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Selective laser melting of WC reinforced maraging steel 300: microstructure characterization and tribological performance

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Abstract: Owing to the limited wear resistance, Maraging steels has limited applications in the harsh environments, leading to a shortened service-life or even failure. In this study, WC particles reinforced maraging steel 300 (MS300) composites were produced by selective laser melting (SLM) to enhance the tribological performance. The microstructure and phase analysis of SLM-produced composites were investigated by SEM-EDS. Dry sliding wear test and microhardness were used to investigate the mechanical properties of WC/MS300 composites. As laser energy density increases, the microhardness drops rapidly and the maximum value of 562.2 HV₂ was measured at the linear energy density of 110 J/m. Dry sliding tests show that the wear rate was decreased by more than 1500 times through the WC reinforcement compared with pure MS300 sample, which emphasizes the superior wear-resistance properties of MS300/WC composite. Besides, the best wear-resistance performance at the improved linear energy density of 125 J/m. The study on wear mechanism shows that formation of carbide layer between WC and MS300 matrix significantly prevent the plowing effect. The present work provides an effectively method to improve the tribological performance MS300 through the reinforcement of WC particle by selective laser melting.

Keywords: Selective laser melting (SLM); WC; Maraging steel 300; tribological behavior

1. Introduction

The selective laser melting (SLM) as an additive manufacturing technology, has been widely applied in different areas [1-3]. SLM technique allows the manufacturing of net-shape components directly from powder material according to the computer-aided design (CAD) model. In this process, the designed model is firstly sliced into a series of layers with constant thickness. Then, the part is fabricated by selectively melting the powder bed by a traverse laser source, and the formed molten pool will experience rapid-solidification at a high cooling rate. This technology enables the fabrication of customized or personalized part within a wide range of areas like biomedical [4], aeronautics and astronautics [5, 6].

Based on the principle of precipitation hardening, the Maraging steel 300 (MS300) exhibits outstanding material properties like high strength, high ductility and superior machinability [7-9]. It has been widely used as engine component, missile parts, aircrafts, oil and gas industries [10, 11]. The application of SLM technique in the fabrication of molding can significantly shorten the preparation period and the production cycle, which enables the rapid innovation and production of molding design. Furthermore, the SLM technique can achieve the production of customized molding with complex geometry. For example, Kempen and Hermann [8, 12] have investigated the mechanical properties and microstructures of Maraging steel 300 (MS300) samples fabricated by SLM, which present high densification rate and mechanical properties. Bai et al. [13] reported the effects of processing parameters and heat treatment on the microstructure and mechanical properties of maraging steel 300 fabricated by SLM. It is reported that the high-frequency friction of molding with counterpart can lead to its shortening of service-life even the failing [14]. The limited wear resistance of Maraging steels has largely restricted its applications in the harsh environments with wear and corrosion. Therefore, it is reasonable to improve the wear behavior by adding the reinforcement particles to the Maraging materials to improve its wear-resistance performance.

Based on the previous literatures, SLM technology has been used to fabricate metal matrix composites (MMCs) with high performance [15-17]. The rapid melting and solidification can result in high mechanical properties and homogenous microstructure of the reinforcement particles such as WC [18, 19], TiB₂ [20-22], WC-CO, Al₂O₃ [23, 24], TiC [25]. For example, Rong et al [18] fabricated the WC/Inconel 718

composite by SLM at different process parameters. It shows that at the high microhardness of 393.2 HV and a lowest coefficient of friction (COF) was obtained at the optimal scan speed of 450 mm/s. Besides, based on AlSi10Mg and nano-scaled TiC ceramic particles, Gu et al [26] fabricated the Al nanocomposites with a novel ring-structured TiC reinforcement. Compared to the unreinforced AlSi10Mg part, the TiC reinforced nanocomposite exhibited elevated microhardness (188.3 HV) and tensile strength with a reduction in elongation. Similarly, Gu et al [25] used nano-TiC reinforcements to improve the microhardness and wear resistance of Ti matrix by SLM. Song et al [27] fabricated Fe-based nanocomposites with micro Fe and nano-SiC particles by SLM, where uniform distribution of SiC particles and homogeneous martensites were observed. Recently, Li et al [20] used nano-TiB₂ decorated AlSi10Mg composite powders to fabricate a dense Al composite samples by SLM, which shows high fracture strength and ductility.

Therefore, in the present work, the MS300 matrix reinforced by micro-sized WC particles were produced by SLM at different linear energy density. The microstructure and phase analysis of the SLM composites were made to understand the role of WC particles during the enhancement of mechanical properties. Meanwhile, the microhardness and wear performance of the produced WC/MS300 composites were investigated to understand the evolution of mechanical properties. Furthermore, the effect of linear energy density on the wear performance of the WC/MS300 composites were studied and discussed.

2. Experimental details

The gas atomized maraging steel 300 powders (EOS GmbH, Germany) with spherical shapes and a size distribution between 33 μm and 40 μm and the irregular WC powders (Jinlu Xiamen China) with a size range of 1-6 μm were used as starting materials. The powder mixture with 85% of MS300 and 15% of WC in weight percentage were mechanically mixed in a tumbling container for 200 mins prior to SLM fabrication. The morphologies of different powders are shown in Fig. 1. The chemical composition of MS300 powder is given in Table. 1. The cross-sectional morphology after etching is shown in Fig. 2 (a), showing homogeneously distributed grains.

Table 1. Chemical composition of MS300 powder.

Element	Fe	Ni	Co	Mo	Ti	Al	Cr	C	Mn	Si
wt%	Rest	17-19	8.5-9.5	4.5-5.2	0.6-0.8	0.05-0.15	≤0.5	≤0.03	≤0.1	≤0.01

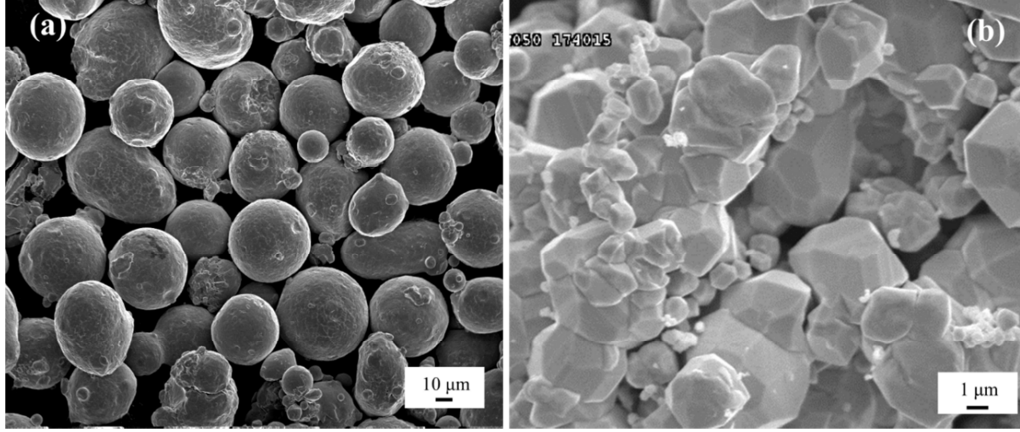


Fig. 1 SEM image of the powder morphologies: (a) MS300, (b) WC.

The sample annotations produced by SLM and the respective operating parameters are given in Table 2. In this work, the WC/MS300 composite were fabricated at different laser powers (190 - 280 W), whereas the pure MS300 sample were fabricated as a comparison object at the laser power of 220 W. It enabled us to investigate the role of WC reinforcement particle in the mechanical properties. A SLM machine (MCP-realizer SLM 250, Germany) was used to fabricate the WC/MS300 composite in this work, which was equipped with an YLR-100-SM single-mode CW ytterbium 400 W fiber laser. The laser beam diameter was set to 40 μm . The SLM process is realized within working chamber (250 mm \times 250 mm \times 280 mm) with an argon atmosphere. The oxygen content was maintained below 0.2% to prevent the oxidation during fabrication. The cubic specimen (10 mm \times 10 mm \times 5 mm) as shown in Fig. 2 (b) were prepared by SLM. For the observation of microstructure evolution, the cross-sectional sample were polished and then etched in a solution of 25 ml HNO_3 , 50 ml HCl and 1 g CuCl_2 in distilled water for 20 s. In addition, the microstructure was then observed by a scanning electron microscope (SEM) equipped with an energy-dispersive spectroscopy (EDS) unit (JSM5800LV, JEOL, Japan). Phase composition of the SLM-fabricated samples was performed by X-ray diffraction (XRD, Siemens, Germany) using a cobalt anticathode ($k=1.78897 \text{ \AA}$) operated at 35 kV and 40 mA.

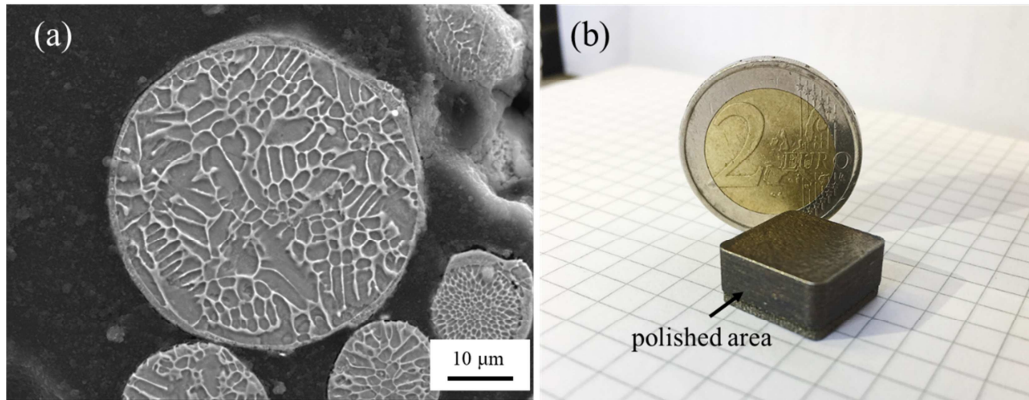


Fig. 2 (a) Cross-section showing the grain structure of the feedstock of MS300 particle after etching, (b) digital photo of the cubic sample.

The porosity level and densification rate were measured and calculated through image analysis (NIH Image J, Software, USA), where the results were the average of 15 analyses. The Vickers hardness was measured using a micro-hardness tester (Leiz-Wetzlar, Germany) at the cross-sections of the composite samples at a load of 2 N and an indentation time of 25 s, and the final value is the mean of 15 measurements.

Table 2 Processing parameters and the designation of the fabricated samples in this work.

Number	Sample	Power (W)	Speed (m/s)	Linear energy density (J/m)
1	WC/MS300	190	2	95
2	WC/MS300	220	2	110
3	WC/MS300	250	2	125
4	MS300	250	2	125
5	WC/MS300	280	2	140

A tribometer implement (CSEM, Switzerland) was used to conduct the dry sliding wear tests at ambient temperature. The surface roughness R_a below $0.15 \mu\text{m}$ was ensured to before the test. The cleaned WC-Co balls with a diameter of 6 mm was used as the counterparts. The used processing parameters for the friction test are given as below: load of 0.5 kg, rotational velocity of 0.1 m/s, sliding distance of 100 m, and a friction radius of 3 mm. After the test, the variation of the coefficient of friction (COF) and the worn morphology observed by SEM and EDS were analyzed to understand the wear performance. To calculate the worn volume, the cross-sectional areas of the worn tracks were measured by an Altisurf 500 profilometer (France), the final value is the mean of 5 measurements.

3. Results and discussion

3.1. Phase analysis

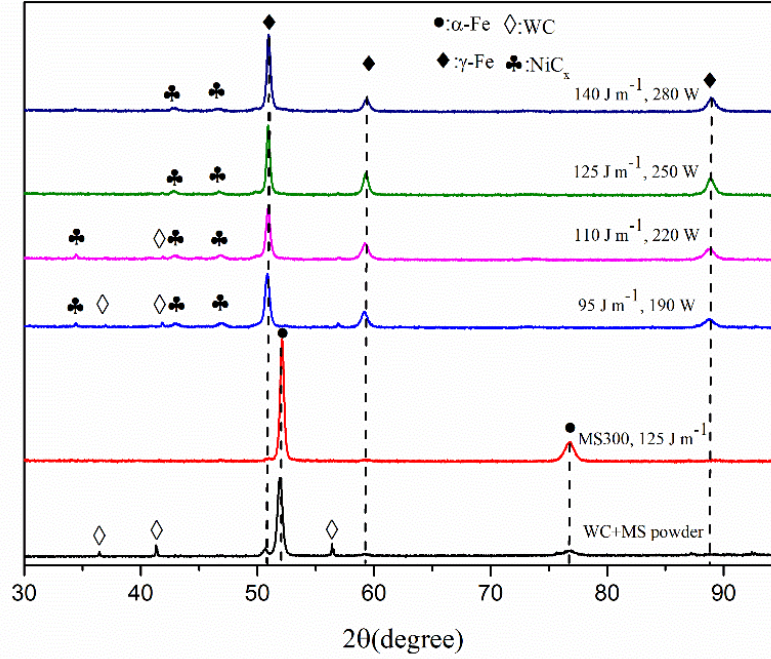


Fig. 3 XRD analysis of the feedstock powder, MS300/WC composite and pure MS300 samples processed by SLM at different laser powers.

Fig. 3 gives the XRD spectra of SLM-processed WC/MS300 composites obtained within a wide range between 30° and 95°. The main diffraction peaks corresponding to α -Fe (bcc) phase were detected in the raw material and MS300 sample [8]. The peaks for WC can also be noticed in the raw powder. Significant difference of XRD spectra can be found in the MS300/WC composite, where the α -Fe in MS300 was transformed to the phase of γ -Fe (fcc). Chemical composition and formation temperature could influence the transformation from high temperature γ -Fe to low temperature α -Fe [28]. In this study, the existence of carbon content could hinder such transformation from γ -Fe to α -Fe. Besides, sub-cooling degree can greatly affect the transformation of γ -Fe to the highly stained α -Fe. As the addition of WC particles with higher conductivity, cooling rate of the molten pool for the composite part was improved compared with the pure MS300. Consequently, γ -Fe was detected on the SLM MS300/WC parts instead of α -Fe in pure MS300 [29, 30]. Furthermore, the XRD characterization within small 2θ (35°-45°) angle should be noticed to

understand the influence of linear energy density on the WC particles. The WC phase is gradually decreased as the increment of laser power. As for the linear energy density higher than 125 J/m, the WC phase cannot be detected any more, which implied the decrement of WC particles. Due to the increasing laser power, the WC particles were partially dissolved in the matrix. The residual amount of WC phase falls out of the measurement range of XRD, which results in the vanishing of WC phase in XRD spectrum.

3.2. Microstructural analysis

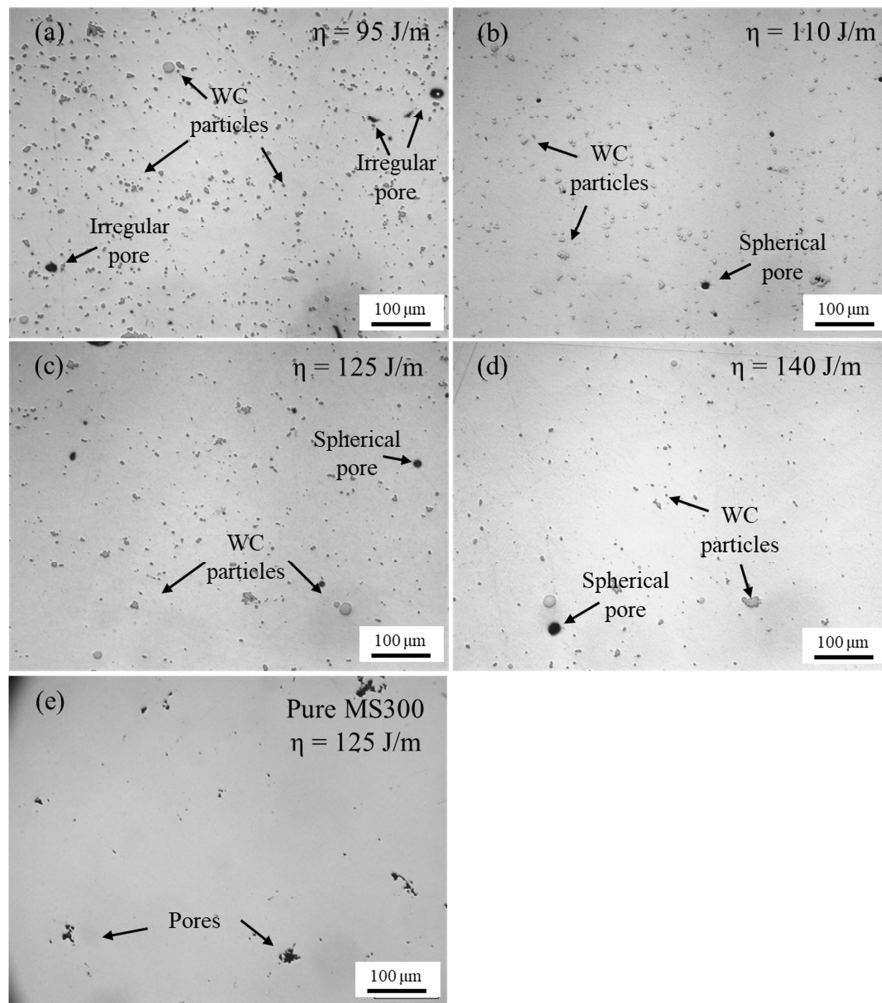


Fig. 4 OM images of cross-sectional microstructure of the SLM-processed: MS300/WC samples at (a) 95 J/m, (b) 110 J/m, (c) 125 J/m, (d) 140 J/m, (e) pure MS300 sample at 125 J/m.

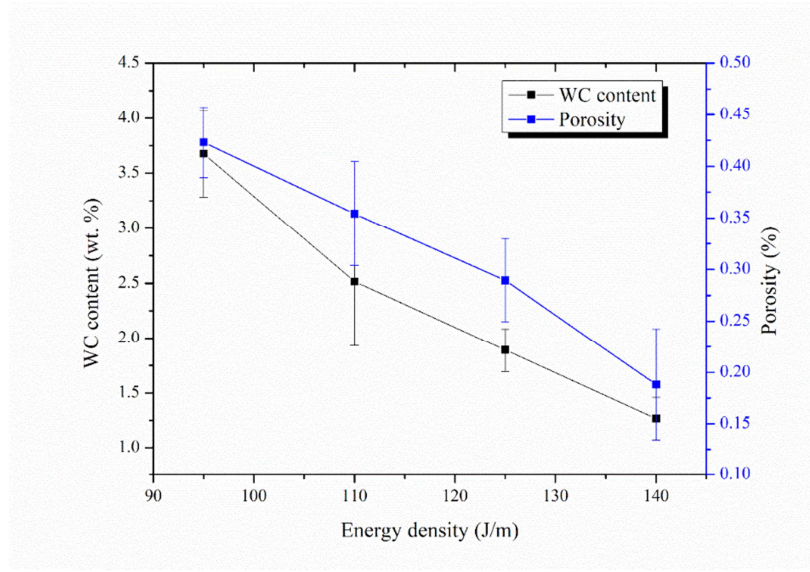


Fig. 5 Evolutions of porosity and the WC content of the WC/MS300 composite fabricated by SLM as a function of linear energy density.

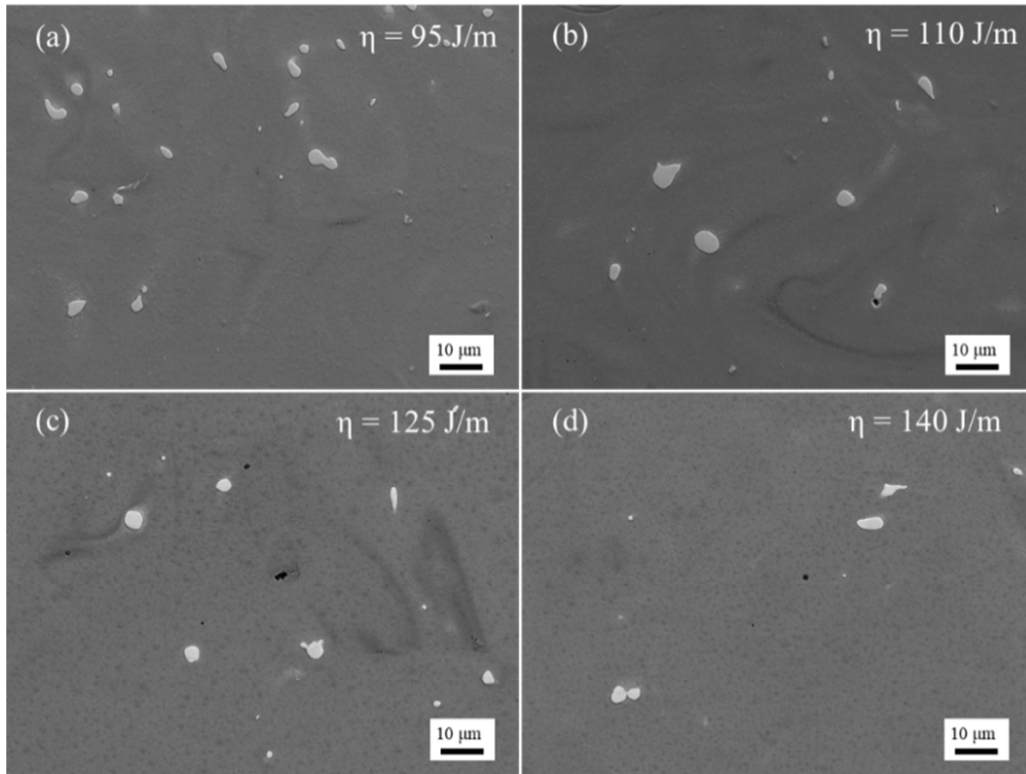


Fig. 6 Distribution of reinforcing particles of WC/MS300 composite fabricated at different laser power densities: (a) 95 J/m, (b) 110 J/m, (c) 125 J/m, (d) 140 J/m.

Fig. 4 shows the microstructure of the SLM processed MS300/WC composites at different energy densities. The uniform distribution of WC particle can be seen in the OM images of MS300/WC samples. It can be noticed that the WC particles are decreasing as a function of the linear energy density, which can be attributed to the enhanced WC dissolution at higher molten pool temperature. Meanwhile, some irregular pores can be seen on the cross-sectional images of the MS300/WC composite and the pure MS300 samples. Based on Fig. 5, it can be observed that the porosity of the composite sample decreases as the laser power increases. Furthermore, the morphology of pores changes from an irregular (Fig. 4 (a)) to a spherical shape (Fig. 4 (b, c and d)). Moreover, the porosity of the pure MS300 sample fabricated at 125 J/m was $0.83 \pm 0.32\%$. As the magnified view is shown in Fig. 6, dense WC/MS300 composites were fabricated at different energy densities with uniform distribution of reinforced particles. It can be noticed that the content of WC particle in the MS300 matrix is decreasing as a function of increasing linear energy density. The MS300 matrix experienced a complete melting under the high-energy input by laser beam, while the WC particles experienced partial dissolution due to their relatively higher melting point. Similar result is also obtained by XRD analysis (Fig. 3), the content ratio of WC particle in the composite was decreasing with the increasing linear energy density. The weight ratio of WC particles in MS300 matrix is evaluated and transformed from the analysis by the ImageJ software and given in Fig. 5. At the linear energy density of 95 J/m, the measured volume content of WC particles is 1.94 % and is far below the volume ratio of 9% in starting materials, which indicates its partial dissolution into matrix. At the higher laser energy density η , less reinforced WC particles can be found in the MS300 matrix due to the enhanced particle dissolution at higher temperature. At the same time, the higher temperature within the molten pool at a higher linear energy density will lead to a stronger Marangoni flow, which is caused by the surface tension gradient proportional to the temperature gradient. The role of Marangoni flow in the SLM process has been widely studied in the numerical model [31, 32]. The increased temperature will reduce the viscosity of molten pool [33], which can promote the flow fluid within the molten pool. Thus, the combined effects of increased Marangoni flow and decreased molten pool viscosity can efficient improve the movement of the reinforced WC particles within the molten pool as well as its solution into the matrix. It can directly lead to the observed WC particles distribution in Fig. 6.

3.3. WC particle evolution

Fig. 7 presents the etched cross-section of the WC particles and the MS300 matrix at different laser energy densities. Generally, it can be noticed that all the WC particles were well inserted in the MS300 matrix. At a relatively low η of 95 J/m, almost no diffusion layer can be seen surrounding the particles (Fig. 7 (a)). As the linear energy density η increased from 95 to 110 J/m, an evident gradient layer can be found around the WC particles in the MS300 matrix, which consists of carbide phase with high hardness. By further increasing the linear energy density to 125 J/m and 140 J/m, the fusion between two adjacent WC particles can be observed due to the gradually increasing temperature and fluid flow within molten pool. At the same time, the diffusion layer is not visible around the WC particles, while the diffusion region of reinforced particles into the matrix material is more evident.

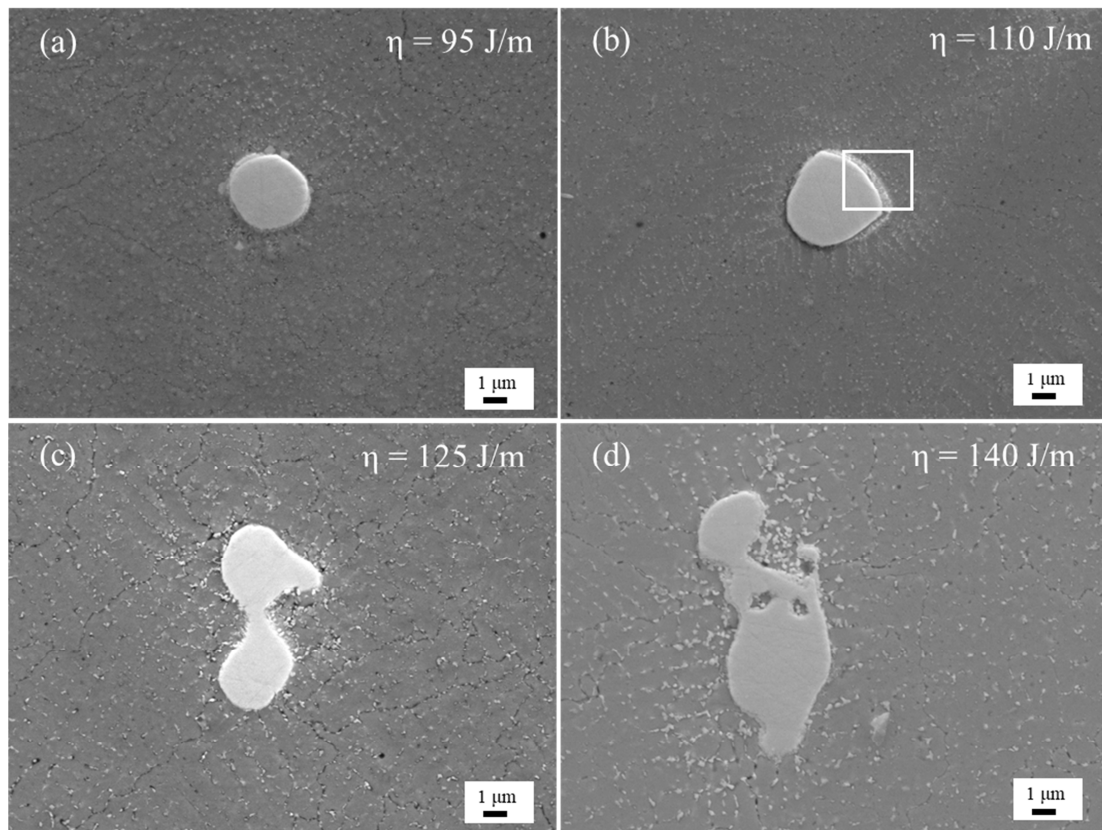


Fig. 7 Morphology of the reinforced WC particle and the diffusion layer around the particle at different energy densities: (a) 95 J/m, (b) 110 J/m, (c) 125 J/m, (d) 140 J/m.

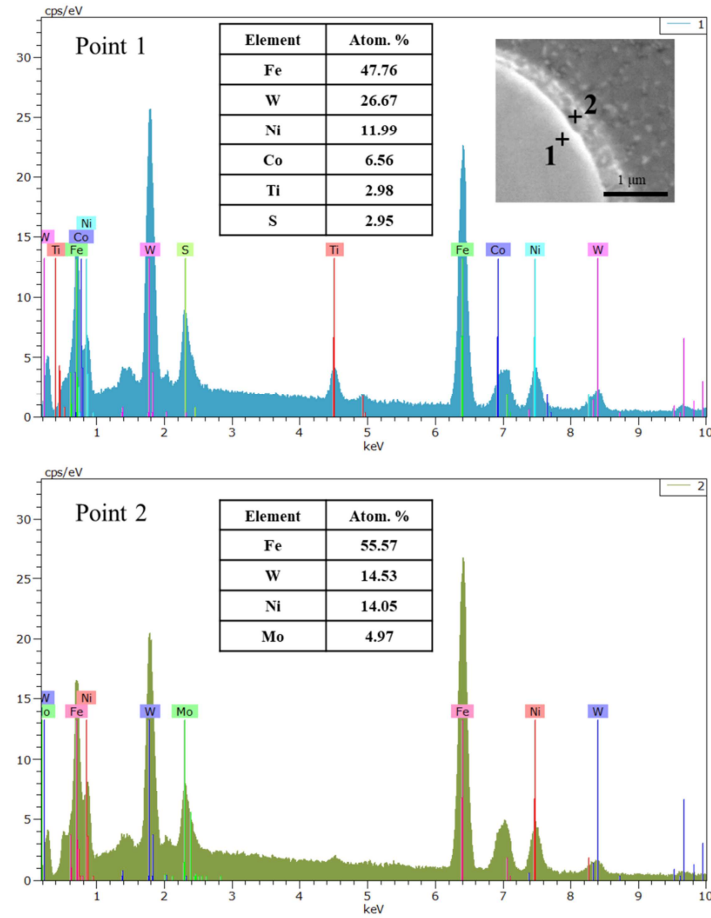


Fig. 8 Atomic percentages of the elements at points 1 and 2 in Fig. 8 (b) by EDS.

To investigate the chemical composition at the diffusion interlayer, the EDS analysis was carried out at the points 1 and 2 marked in magnified view in Fig. 7 (b). Fig. 8 presents the corresponding EDS results, showing the chemical composition at the diffusion region. It can be noticed that the magnified dissolution region can be seen at the periphery of WC particle imbedding in the MS300 matrix in Fig. 8 (a). Furthermore, continuous shell and disconnected globular dendrite are seen surrounding WC particle. Based on our previous study [34], the W elements were uniformly diffused from the reinforcement particle into matrix through this interlayer. Thus, based on the related studies, it is reasonable to suggest that the carbides ($\text{Fe}_3\text{W}_3\text{C}$, $\text{Fe}_4\text{W}_2\text{C}$ or $\text{Fe}_6\text{W}_6\text{C}$ [35]) were formed through the in-situ reaction between the WC particle and the element from matrix. The release and solution of W and C into the molten pool during SLM process can promote the solid solution hardening through the dislocation by precipitates [36]. Furthermore, the partial dissolution of WC particle with good wettability can effectively increase its contact

interface in the matrix molten pool, which can further improve the mechanical strength of composite. In addition, it can be found that at the point 1 of the gradient interface, the atomic percentages of Fe and Ni are 47.76 and 11.99 respectively, as the one of W is 26.67. The samples were covered by a thin conductive layer by carbon sputtering prior to the SEM observation, so that the detection of C element in EDS is not available in this work. Meanwhile, similar element types were found at the point 2 of diffusion layer with large difference in the atomic percentage from the point 1. In the diffusion layer, the W element possessed an atomic percentage of 14.57, which showed a decreased concentration compared to the gradient interface. Meanwhile, the EDX results show that the content of Fe element gradually increases from the WC particle to the matrix. Such features demonstrate a typical gradient interface between particle and matrix. Besides, it should be noticed that the WC particle has experienced a partial dissolution, resulting in a distinct shape change. Different from the original sharp-shaped features, the WC particles embedded in matrix demonstrate rounded shapes caused by the partial dissolution.

3.4. Microhardness

Fig. 9 shows the microhardness values of the WC/MS300 composite sample fabricated at different linear laser energy density of η . By increasing the applied η from 95 J/m to 110 J/m, the average microhardness value was slightly enhanced from 534.1 ± 52.4 HV2 to 562.2 ± 62.8 HV2. Such microhardness enhancement was attributed to the more evident densification effect at higher linear energy density. By further improving the implemented η to 125 J/m, an apparent microhardness drop can be seen. At the applied η of 140 J/m, the microhardness continues to decrease to 478.2 ± 33.2 HV2. Such decrement of microhardness at high linear energy density is considered as the combination effects of the decline of the metastable stiff phase and the coarsened microstructure. As indicated in the XRD analysis in Fig. 3, the metastable stiff phase of NiC_x is decreasing as the increment of linear energy density. Meanwhile, as shown in Fig. 7, at the relatively high linear energy density, the overheating of the molten pool leads to the formation of over coarsened microstructure. Thus, it is indicated that the microhardness of the composites is directly influenced by the phase, grain size and defects in the matrix [18, 27]. By comparing the WC/MS300 composites with the pure MS300 sample fabricated at the linear energy density of 125 J/m, it can be found that the reinforcement of WC particles can significantly increase the microhardness through the formation of the metastable stiff phase (NiC_x).

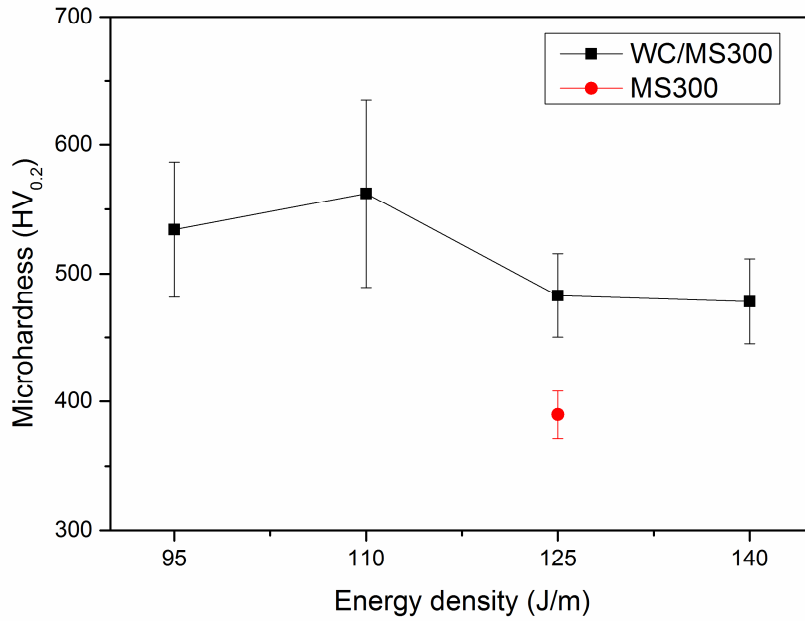


Fig. 9 Microhardness of the WC/MS300 composites prepared at different energy densities and pure MS300 sample at 125 J/m.

3.5. Wear rate and morphology

3.5.1. Wear rate and coefficient of friction (COF)

Fig. 10 and Fig. 11 show the evolution of COF with sliding distance and attendant wear rates for the WC/MS300 composites at different energy densities and for pure MS300 sample at 125 J/m. By comparing the WC/MS300 composites produced at different energy densities, the applied laser linear energy η played a key role in influencing the wear behavior of these composites. As can be seen from the COF curves, all WC/MS300 samples demonstrated stable variations with few fluctuations at the beginning stage. At a lower η of 95 J/m, the average COF value was 0.516 ± 0.037 , while the attendant wear rate reached as much as $42.27 \times 10^{-6} \text{ mm}^3/(\text{N} \cdot \text{m})$. However, a severe fluctuation of COF value was observed in Fig. 10. By increasing the η from 110 J/m to 125 J/m, the COF value increases from 0.54 to 0.53 and the resultant wear rate was reduced from $21.08 \times 10^{-6} \text{ mm}^3/(\text{N} \cdot \text{m})$ to $0.29 \times 10^{-6} \text{ mm}^3/(\text{N} \cdot \text{m})$, where fewer COF fluctuation can be found.

On the other hand, the pure MS300 sample is found to possess an evidently higher COF value and wear rate value ($472 \times 10^{-6} \text{ mm}^3/(\text{N} \cdot \text{m})$) than that of WC/MS300 composite ($0.29 \times 10^{-6} \text{ mm}^3/(\text{N} \cdot \text{m})$) fabricated at the same linear energy density. Such significant wear rate decrease (more than 2500 times) can further emphasize the role of reinforcement WC particles in the MS300 matrix in the wear performance. However, while the laser linear energy density η was further increased to 140 J/m, the COF value increased to 0.54 and wear rate increased to $1.65 \times 10^{-6} \text{ mm}^3/(\text{N} \cdot \text{m})$. It is obvious that the wear performance of the SLM-fabricated WC/MS300 composite is strongly affected by the applied linear energy density. The aggregated reinforcing particles also had significant influence on the unsteady COF values during the initial stage of friction test. At a relatively low energy input, the low wear behavior is mainly limited by the increase porosity level and bigger-sized WC. At a higher linear energy density, the densified microstructure and uniformly distributed reinforcement particles can result in a steady and lower COF value and lower wear rate.

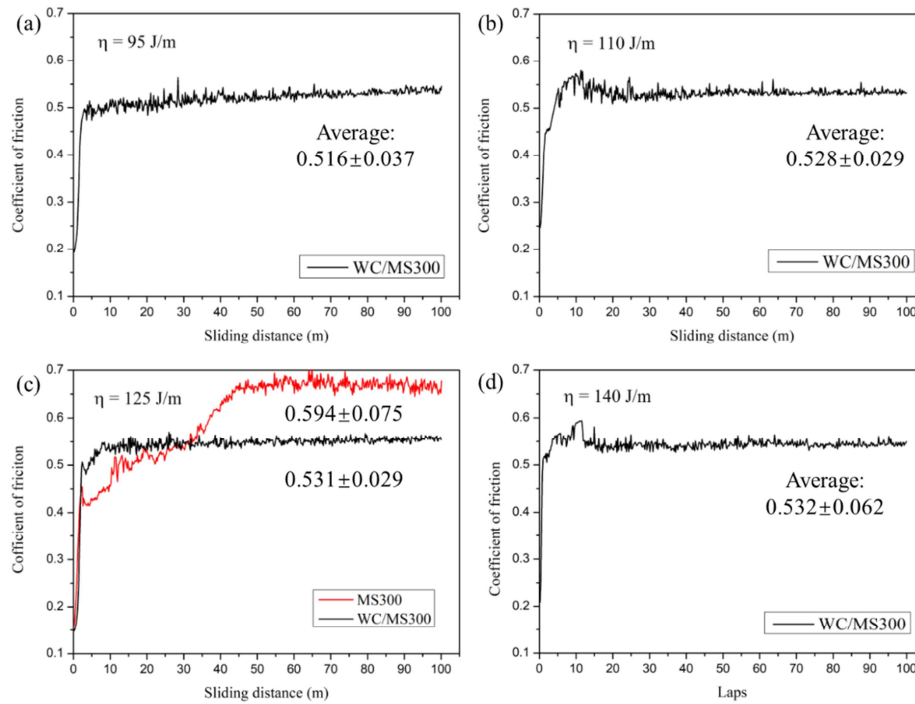


Fig. 10 COF value evolution of the samples prepared at different linear energy densities: (a) WC/MS300, 95 J/m, (b) WC/MS300, 110 J/m, (c) WC/MS300 and MS300, 125 J/m, (d) WC/MS300, 140 J/m.

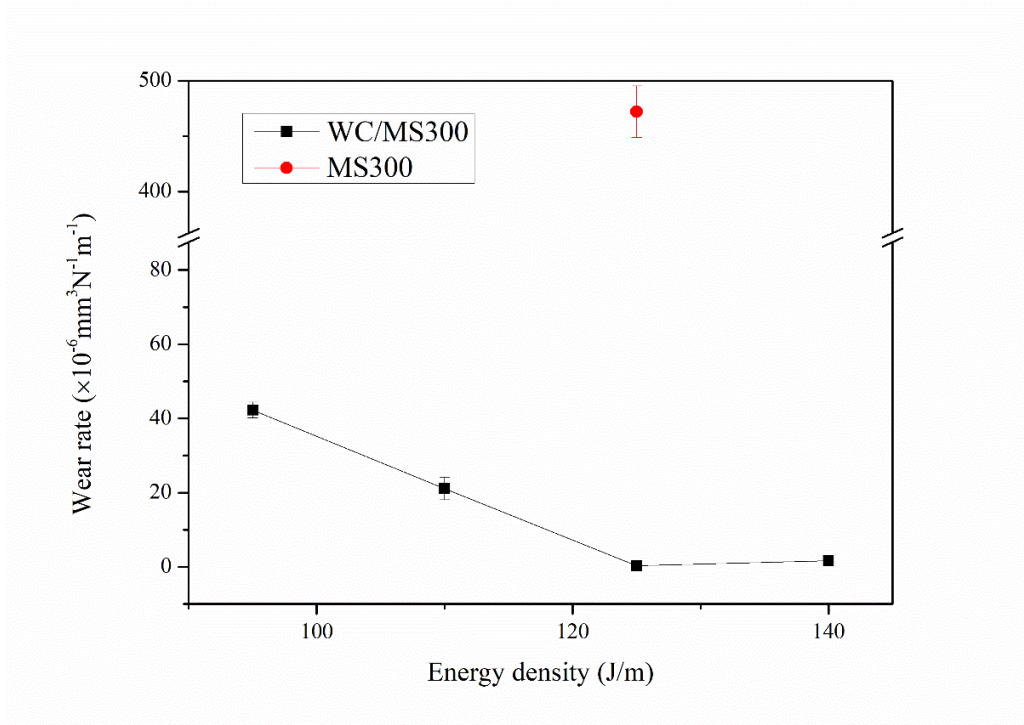


Fig. 11 Wear rate of the WC/MS300 composites and pure MS300 sample prepared at different linear energy densities.

3.5.2. Wear trace

To provide a profound insight into the wear mechanism of WC/MS300 composite, the worn morphology was observed. Fig. 12 shows the characteristic morphologies of worn surfaces at different linear energy densities η . As shown in Fig. 12(a), parallel grooves were found on the worn surface fabricated at 95 J/m, which is the sign of abrasion wear. The rough and porous worn surface was scattered by shattered WC debris. Furthermore, accompanying gathering pores, the interface region between the WC and the matrix experienced evident cracking during the sliding process. By increasing η to 110 J/m, shallower grooves were found on the surface of worn tracks, as well as the existence of WC reinforcement particles with relatively complete shape. Such phenomena indicate that the bonding strength of the interface between WC and the matrix was reasonably enhanced owing to the higher linear energy density. However, the carbide stiff interface with uneven distribution was easily smashed when suffering the shear stress at the plowing surface, which is given in the magnified view (Fig. 13 (b)).

By increasing the η to 125 J/m, a smoothing and densified tribolayer was formed (Fig. 12 (c)). No sign of plowing nor prominent WC particles can be seen on the worn surfaces. Such phenomena indicate the change of wear mechanism from form abrasion to adhesion at a higher linear energy density. Meanwhile, more W and C atoms from WC particle surface were released into the molten pool as can be seen from the Fig. 7. It promotes their reaction with the major elements in MS300 like Fe and Ni, which leads to the formation of the thick and even carbide layer around the WC particles. As the main composition of the interfacial layer, the carbide layer observed in Fig. 7 can lead to the desired bonding between the WC particles and the matrix. As a general understanding, the wear performance strongly depends on the bonding strength between the reinforcements and the matrix. Therefore, the plowing effect of the WC particles on the sliding surface that possess excellent bonding with the matrix was significantly decreased. Besides, as shown in Fig. 13 (c), the WC particles and the in-situ formed interfacial layer did not experience splitting or cracking during the test. On the contrary, the WC particles and the interfacial layer exhibit a remarkable tendency of adhesion and undergo strain hardening, which promotes the formation of the tribolayer over the entire surface and the enhancement of the wear resistance ability. The evolution of wear rate at different linear energy density given in Fig. 11 also suggests such transition of wear resistance performance. Meanwhile, as shown in Fig. 14, the EDS mapping was made to identify the reinforcement WC particles on the worn surface. It is indicated that partial WC particles were partially dissolved in the matrix of MS300, and W element is randomly distributed in the MS300 matrix, which promotes the improvement of mechanical properties.

As for the WC/MS300 composite fabricated at the highest η value of 140 J/m, the worn surface consists of a number of fragments on the tribolayer, with the fracture of the WC particles and the collapses of the interfacial layer. The similar spalling and cracking on the tribolayer surface were in consistence with the close wear rate value for the composites at 125 and 140 J/m given in Fig. 11. As shown in Fig. 13(d), at the highest energy input, the higher porosity and coarsened microstructure lead to the cracking and peeling of the tribolayer on the worn surface. As a determining factor for wear resistance [37], microhardness has a great influence on the wear performance of the WC/MS300 composite (see Fig. 9 and Fig. 11).

1 As for the pure MS300 sample produced at 125 J/m, much wider and larger wear trace can be found in
2 Fig. 12(e) through the comparison with WC/MS300 composite at the same linear energy density.
3 Meanwhile, the worn morphology of pure MS300 sample is noticed to be consisting of a number of
4 fragments and spalling of the interfacial layer can be found which agrees with the evident higher wear rate
5 value and COF value. With the absence of reinforcement WC particles, severe deformation and material
6 peel-off during the sliding test, which indicates that abrasion was the major wear mechanism [38].

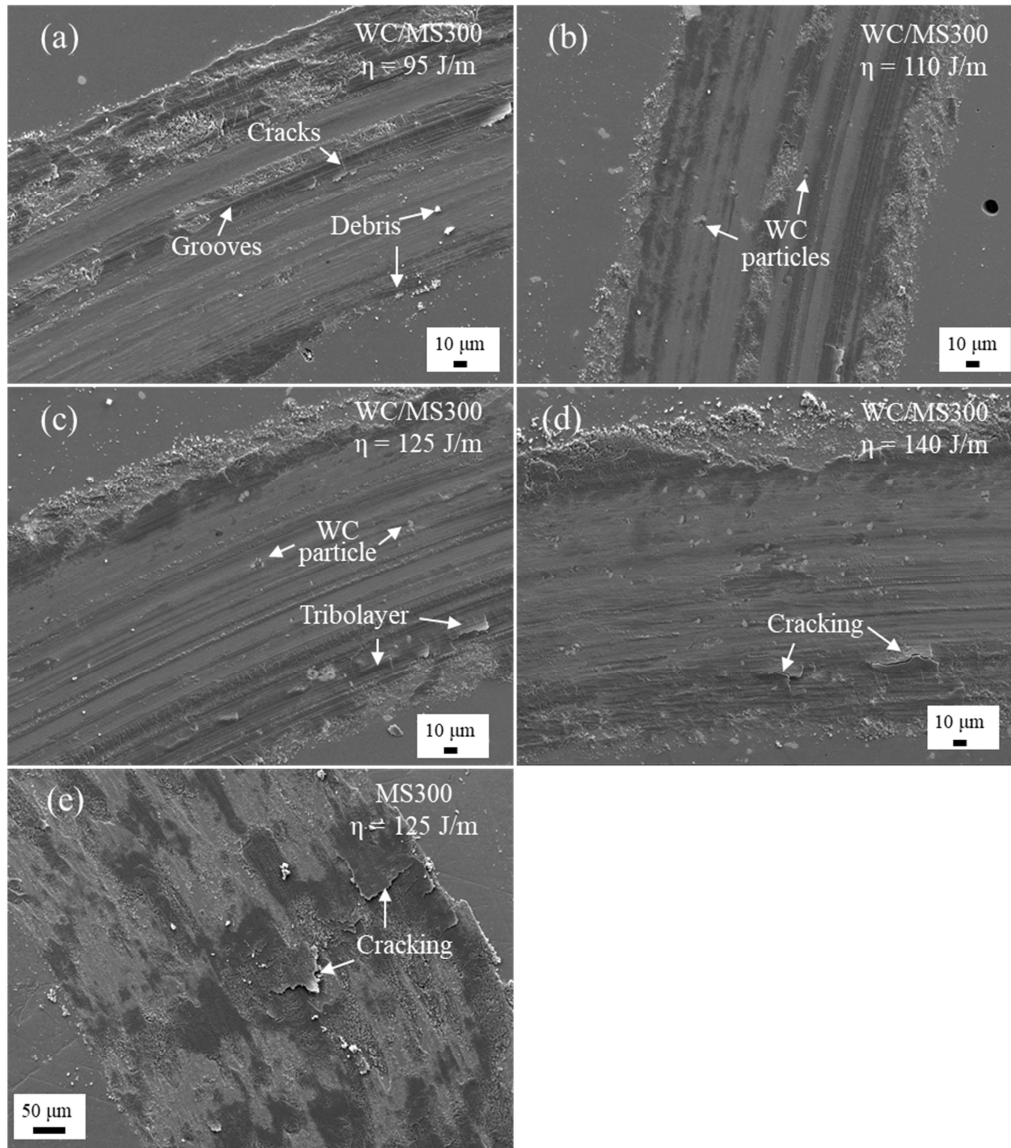
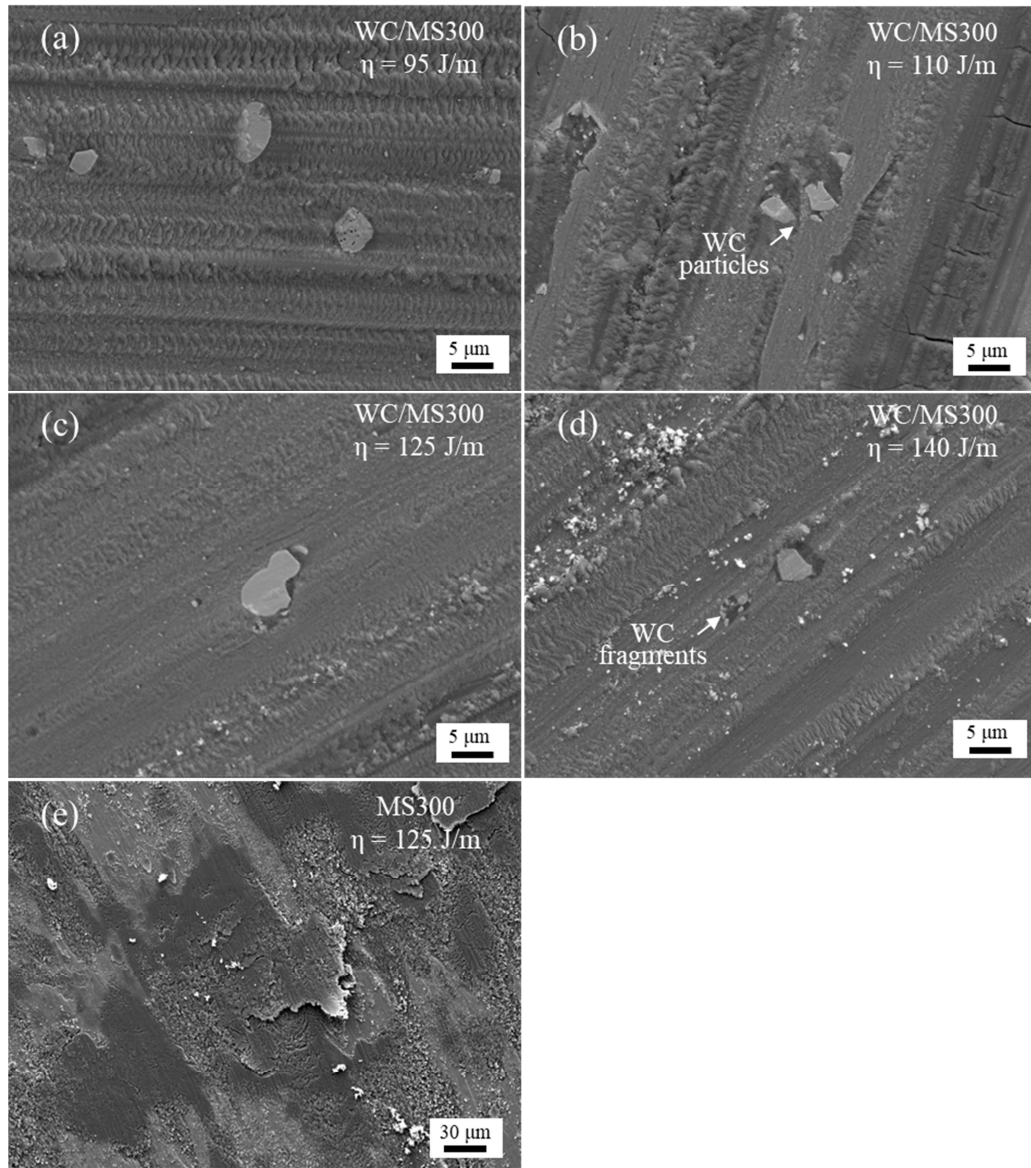


Fig. 12 Worn surface morphologies of the samples prepared at different linear energy densities: (a) WC/MS300, 95 J/m, (b) WC/MS300, 110 J/m, (c) WC/MS300, 125 J/m, (d) WC/MS300, 140 J/m, (e) MS300, 125 J/m,



1

2 Fig. 13 Worn tracks at high magnification of the samples prepared at different linear energy densities: (a)

3 WC/MS300, 95 J/m, (b) WC/MS300, 110 J/m, (c) WC/MS300, 125 J/m, (d) WC/MS300, 140 J/m, (e)

4 MS300, 125 J/m.

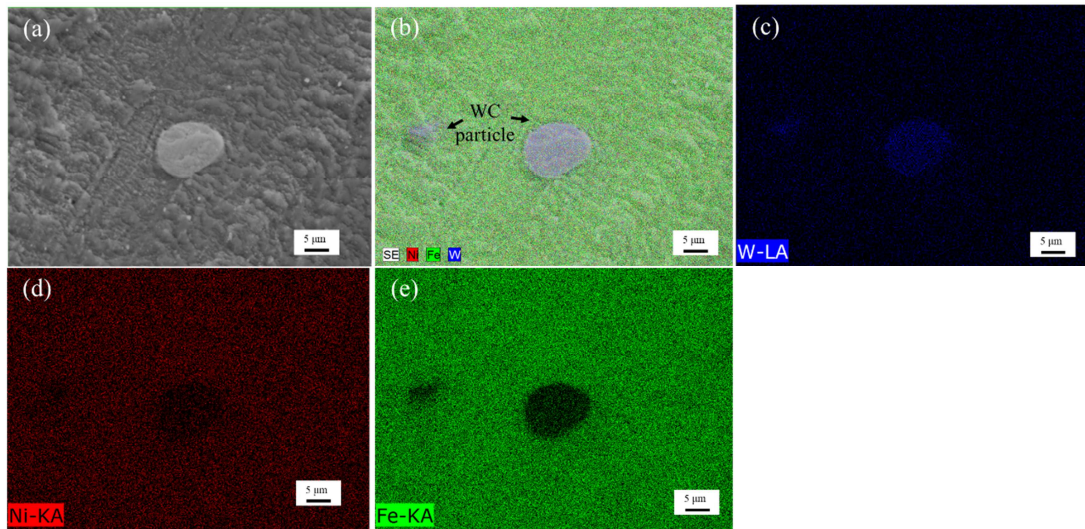


Fig. 14 EDS analysis result of the single WC particle remained on the worn surface of the WC/MS300 fabricated at 125 J/m.

4. Conclusion

In this work, the maraging steel 300 composites reinforced by the WC particles were fabricated by SLM under different laser powers. The present work seeks to provide an effectively method to improve the tribological performance MS300 through the reinforcement of WC particle by selective laser melting.

- 1) The microstructure observation shows that gradient interface as well as diffusion interlayer surrounding the WC particles in MS300 matrix. However, with the increasing linear energy density, the WC particles experienced more sufficient dissolution and such gradient interface was disappeared.
- 2) After slight increase from 95 J/m, the microhardness rapidly drops as the linear energy density further increases above 110 J/m owing to the linear energy density induced transition of the phase, grain size and defects in the matrix.
- 3) Friction test shows that an improved wear-resistance at higher linear energy density, whereas the lowest wear rate value and best wear resistance performance was obtained at the linear energy density of 125 J/m. The wear mechanism during was changed from abrasion to adhesion as the linear energy density increases.

4) By comparing the WC/MS300 composite with the pure MS300 at the same linear energy density of 125 J/m, it can be found that the role reinforced WC particles can evidently increase the microhardness as well as the wear performance.

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