudha Kumari and Ms Malar Vadani GT for helping in the experimentation. This work was supported by the prestigious Prime Minister's Research Fellowship (PMRF ID-2902491) scheme by the Ministry of Education, Government of India. The authors would like to thank the Department of Science and Technology (DST) India, through the project "Advanced Manufacturing Technology - Centre of Excellence (AMT-CoE) on Degradation Resistant Thermal Spray Coatings Engineered for Indigenous Industrial

Applications (DST/TDT/AM/2022/143) established at IIT Ropar. The cold spray system used for this study was established through MHRD-DST funded Uchhatar Avishkar Yojana (UAY, IITRPR_001). We are also grateful to Indian Institute of Technology Ropar including Central research facility (IIT Ropar) for providing the research facilities to carry out this work. The Authors at University of Turku also acknowledges financial support from GREEN-BAT (352517), co-funded by the Research Council of Finland and the European Union under the M-ERA-NET 2021 framework, as well as the SOLACE (DNR 360540) Academy research fellowship, funded by the Research Council of Finland and the City of Turku for supporting Prof. Ganviirs tenure-track grant. The authors also acknowledge the Finnish Digital Design and Manufacturing Infrastructure (FiDiEm) for access to the experimental facilities.

Data availability

Data will be made available on request.

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