Microstructure and tribological property of Ni–MoS$_2$ composite coatings prepared by ultrasonic and mechanical stirring electrodeposition

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Abstract

To optimize the deteriorated mechanical properties of the loose MoS$_2$ coatings, ultrasonic oscillation was employed to change the deposition mode of MoS$_2$ particles, hence their structure and mechanical properties. The electrodeposition of Ni–MoS$_2$ self-lubricating composite coating was assisted by magnetic stirring (Mag-NM) or ultrasonic oscillation (US-NM), respectively. A comparison of coating structure, depositing mechanism, and friction performance was investigated by means of XRD and SEM. Moreover, the influence of MoS$_2$ concentration on the coating morphology and the friction performance was explored. The results demonstrated that the Ni–MoS$_2$ composite coatings have superior antifriction performance compared to pure Ni coating. Comparing to the porous structure of Mag-NM coatings, the US-NM coatings obtained compact structures with well-dispersive MoS$_2$ particles. There is 80.4% hardness reduction occurred on the Mag-NM coatings, but only 8.8% reduction of the US-NM coatings. With the increase of MoS$_2$ concentration, a clear and continuous wear trace was formed on the Mag-NM coatings. The protrusion grinding indicates the excellent wear resistance of the US-NM coatings.

1. Introduction

The self-lubricating composite coating is a major area of interest within the field of solid lubrication [1]. With insoluble particles co-deposited as lubricants, outstanding wear resistance and self-lubrication can be achieved [2–4], which plays a key role in improving machine efficiency, energy conservation [5]. The two-dimensional material MoS$_2$ with unique lamellar structure performs excellent low friction property associated with easy shearing of the weak interlayer bonds (van der Waals) [5–7]. The Ni coatings incorporated with MoS$_2$ act as important alternative materials for new generation high-performance gas turbine engines [7] due to the excellent stability of MoS$_2$ particles in high temperature and high vacuum service environment [2, 8].

Much of the current literature on deposited MoS$_2$ self-lubricating coating pays particular attention to MoS$_2$ concentration [2, 5], particle structure [6, 9, 10] as well as its friction performance at elevated temperature [4, 7, 11]. Cardinal et al. [2] found that the friction coefficient of the pulse electrodeposited Ni–W coatings were reduced to half its value when the concentration of MoS$_2$ was low. In contrast, Ni–W nanostructured coatings with high MoS$_2$ content shew high surface roughness and poor frictional behavior due to a porous sponge-like structure. Moreover, the problem of MoS$_2$ particles agglomeration and sedimentation in composite electrodeposition have great impacts on coating structure and uniformity [12, 13]. It has been proposed that the deposit priority of the semiconducting MoS$_2$ particles mostly accounts for its porous structure [1, 14, 15]. The addition of cationic surfactant [3, 4, 10], preventing the particle agglomeration by charge repulsion of the...
positive ion adsorbed to the surface of MoS₂ [1], becomes the most common solution. Meanwhile, magnetic stirring [9, 13, 14] is often applied to keep the suspension of the MoS₂ particles.

Ultrasonic oscillation is widely used as a cleaning and dispersing method to synthesize composite coatings co-deposited with ceramic particles, such as Al₂O₃ [16], TiO₂ [17], SiC [18], and IF-WS₂ [10]. Garcia-Lecina [13] et al have conducted a comprehensive study of the effect of mechanical and ultrasound agitation on the properties of electrodeposited Ni–Al₂O₃ composite coatings. Compared with the magnetic stirring deposition coating, there was more Al₂O₃ incorporated into the ultrasonic coatings with better particle dispersion, resulting in higher hardness and better tribological performance. The impacts ultrasonic oscillation to enhance mass transportation, to optimize particles dispersion and to accelerate particles deagglomeration have been revealed by lots of research [12, 13, 19]. Nevertheless, how ultrasonic oscillation acts on the deposit priority of the semiconducting MoS₂ particles and the improvement of friction performance of MoS₂ incorporated composite coating are seldom point out.

In the present work, the Ni–MoS₂ composite films were prepared by electrodeposition under magnetic stirring and ultrasonic oscillation, and the influence of agitation system variation on the deposition mechanism of MoS₂ has been studied. The structure and morphology of the electrodeposited Ni–MoS₂ composite coatings were studied using x-ray diffraction, scanning electron microscopy. The hardness and wear of the composite coatings were characterized to evaluate the mechanical property and self-lubrication. The frictional behavior of Ni–MoS₂ were analyzed.

2. Experimental methods

2.1. Preparation of Ni–MoS₂ coatings
Ni–MoS₂ composite coatings were deposited on the steel substrate by electrolytic deposition. The soluble ingredients of the electrolyte solution were added in sequence shown in table 1. Na (OH) (0.1 mol·L⁻¹) solution was used to adjust pH to 4 [1, 10] after each ingredient fully dissolved. Then MoS₂ particles were added to the solution, with 1 h magnetic stirring. The MoS₂ are irregular tabular particles with distinctive lamellar feature (figure 1), with the size ranging from 1 um to 4um. The MoS₂ concentration is set as 1, 1.5, 2 g·L⁻¹, respectively. After the MoS₂ particles completely wet and suspended, another ultrasonic oscillation treatment was applied for 1 h to reduce agglomeration and enhance the particle suspension [13, 20].

The coatings were deposited on a planar, 5 mm thick 45 steel substrate with a diameter of 15 mm, which was polished using 5000 silicon carbide papers. Afterward, the polished substrate was rinsed in rectified alcohol and then dried with a blower before electrodeposition. The matrix is clamped in a circular mold made of polytetrafluoroethylene with an exposed area of 1 cm². A platinum anode (15 mm × 15 mm) was positioned at 2 cm distance parallel to the cathode surface.

Composite electrodeposition was carried out at a constant current density of 4.0 A·dm⁻² [1] for 60 min using ITE6411 high precision DC source. Mechanical agitation and ultrasonic oscillation are respectively provided by DF-101S thermostatic heating magnetic stirrer and KH2200DB CNC ultrasonic cleaner, respectively. The deposition temperature of Ni–MoS₂ coatings was maintained at 45 ± 5 °C. The detailed experimental parameters were listed in table 1.

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**Table 1. Electrolyte composition and electrodeposition parameters for Ni–MoS₂ composite coating.**

| Composition                               | Parameters        |
|-------------------------------------------|-------------------|
| NiSO₄·6H₂O                                 | 262.85 g·L⁻¹      |
| H₂BO₃                                     | 30.9 g·L⁻¹        |
| Saccharin                                 | 1 g·L⁻¹           |
| 1, 4-butyndiol coumarin                   | 0.2 g·L⁻¹         |
| Dodecyltrimethylammonium                  | 0.028 g·L⁻¹       |
| Bromide(DTAB)                             | 1 g·L⁻¹           |
| pH                                        | 4 ± 0.2           |
| Temperature                               | 45 ± 5°C [13]     |
| Current density                           | 4 A·dm⁻²          |
| Time                                      | 60 min            |
| Mechanical stirring                       | 120 rpm [5]       |
| Ultrasonic oscillation                    | 40KHz, 60 W·cm⁻²  |

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2.2. Characterization
The phase determination of the deposits was investigated by means of D/MAX2500V x-ray diffraction (XRD) using Cu-Kα radiation, and the working voltage and current were 60 kV and 80 mA, respectively. The surface and cross-section morphology, and wear scars of the coatings were examined by a JSM-6490LV scanning electron microscopy (SEM). The chemical composition of the coatings was analyzed from cross-sections photomicrographs using an XFlash6160 energy dispersive spectrometer (EDS) coupled to the SEM. Meantime, Image J software was employed to measure the thickness of the Ni–MoS2 coatings from the cross-section images.

The hardness of Ni–MoS2 self-lubricating coatings was evaluated on a slightly polished surface section using an HVS-1000 micro indenter with an applied load of 100 g and dwell time of 15 s. Seven readings were taken from each sample and the values of the hardness were averaged. Wear tests were performed using a reciprocating tribometer (CFT-I) at room temperature. A 15GCr steel ball with a diameter of 5 mm was used as a counterface material. The wear tests were conducted under the load of 320 g for 15 min, with a reciprocating length of 5 mm and friction speed of 150 \( \cdot \) m\(^{-1}\).

3. Results

3.1. XRD of Ni–MoS2 coatings
X-Ray Diffraction (XRD) was employed to study the phase of Ni-MoS2 coatings (figure 2). It is obvious that the MoS2 peaks intensity of the deposited Ni–MoS2 coatings with Magnetic stirring (Mag-NM) is stronger than that of the deposited Ni–MoS2 coatings accompanied by the ultrasonic oscillations (US-NM), indicating a higher content of MoS2 co-deposition [2]. The MoS2 peaks of the Mag-NM coatings are intensified when the MoS2 concentration increase from 1 g\( \cdot \)L\(^{-1}\) to 2 g\( \cdot \)L\(^{-1}\), while no obvious change of the MoS2 peaks intensity was observed in the US-NM coatings. In comparison to the Mag-NM coatings, for there are fewer large agglomerates incorporated into the deposits with ultrasonic deagglomeration, it occurs a reduction in particle content in the US-NM coatings, corresponding to the weak MoS2 peak intensity [12]. During the direct current deposition process, the mass transport enhancement caused by ultrasound are greater than that of the increase of MoS2 concentration [19]. Therefore, the content of incorporated MoS2 particles maintains in the US-NM coatings. The dominant peaks of MoS2 locates at 14° and 39°, respectively corresponding to the (002) and (100) plane. The (002) plane demonstrates a parallel orientation to the substrate surface, where the active edge sites of MoS2 are well protected [21, 22]. Moreover, the tangential motion of the crystalline layers of MoS2 can be easily achieved at the parallel orientation, providing the Ni–MoS2 coatings a good lubrication property as well as better oxidation resistance. However, the MoS2 peak of (100) plane represents a vertical orientation perpendicular to the substrate, which largely accounts for the porous and columnar structure of Mag-NM coatings [22]. Scherrer equation was applied to calculate the average grain sizes of Ni–MoS2 composite coatings using the dominant Ni (111) peak at 45° [5]. The average crystal size has been calculated as 21.8 nm for the Mag-NM coating and 19.4 nm for the US-NM coatings (MoS2 concentration is 1 g\( \cdot \)L\(^{-1}\)). It is illustrated that the acoustic cavitation phenomenon in the ultrasonic environment contributes to refining crystal grain [13].
3.2. Morphology and structures of Ni–MoS₂ coatings

The surface morphology of the Mag-NM coatings and the US-NM coatings with different MoS₂ concentration was analyzed by SEM (figure 3). Distinct surface topographies were observed for Mag-NM coatings and US-NM coatings. The Mag-NM coatings are formed by the piling up of lamellar MoS₂ under the magnetic stirring treatment. Ni ions preferentially deposit on the surface of MoS₂ for the adhesion between these mutually contacted MoS₂ particles (figures 3(a)−(c)), therefore, the formation of Mag-NM coatings is often referred as dendrite growth process [14]. There are lots of dendritic gaps observed in Mag-NM coatings. As the MoS₂ concentration increases, these gaps gradually shrink since high MoS₂ concentration offers more growing points of dendrite at the substrate and promote the extension of dendrite branches [5, 9]. In contrast with the loose structure of Mag-NM coatings, the US-NM coatings exhibit unique surface feature with irregular protrusions (figures 3(d)−(f)). The size of the irregular protrusions increases when the MoS₂ concentration increases from 1 g L⁻¹ to 1.5 g L⁻¹ US-NM coating. When MoS₂ concentration reaches to 2 g L⁻¹, no protrusions were distinctly identified but structures with flake feature were obviously observed (inset of figure 3(f)).

The cross-section images of the composite coatings (the MoS₂ concentration is 1 g L⁻¹) furtherly indicate the different growth mechanism of Mag-NM coating and US-NM coating. The Mag-NM coating consists of an initial Ni-rich layer (∼6 um thick) adhered to the substrate, and dendritic epitaxial Ni-based coating with lamellar MoS₂ particles incorporated (∼72 um). The agglomerated MoS₂ particles and their stacking structure
are easily distinguished in figure 4(a). The dendritic gaps (the black area) are obviously observed in the cross-section image of Mag-NM coating, further demonstrating its high porosity and loose structure. The area-scanning composition analysis of the Mag-NM coating indicates a relatively high MoS2 content of 6.02 wt%, in contrast to the 2.44 wt% of US-NM coating. The incorporated MoS2 particles possess small volume and are well dispersed by the application of ultrasonic oscillation. As for the Fe element detected in the Ni–MoS2 coatings, there is replacement reaction happened between Fe (from the 45 steel) and Ni (deposition solution) once the sample being placed in the solution. Consequently, a small number of Fe element exist in the solution, hence in the composite coatings. Without continuous dendritic gaps and large agglomerates of MoS2 particles, the US-NM coating is more uniform and compact than the Mag-NM coating, seen as figure 4(b). The US-NM coating exhibits good adhesion, low surface roughness and larger thickness (≈80 um), with few pores and defects observed [23].

3.3. Deposition mechanism analysis

The schematic diagram of the deposition mechanism of the Mag-NM and the US-NM coatings is illustrated in figure 5. Concerning the deposition mechanism of Mag-NM coating, the semi-conductivity of MoS2 particles have a significant impact on it [1, 5]. The addition of semi-conductive MoS2 particles disturbs the current distribution, making the electrolytic current concentrating in the vicinity of MoS2 particles [5, 24]. Therefore, there is a bigger electric field formed around MoS2 to speed up its migration to the cathode, as well as the migration of Ni2+ ions towards MoS2 particles. Not only Ni2+ ions easily accept electrons and preferably deposit on the incorporated MoS2 [5, 25], but other MoS2 particles are also more likely attracted to its surface. This can be attributed to the dendrite growth structure that makes MoS2 particles hard to be incorporated in Ni electroplating [14], presenting a trend of particle agglomeration of the pre-dispersed MoS2 [19]. On account of the irregular shape of MoS2 particles, particles pile up forming dendritic branches, leaving a large amount of stacked gap behind. The agglomerated MoS2 particles with bigger volume would further exacerbate the nonuniform distribution of current [26], preventing the reduction of Ni ions on the stacked gap, finally forming many pores and defects in the Mag-NM coatings. The epitaxial growth of dendritic branches and high porosity

![Figure 4. SEM cross-sectional image of (a) Mag-NM coating and (b) US-NM coating.](image-url)
make the Mag-NM coatings having a loose structure. Other than the magnetic agitation, there is a cavitation effect in the ultrasonic environment [13, 19]. High temperatures and pressures (around 5000 K and 1000 atm, respectively) are created by the violently collapses of the liquid cavity, forming high-energy liquid jet streams around 400 km·h\(^{-1}\) [13, 27]. The high-energy liquid jet streams could suppress the agglomeration trend of the MoS\(_2\) particles, to keep the MoS\(_2\) particles in a small volume, and finally to be covered up by the reduction of Ni ions [10, 25]. The Ni deposition on the surface of MoS\(_2\) particles can reduce the irregularity of MoS\(_2\) particles, therefore eliminating the possible gaps during the deposition process. Small MoS\(_2\) particles, corresponding to a lower degree of current disturbance, offer Ni ions an opportunity to deposit uniformly and efficiently [10]. Moreover, these high-energy liquid jet streams could scatter these mutual absorbed MoS\(_2\) particles, reducing the continuous deposition of particles and optimizing the particles dispersity in US-NM coating [10, 19]. Therefore, the US-NM coatings acquire a compact structure, with well-dispersed MoS\(_2\) and high Ni content.

### 3.4. Mechanical properties

#### 3.4.1. Hardness

The hardness of the Mag-NM and the US-NM coatings with different MoS\(_2\) concentrations is shown in figure 6 with pure Ni coating as a control sample. It is shown that the Ni–MoS\(_2\) coatings exhibit hardness reduction compared to the Ni coating. The hardness of the US-NM coatings [10, 18] is distinctively higher than that of the Mag-NM coatings. At the MoS\(_2\) concentration of 2 g·L\(^{-1}\), there is about 80% hardness reduction [2] (from 497.56 HV\(_{0.1}\) to 97.62 HV\(_{0.1}\)) of the Mag-NM compared to the Ni coating [18, 28] whereas the hardness of the US-NM coating indicates slightly reduction (453.92 HV\(_{0.1}\)). Moreover, obvious decrease is observed in the hardness of Mag-NM coatings as the MoS\(_2\) concentration increases. On the contrary, the MoS\(_2\) concentration has less effect on the hardness of US-NM coatings. The variation of hardness reduction is largely attributed to the
difference in coating structure. The porous sponge-like structure is largely accounting for the decline of Mag-NM coating hardness, which is reported in a previous study [2]. As the MoS2 concentration increases, more MoS2 particles are mutually attracted, the nonuniform current allots more Ni ions to deposit on the MoS2 surface, leaving little Ni ions reduced in the contact area to consolidate the particle adhesion. The decrease of Ni content, which plays the supporting and binding roles, greatly weakens the load-carrying property of the Mag-NM coatings, resulting in a hardness reduction. As for the US-NM coatings, with the absence of agglomerated MoS2 particles and a lower degree of current disturbance, plenty of Ni ions are uniformly and efficiently deposited to form the compact US-NM coatings, therefore obtained a relatively high hardness.

3.4.2. Tribological properties

The friction coefficients of the Ni–MoS2 coatings with different MoS2 concentrations were presented in figure 7, with that of Ni coating as a control sample. It can be found that the friction coefficients of pure Ni coating [18, 28] keep stable at 0.92 after a running-in stage. In contrast, there is more than 50% reduction of friction coefficients of the Ni–MoS2 composite coatings [2, 5], indicating that the incorporated MoS2 particles indeed performs good friction reduction effect. The friction coefficients of the Mag-NM coatings descend with the increase of MoS2 concentration. The average friction coefficients are 0.38, 0.30 and 0.17, corresponding to the Mag-NM coatings with the concentration of 1, 1.5, 2 g·L\(^{-1}\), demonstrating gradually improved antifriction. Except for some fluctuations for the 2 g·L\(^{-1}\) Mag-NM coatings, these friction coefficient curves are nearly parallel to each other, which illustrates the same friction behavior happened during the reciprocating slide process. In contrast, the friction coefficients of the US-NM coatings ascend following the increase of MoS2 concentration. The average friction coefficient of the US-NM coating increases from 0.17 at 1 g·L\(^{-1}\) to 0.28 at 1.5 g·L\(^{-1}\) MoS2 concentration with the same growth rate. The friction coefficient of the 2 g·L\(^{-1}\) US-NM coating starts at 0.20 and ends at 0.41, with a faster-ascending speed and fluctuations after 9 min reciprocating slide, indicating the antifriction deterioration of the US-NM coatings and the change of friction behavior.

After a 15 min wear test at the normal load of 320 g, the morphologies of wear tracks of Mag-NM and US-NM coatings with different MoS2 concentration are imaged under SEM (figure 8). As for the Mag-NM coatings, the grinding trace is shown as the dark-gray polished area on the surface. With the increase of MoS2 concentration, the area of the dark-gray polished surface gets enlarged and its continuity gets optimized. A complete wear scar with a clear boundary line can be seen in the 2 g·L\(^{-1}\) Mag-NM coating (figure 8(c)). In contrast, the dark-gray wear scar is sparsely distributed, mainly located in the protrusions on the surface of US-NM coatings (figures 8(d)–(f)). With higher MoS2 concentration and bigger round-cell protrusions, the worn area gradually gets enlarged, but no continuous wear mark is observed. Therefore, there is nearly no damage caused at the underneath layer below these protrusions. The origin feature of the US-NM coatings is maintained by the protection of these surface protrusion.

With the normal force applied, a large amount of dendritic branches fracture during the reciprocating slide process. On the one hand, the fractured branches are used to pad these dendritic gaps to form a compact wear trace. On the other hand, the incorporated MoS2 particles of the parallel orientation are subject to the shear force [22], forming the tribofilm with reduced friction [4]. With the increase of MoS2 concentration and the concomitant decrease of hardness, the uncompacted particles (figure 9(a)) vanish and the polished surface gets smoother (figure 9(b)). As for the US-NM coatings, the worn plane of the surface protrusions is obviously
observed in figure 9(c), with debris scattered. The virgin surface of the underneath layer is maintained, indicating that the grinding only occurs at the surface protrusions. Plastic deformation happened on the surface protrusions with distinct micro-scuffing along the sliding direction, indicating the aggravated adhesive wear of the US-NM coatings [4]. Flake characteristic of stacked MoS2 particles became fragments around the rubbed protrusion (figure 9(d)). The area-scanning composition analysis of the wear scar reveals the transfer of Fe element, demonstrating a metal-metal direct contact between the US-NM coating and the GCr15 ball surface [5]. The stacked MoS2 particles form large and loose particle cluster around worn protrusions after a couple of periods, acting as the reserved lubricant to supply lubricants persistently during the whole dry sliding process [29].

4. Discussion

The lamellar MoS2 are stacking in repeated unit layers of (S–Mo–S) wherein the molybdenum atoms are sandwiched between layers of sulfur atoms with covalent bonding [6]. Such a strong ionic bond between S and Mo provides the lamellar MoS2 a high resistance to asperities penetration, which promises the wear resistance of the MoS2 lubricating film. What’s more, the considerable thermal expansion of lamellar MoS2 generated from the friction heat can offset part of the exerted compressive stress during the sliding, easing the damage to the tribofilm [30]. While the interaction between the (S–Mo–S) unit layers is loosely bound by the weak van der Waals forces [2]. With the GCr15 ball loaded, the interplanar shear of the adjoining S–S layers occurs easily, which is responsible for the excellent lubricity of Ni–MoS2 coatings [31]. Moreover, compressive strains happen between the MoS2 layers during the sliding, making charge transfer from Mo to S atom. With more charges around the S atomic plane, there is stronger electrostatic coulombic repulsive interaction between the MoS2 sheets, resulting in lower friction [32].

Owing to the its porous structure and the poor bearing performance, the concomitantly reduced hardness makes the Mag-NM coatings soft films. The direct grinding of the hard couple is avoided by the soft-hard coordination between the soft Mag-NM coatings and the GCr15 counterpart ball [4]. Therefore, the Mag-NM coatings are qualified with good antifriction performance. Contrary to the soft Mag-NM coatings, the high Ni content and compact structure equip the US-NM coatings with relatively high hardness. Consequently, there is excellent wear resistance of the US-NM coatings [10, 33].

The friction force is the product of the real contact area and the shear strength of lubricating coatings [33, 34]. A high elastic modulus is essential to the shear strength of the Ni–MoS2 coatings. With the increase of the MoS2 content, the elastic modulus of the Ni–MoS2 coatings decrease [34]. Moreover, the contact area has an inverse ratio to the elastic modulus, where a high elastic modulus corresponded to a small contact area. Owing to a relative higher MoS2 content of the Mag-NM coatings, their elastic modulus decrease, hence the shear strength. Ascribed to the low shear strength, the low friction coefficients of the Mag-NM coatings are seen in figure 7(a). Furthermore, there are large deformation and wear of the Mag-NM coatings accounting for their
decreased elastic modulus. It is obvious that the integrity of the formed MoS₂ lubricating film reflects the lubricating ability of the Mag-NM composite coatings [34], which corresponds to the successively reduced friction coefficient (figure 7(a)). Due to the rough and inhomogeneous surfaces of the Mag-NM coatings, the friction curves are not smooth. In contrast, the US-NM coatings have high elastic modulus for their low MoS₂ content, hence high shear strength. With the approximate mechanical property of the US-NM coatings, the friction coefficients of the US-NM coatings mainly depend on the contact area [33]. The surface protrusions of the US-NM coatings can provide load and anti-wear supports during the sliding contact [29]. Owing to the protrusions worn rather than plane worn, the shrink of contact area accounts for the lower friction coefficients of the US-NM coatings compared with that of pure nickel coating. With the increase of MoS₂ concentration, surface protrusions get coarsening and the grinding area of US-NM coatings get enlarged correspondingly (figures 8(a)–(c)). Therefore, the friction coefficients ascend with the increased MoS₂ concentration. For 2 g·L⁻¹ US-NM coating, the observed flake characteristic on the surface protrusions is formed by the lamellar MoS₂ particles. Their excellent lubricity provides the 2 g·L⁻¹ US-NM coating a low friction coefficient in the former 9 min. The deteriorated mechanical properties of these protrusions lead to rapid climb of its friction coefficient (figure 7(b)).

![Figure 9. SEM of wear scars of Ni–MoS₂ composite coating prepared at MoS₂ (a), (c) 1 g·L⁻¹, (b), (d) 2 g·L⁻¹, (a), (b) under magnetic stirring, (c), (d) under ultrasonic agitation and (e), (f) EDS.](image)
5. Conclusion

(1) The particle agglomeration and priority deposition of semi-conductive MoS$_2$ results in a porous sponge-like structure for the Mag-NM coatings by the epitaxial growth of dendritic branches. Assisted with ultrasound, the relatively optimized uniformity of current distribution allows more Ni ions to uniformly and efficiently deposit, forming the compact US-NM coatings with well-dispersive MoS$_2$ particles and surface protrusions.

(2) The porous sponge-like structure accounts for the 80.4% decline (from 497.56 HV$_{0.1}$ to 97.62 HV$_{0.1}$) of Mag-NM coating hardnes. With well-dispersed MoS$_2$ particles and a compact structure, the US-NM coatings maintain 453.92 HV$_{0.1}$ with 2 g L$^{-1}$ MoS$_2$.

(3) The Ni–MoS$_2$ composite coatings demonstrate superior antifriction property compared to pure Ni coating. With the increase of MoS$_2$ concentration, the friction coefficients of the Mag-NM coatings decrease in turn with the reduced shear strength. As for the US-NM coatings, they exhibit low level but an upward trend of friction coefficients due to the enlarging grinding area with the increase of the MoS$_2$ content.

(4) The shear strength of the Mag-NM coatings is reduced with high MoS$_2$ content, which accounts for good antifriction for the Mag-NM coatings. Plastic deformation happens on the surface protrusions of the US-NM coatings with the underneath layer well protected. The distinct micro-scuffing indicates the aggravated adhesive wear of US-NM coatings.

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Compliance with ethical standards

The authors declare that they have no conflict of interest.

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