Optimization of several surface treatment processes for alleviating fretting damage of a locking pin

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Abstract: The operational safety and reliability of a variable gauge train are affected by the anti-fretting wear performance of the locking mechanism. The main purpose of this study is to optimize the surface treatment process for a locking pin material under actual service conditions to alleviate fretting damage. Based on the two basic principles of surface strengthening and friction reduction, a substrate (AISI 4135 steel) surface was treated by laser quenching (LQ), plasma nitriding (PN), and bonded MoS₂ coating. Systematic fretting wear tests were conducted, and the wear behavior and damage mechanism of various treated surfaces were comprehensively investigated. The results indicate that the wear resistances of the LQ- and PN-treated surfaces were significantly improved, and their main wear mechanisms were abrasive wear, delamination, and oxidation wear. The MoS₂ coating exhibits the lowest friction coefficient and energy dissipation due to its self-lubricating property, but it incurs the highest wear rate and failure in the form of plastic deformation. Furthermore, the rough compound layer with a high hardness on the PN-treated surface is conducive to the formation and maintenance of the third-body contact at the fretting interface, consequently resulting in a significant reduction in wear. An optimal surface treatment process for alleviating fretting damage of the locking pin is recommended via comprehensive evaluation, which provides a reference for the anti-fretting protection of related mechanical components.

Keywords: locking pin; fretting wear; surface treatment; dynamic response; wear mechanism

1 Introduction

Railways, as the most widely used type of land transportation, play a crucial role in promoting the development of economies and trade worldwide [1]. The existence of various types of rail gauges in different countries and regions has hampered further improvements in the comprehensive efficiency of railway transport. However, the emergence of variable gauge trains has considerably solved this problem. At present, multiform variable gauge trains have been produced and put into service around the world [2], and China has developed a high-speed variable gauge

bogie system (see Fig. 1(a)) for the 1,435/1,520 mm gauge switch. As one of the key components of the bogie, the locking mechanism (as shown in Fig. 1(b)) has the function of a wheel track switch and mechanical locking, and its running reliability strongly determines the safe operation of the trains. During the process of wheel track switching, the locking pin will be jacked up by a ramp track (the spring will be compressed), and will be separated from the matching locking block. Meanwhile, the wheels will be driven by the guiding device to move along the axle to a prescribed gauge. Thereafter, the bogie will be moved away from the ramp, the spring will be reset to make

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Fig. 1 Diagrams of (a) bogie structure for variable gauge train and (b) service conditions of the locking mechanism.

the locking pin and the locking block re-engage in another position. However, given the clearance fit of the locking mechanism and the alternating loads due to mechanical vibration during the running process, the fretting damage [3–5] always occurs on the matching tooth surface of the locking pin. Based on the actual operating conditions of a train and the relevant standards [6, 7], the lateral load applied on the locking mechanism can reach 65.6 kN when the variable gauge train is running along a curve. As shown in Fig. 2, the maximum contact stress (601.9 MPa) on the contact surface of the locking pin is obtained by finite element analysis. Moreover, as the relative slip components C_{slip1} (longitudinal) and C_{slip2} (lateral) of the mating interface at each node are extracted, the total relative slip of each unit is calculated and the maximum value is 100 μm . Studies have confirmed that the fretting damage in the key mechanical structures of the bogie strongly threatens the safety of the high-speed trains [8, 9]. Therefore, to ensure the operational reliability of the locking mechanism, it is crucial to implement surface engineering design of the friction pair material and perform research on the fretting wear behavior

under actual service conditions.

The locking pin is made of AISI 4135 steel (similar to 35CrMo steel), which is a common engineering material and has been widely used in various mechanical components due to its excellent mechanical properties. The researches on fretting wear mechanism for the kind of alloy steels under various factors provide lots of knowledge of the anti-fretting protection for the locking mechanism [10–12]. Among the various measures to alleviate fretting damage [13], surface modification as an efficient and economical solution has been widely used in the industrial field [14, 15]. In this study, optimizing a suitable surface treatment method is essential to protect the locking pin from fretting damage.

In numerous surface modification methodologies, plasma nitriding (PN) as a thermochemical diffusion process, has the advantages of low cost, friendly environment, and excellent wear resistance. The surface hardness of ferrite steel has been found to increase significantly after PN-treated, yielding $\epsilon\text{-Fe}_{2.3}\text{N}$ and $\gamma'\text{-Fe}_4\text{N}$ phases [16, 17]. Meanwhile, PN treatment of different engineering materials has been proven to enhance fatigue properties [18, 19] and to reduce

Fig. 2 (a) Finite element model of the locking mechanism and maximum contact stresses of (b) locking pin and (c) locking block.

surface wear [20–22]. Moreover, laser quenching (LQ), as an economical and efficient surface heat treatment, is a rapid austenitizing process [23]. The martensite structure generated by LQ treatment significantly enhances the hardness and strength of a substrate surface, and a residual compressive stress layer is also beneficial for resisting fretting wear [24, 25]. According to the research of Wang et al. [26], the wear resistance of railway wheel-rail material improved after LQ treatment, and the wear volumes of the wheel and rail materials were reduced by 92.9% and 80%, respectively. Lesyk et al. [27] proved that diode laser heat treatment enhanced the hardness and wear resistance of tool steel. In addition, in contrast to the method of improving the surface strength and hardness, solid lubricant coatings have also been proven to effectively reduce material wear owing to their antifriction effect [28, 29]. Luo et al. [30, 31] conducted detailed studies on the wear behavior of MoS₂ coating in the form of torsion and rotational fretting wear and confirmed that the solid lubricant coating could alter the fretting state and reduce the wear of a substrate.

Previously, the wear resistances of steels after surface treatment, such as PN and LQ, have been successively studied. However, these researches mainly focused on the ball/pin on disk or reciprocating sliding wear mode [15, 22, 23], and the obtained data, such as the dynamic response and dissipated energy of the friction interface, are scarce. Moreover, comparative investigations of fretting wear behavior and damage mechanisms of various treatment surfaces based on engineering applications have rarely been reported.

In this study, AISI 4135 steel was treated by three economical and reliable surface modification techniques from two aspects of improving surface strength and reducing surface friction. Fretting wear tests were performed based on the actual service conditions. The wear behavior, dynamic response, energy dissipation, and damage mechanism of various treated surfaces

were investigated. The results can provide theoretical support and application guidance for the design of surface modification of the locking mechanism.

2 Materials and experimental procedures

2.1 Materials and surface treatment

In this study, the commercial material AISI 4135 steel was used as the substrate, and its chemical composition were 0.35 wt% of C, 0.32 wt% of Si, 0.47 wt% of Mn, 0.87 wt% of Cr, 0.83 wt% of Mo, and Fe balance. The main mechanical properties are listed in Table 1. After quenched-tempered heat treatment (heated to 850 °C for 30 min, quenched in oil, and then tempered at 600 °C for 60 min), the substrate was machined into samples with dimensions of 25 mm × 12 mm × 10 mm with a surface roughness of approximately 0.03 µm. Prior to surface treatment, the specimens were ultrasonically cleaned in acetone and dried in nitrogen. Subsequently, the prepared substrate specimens were separately treated by LQ, PN, and bonded MoS₂ coating. The process parameters of the three treatment methods are determined based on the properties of the locking pin material and the pre-test results.

A CO₂ cross-flow laser was used to conduct LQ on the pretreated samples, and a hardening layer was formed on the surface. The laser power was 1.9 kW, the dimensions of the laser beam were 10 mm × 2 mm on the focal plane, and the scanning speed was 300 mm/min.

Additionally, PN treatment was performed using an LDM-200 PN furnace. First, the samples were placed in the furnace as a cathode, and the vacuum degree in the chamber was adjusted to reach 300–350 Pa. The surfaces of the samples were cleaned for 20 min with hydrogen ion sputtering at 700 V. Then, a mixture of 15% N₂ and 85% H₂ was injected to maintain a pressure of 200 Pa in the chamber, and the nitriding

Table 1 Main mechanical properties of AISI 4135 steel and SiC ceramic ball.

Material	Tensile strength (MPa)	Yield strength (MPa)	Young's modulus (GPa)	Hardness (HV)	Elongation (%)
AISI 4135	≥985	≥835	206	240–269	≥14
SiC	—	—	400	1,650	—

process was implemented at 530 °C for 15 h. Finally, low-flow H_2 was continued to be fed into the furnace until the temperature was below 100 °C.

The preparation of the bonded MoS_2 coating includes the following steps. MoS_2 particles with a grain diameter of 0.5–1 μm were uniformly dispersed in the adhesive (polyamide-imide and epoxy resin system) to form the mixture. The surfaces of the substrate samples were sanded in a random direction until rough ($R_a = 2 \pm 0.5 \mu m$) and then coated with the mixture via spraying. Finally, the coating was solidified by heating to 150 °C for 30 min and then to 170 °C for 60 min.

2.2 Experimental method and parameters

The fretting wear test was performed using a multifunctional fretting wear test system (MFTS-01) developed by Southwest Jiaotong University. A structural schematic and photograph of the test system are shown in Figs. 3(a) and 3(b), respectively. The reciprocating tangential motion is driven by a piezoelectric actuator, and the displacement control precision reached $\pm 0.2 \mu m$ by the feedback of a grating encoder. Meanwhile, the force closed-loop control of the normal load (F_n) is realized by a servo loader and a load sensor, with a control accuracy of $\pm 0.3 N$. All signals are acquired by the multi-channel data acquisition unit, and then analyzed and processed by the host computer, consequently transmitted to the servo control unit to drive the actuators. Meanwhile, the tangential force (F_t) and actual displacement (D) during the test were collected dynamically in real-time; hence, the typical F_t – D curves were automatically generated. As shown in Fig. 3(c), fretting wear

tests were completed under the sphere–plane contact mode.

To simulate the wear state of the material under diverse vibration amplitudes, the fretting displacements (D) were controlled at 15, 25, and 50 μm . Based on the actual contact stress state of the locking pin and the experimental requirements, the normal load (20 N) is determined by the Hertz contact calculation, and the contact stress on the substrate is 841.1 MPa. In the present study, a high-stability SiC ceramic ball with a diameter of 10 mm was used as the counterpart, and its main parameters are listed in Table 1. The summarized test parameters are shown in Table 2, and each group of experiments was repeated three times.

2.3 Characterization of treated surfaces and wear scars

The phase structures of the specimens were analyzed using X-ray diffraction (XRD, D8 ADVANCE A25X, Bruker, Germany). A Vickers microhardness tester (MVK-H21, Akashi, Japan) was used to measure the hardness values of the sample surface and cross-section. Furthermore, the profile structures of the surface-treated samples were observed using scanning electron microscopy (SEM, JSM-6610, JEOL, Japan) with an accelerating voltage of 20 kV.

After completion of the fretting wear tests, the morphologies of the fretting scars were measured using SEM and white light interferometry (Contour GT, Bruker, Germany). The chemical states of the elements on the worn surfaces were surveyed via X-ray photoelectron spectroscopy (XPS, ESCALAB 250 Xi, Thermo, USA) with an Al $K\alpha$ X-ray source.

Fig. 3 (a) Structure of the fretting test system (1: gantry system, 2: piezoelectric actuator, 3: friction sensor, 4: linear guide, 5: load, 6: servo loader, 7: grating displacement sensor, 8: linear guide); (b) photo of the test module; and (c) sphere–plane contact model.

Table 2 Main test parameters in the study.

Test parameter	
Displacement amplitude (D , μm)	15, 25, 50
Normal load (F_n , N)	20
Frequency (Hz)	2
Number of cycles (N)	$1-5 \times 10^3$
Temperature and humidity	$23 \pm 3^\circ\text{C}$ and $50\% \pm 5\%$
Medium environment	Dry condition

The binding energy of the contaminated carbon (C 1s: 284.8 eV) as the benchmark and the peak value of the target element was calibrated with a resolution of approximately ± 0.3 eV.

3 Results

3.1 Material characterization

The three-dimensional morphologies of the sample surfaces are shown in Fig. 4(a). Compared with the substrate, the LQ sample has an equivalent surface roughness. However, due to plasma sputtering, the PN-treated surface was clearly rougher ($R_a = 0.64 \mu\text{m}$) with a uniform texture. Similarly, the surface of

the MoS_2 coating had an irregular texture, and the maximum roughness reached $R_a = 1.05 \mu\text{m}$.

XRD phase analysis was performed on the surfaces of the treated specimens, and the results are shown in Fig. 4(b). Sharp and strong diffraction peaks can be observed at 2θ angles of 44.46° , 64.75° , and 82.11° on the surfaces of the substrate and LQ samples, corresponding to the α -Fe phase with high crystallinity. A compound layer of ϵ - $\text{Fe}_{2.3}\text{N}$ and γ '- Fe_4N was observed on the surface of the PN sample [32], but the peak reflection of α -Fe was not observed. For the MoS_2 coating, a remarkable peak (corresponding to the (002) lattice plane of MoS_2 phase) appeared around $2\theta = 13^\circ$ in the XRD pattern. In addition, other characteristic peaks match well with the JCPDS data (JCPDS 37-1492), confirming the polycrystalline and hexagonal structure of the MoS_2 particles [33].

As illustrated in Fig. 4(c), the surface and cross-sectional microhardness test results were obtained. The hardness values of the surface and cross-section of the substrate remained stable between 250–300 HV. In contrast, those of the LQ and PN samples were significantly enhanced, reaching 565 and 552 HV, respectively. The hardness gradient was apparent on the cross-section of the LQ sample, and the depth of

Fig. 4 (a) Three-dimensional morphologies of sample surfaces; (b) XRD patterns of samples; and (c) microhardness results of surface and profile of samples.

the hardening layer reached approximately 0.3 mm. However, the depth of diffusion layer for the PN sample was 0.42 mm, and the hardness of the cross-section linearly decreased with the depth. In addition, owing to the interlaminar slippage of the MoS₂ crystals, the surface layer (≤ 0.03 mm) hardness of the MoS₂ coating was only approximately 48 HV.

The cross-sections of the samples were prepared (polished and then corroded with 3% alcohol nitrate solution for 10 s) and SEM images were obtained. Compared with the uniform phase structure of the substrate (Fig. 5(a)), the martensite structure [34] is generated after LQ treatment (Fig. 5(b)), which is an important factor for enhancing the hardness of the material. In addition, Fig. 5(c) shows that the surface of the PN sample was uniformly covered by micron-sized spherical iron nitride, while a dense compound layer of approximately 8 μ m was formed on the surface and diffused to the interior in a reticular manner. As shown in Fig. 5(d), the thickness of the bonded MoS₂ coating was approximately 28 μ m, and the randomly stacked micron-sized particles resulted in a rough surface.

3.2 Dynamic response

3.2.1 F_t - D - N curves

The F_t - D - N curves with different tangential displacements are shown in Fig. 6, which reflect the

fretting regimes of the contact interface. At $D = 15$ μ m in all test cycles, the F_t - D curves of the MoS₂ coating are open, as parallelograms; specifically, fretting operated in the slip regime (SR). In contrast, the F_t - D curves of the substrate and the other surface-treated samples changed rapidly from parallelogram to linear with the increase in test cycles. This indicates that the contaminated film (oxide film, adsorption film) on the contact interface was removed in the initial test cycles, the imposed displacement was mainly absorbed by the elastic deformation of the contact material, and the fretting wear operated in the partial slip regime (PSR) [4]. For $D = 25$ and 50 μ m, the F_t - D curves of the substrate and the three treated samples were all quasi-rectangular. It is clear that the imposed displacement was mainly dominated by the plastic deformation of the contact surface; thus, fretting was operated in the SR.

3.2.2 Energy dissipation

Referring to Fig. 7(a), trapezoidal integration was performed on the single-cycle F_t - D curve (Eq. (1)), and the energy dissipation of a single cycle (E_s) was obtained. Furthermore, the accumulated energy dissipation (E_d) of the entire wear process was acquired via Eq. (2),

$$E_s = 2 \left(\int_{-\delta_0}^{\delta_m} f d\delta - \int_{\delta_0}^{\delta_m} f d\delta \right) \quad (1)$$

Fig. 5 Surface and cross-sectional SEM images of (a) substrate; (b) LQ; (c) PN; and (d) bonded MoS₂ coating.

Fig. 6 F_t - D - N curves of different samples under various fretting displacements.

Fig. 7 (a) Calculation method of energy dissipation and (b) cumulative and single-cycle energy dissipation of different samples.

$$E_d = \sum_{s=1}^{5,000} E_s \quad (2)$$

As shown in Fig. 7(b), E_d of the surface-treated samples increased significantly with an expansion of the displacement. Compared with the substrate, in different fretting regimes, E_d of the three surface-treated specimens decreased to different degrees. The self-lubricating properties of the MoS₂ coating resulted in the minimum E_d , especially at $D = 50 \mu\text{m}$, which was approximately 67% of that of the substrate. In

addition, E_s in the steady state was positively correlated with the increasing trend of fretting displacement, and the dissipated energy of the substrate and surface-treated samples followed the descending order of substrate, LQ, PN, and MoS₂ coating.

3.3 Coefficient of friction (COF)

The COF is the ratio between the friction force (F_t) and normal load (F_n), which plays an essential role in evaluating fretting wear behavior. Figures 8(a)–8(c) show the evolutions of the COFs for various treated

samples with different displacements. For $D = 15 \mu\text{m}$ (Fig. 8(a)), the COF varied with the test cycles and comprised three stages: initial; ascending; and steady. In the initial stage (before the 10th cycle), owing to the presence of the contaminated film, the fretting interface had a low COF (approximately 0.2). After different periods of the ascending stage, the COF of each sample reached a stable value, and followed the descending order of substrate, LQ, PN, and MoS_2 coating. The COF of the PN sample was stable after running-in for 1,600 cycles because of its texture and high hardness. In contrast, the COF of the MoS_2 coating increased slowly from 0.18 to approximately 0.38, showing an excellent fretting wear resistance.

As the fretting displacement increased to 25 and $50 \mu\text{m}$, a complex evolution of the COF appeared in the SR, as shown in Figs. 8(b) and 8(c). The initial stages of the substrate and LQ sample were significantly shortened, and the COF decreased and fluctuated after reaching the maximum value until the steady state was reached around 3,000 cycles. This occurred because the increased friction led to the aggravation of wear, and fretting debris was constantly produced

and gradually carried the normal load, which promoted the decrease in the COF. Moreover, when the formation and discharge of the third bodies reached a dynamic equilibrium, the COF was in a steady state. At the initial stage of $D = 50 \mu\text{m}$, the contaminated film (oxide film and adsorption film) that exist in the contact interface was rapidly destroyed due to the high-hardness LQ treated surface, the adhesion and plastic deformation of the mating materials occurred. Therefore, the COF of the LQ sample increased rapidly and is slightly higher than that of the substrate. The COF of the PN sample maintained a long period of the low-friction ascending stage (approximately 800–900 cycles) under the displacements of 25 and $50 \mu\text{m}$, which was inseparable from the rough surface of the nitriding layer. The roughness peak of the PN sample surface carried the normal load and had a higher Hertz contact stress than those of the other specimens. Then, the plastic deformation and adhesion of the friction pair induced the decrease of relative slip, and fretting wear tended to PSR; therefore, it had a smaller COF. In addition, compared with the other samples, the interlaminar slip characteristic of MoS_2

Fig. 8 Analysis of the COF for (a) $D = 15 \mu\text{m}$, (b) $D = 25 \mu\text{m}$, and (c) $D = 50 \mu\text{m}$; and (d) comparisons of the steady-state COF.

crystals contributed to the lower COF under different fretting displacements [35]. However, due to the accumulation of the work hardening and micro-defects upon fretting wear, the self-lubricating effect of the coating gradually weakened, consequently leading to the increase of COF with cycles.

In order to deeply analyze the mechanical response of various treated surfaces on fretting wear, the steady-state COFs determined by averaging COF from the steady stage of friction coefficient curves were obtained, as shown in Fig. 8(d). For the MoS₂ coating with excellent self-lubricating properties, the lowest COF was observed. In addition, although the surface hardness of the PN sample was slightly lower than that of the LQ sample, the lower steady-state COF demonstrated that the texture on the PN-treated surface played a pivotal role in reducing friction. Therefore, in terms of the antifriction effect, the treated surfaces were superior to the substrate, and the MoS₂ coating was the best, followed by the PN sample.

3.4 Wear analysis

3.4.1 Damage observations

The wear morphologies of the substrate and various surface-treated samples in the PSR ($D = 15 \mu\text{m}$) were

obtained, as shown in Fig. 9. As illustrated in Figs. 9(a) and 9(b), typical fretting rings appeared on the surfaces of the substrate and LQ sample. Compared with the substrate, the LQ sample had a smaller scar area, a narrower slip band (approximately $35 \mu\text{m}$), and no obvious heart damage due to its higher surface hardness. In Fig. 9(c), the rough texture of the PN sample resulted in an irregular fretting scar, uneven contact area, and slight furrows. In addition, as shown in Fig. 9(d), severe plastic deformation and layered stacking occurred at the contact interface of the MoS₂ coating, and flaky peeling appeared.

For the SR of fretting ($D = 50 \mu\text{m}$), the three-dimensional and microscopic wear morphologies of different samples are shown in Fig. 10. As shown in Fig. 10(a), the LQ and PN samples with enhanced surface hardness exhibited less abrasion than the substrate. In contrast, a deep pit formed on the MoS₂ coating owing to its severe plastic deformation. In Figs. 10(b) and 10(c), obvious furrows and delamination existed on the surfaces of the substrate and LQ sample, accompanied by a mass of wear debris. In contrast, the contact area of the PN sample was slightly damaged, with shallow furrows in the heart and delamination at the edge of the scar (Fig. 10(d)). However, owing to plastic flow and surface work hardening in the

Fig. 9 Wear morphologies of different treated surfaces in fretting PSR.

Fig. 10 Wear morphologies of different treated surfaces in fretting SR. (a) Three-dimensional morphologies; (b) substrate; (c) LQ; (d) PN; and (e) bonded MoS₂ coatings.

contact area of the MoS₂ coating, cracks and flaking occurred, and the internal MoS₂ particles were exposed (Fig. 10(e)).

3.4.2 Abrasion loss

The variation in the wear scar depth of different samples is shown in Fig. 11(a). With an increase in fretting displacement, the depth and width of the fretting scar gradually expanded. The wear depths of the LQ and PN samples were the lowest in various

fretting regimes, followed by the substrate. Furthermore, the wear depth positively correlates with the surface hardness of the samples.

As shown in Fig. 11(b), the wear volumes and areas of the four samples under different fretting displacements were comprehensively compared. In different fretting regimes, the PN sample had the lowest wear volume, followed by the LQ sample, and then the substrate. In particular, at the displacement of 50 μm, the volume loss of the PN sample was

Fig. 11 (a) Cross-sectional profiles of fretting wear scars and (b) comparison of wear volumes and wear areas.

reduced by 41% compared with that of the substrate. Similarly, the comparison results of the scar area are consistent with the results of wear volume. However, it is noteworthy that the wear area of the PN sample was slightly larger than that of the LQ sample, due to the surface texture. For the MoS₂ coating, owing to its low surface hardness and inter-crystal slip characteristics, the wear volume and area were the largest under the same experimental conditions.

3.5 Tribochemical analysis

In the SR ($D = 50\ \mu\text{m}$), the tribochemical states on worn scars of different treated surfaces were characterized. The binding energies of all elements obtained by XPS were calibrated, and the convolution was performed using a Shirley background.

Figures 12(a)–12(c) show the Fe 2p high-resolution spectral analyses of the substrate, LQ, and PN samples, respectively. Considering the substrate (Fig. 12(a)),

there were two Fe 2p_{3/2} binding energy peaks at 710.62 and 712.3 eV, and two Fe 2p_{1/2} binding energy peaks at 723.7 and 725.9 eV, respectively [36]. Among them, the two peaks at 710.62 and 723.7 eV may be attributed to the binding energy of Fe–O in FeO (Fe²⁺), while those at 712.3 and 725.9 eV are assigned to Fe³⁺ in Fe₂O₃. In addition, two peaks at 706.75 and 719.9 eV were found in the Fe 2p high-resolution spectrum of the LQ sample (Fig. 12(b)), indicating the presence of Fe⁰ after abrasion.

Integral statistics were performed for the percentage of peak area in the Fe 2p high-resolution spectrum, and the results are shown in Table 3. Previous studies have indicated that Fe atoms on the surface of the sample form an oxidation film (mainly FeO) in the air, and the oxidation reaction is activated by friction to promote the transformation of FeO into Fe₂O₃ [36, 37]. Therefore, the three samples were oxidized to varying degrees in fretting wear, of which the content of Fe³⁺

Fig. 12 High-resolution Fe 2p spectra for the scar of (a) substrate, (b) LQ sample, and (c) PN sample; and (d) high-resolution Mo 3d spectra for the scar of bonded MoS₂ coating.

Table 3 Peak positions of Fe 2p XPS results and corresponding area ratios of different fretting interfaces.

Sample	Peak position (± 0.3 eV)						Area ratio (%)		
	Fe metal (Fe^0)		FeO (Fe^{2+})		Fe_2O_3 (Fe^{3+})		Fe (Fe^0)	FeO (Fe^{2+})	Fe_2O_3 (Fe^{3+})
	Fe^0 2p _{3/2}	Fe^0 2p _{1/2}	Fe^{2+} 2p _{3/2}	Fe^{2+} 2p _{1/2}	Fe^{3+} 2p _{3/2}	Fe^{3+} 2p _{1/2}			
Substrate	—	—	710.62	723.72	712.3	725.9	0	24.06	75.94
LQ	706.75	719.9	709.02	722.62	712.18	725.78	13.24	31.14	55.62
PN	—	—	710.54	723.14	712.13	725.73	0	30.77	69.23

in the scar of the substrate was the highest, reaching 75.94%. In contrast, the Fe^{3+} contents in the wear products of the LQ and PN samples were reduced to 55.62% and 69.23%, respectively. The fact that Fe metal was present in the worn scar of the LQ sample may be owing to the possibility that fresh matrix exposed to the scar surface due to material spalling has not been completely oxidized.

Figure 12(d) illustrates the Mo 3d high-resolution spectra of the scar on the MoS_2 coating. Evidently, the binding energy peaks of MoS_2 appeared at 229.4 and 232.5 eV, while those of MoO_3 were located at 233.0 and 235.8 eV [38, 39]. The formation of MoO_3 was caused by the oxidation reaction of Mo^{4+} to Mo^{6+} during the wear process [40].

4 Discussion

Based on the structural features and service conditions of the locking pin, combined with the dynamic response on the fretting wear, it is determined that the fretting damage of the mechanical component is mainly located in the SR. Therefore, the damage mechanism and anti-wear properties of various treated surfaces for the fretting displacement of 50 μm are discussed comprehensively herein, and an optimal surface treatment method suitable for this component is recommended.

4.1 Wear mechanism of various treated surfaces

Combined with the damage morphologies (Fig. 10) and tribochemical states (Fig. 12), the wear mechanism of different samples is summarized. Various degrees of furrows and delamination, as well as oxidation products, were detected in the abrasion scars of the substrate and the two wear-resistant surfaces (LQ- and PN-treated surfaces). This indicates that their wear

mechanism was mainly due to the combination of abrasive wear, delamination, and oxidation wear. In addition, work hardening and flaky peeling were apparent in the scar of the MoS_2 coating, and its wear mechanism was plastic deformation and peeling, accompanied by oxidative wear.

4.2 Evaluation of anti-fretting damage performance

Based on the functional relationship between dissipated energy (Fig. 7(b)) and wear volume (Fig. 11(b)), as shown in Fig. 13(a), the energy-based wear rate was obtained [41]. Previous studies have confirmed that wear amount is inversely proportional to the hardness of a material [42]. In the case of a small fretting displacement, severe strain hardening occurred on the contact surface; therefore, a lower wear rate was observed in the PSR ($D = 15 \mu\text{m}$). Compared with the substrate and antifriction coating, the two wear-resistant surfaces with high hardness contributed to their lower wear rates. Moreover, the PN sample exhibited the best wear resistance owing to the wear reduction effect of the surface texture.

From Fig. 13(b), the PN-treated surface displayed the best comprehensive performance in the SR of fretting, followed by the LQ-treated surface. In contrast, the martensite structure with high hardness for LQ-treated surface lower the plastic flow capability of the material, which will make cracks being prone to be initiated and propagated [43]. Although the self-lubricating properties of the MoS_2 coating contributed to the lowest COF and dissipated energy, the low stiffness will inevitably expand the fit clearance of the locking mechanism, which may aggravate the impact and vibration of the components in actual working conditions. In summary, PN is the preferred surface treatment method for locking pin materials.

4.3 Alleviating fretting damage mechanism of the PN-treated surface

To demonstrate the reason why the PN-treated surface alleviated the fretting damage, an idealized model was established, as shown in Fig. 14. As mentioned above (Section 3.1), the PN-treated surface is covered with high-hardness spherical ferro-nitride ($\text{Fe}_{2.3}\text{N}$ and Fe_4N), which increases the roughness while strengthening the surface. The surface texture results in higher contact stress, and the material adhesion at the contact interface increased. Furthermore, part of the imposed displacement was coordinated by the elastic deformation of the material, which reduced the damage. Meanwhile, as the third bodies (Fig. 14(a)),

the ferro-nitride particles exfoliated quickly and were collected by the texture, which plays the role of bearing and lubrication. Referring to the wear morphologies in Fig. 14(c), ring-shaped lamellae were formed on the edge of the scar by stacking and rolling (Fig. 14(b)), which retarded the discharge of debris from the contact area, thereby consolidating the wear-reducing effect of the third bodies (Fig. 14(d)).

In addition, the coarse porous compound layer provided a favorable environment for the storage and release of lubrication medium. The PN-treated surface may exhibit better anti-fretting effects under lubrication conditions, which needs to be further explored in the follow-up studies.

Fig. 14 Mechanism of reducing fretting wear on PN-treated surface.

5 Conclusions

Three surface treatments were applied to the locking pin material (AISI 4135 steel), the fretting wear experiments and various characterization analyses were performed. The wear behavior and damage mechanisms of the various treated surfaces were investigated. Several conclusions can be drawn from the results and discussion.

1) The surface hardness of laser quenching (LQ) and plasma nitriding (PN) samples was enhanced, and a lower fretting wear rate was obtained. The main wear mechanisms of the two strengthened surfaces were abrasive wear, delamination, and oxidation wear, while the damage of PN surface is slight.

2) The MoS₂ coating had the lowest coefficient of friction (COF) and dissipation energy in different fretting regimes, but the highest wear rate. The main types of failure were plastic deformation, delamination, and oxidative wear.

3) PN is preferred as a surface-strengthening measure for locking pin materials. The rough compound layer causes fretting to tend to the partial slip state, forming and maintaining a third-body bed that mitigates wear.

4) This research provides a reference for the surface protection of related mechanical structures in the engineering field to alleviate fretting damage.

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