

Full length article

Exceptional tribocorrosion resistance achieved by architecting multi-interfaces in soft/hard carbon-based coatings

Longwen Shang^{a,b}, Silong Zhang^a, Hao Li^a, Yanjin Li^{a,b}, Xin Zhang^a, Rende Chen^a, Kazuhito Nishimura^a, Peng Guo^{a,*}, Aiyang Wang^{a,b,**}

^a State Key Laboratory of Advanced Marine Materials, Zhejiang Key Laboratory of Extreme-environmental Material Surfaces and Interfaces, Ningbo Institute of Materials Technology and Engineering, Chinese Academy of Sciences, Ningbo 315201, China

^b Center of Materials Science and Optoelectronics Engineering, University of Chinese Academy of Sciences, Beijing 100049, China

ARTICLE INFO

Keywords:

Amorphous carbon coating
Alternating multilayer
Tribocorrosion
Damage mechanism

ABSTRACT

Amorphous carbon (a-C) coatings are ideal surface protection materials for critical moving components in marine engineering. However, conventional metal doping and multilayer strategy cannot address the trade-off between high-load capacity and anti-tribocorrosion performance of a-C. Based on various a-C layers with similar compositions and high corrosion resistance, a soft/hard alternating diamond-like carbon (DLC)/graphite-like carbon (GLC) multilayer structure, namely (DG)₁₀ coating, was proposed. Through a detailed investigation into crack propagation during the tribocorrosion process, results showed that, under both high (20 N) and low (10 N) loads, the (DG)₁₀ coating demonstrated the lowest coefficient of friction (~ 0.06) and wear rate on the order of 10^{-8} mm³/N·m. Under low load, all three coatings could maintain their structural integrity, and the topmost carbon layer played a dominant role. While, under high load, the DLC coating showed brittle spallation due to its poor plastic deformation capability, and the GLC coating exhibited layer-by-layer shear removal and rapid consumption. For the (DG)₁₀ coating, the multilayer structure reduced the crack propagation rate and retarded the formation of through-thickness cracks, so its damage mechanism was transformed from abrupt failure to progressive damage, and the coating only exhibited slight localized spallation within prolonged sliding time (6 h).

1. Introduction

Key moving components used in marine engineering, such as cylinders, piston pumps, sealing components and bearings, typically face the combined challenges of chemical corrosion and mechanical wear [1–3]. Meanwhile, their premature degradation and serious damage pose a significant threat to the operational stability and service life of marine engineering equipment [4,5]. To ensure the long-term reliability of critical components, high-performance coatings with low friction, high wear and corrosion resistance are essential [6]. Amorphous carbon (a-C) coatings composed of both diamond-structured sp³-C atoms and graphite-structured sp²-C atoms meet these demands and are thus regarded as ideal surface protection materials [7,8]. For example, the previous study [9] has reported that a-C coatings had both low coefficient of friction (COF) of 0.09, and wear rate of 1.5×10^{-8} mm³/N·m in 3.5 wt% NaCl solution, and that these coatings could effectively protect

metal substrates in northwestern Pacific Ocean with a depth of 5942 m for 342 days.

According to sp²/sp³ ratio, a-C coatings can be broadly categorized into diamond-like carbon (DLC) coatings with main content of sp³ bonds, which have high hardness and excellent wear-resistant performance, and graphite-like carbon (GLC) coatings with main content of sp² bonds, which are characterized by self-lubrication performance [10, 11]. Both coatings have been attempted for corrosion and wear protection, but showed various tribological performance. For example, Zhang et al. [12] reported that DLC exhibited excellent tribocorrosion resistance under a lower load (5 N, ~ 1.36 GPa) but failed rapidly due to brittle spalling under a higher load (10 N, ~ 1.72 GPa). In contrast, the highly plastic GLC could provide good protection under the higher load, but its low hardness and graphite-like characteristics resulted in poor wear resistance under a lower load.

In practice, various multilayer or composite structures are designed

* Corresponding author.

** Corresponding author at: State Key Laboratory of Advanced Marine Materials, Zhejiang Key Laboratory of Extreme-environmental Material Surfaces and Interfaces, Ningbo Institute of Materials Technology and Engineering, Chinese Academy of Sciences, Ningbo 315201, China.

E-mail addresses: guopeng@nimte.ac.cn (P. Guo), aywang@nimte.ac.cn (A. Wang).

<https://doi.org/10.1016/j.triboint.2026.112052>

Received 6 March 2026; Received in revised form 2 April 2026; Accepted 13 April 2026

Available online 15 April 2026

0301-679X/© 2026 Elsevier Ltd. All rights are reserved, including those for text and data mining, AI training, and similar technologies.

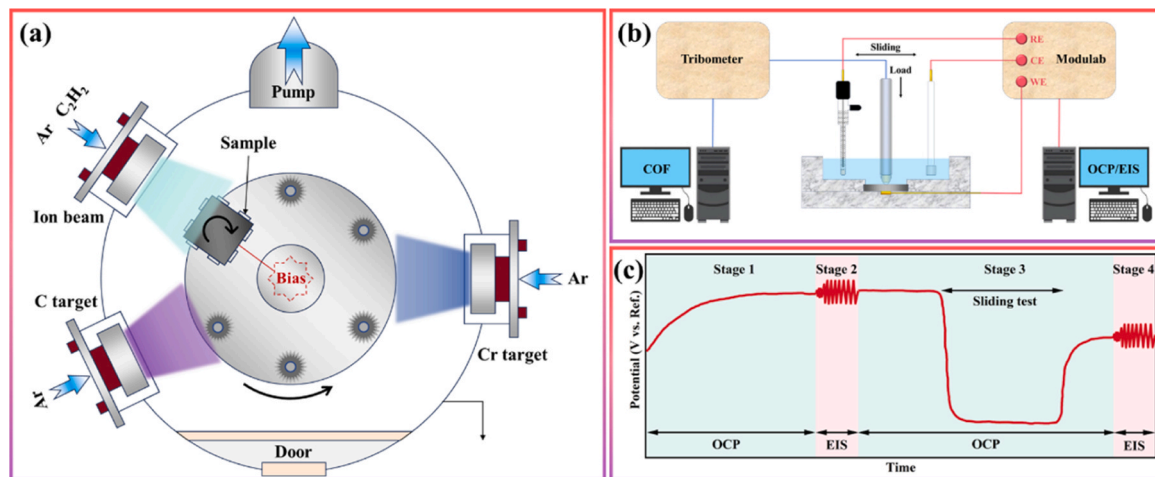


Fig. 1. Schematic diagrams of (a) ion beam composite multi-target magnetron sputtering system, (b) three-electrode system and (c) program settings during tribocorrosion test.

for DLC and GLC to achieve a high load-bearing capacity, along with friction reduction and wear resistance. Zhao et al. [13] demonstrated that an alternating DLC/CrN multilayer design markedly improved wear resistance, exhibiting a reduced COF from 0.1 to 0.055 and a low wear rate of $2 \times 10^{-7} \text{ mm}^3/\text{N}\cdot\text{m}$ in an atmospheric environment. Zhang et al. [14] employed a WC/W-DLC/DLC gradient composite strategy and achieved long-term protection with a sliding distance exceeding 2100 m under a 5 N load in 3.5 wt% NaCl solution. While, for metal/a-C composite coatings, the susceptibility of interlayer metal components to corrosion can break structural integrity of a-C coatings over long-term service and may result in aggravated tribocorrosion failure [15].

Inspired by the aforementioned research, the alternating soft carbon/hard carbon (C/C) multilayer coating may address related issues, from the perspective of inhibiting corrosion and enhancing strength and toughness. Primarily, this multilayer structure can hinder the diffusion of corrosive media [16,17]. For example, Jiang et al. [18] pointed out that, compared with the pristine GLC, the alternating coating decreased the corrosion current density from $2.43 \times 10^{-8} \text{ A}\cdot\text{cm}^{-2}$ to $7.8 \times 10^{-9} \text{ A}\cdot\text{cm}^{-2}$. Additionally, the multilayer structure can reduce compressive stress and inhibit crack propagation [19–21]. For instance, Usman et al. [22] confirmed that the optimized multilayer coatings exhibited a high load-bearing capacity (80 N, 3.4 GPa) and a low wear rate of $7.9 \times 10^{-8} \text{ mm}^3/\text{N}\cdot\text{m}$. Similarly, Guo et al. [23] prepared multilayer hydrogenated a-C coatings and demonstrated their exceptional durability over 100,000 cycles with low COF of 0.005 under 3.19 GPa. While, under tribocorrosion conditions, the existing corrosion-wear interaction for a-C coatings with metal components is no longer prominent, the failure of C/C and coating-substrate interfaces may determine the service performance. However, limited by insufficient in-situ and micro-scale analysis, microstructure evolution of the two interfaces remains unclear. It is one open challenge to establish design principles for tribocorrosion resistance multilayer a-C and clarify the underlying damage mechanisms.

In this work, we prepared DLC/GLC multilayer coatings via the direct current magnetron sputtering (DCMS) and linear ion source (LIS) alternating deposition process, with single DLC and GLC coatings serving as control groups. The microstructure and tribocorrosion behavior of the three coatings were compared. Furthermore, based on in-situ electrochemical and wear track investigation from different tribocorrosion stages, the role of interfacial characteristics on coating spalling modes and damage mechanism was discussed.

2. Experimental details

2.1. Coating deposition

As shown in Fig. 1 (a), C/C multilayer coatings were prepared using the self-developed deposition system. Before the experiment, 17–4 PH stainless steel pieces ($\Phi 17 \times 3 \text{ mm}$) were ultrasonically cleaned in ethanol and acetone for 10 min respectively, followed by drying with a high-pressure air gun. Then the substrates were placed in the vacuum chamber. When the vacuum was below $4 \times 10^{-5} \text{ Torr}$, LIS was used to etch the substrate surface with highly ionized Ar⁺. Subsequently, the Cr target (size of $\Phi 100 \times 5 \text{ mm}$ and purity of 99.95%) was used to deposit a Cr layer as the transition layer via DCMS, with the Cr target current of 1.2 A, the substrate bias voltage of -300 V , Ar flow rate of 60 sccm, and the deposition time of 15 min. Next, the DCMS-LIS alternating sputtering program was initiated to prepare the multilayer coatings. The structure was configured with the soft GLC layer deposited first, followed by the hard DLC layer, cycling for 10 periods and totaling 20 layers. In particular, these multilayer coatings were designated as (DG)₁₀ coatings for short. The single GLC layers were deposited by DCMS with the graphite target (size of $400 \times 100 \times 8 \text{ mm}$ and purity of 99.99%), with the following parameters as the target current of 3 A, the substrate bias voltage of -200 V , the Ar flow rate of 60 sccm, and the deposition time of 15 min. As for the single DLC layers, LIS was operated at 0.2 A with the C₂H₂ flow rate of 34 sccm. The substrate bias voltage and the deposition time were maintained at -100 V and 6 min, respectively. Additionally, the deposition times for the single DLC and GLC coatings were 120 min and 300 min, respectively, with all other parameters identical to those described above. More details can be found in our previous work [12].

2.2. Coating characterization

The surface morphology of the as-deposited coatings and wear tracks after tribocorrosion test were observed using a fluorescence metallographic microscope (NMM-800RF, China). The microstructure of the coating was observed using a thermal field scanning electron microscope (SEM, Zeiss Sigma 300, Germany). After tribocorrosion test, the wear track was subjected to ion beam cutting by a focused ion beam scanning electron microscope (FIB-SEM, Auriga, Germany), and the fine structure and composition of the cross-sectional sample were analyzed using a transmission electron microscope (TEM, Talos F200x, USA). The carbon bond structure of the coatings was characterized by a confocal micro-Raman spectrometer (Renishaw inVia Reflex, UK), with the excitation source selected as a visible light source at a wavelength of

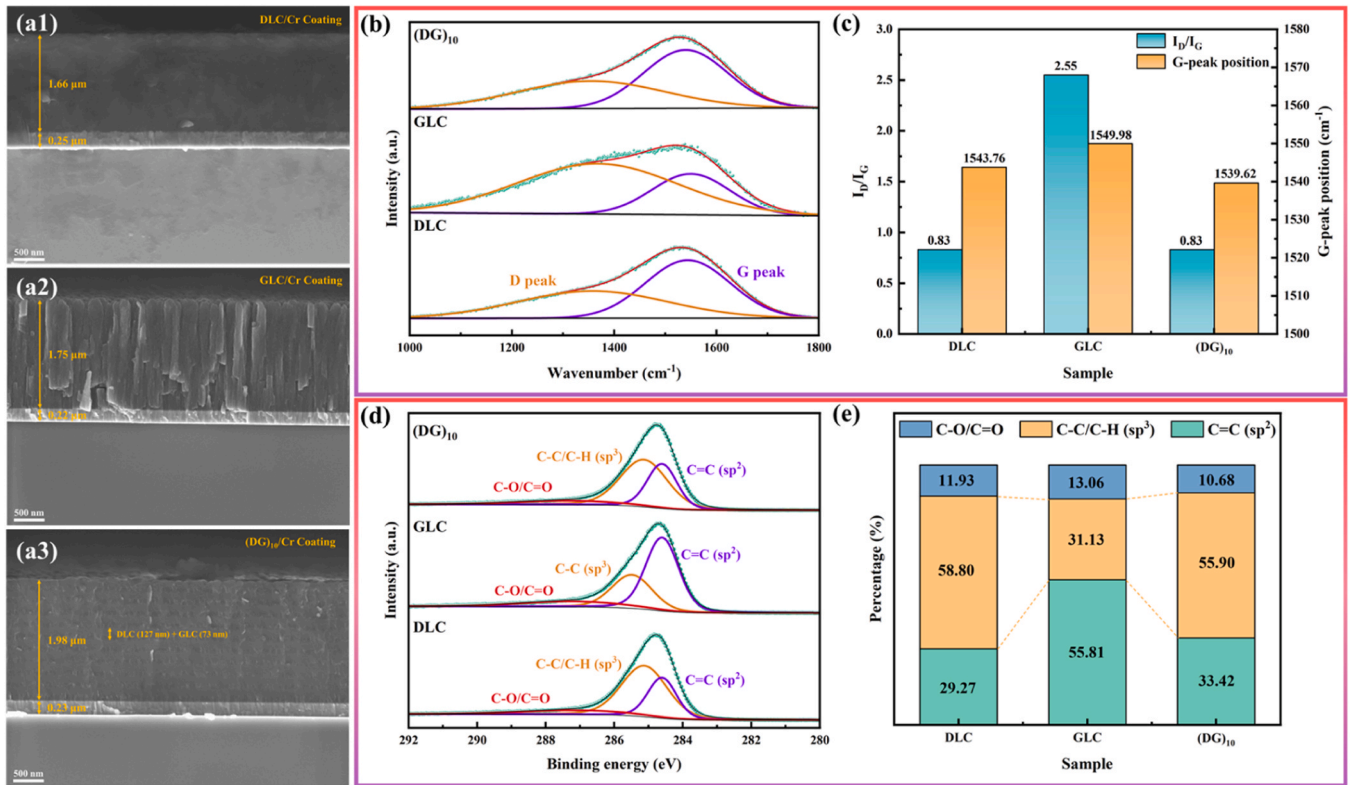


Fig. 2. (a1-a3) Cross-sectional SEM images, (b, c) Raman spectra, and (d, e) XPS spectra of the DLC, GLC, and (DG)₁₀ coatings.

532 nm. The elemental composition and bonding structure were characterized using an X-ray photoelectron spectrometer (AXIS SUPRA, UK). The hardness and modulus of the coating were evaluated by a nanoindenter (G200, USA) in a continuous stiffness measurement mode, with the indentation depth set to 1/10 of the coating thickness. Each coating was tested at 8 points, and the average value from the stable region was selected as the test result to reduce errors. Meanwhile, the brittleness of the coating was qualitatively evaluated using a Cube indenter. The curvature radii of the substrate before and after coating deposition were measured using a film stress tester (FST5000, China), and the residual stress was calculated by the Stoney equation [24]:

$$\sigma = \frac{E_s t_s^2}{6(1 - \nu_s)} \frac{1}{t_f} \left(\frac{1}{R_f} - \frac{1}{R_s} \right) \quad (1)$$

where E_s and ν_s represent the Young's modulus and Poisson's ratio of the substrate, t_s and t_f are the thicknesses of the substrate and coating, R_s and R_f correspond the curvature radii of the substrate before and after coating deposition.

2.3. Electrochemical corrosion and tribocorrosion tests

Both corrosion and tribocorrosion tests were conducted in 3.5 wt% NaCl solution. All electrochemical measurements employed a classic three-electrode system, with the coated samples as the working electrode, a platinum sheet as the counter electrode, and an Ag/AgCl electrode as the reference electrode, as shown in Fig. 1 (b).

For electrochemical corrosion test, the potentiodynamic polarization (PDP) and potentiostatic polarization (PSP) test data were collected using a Gamry (Reference 620, USA) electrochemical workstation. Prior to testing, the coatings were immersed in 3.5 wt% NaCl solution for 2 h to achieve a stable open-circuit potential (OCP), ensuring their steady electrochemical state. In the PDP test, a continuously increasing potential was applied to obtain the current density-potential curve. Subsequently, the corrosion potential and current density of the coating

were determined through Tafel fitting. The scanning range for PDP test was -0.4 V to 1.0 V (vs. Ref.), with a scanning rate of 1.0 mV/s. In the PSP test, a constant potential was applied to obtain the current density-time curve. The constant potential for PSP test was 0.7 V (vs. Ref.) with a testing duration of 150 min.

For the tribocorrosion test, an electrochemical workstation (ModuLab XM ECS, China) and a reciprocating tribometer (MFT5000, Rtec, USA) were used, as shown in Fig. 1 (b). Here, an Al_2O_3 ball with a diameter of 6 mm was selected as the counterpart to ensure chemical inertness and avoid elemental repetition at the friction contact interface [25–27]. The sliding distance and speed were set to 4 mm and 20 mm/s, respectively, under various normal loads and duration conditions, namely 10 N-1 h, 20 N-1 h, and 20 N-6 h. In particular, estimated contact pressures are summarized in Table S1.

The entire tribocorrosion process could be divided into four stages (Fig. 1 (c)). At stage 1, the OCP value was monitored until it stabilized, with the potential fluctuation on the coating surface remaining within ± 20 mV. At stage 2, electrochemical impedance spectroscopy (EIS) test was performed to evaluate the corrosion resistance of the coatings prior to tribocorrosion. The frequency range for EIS test was 10^5 – 10^{-2} Hz, with a sinusoidal AC potential perturbation of ± 10 mV applied. During stage 3, the sliding friction test was initiated once the OCP value stabilized again, after that, the OCP value was continuously monitored until it recovered and stabilized. In the final stage 4, EIS test was performed again to evaluate the corrosion resistance.

The 3D profile of the wear track on the coating surface was measured by a 3D optical profiler (UP-Lambda, USA), and the wear rate was calculated as follows:

$$W = \frac{S \times l}{N \times L} \quad (2)$$

where S is the wear cross-sectional area, l is the wear track length, N is the applied normal load, and L is the total sliding distance.

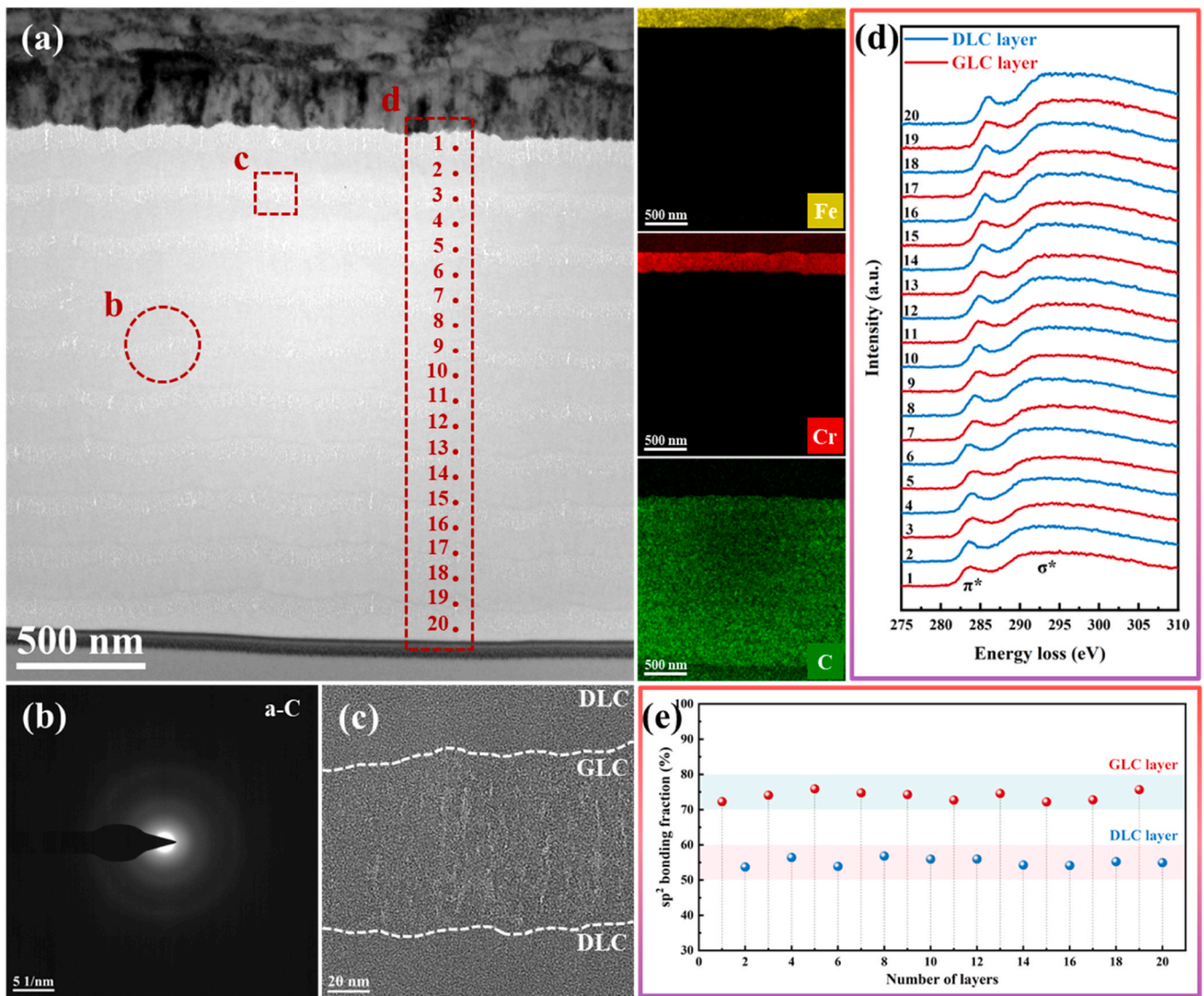


Fig. 3. (a) Cross-sectional TEM image and EDS mapping, (b) SAED, (c) HRTEM image, and (d, e) EELS spectra of single layers and sp^2 content distribution of the (DG)₁₀ coating.

3. Results and discussions

3.1. Micromorphology and structure

Fig. 2 shows the cross-sectional morphology, Raman spectra, and XPS spectra of the DLC, GLC and (DG)₁₀ coatings. In Fig. 2 (a1-a3), the total thicknesses of the three coatings were 1.91 μm , 1.97 μm , and 2.21 μm , respectively. The DLC coating exhibited a dense and smooth structure, while the GLC coating showed distinct columnar characteristic. In the (DG)₁₀ coating, the thicknesses of the single DLC and GLC layers were approximately 127 nm and 73 nm, respectively, the introduction of DLC layer effectively inhibited the columnar grain growth of GLC layer, showing a multilayered dense morphology. In particular, the surface morphologies of the three coatings are shown in Fig. S1.

From Raman spectra and their fitting results in Fig. 2 (b, c), three coatings had typical a-C characteristic curves with two broadened peaks, namely D and G peaks, within the range of 1000–1800 cm^{-1} [28]. For the DLC coating, its I_D/I_G ratio was 0.83 and G peak located at 1543.76 cm^{-1} , which were lower than those of the GLC coating, indicating a lower sp^2 content of the DLC coating. In Fig. 2 (d), the XPS C 1 s spectra revealed carbon hybridization peaks at 284.6 eV, 285.2 eV, and 287.0 eV, corresponding to C=C (sp^2) bonds, C-C/C-H (sp^3) bonds, and

C-O/C=O bonds, respectively. The presence of oxygen atoms was primarily associated with residual oxygen in the vacuum chamber during coating deposition or exposure of the samples to air [29]. From Fig. 2 (e), the sp^3 bond content of the DLC coating (58.80%) was higher than that of the GLC coating (31.13%), suggesting the lower graphitization of the DLC coating. As the top layer of the (DG)₁₀ coating was DLC, the (DG)₁₀ coating had similar Raman and XPS results.

To investigate microstructure of DLC/GLC interface, the cross-sectional area was further characterized by TEM-EELS. In Fig. 3 (a), the (DG)₁₀ coating exhibited a periodic structure with alternating GLC and DLC layers and they were tightly bonded at the interfaces. The corresponding EDS mapping and SAED pattern (Fig. 3 (b)) confirmed the stable and uniform composition and structure along the thickness direction. While, from HRTEM image (Fig. 3 (c)), some elongated structural regions were present within the GLC layer, in contrast to the structure of the DLC layer. Fig. 3 (d) displays the C-K edge EELS spectra of each single layer, including a π^* peak around 285 eV and a broad σ^* state absorption edge in the range of 290–305 eV [30]. The π^* and σ^* features, originating from sp^2 -bonded carbon and sp^3 -hybridized carbon respectively, could be used to compare the bonding characteristics in a-C coatings. And the sp^2 -C bond content could be estimated using the following Eq. 3 [31–33]:

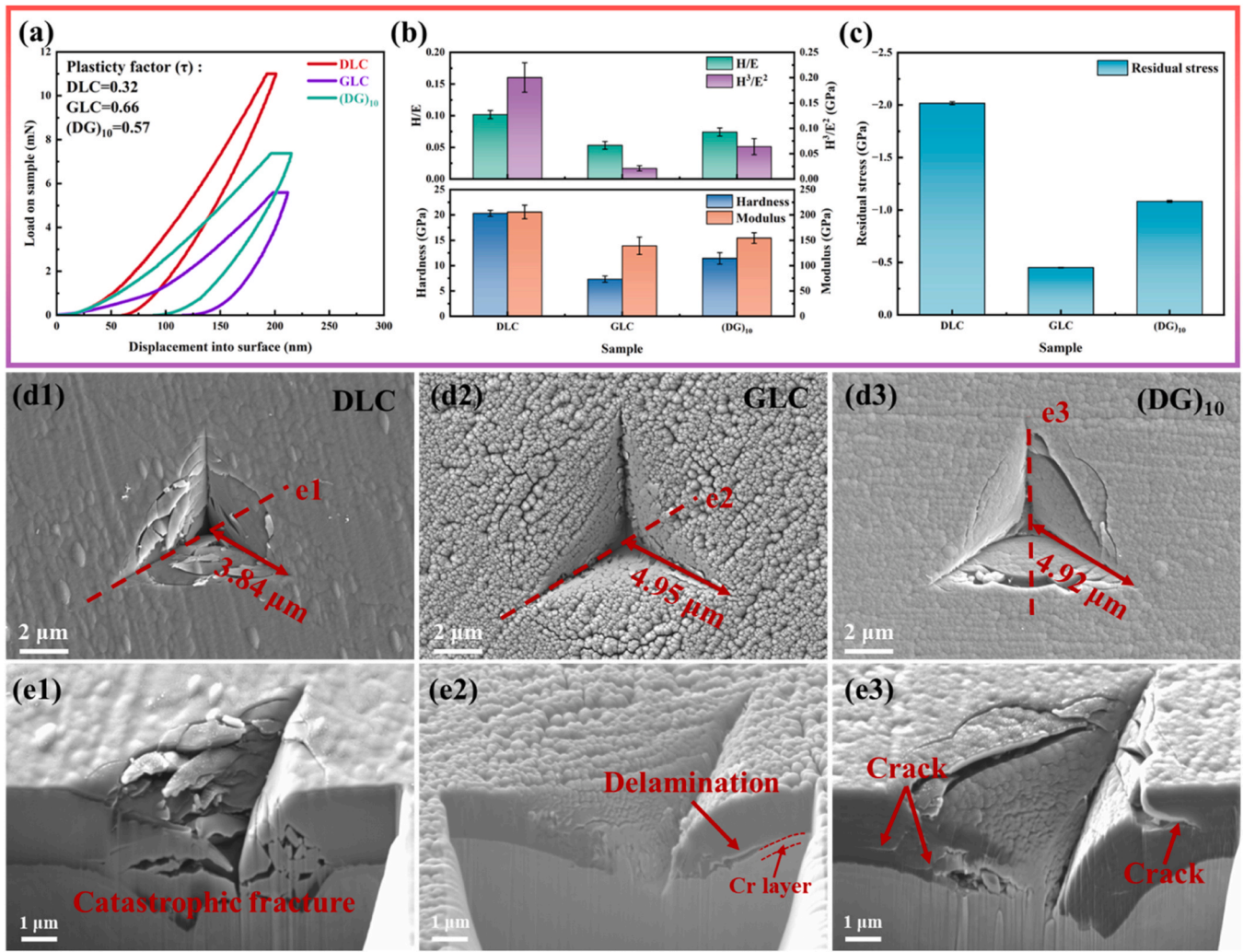


Fig. 4. (a) Nanoindentation load-displacement curves, (b) hardness and modulus values, (c) residual stress, (d1-d3) surface and (e1-e3) cross-sectional morphologies of Cube indentations under a 200 mN load of the DLC, GLC, and (DG)₁₀ coatings.

$$\frac{(\pi^*/\sigma^*)_{a-c}}{(\pi^*/\sigma^*)_{std}} = \frac{3x}{4-x} \quad (3)$$

where the π^* intensity corresponds to the Gaussian integral area at 285 eV, and the σ^* intensity represents the integral area within the energy window of 290–305 eV. Highly oriented pyrolytic graphite (HOPG, 100% sp² C bonds) is selected as the standard sample. In Fig. 3 (e), each DLC and GLC single layer had a stable sp² carbon bond content, approximately 55.12% and 73.91%, respectively.

3.2. Mechanical properties

As shown in Fig. 4, nanoindentation and Cube indentation tests were performed on the DLC, GLC, and (DG)₁₀ coatings. From load-displacement curves in Fig. 4 (a), the plasticity factor (τ) can be calculated [34]:

$$\tau = \frac{W_p}{W_t} \quad (4)$$

where W_p denotes the plastic deformation work, calculated from the integral area enclosed by the loading-unloading curve. W_t represents the total deformation work, characterized by the integral area up to the maximum displacement. The τ value ranges from 0 (purely elastic) to 1 (purely plastic). The τ value of the DLC coating (0.32) was much lower

than that of the GLC coating (0.66), and the (DG)₁₀ coating had a moderate level, since its τ value could be affected by both DLC and GLC layers under large indentation depth of 200 nm.

In Fig. 4 (b), the DLC coating had high hardness and modulus, compared with the GLC coating. With the highest H/E and H³/E², the DLC coating had a low probability of plastic deformation under impact loading [35]. Besides, compared to the DLC coating, the residual stress of the (DG)₁₀ coating was effectively relieved with a reduction of approximately 50% due to the introduction of the GLC layers and the alternating composite structure, as shown in Fig. 4 (c).

The above-mentioned properties could greatly affect the adhesion strength and interfacial failure under high load. From damage area along the edges of the Cube indentations in Fig. 4 (d1-d3), distinct nest-like cracks and coating spallation were observed for the DLC coating, indicating its severe brittle fracture. By comparison, the GLC coating showed intact indentation edge and granular morphology. For the (DG)₁₀ coating, it had similar characteristic as the DLC coating, while the nest-like cracks and spallation were significantly alleviated.

From cross-sectional morphologies of Cube indentations (Fig. 4 (e1-e3)), the DLC coating exhibited catastrophic fracture, accompanied by severe deformation at the coating-substrate interface. In contrast, the GLC coating remained internally intact but suffered from coating-substrate interfacial delamination. Notably, for the (DG)₁₀ coating, the propagating cracks underwent significant lateral deflection and ultimately terminated.

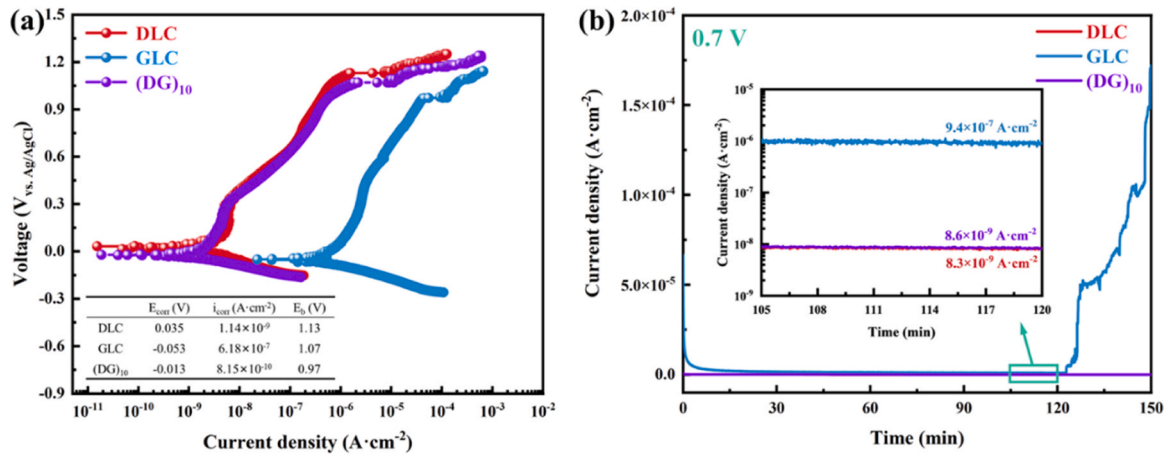


Fig. 5. (a) PDP curves and (b) PSP curves of the DLC, GLC, and (DG)₁₀ coatings.

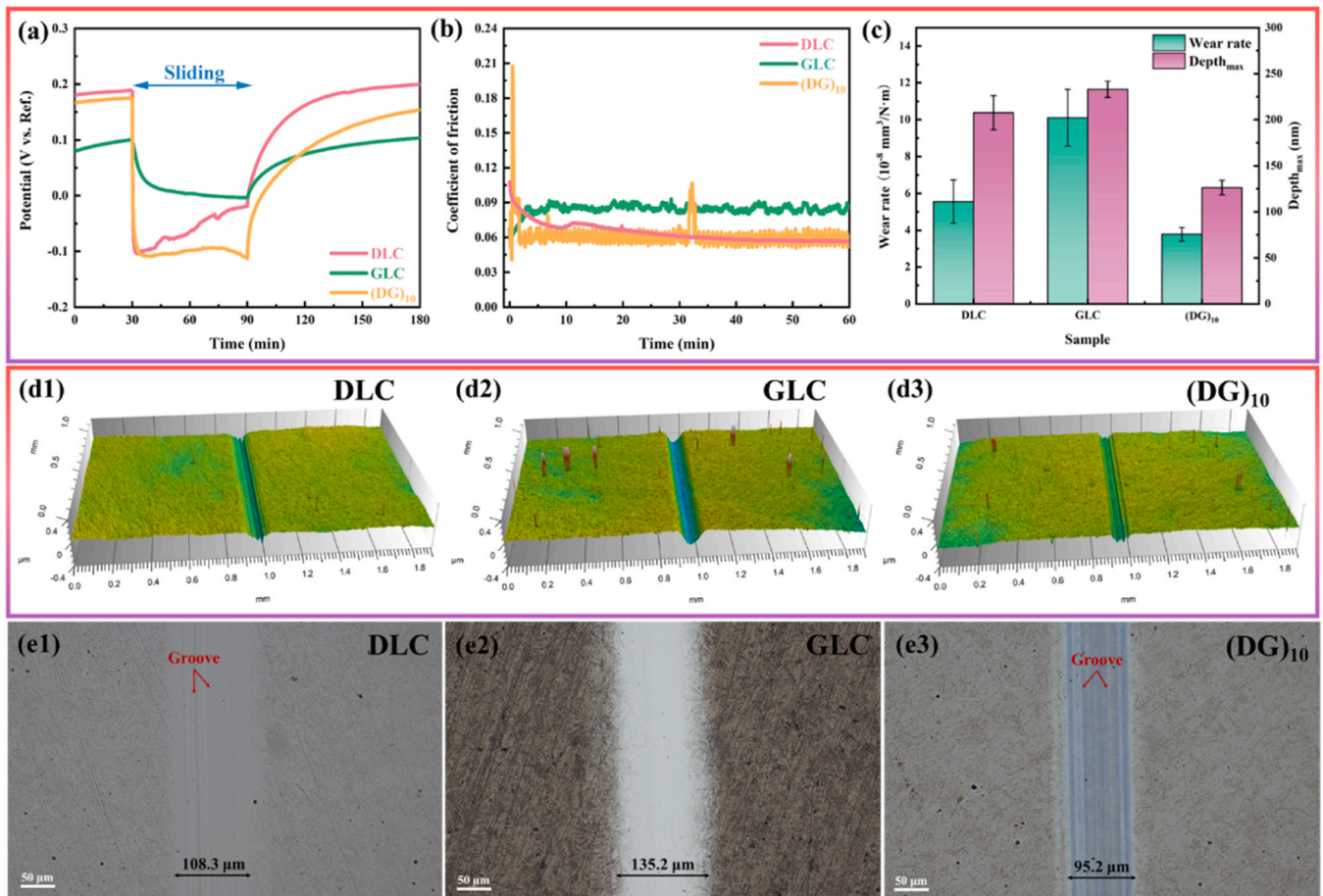


Fig. 6. (a) OCP, (b) COF, (c) wear rate and maximum wear depth, (d1-d3) 3D wear track profiles, and (e1-e3) wear track optical morphologies of the DLC, GLC, and (DG)₁₀ coatings under 10 N-1 h test condition.

3.3. Electrochemical properties

In general, three parameters are commonly used to evaluate the corrosion protection performance of coatings, namely corrosion potential (E_{corr}), corrosion current density (i_{corr}), and breakdown potential (E_b). Specifically, E_{corr} indicates the tendency for corrosion to occur, i_{corr} represents the corrosion rate, and E_b signifies the resistance to pitting corrosion [36,37].

From the PDP curves in Fig. 5 (a), similar slopes were observed for

the three coatings in the cathodic region, indicating the identical cathodic reaction process. In the anodic region, the passivation region was associated with the anodic passivation of the Cr interlayer and substrate exposed due to coating defects [38]. Compared with the GLC coating, the DLC coating exhibited higher E_{corr} (0.035 V) and E_b (1.13 V), as well as lower i_{corr} (1.14×10^{-9} A·cm⁻²), demonstrating superior corrosion protection performance. The (DG)₁₀ coating had moderate E_{corr} (-0.013 V) and E_b (1.07 V). Meanwhile, its i_{corr} (8.15×10^{-10} A·cm⁻²) was further reduced compared with that of the

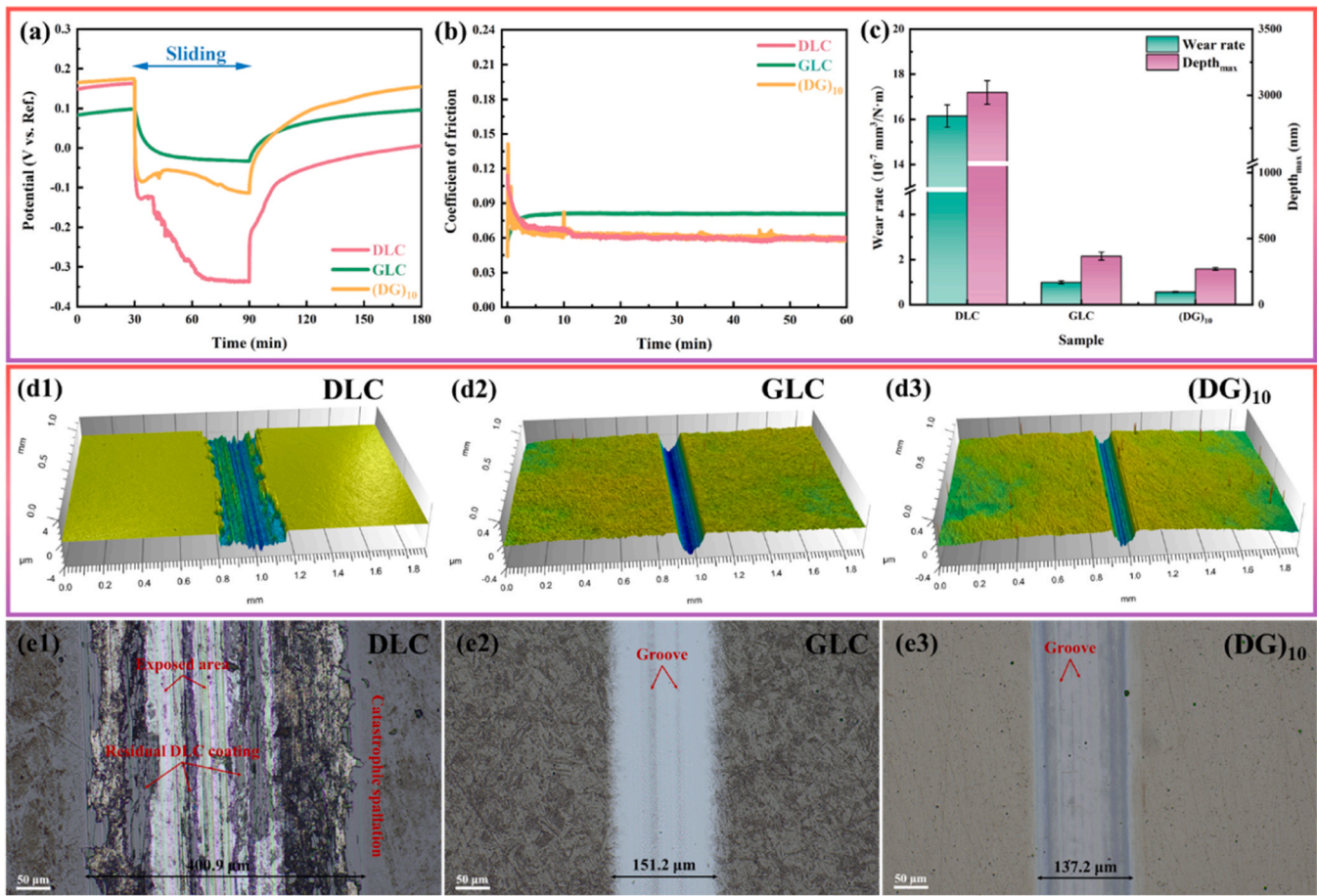


Fig. 7. (a) OCP, (b) COF, (c) wear rate and maximum wear depth, (d1-d3) 3D wear track profiles, and (e1-e3) wear track optical morphologies of the DLC, GLC, and (DG)₁₀ coatings under 20 N-1 h test condition.

DLC coating, suggesting that the discontinuous alternating growth could effectively reduce through-thickness defects.

Since the a-C was inert, its current response originated from the interaction of the corrosive medium with the substrate or the Cr interlayer through defects. So, under potentiostatic conditions, the current density-time could evaluate the resistance of the a-C coatings to pitting corrosion. In Fig. 5 (b), PDP tests at 0.7 V were conducted for 150 min. Both the DLC and (DG)₁₀ coatings exhibited superior corrosion protection performance. During the stable stage of 105–120 min, their average current densities were around $8.5 \times 10^{-9} \text{ A}\cdot\text{cm}^{-2}$, about two orders of magnitude lower than that of the GLC coating. However, after 123 min, a sharp rise in the current density appeared in the GLC coating, indicating the initiation of substrate pitting and its breakdown.

3.4. Tribocorrosion properties

Fig. 6 shows the tribocorrosion behavior of the DLC, GLC, and (DG)₁₀ coatings under the 10 N-1 h test condition. Fig. 6 (a) recorded their OCP evolution throughout the entire tribocorrosion process, which could clearly be divided into three stages: pre-loading, loading, and post-unloading. During the pre-loading stage, all coatings were required to maintain a stable initial OCP value. The initial OCP value of the GLC coating was lower than those of the other coatings, indicating a higher corrosion tendency and lower electrochemical stability. Then, in the loading stage, all OCP dropped abruptly as sliding test began, and the GLC coating exhibited a much smaller decrease compared with the DLC and (DG)₁₀ coatings, indicating that its corrosion resistance was less affected by the mechanical-tribological coupling. Subsequently, the OCP of the GLC and (DG)₁₀ coatings entered a plateau, whereas that of the

DLC coating just showed an upward trend due to competition between the corrosion rate and passivation rate of substrate or Cr interlayer [39]. After unloading, the OCP of all samples recovered rapidly, indicating a re-passivation process of the exposed substrate or Cr interlayer [40].

In Fig. 6 (b), compared with the higher COF value of the GLC coating (~ 0.085), the COF values of the DLC and (DG)₁₀ coatings were only 0.06, exhibiting the characteristics of the topmost DLC layer. Fig. 6 (c) shows the wear rate and maximum wear depth of the three coatings. Compared with the DLC coating ($5.56 \times 10^{-8} \text{ mm}^3/\text{N}\cdot\text{m}$, 208 nm) and GLC coating ($1.01 \times 10^{-7} \text{ mm}^3/\text{N}\cdot\text{m}$, 233 nm), the (DG)₁₀ coating had the minimal wear rate of approximately $3.79 \times 10^{-8} \text{ mm}^3/\text{N}\cdot\text{m}$ and wear depth of 126 nm, respectively. From the 3D profiles of the wear tracks in Fig. 6 (d1-d3), many grooves appeared for the DLC and (DG)₁₀ coatings, while the GLC coating just exhibited smooth morphological characteristics, consistent with the optical morphologies in Fig. 6 (e1-e3).

To evaluate the load-bearing capacity of the three coatings, Fig. 7 compared their tribocorrosion behavior under a higher load (20 N-1 h). In Fig. 7 (a), the OCP of the DLC coating continuously decreased after 9 min of sliding and stabilized at -0.33 V after 35 min, combined with the 3D wear track profile (Fig. 7 (d1)) and optical morphology (Fig. 7 (e1)), confirming its catastrophic spallation. The direct contact between the counterpart ball and the stainless-steel substrate also led to a sharp increase in the wear rate and wear depth, reaching as high as $1.62 \times 10^{-6} \text{ mm}^3/\text{N}\cdot\text{m}$ and 3020 nm (Fig. 7 (c)). However, its COF value remained a low level of 0.06, which was likely related to the lubrication provided by the residual DLC within the wear track [41]. In contrast, no obvious spallation occurred in the GLC and (DG)₁₀ coatings. Specifically, the (DG)₁₀ coating exhibited the lowest wear rate of

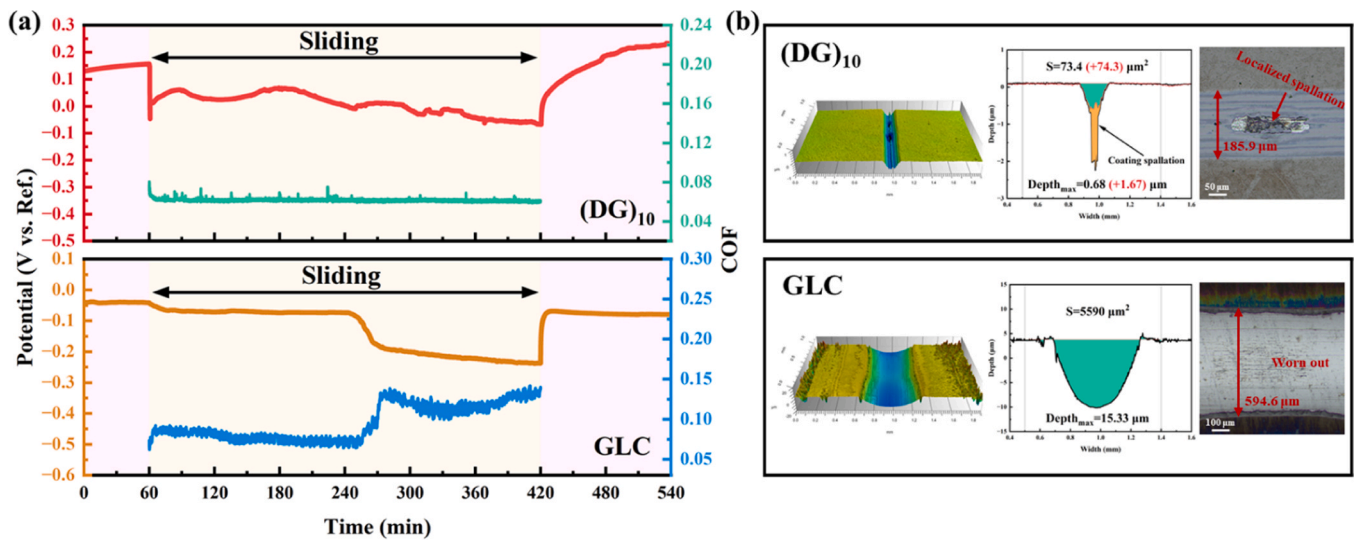


Fig. 8. (a) OCP and COF, and (b) wear track morphologies of the (DG)₁₀ and GLC coatings under 20 N-6 h test condition.

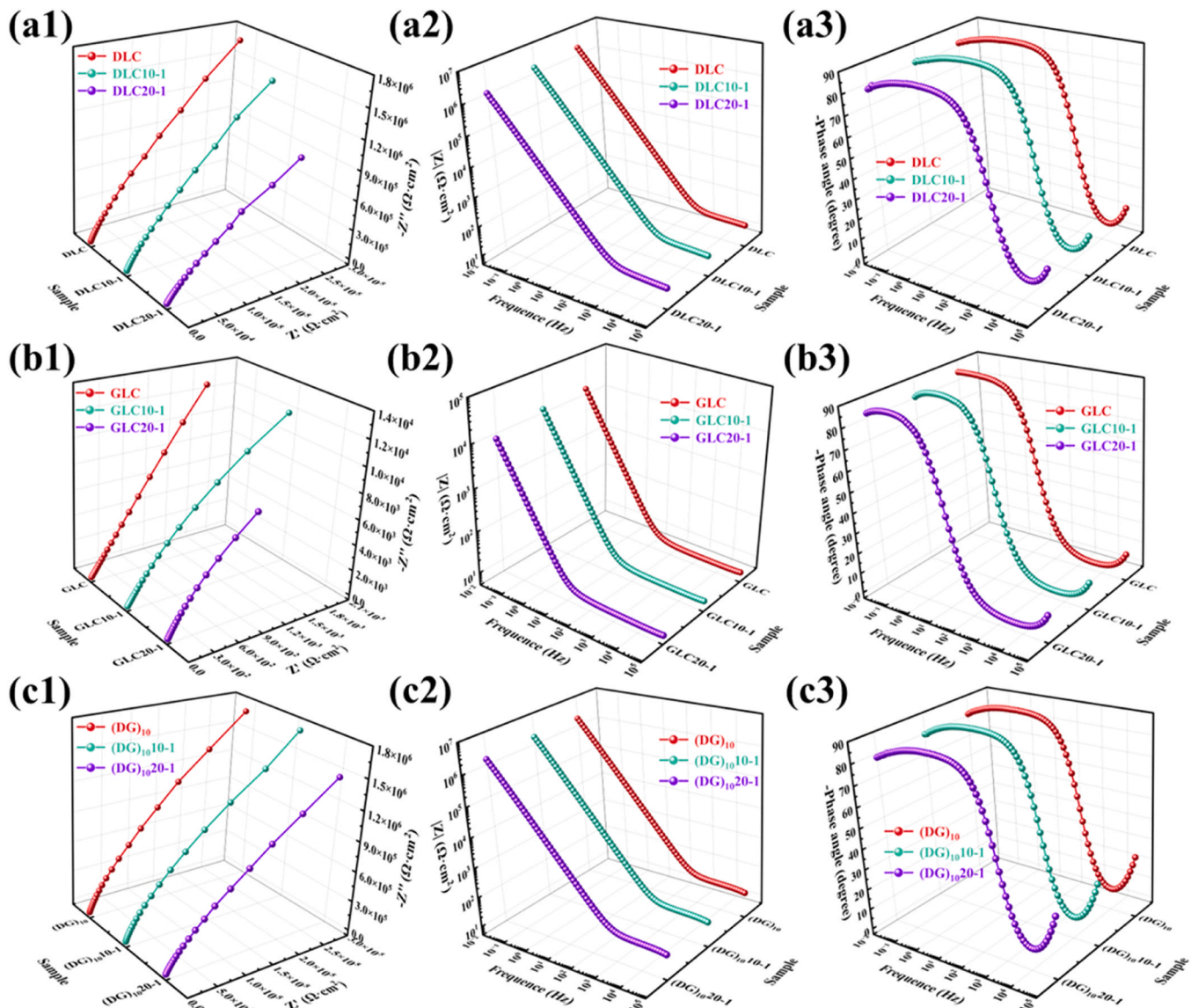


Fig. 9. Nyquist and Bode impedance plots of (a1-a3) DLC, (b1-b3) GLC, and (c1-c3) (DG)₁₀ coatings before and after tribocorrosion at 10 N-1 h and 20 N-1 h.

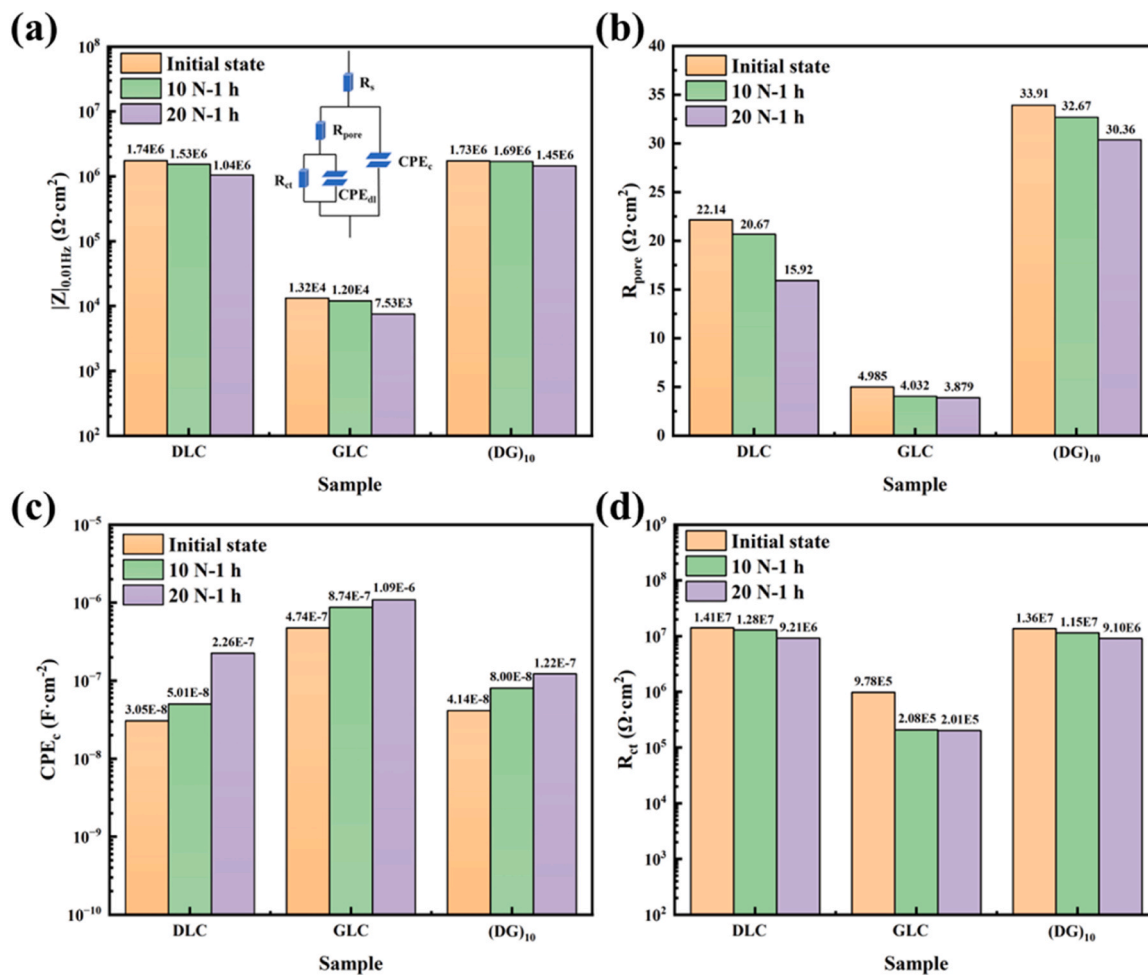


Fig. 10. EIS fitting results of the DLC, GLC, (DG)₁₀ coatings before and after tribocorrosion at 10 N-1 h and 20 N-1 h: (a) $|Z|_{0.01 \text{ Hz}}$, (b) R_{pore} , (c) CPE_c , and (d) R_{ct} .

$5.63 \times 10^{-8} \text{ mm}^3/\text{N} \cdot \text{m}$ and wear depth of 272 nm.

To further evaluate the service life of the coating under high load, Fig. 8 compared tribocorrosion behavior of the (DG)₁₀ and GLC coatings under the 20 N-6 h condition. In Fig. 8 (a), the OCP of the (DG)₁₀ coating exhibited an overall slight downward trend with significant fluctuation. Due to some local spallation (Fig. 8 (b)), the continuous exposure and passivation of the stainless-steel substrate led to the instability of the OCP. However, despite the local spallation, its COF value remained around 0.06. Specifically in Fig. 8 (b), for the non-spalled regions of the wear track, the cross-sectional area and maximum wear depth were $73.4 \mu\text{m}^2$ and $0.68 \mu\text{m}$, which could reach $147.7 \mu\text{m}^2$ and $2.35 \mu\text{m}$ at the spalled regions, respectively, indicating the explosion of local substrate.

For the GLC coating, its OCP and COF kept stable during 190 min of sliding. After 215 min, the OCP value stabilized at a low level of -0.22 V , and the COF value increased from 0.075 to 0.11. Combined with the wear track morphology in Fig. 8 (b), the GLC coating completely failed, leading to severe wear of substrate, with the wear track cross-sectional area and wear depth of $5590 \mu\text{m}^2$ and $15.33 \mu\text{m}$, respectively.

Obviously, the single-layer DLC coating failed rapidly due to catastrophic spallation under the high load of 20 N. The single-layer GLC coating exhibited better wear resistance than DLC under high load, but with short service life. The (DG)₁₀ coating demonstrated excellent tribocorrosion resistance under various load conditions.

3.5. In-situ EIS before and after tribocorrosion test

Under tribocorrosion conditions, the deterioration of corrosion

resistance is one key factor accelerating coating failure. Fig. 9 illustrates the in-situ EIS evolution of the DLC, GLC, and (DG)₁₀ coatings under different conditions (initial state, 10 N-1 h, and 20 N-1 h). From the Nyquist plots in Fig. 9 (a1, b1, c1), after tribocorrosion test, the capacitive arc radius of all three coatings decreased and exhibited a negative correlation with the applied load. Similarly, the low-frequency impedance modulus ($|Z|_{0.01 \text{ Hz}}$) followed the same variation behavior in the Bode impedance plots (Fig. 9 (a2, b2, c2)). From the Bode phase angle plots (Fig. 9 (a3, b3, c3)), low-frequency phase angle decreased, while the phase angle in the medium-to-high frequency range (10^2 – 10^5 Hz) exhibited negligible change. The above suggested that corrosion primarily occurred at the coating-substrate interface, where more corrosive media after sliding enhanced the low-frequency electrode reactions [42]. Notably, the (DG)₁₀ coating exhibited the smallest reduction in both the capacitive arc radius and $|Z|_{0.01 \text{ Hz}}$ compared with the other two coatings, demonstrating its superior tribocorrosion resistance.

As shown in Fig. 10, the EIS results were fitted and analyzed using ZSimpWin software. Fig. 10 (a) shows the low-frequency impedance modulus of the three coatings and the equivalent circuit model with two time constants. $|Z|_{0.01 \text{ Hz}}$ is an electrochemical parameter commonly used to directly evaluate the corrosion resistance of coatings, where a lower value corresponds to poorer protective performance [15]. Compared with the DLC (12.1%, 40.2%) and GLC coatings (9.1%, 43.0%), the (DG)₁₀ coating exhibited a $|Z|_{0.01 \text{ Hz}}$ reduction of only 2.3% and 16.2% after tribocorrosion at 10 N-1h and 20 N-1h, respectively. This indicated that the (DG)₁₀ coating underwent the least corrosion resistance degradation induced by sliding. In the equivalent circuit model, R_s , R_{pore} , and R_{ct} represent the solution resistance between the

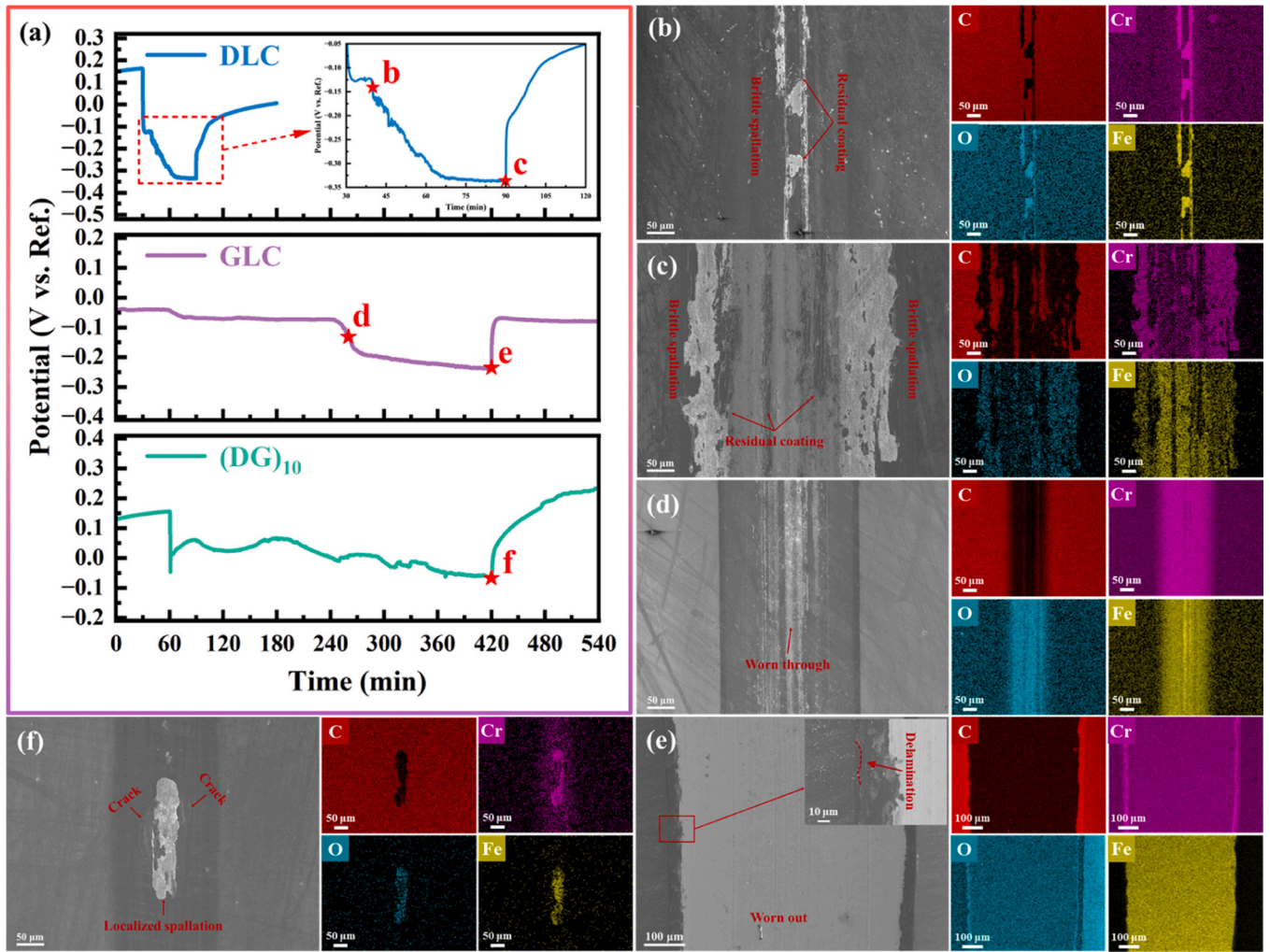


Fig. 11. (a) OCP evolution of the DLC, GLC, and (DG)₁₀ coatings under 20 N. The wear track surface morphologies at different tribocorrosion stages: (b) DLC-10 min, (c) DLC-1 h, (d) GLC-200 min, and (e) GLC-6 h, (f) (DG)₁₀-6 h.

working and reference electrodes, the pore resistance in the coating, and the charge transfer resistance of the double layer at the coating-substrate interface, respectively. Since the coating inevitably contains pores or defects, a constant phase element (CPE) is used instead of an ideal capacitor (C), where CPE_c and CPE_{dl} correspond to the coating capacitance and the double-layer capacitance, respectively [43]. The fitting results and accuracy are summarized in Table S2. R_{pore} reflects the barrier ability of the coating against corrosive media, with a higher R_{pore} indicating fewer pores or defects [44,45]. The initial R_{pore} of the DLC, GLC, and (DG)₁₀ coatings were 22.14, 4.985, and 33.91 $\Omega \cdot \text{cm}^2$, respectively, indicating the poorest corrosion resistance of the GLC coating (Fig. 10 (b)). Similarly, a larger CPE_c reflects stronger permeability of corrosive media within the coating. In Fig. 10 (c), the GLC coating exhibited the highest initial CPE_c ($4.74 \times 10^{-7} \text{ F} \cdot \text{cm}^{-2}$), compared with the DLC ($3.05 \times 10^{-8} \text{ F} \cdot \text{cm}^{-2}$) and (DG)₁₀ coatings ($4.14 \times 10^{-8} \text{ F} \cdot \text{cm}^{-2}$). Therefore, the variations in R_{pore} and CPE_c were associated with the number and size of pores, defects, and microcracks within the coating. However, after tribocorrosion under the high load of 20 N, the R_{pore} of the DLC, GLC, and (DG)₁₀ coatings decreased by 28.1%, 22.2%, and 10.5%, while their CPE_c increased by approximately sixfold, 1.3-fold, and twofold, respectively. Compared with the GLC coating, the highly brittle DLC coating was more prone to rapid microcrack propagation under the high load, leading to the formation of more permeation channels. In contrast, the C/C multilayer structure effectively suppressed the direct initiation of microcracks and retarded the

infiltration of corrosive media. The above results could also be interpreted based on Eq. 5 [46]:

$$CPE_c = \frac{\gamma \epsilon_0 \epsilon (1 - P) A}{d} \quad (5)$$

where γ represents the surface roughness of the coating, ϵ_0 and ϵ denote the dielectric constants of vacuum and the coating material, respectively, P is the porosity of the coating, A refers to the effective exposed area of the coating, and d stands for the thickness of the coating.

During tribocorrosion test, the DLC coating generated more porous channels, consequently, the effective exposed area of the coating increased, resulting in a more pronounced change in CPE_c [47]. R_{ct} typically represents the charge transfer resistance of electrochemical reactions occurring at the coating-substrate interface, where a decrease in R_{ct} indicates an accelerated interfacial reaction rate and degradation of the coating performance [48]. In Fig. 10 (d), the GLC coating exhibited the lowest initial R_{ct} , and the applied load significantly accelerated the interfacial corrosion rate, leading to its severe performance degradation, since the GLC coating has better electrical conductivity. However, the (DG)₁₀ coating remained at a high R_{ct} level with minimal variation, which may be attributed to the overall densification of the coating and the shielding effect of the insulating DLC layer.

Based on the EIS results, the (DG)₁₀ coating exhibited superior corrosion resistance compared with the single-layer DLC and GLC coatings. Meanwhile, the slight changes induced by loading and sliding

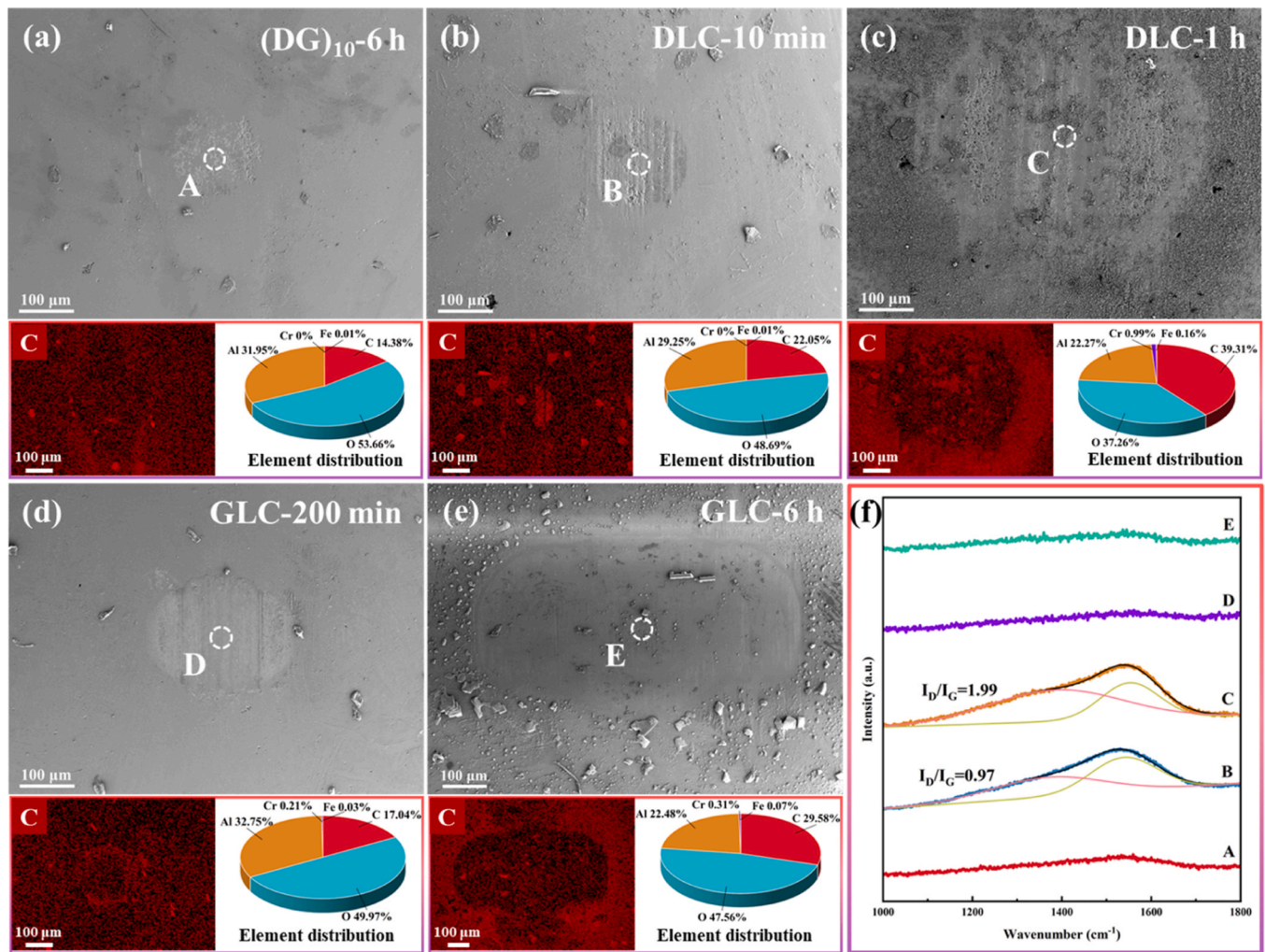


Fig. 12. Wear scar morphologies and element distributions of Al₂O₃ counterpart balls under a 20 N load at various tribocorrosion stages: (a) (DG)₁₀-6 h, (b) DLC-10 min, (c) DLC-1 h, (d) GLC-200 min, and (e) GLC-6 h. (f) Raman spectra from the wear scar centers.

demonstrated the excellent stability of the (DG)₁₀ coating.

3.6. Wear track and wear scar at different tribocorrosion stages

Fig. 11 shows surface morphology of wear tracks at different tribocorrosion stages under 20 N. The (DG)₁₀ coating only suffered slight localized spallation, whereas the DLC and GLC coatings underwent complete failure, and the continuous decrease in OCP indicated the onset of coating spallation, as shown in Fig. 11 (a).

Given the distinct spallation of the DLC and GLC coatings, the wear tracks at the initial stage of OCP decline were selected to study their failure modes. The OCP curves of the three coatings in Fig. 7(a) and Fig. 8(a) were replotted in one figure (Fig. 11 (a)), and the five points of interest were marked with red five-pointed stars. For the DLC coating at initial spallation (Fig. 11 (b)) and post-tribocorrosion (Fig. 11 (c)) stages, the wear tracks indicated its brittle spallation characteristic, and exhibited avalanche type spallation at the end of tribocorrosion test. Furthermore, some residual DLC coating fragments were observed within the wear track in Fig. 11 (c), which may still maintain a stable COF value. For the GLC coating, at the initial failure stage (Fig. 11 (d)), its wear track center showed some uniform and continuous strip-shaped exposed regions. After tribocorrosion (Fig. 11 (e)), the wear track exhibited smooth edges with obvious delamination, suggesting its layer-by-layer shear removal characteristic. For the (DG)₁₀ coating after 6 h tribocorrosion test in Fig. 11 (f), its wear track had some obvious

localized brittle spallation in the center area, with numerous cracks and debris along its edges. The corresponding EDS mapping confirmed the exposure of the substrate.

Notably, the failure mode of the (DG)₁₀ coating was primarily determined by the DLC layers, but the introduction of the GLC layers significantly delayed the onset of catastrophic coating spallation, as evidenced by the overall morphology of the wear track on the (DG)₁₀ coating (Fig. S2).

Fig. 12 shows the corresponding wear scar morphologies of the Al₂O₃ counterpart balls. First, all wear scars could be divided into the frictional contact center and the wear debris accumulation region. In Fig. 12 (a), for the (DG)₁₀ coating, it had the smallest wear scar, about 160 μm in diameter. Some slight grooves and fine pits appeared on the central region. Meanwhile, EDS mapping revealed a few bright red spots with C element, indicating little adhesion of carbon-based transfer films. Similarly, for the DLC coating in Fig. 12 (b, c), distinct grooves and dense pits were observed inside the wear scar. As the tribocorrosion test proceeded, the wear scar became much larger, and more carbon-based material and Cr and Fe elements accumulated in the central region. As for the GLC coating in Fig. 12 (d, e), few grooves, pits, or flaky transfer films were present inside the wear scar at the initial spallation stage. Once the counterpart ball wore through the GLC coating and contacted the substrate, the wear scar rapidly expanded, and the grooves, pits, and flaky transfer films disappeared. Fig. 12 (f) shows the Raman spectra at the contact centers of the wear scars. No distinct a-C characteristic peaks

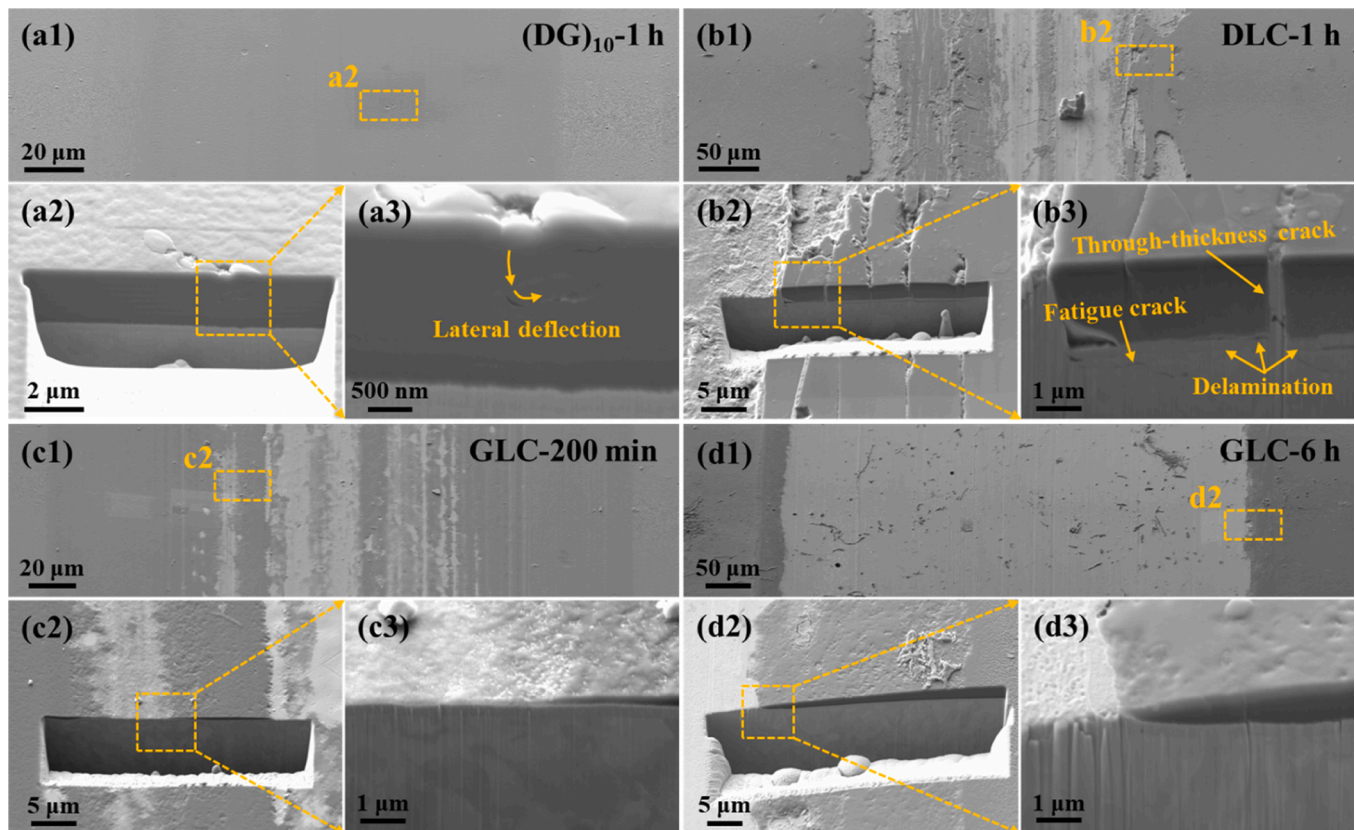


Fig. 13. Surface and cross-sectional morphologies of wear tracks at different tribocorrosion stages under a 20 N load: (a1-a3) (DG)₁₀-1 h, (b1-b3) DLC-1 h, (c1-c3) GLC-200 min, and (d1-d3) GLC-6 h.

were observed for the (DG)₁₀ and GLC coatings. Meanwhile, for the DLC coating, the I_D/I_G values at initial spallation and post-tribocorrosion stages were 0.97 and 1.99, respectively, both higher than the pre-tribocorrosion value of 0.83 (Fig. 2(c)), due to the graphitization transition during transfer film formation [49].

3.7. Crack propagation during the tribocorrosion process

During the tribocorrosion process, the initiation of coating cracks can directly alter its structural integrity and corrosion protection performance, resulting in various failure behaviors. Fig. 13 illustrates the surface and cross-sectional morphologies inside the wear tracks at different tribocorrosion stages under 20 N. After 1 h, the (DG)₁₀ coating only exhibited slight grooves and pits (Fig. 13 (a1)), in contrast to the extensive brittle spallation of the DLC coating (Fig. 13 (b1)). From cross-sectional morphologies of the DLC coating, many large-scale cracks rapidly propagated to the coating-substrate interface and developed into through-thickness cracks in the substrate (Fig. 13 (b2, b3)). In contrast, with the soft/hard alternating structure in the (DG)₁₀ coating, the few cracks existed mainly in the subsurface layers of the coating, due to lateral deflection (Fig. 13 (a2, a3)), and the C/C and coating-substrate interfaces beneath remained intact. For the GLC coating (Fig. 13 (c1, d1)), it was gradually removed throughout the sliding process. However, the coating-substrate interface remained well-bonded, and no fatigue cracks were observed inside the coating or substrate (Fig. 13 (c3, d3)).

To further evaluate the crack propagation behavior of the (DG)₁₀ coating, its wear track morphologies after 6 h test are shown in Fig. 14. Clearly, some localized spallation areas around several hundred microns in size appeared in the wear track (Fig. 14 (a)). Fig. 14 (b) and (c) show the cross-sectional morphologies of spalled regions b and c, which correspond to the spallation areas marked in Fig. 14 (a). Similar to the crack propagation of the DLC coating in Fig. 13 (b2) and (b3), through-

thickness cracks formed in the (DG)₁₀ coating, and the delamination along the coating-substrate interface and fatigue cracking inside the substrate led to its spallation.

From unspalled region d with cracks in Fig. 14 (a), the corresponding FIB-STEM image (Fig. 14 (d, e)) exhibited a longitudinal crack extending from the coating surface to the substrate, with a length of approximately 1800 nm and a maximum width of 158 nm. Meanwhile, three DLC/GLC bilayers were removed, corresponding to an approximate thickness loss of 600 nm, which was comparable to the maximum wear depth of the unspalled regions in Fig. 8 (b). From edges of the through-thickness crack, the DLC and GLC layers were well bonded, while a microcrack extended into the substrate. In Fig. 14 (e), Fe and O enrichment can be observed at the edges of the through-thickness crack. Besides, within the microcrack between Cr layer and substrate, which was marked with a yellow border as region f in Fig. 14 (d), Fe₃O₄ was identified. This indicated that the corrosive medium could rapidly penetrate along these through-thickness cracks to reach the coating-substrate interface and even the substrate, thus accelerating the damage and spallation of the (DG)₁₀ coating.

3.8. Failure modes and damage mechanism

Previous research had confirmed that, during tribocorrosion test, DLC failed rapidly via brittle spallation under a higher load, while GLC showed poor wear resistance due to its low hardness and graphite-like characteristics [12]. Here, with the DLC/GLC multilayer structure, the (DG)₁₀ coating demonstrated excellent tribocorrosion resistance under various load conditions and longer service life, as shown in Figs. 6, 7 and 8.

Generally speaking, the tribocorrosion behavior of carbon-based coating exhibits abrupt or progressive failure characteristic. Given the high stability of the carbon-based coating, the evolution of surface and

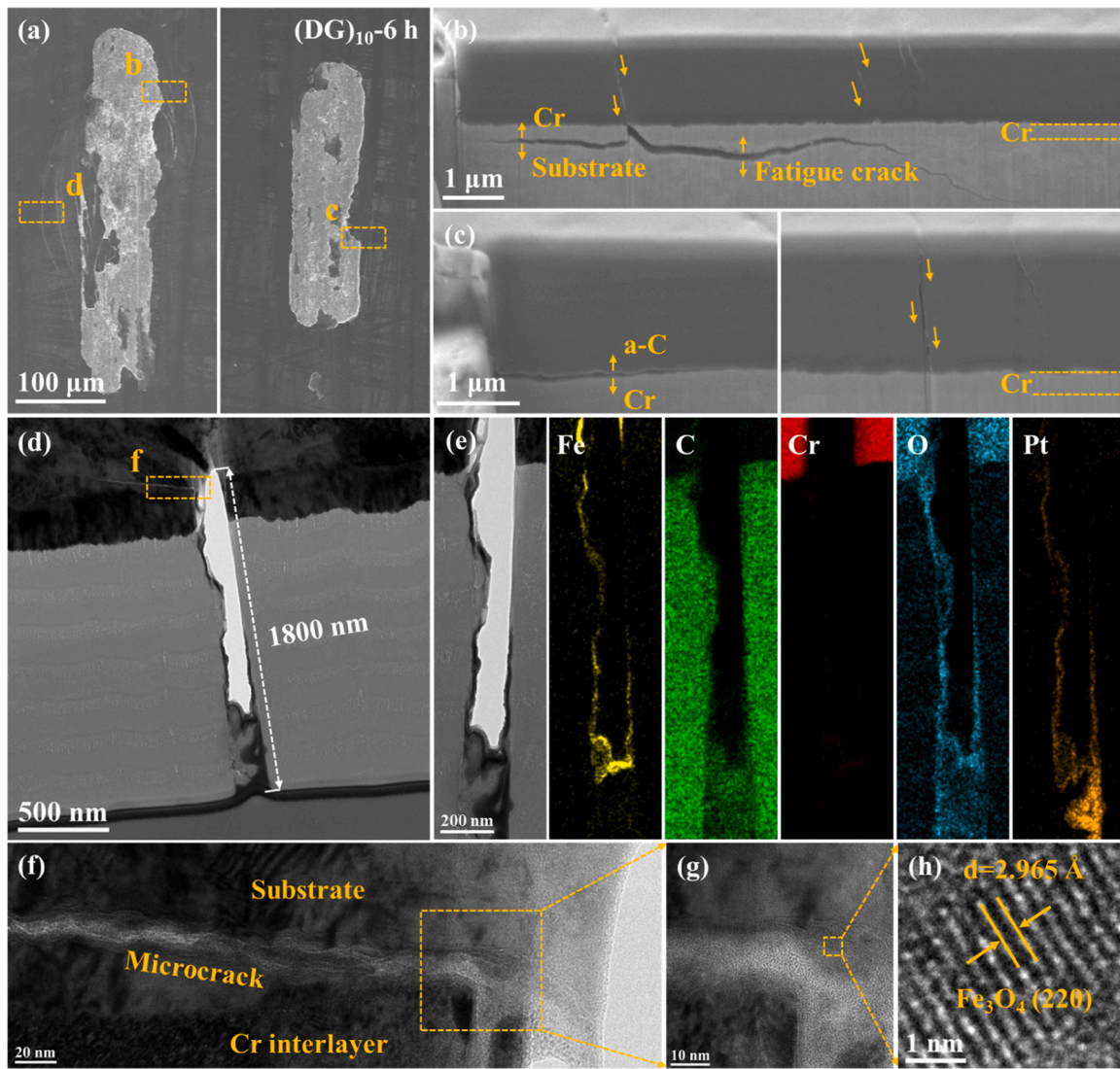


Fig. 14. Surface and cross-sectional morphologies of the (DG)₁₀ coating after tribocorrosion test at 20 N-6 h: (a) localized spallation morphology, (b, c) cross-sectional morphologies of spalled region b and c, (d, e) cross-sectional characteristics and EDS mapping of the through-thickness crack from unspalled region d, (f) morphology of the microcrack at the coating-substrate interface, and (g, h) HRTEM images within the microcrack between Cr layer and substrate.

interfacial characteristics became the two key factors. Under the lower load of 10 N condition, all three coatings were sufficient to withstand

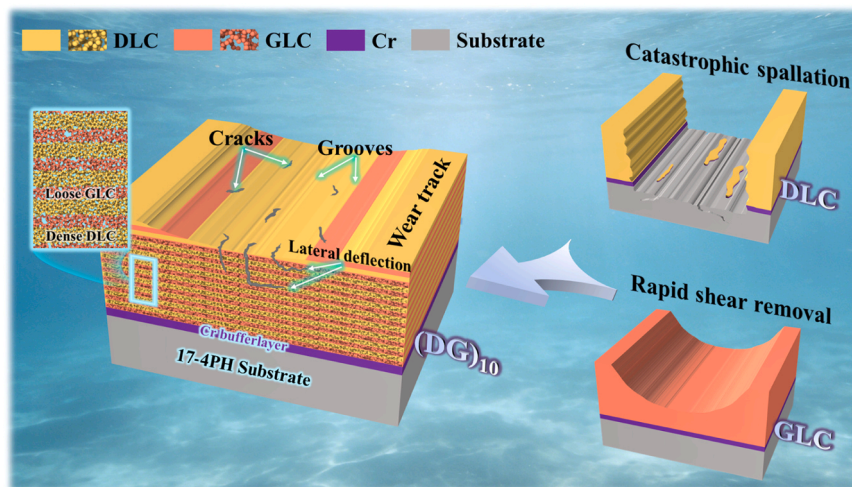


Fig. 15. The schematic diagram of damage mechanism for the DLC, GLC, and (DG)₁₀ coatings.

the external forces and maintain their structural integrity. In this case, both the DLC and (DG)₁₀ coatings showed similar COF and lower wear rate, whereas the GLC coating had a larger wear rate, as shown in Fig. 6, since the topmost carbon layer played a dominant role. As the applied load increased to 20 N, the brittle spallation of the DLC coating occurred (Fig. 13 (b)), due to its poor plastic deformation capability (Fig. 4). Consequently, the direct contact between the counterpart ball and the substrate led to the largest wear rate (Fig. 7). Meanwhile, carbon-based transfer films remained firmly adhered to the frictional contact center of the wear scar (Fig. 11 (c)), which provided lubrication and maintained the stable COF of the DLC coating. As for the GLC coating, its high plasticity and graphite-like characteristics resulted in layer-by-layer shear removal (Fig. 13 (c)). The substrate was not exposed until the GLC coating was completely worn through, so GLC coating exhibited better short-term tribocorrosion protection (Fig. 7). However, the rapid shear consumption of GLC also explained its poor wear resistance with longer test time (Fig. 8 and Fig. 13 (d)).

In contrast, the (DG)₁₀ coating remained relatively intact after 6 h tribocorrosion test under 20 N. First, the (DG)₁₀ coating had the excellent plastic deformation capability (Fig. 4). Its soft/hard alternating structure reduced the crack propagation rate and retarded the formation of through-thickness cracks. Hence, severe brittle fracture of the DLC coating under high load could be avoided, and the (DG)₁₀ coating exhibited excellent anti-tribocorrosion properties under various loads and test time. In this case, progressive damage from DLC/GLC and coating-substrate interfaces will be primarily responsible for the (DG)₁₀ coating failure. Specifically, the DLC/GLC interfaces caused lateral deflection of propagating cracks (Fig. 13 (a)), effectively preventing coating-substrate interfacial delamination under lower load, and exhibiting its best short-term protective performance under higher load. Notably, such crack deflection could still prolong the diffusion path of corrosive media (Fig. S3). In addition, more DLC/GLC interfaces resulted in excellent corrosion resistance (Fig. 5) and high stability (Fig. 10). Consequently, the corresponding damage mechanism of the three coatings are illustrated in Fig. 15.

However, during prolonged high-load sliding test, through-thickness cracks still developed in the (DG)₁₀ coating, exposing the coating-substrate interface to corrosive media (Fig. 14), since the insertion of GLC layer was unable to completely change the crack and spallation characteristic of DLC coating (Figs. 4 and 14). This resulted in slight localized brittle spallation and a slow decrease in OCP values of the (DG)₁₀ coating during tribocorrosion test (Fig. 11). Though the failure mode of the (DG)₁₀ coating remained brittle spallation due to the brittle characteristic of DLC layers, the application of soft/hard alternating structure effectively delayed the onset of coating spallation.

4. Conclusion

In this work, a tribocorrosion-resistant (DG)₁₀ coating with a soft/hard alternating pure C/C structure was successfully fabricated via the DCMS-LIS alternating cyclic sputtering technique. The tribocorrosion behavior of the (DG)₁₀ coating under various conditions, as well as the role of interfacial characteristics on its damage mechanism were investigated. Compared with the DLC and GLC coatings, the (DG)₁₀ coating exhibited excellent anti-tribocorrosion properties, with the lowest COF of 0.06 and a wear rate on the order of 10^{-8} mm³/N·m under both high (20 N) and low (10 N) loads. Under a low load of 10 N, all three coatings could maintain their structural integrity. Owing to the dominant role of the topmost carbon layer, the (DG)₁₀ coating exhibited tribological properties similar to those of the DLC coating. Under a high load of 20 N, the onset of spallation in the (DG)₁₀ coating was effectively delayed. The multi-interfaced soft/hard alternating structure prevented the rapid catastrophic fracture of the coating under high load by reducing the crack propagation rate and suppressing the formation of through-thickness cracks. However, the progressive damage from the DLC/GLC and coating-substrate interfaces became the primary cause of

its failure with increasing test time. Together, the findings not only reveal the damage mechanism of the soft/hard alternating pure C/C coating during tribocorrosion, but also provide a feasible strategy for developing high-performance protective coatings tailored to the variable-load service requirements in harsh marine environments.

CRediT authorship contribution statement

Longwen Shang: Writing – original draft, Validation, Investigation, Formal analysis, Data curation, Conceptualization. **Silong Zhang:** Writing – review & editing, Software. **Hao Li:** Writing – review & editing, Validation. **Yanjin Li:** Data curation. **Xin Zhang:** Methodology, Investigation. **Rende Chen:** Funding acquisition. **Kazuhito Nishimura:** Funding acquisition. **Peng Guo:** Writing – review & editing, Funding acquisition. **Aiying Wang:** Writing – review & editing, Supervision, Funding acquisition, Conceptualization.

Declaration of Competing Interest

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

Acknowledgements

This work was financial supported by the Strategic Priority Research Program of the Chinese Academy of Sciences (XDB1210402), National Science Foundation of China (U24A2030), Leading Innovative and Entrepreneur Team Introduction Program of Zhejiang (2024R01004), Major Special Project of Ningbo (2024Z134), and Natural Science Foundation of Ningbo (2025QL023, 2024J081).

Appendix A. Supporting information

Supplementary data associated with this article can be found in the online version at doi:10.1016/j.triboint.2026.112052.

Data availability

Data will be made available on request.

References

- [1] Xu R, Wu D, Wang Z, Deng Y, Cheng Q, Pang H, Wei W, Liu Y. Knowledge mapping analysis of the track and hotspot of water lubrication: a scientometrics review. *Friction* 2023;11(9):1557–91.
- [2] Liu L, Chen H, Li R, Hao X, Zhao X. Structural, dry sliding wear and tribocorrosion behaviors of the Mo-Si-B intermetallic alloys. *Tribol Int* 2025;201:110247.
- [3] Yan H, Shi L, Ling B, Chen H, Zhang B, Zhang L. Dual-stimuli triggered lubrication and barrier: a synergistic strategy for high-performance tribocorrosion protection in MXene-based epoxy coatings. *ACS Appl Mater Interfaces* 2025;17(46):63836–51.
- [4] Liu Y, Zhu L, Ma G, Lu Y, Wei A, Yang C, Huang Y, Liu M, Wang H. Tribocorrosion behaviours and mechanisms of Inconel 625–nano/micro Al₂O₃ composite coatings prepared by plasma enhanced high-velocity arc spraying. *Corros Sci* 2025;256:113221.
- [5] Zhou X, Du H, Ma G, Wang Z, Yang W, Nishimura K, Guo P, Wang A, Ke P. HiPIMS-deposited Cr_xN coatings for marine applications: controllable microstructure and tribocorrosion behavior. *Corros Commun* 2025;20:24–32.
- [6] Wang C, Zhang X, Lu Y, Lu X, Hao J. Tailoring long-life, wear and corrosion resistant a-C:H coating for marine applications by PVD combined HC-PECVD co-deposition technology. *Surf Coat Technol* 2025;513:132450.
- [7] Chen Z, Wei X, Lu S, Ding J, Du N, Feng C, Chen K, Guo P, Lee K-R, Zhang G, Li X. Surface *in-situ* graphitization and properties of amorphous carbon film induced by laser irradiation. *Friction* 2025;13(6):9440977.
- [8] Sun Q, Mao X, Yao Y, Li J, Shen Z, Tu K-N, Liu Y. Effect of WC thickness on the microstructure and properties of WC-C/DLC coated 304 steel. *Appl Surf Sci* 2025;704:163418.
- [9] Liu Y, Ma G, Ma X, Li H, Guo P, Wang A, Ke P. Corrosion and tribocorrosion performance degradation mechanism of multilayered graphite-like carbon (GLC) coatings under deep-sea immersion in the western Pacific. *Corros Sci* 2024;239:112418.

- [10] Li Z, Guan X, Wang Y, Li J, Cheng X, Lu X, Wang L, Xue Q. Comparative study on the load carrying capacities of DLC, GLC and CrN coatings under sliding-friction condition in different environments. *Surf Coat Technol* 2017;321:350–7.
- [11] Yan M, Wang X, Zhang S, Zhang S, Sui X, Li W, Hao J, Liu W. Friction and wear properties of GLC and DLC coatings under ionic liquid lubrication. *Tribol Int* 2020; 143:106067.
- [12] Zhang Y, Li H, Cui L, Yang W, Ma G, Chen R, Guo P, Ke P, Wang A. Comparative study on tribocorrosion behavior of hydrogenated/hydrogen-free amorphous carbon coated WC-based cermet in 3.5 wt% NaCl solution. *Corros Sci* 2024;227: 111738.
- [13] Zhao Y, Xu F, Yuan L, Zhao W, Zhang H, Zuo D. Microstructural, mechanical and tribological performances of DLC/CrN multilayer films with different modulation period. *Diam Relat Mater* 2025;154:112163.
- [14] Zhang Y, Zhou X, Chen R, Yang W, Guo P, Nishimura K, Li X, Wang A. Ultrahigh tribocorrosion resistance of hydrogenated amorphous carbon coating via trace gradient W incorporation. *Corros Sci* 2024;240:112504.
- [15] Ma H, Cui Y, Liu R, Ke P, Wang F, Liu L. Investigation of the corrosion-wear failure mechanism of multilayer Cr/GLC coatings under hydrostatic pressure-tribocorrosion alternation. *Corros Sci* 2024;240:112493.
- [16] Xiao Q, Yang Z, Xiao Y, Guo Y, Hu L, Hu Q, Liu H, Guo X, Guo K, Wang Q. Enhancement of mechanical, tribological and corrosion properties of ta-C coatings through soft/hard multilayer structures. *Diam Relat Mater* 2025;152:111972.
- [17] Liu Y, Li S, Li H, Ma G, Sun L, Guo P, Ke P, Lee K-R, Wang A. Controllable defect engineering to enhance the corrosion resistance of Cr/GLC multilayered coating for deep-sea applications. *Corros Sci* 2022;199:110175.
- [18] Jiang Z, Wei J, Li H, Chen R, Yang W, Ma G, Nishimura K, Guo P, Wang A. Enhanced corrosion resistance of graphite-like carbon coatings via dense ta-C interlayer encapsulating. *Diam Relat Mater* 2024;146:111139.
- [19] Wang Z, Zhang Z, Li J, He Y, Li Y. Bias-induced sp^3 bonding gradient design for enhancing the frictional bearing capacity of superhard Ta-C coating. *Tribol Int* 2026;219:111780.
- [20] Lin Y, Zia A, Zhou Z, Shum W, Li Y. Development of diamond-like carbon (DLC) coatings with alternate soft and hard multilayer architecture for enhancing wear performance at high contact stress. *Surf Coat Technol* 2017;320:7–12.
- [21] Wang C, Zhang X, Lu Y, Hao J. Multilayer structure design for preparing ultralow wear amorphous carbon films. *Tribol Int* 2024;196:109729.
- [22] Usman M, Zhou Z, Zia A, Li K. Designing hydrogen-free diamond like multilayer carbon coatings for superior mechanical and tribological performance. *Tribol Int* 2024;192:109211.
- [23] Guo J, Chen X, Deng W, Qi W, Luo T. Structure-property correlations in multilayer hydrogenated amorphous carbon films enabling long-lasting stable superlubricity under ultra-high contact stress. *Carbon* 2025;245:120797.
- [24] Guo C, Pei Z, Fan D, Liu R, Gong J, Sun C. Predicting multilayer film's residual stress from its monolayers. *Mater Des* 2016;110:858–64.
- [25] Zhang T, Jiang F, Liao T, Dneg Q, Li S, Wang Y, Leng Y. Tribological behavior of diamond like carbon film sliding against CoCrMo or Al_2O_3 in air and water environment. *Tribol Int* 2016;95:456–61.
- [26] Liu A, Deng J, Cui H, Chen Y, Zhao J. Friction and wear properties of TiN, TiAlN, AlTiN and CrAlN PVD nitride coatings. *Int J Refract Met Hard Mater* 2012;31:82–8.
- [27] Mo J, Zhu M, Lei B, Leng Y, Huang N. Comparison of tribological behaviours of AlCrN and TiAlN coatings—Deposited by physical vapor deposition. *Wear* 2007; 263(7–12):1423–9.
- [28] Paik N. Raman and XPS studies of DLC films prepared by a magnetron sputter-type negative ion source. *Surf Coat Technol* 2005;200(7):2170–4.
- [29] Nakao S, Yukimura K, Nakano S, Ogiso H. DLC Coating by HiPIMS: the influence of substrate bias voltage. *IEEE Trans Plasma Sci* 2013;41(8):1819–29.
- [30] Chen X, Zhang C, Kato T, Yang X, Wu S, Wang R, Nosaka M, Luo J. Evolution of tribo-induced interfacial nanostructures governing superlubricity in a-C:H and a-C:H:Si films. *Nat Commun* 2017;8(1):1675.
- [31] Galvan D, Pei Y, Th J, De Hosson M, Cavaleiro A. Determination of the sp^3 C content of a-C films through EELS analysis in the TEM. *Surf Coat Technol* 2005; 200:739–43.
- [32] Wang N, Komvopoulos K. The multilayered structure of ultrathin amorphous carbon films synthesized by filtered cathodic vacuum arc deposition. *J Mater Res* 2013;28(16):2124–31.
- [33] Xie J, Komvopoulos K. The effect of Argon ion irradiation on the thickness and structure of ultrathin amorphous carbon films. *J Appl Phys* 2016;119(9):095304.
- [34] Beake B. The influence of the H/E ratio on wear resistance of coating systems – Insights from small-scale testing. *Surf Coat Technol* 2022;442:128272.
- [35] Hassani S, Bielawski M, Beres W, Martinu L, Balazinski M, Klemberg-Sapieha JE. Predictive tools for the design of erosion resistant coatings. *Surf Coat Technol* 2008;203(3):204–10.
- [36] Zhao Y, Xu F, Xu J, Li D, Sun S, Gao C, Zhao W, Lang W, Liu J, Zuo D. Effect of the bias-graded increment on the tribological and electrochemical corrosion properties of DLC films. *Diam Relat Mater* 2022;130:109421.
- [37] Ramos B, Vicente F, Hammes G, Bendo T, Binder C. Enhancing corrosion and tribology performance of stainless steel with DLC coatings: effects of doping and multilayered structures. *Surf Coat Technol* 2024;477:130334.
- [38] Liu Y, Li S, Zhou X, Guo P, Chen R, Wei J, Wang A, Ke P. Enhanced anti-tribocorrosion property of a-C film under high hydrostatic pressure by high power pulsed magnetron sputter (HiPIMS). *J Mater Res Technol* 2024;28:3052–67.
- [39] Yan Y, Neville Y, A, Dowson D. Tribo-corrosion properties of cobalt-based medical implant alloys in simulated biological environments. *Wear* 2007;263(7):1105–11.
- [40] Li L, Liu L, Li X, Guo P, Ke P, Wang A. Enhanced Tribocorrosion Performance of Cr/GLC multilayered films for marine protective application. *ACS Appl Mater Interfaces* 2018;10(15):13187–98.
- [41] Li H, Liu L, Guo P, Sun L, Wei J, Liu Y, Li S, Wang S, Lee K-R, Ke P, Wang A. Long-term tribocorrosion resistance and failure tolerable of multilayer carbon-based coatings. *Friction* 2022;10(10):1707–21.
- [42] Wang Z, Zhang J, Han X, Li Q, Wang Z, Wei R. Corrosion and salt scale resistance of multilayered diamond-like carbon film in CO_2 saturated solutions. *Corros Sci* 2014; 86:261–7.
- [43] Liu Y, Wang J, Liu L, Li Y, Wang F. Study of the failure mechanism of an epoxy coating system under high hydrostatic pressure. *Corros Sci* 2013;74:59–70.
- [44] Kim C, Cho S, Yang W, Karayan A, Castaneda H. Corrosion behavior of Al-Si-Mg coated hot-press-forming steel. *Corros Sci* 2021;183:109339.
- [45] Li J, Meng F, Liu L, Cui Y, Liu R, Zheng H, Wang F. Effect of nano- Fe_2O_3 /graphene oxide hybrids on the corrosion resistance of epoxy coating under alternating hydrostatic pressure. *Corros Commun* 2022;5:62–72.
- [46] Liu C, Bi Q, Leyland A, Matthews A. An electrochemical impedance spectroscopy study of the corrosion behaviour of PVD coated steels in 0.5 N NaCl aqueous solution: part II. EIS Interpret *Corros Behav Corros Sci* 2003;45(6):1257–73.
- [47] Ma H, Liu R, Cui Y, Ke P, Wang F, Liu L. The effect law of different hydrostatic pressures on the failure of multilayer Cr/GLC coatings in 3.5 wt% NaCl solution. *Corros Sci* 2023;217:111120.
- [48] Meng F, Zhang T, Liu L, Cui Y, Wang F. Failure behaviour of an epoxy coating with polyaniline modified graphene oxide under marine alternating hydrostatic pressure. *Surf Coat Technol* 2019;361:188–95.
- [49] Guo F, Dong G, Qin L, Zhang D, Guo J. Tribological performance of DLC coating under aqueous solutions. *Lubr Sci* 2019;31(6):262–72.