



REGULAR ARTICLE

The Influence of the Structure and Properties of the Transition Zone Between the Matrix and Filler on the Destruction of Composite Coatings During Friction

M.V. Kindrachuk¹ , Yu.I. Bohomol², T.S. Cherepova³, N.M. Stebeletska^{4,*} , V.V. Kharchenko¹,
A.L. Hlovyn⁵

¹ National University "Kyiv Aviation Institute", 03058 Kyiv, Ukraine

² National Technical University of Ukraine "Igor Sikorsky Kyiv Polytechnic Institute", 03056 Kyiv, Ukraine

³ Kurdyumov Institute for Metal Physics, Ukrainian National Academy of Sciences, 03142 Kyiv, Ukraine

⁴ Separate Subdivision of the National University of Life and Environmental Sciences of Ukraine "Berezhany Agrotechnical Institute", 47501 Berezhany, Ukraine

⁵ Separate Structural Subdivision Berezhany Vocational College National University of Life and Environmental Sciences of Ukraine", 47501 Berezhany, Ukraine

(Received 28 February 2026; revised manuscript received 18 April 2026; published online 29 April 2026)

A study was conducted on the wear processes of nickel-based composite electrolytic coatings (CEC) under friction conditions without lubrication, taking into account changes in mechanical properties in the transition zone, the effect of diffusion chromium plating, and the dispersibility of the TiC carbide phase. The diffusion zone formed in the surface layer at the "coating-base" boundary ensured an increase in the adhesive strength of the CEC with the base and contributed to an improvement in the mechanical characteristics of the nickel matrix and the alloyed nickel matrix. The wear mechanisms were analyzed based on energy theory, according to which, in the contact zone, the energy of plastic deformation is converted into the energy of dislocations in the friction area, which causes the solid phase to crumble. The mechanism of destruction of the surface layers of the friction pair was evaluated based on the analysis of theoretical and experimental research data within the energy concept of the formation of wear particles in their near-surface zones. The influence of the mechanical properties of the material on the size of wear particles was shown. It has been established that the size of the spalling particles is determined by the elastic modulus of the substance, and their ability to penetrate the friction zone is determined by the specific cohesion energy.

Keywords: Composite electrolytic coatings, Friction, Wear particles, Transition zone, Wear resistance, Diffusion chromium plating, Structure.

DOI: [10.21272/jnep.18\(2\).02003](https://doi.org/10.21272/jnep.18(2).02003)

PACS numbers: 81.40.Pq, 62.20.Qp, 46.55.+d,
81.40.Gh, 61.66.Dk

1. INTRODUCTION AND STATEMENT OF THE PROBLEM

The relevance of the task of surface strengthening of structural materials is determined by the increase in the reliability and durability of machine parts, the service life of which is mainly determined by their resistance to the corresponding type of contact interaction. The effectiveness of solving the problem of contact destruction is directly related to the level of protection of the part surface [1-3].

Theoretical and experimental methods of studying the wear resistance of materials have shown that destruction is the final phase of a multi-stage complex damage process that develops at the atomic and mesoscopic levels much faster than the appearance of externally visible signs of destruction and is a continuous process of increasing the internal energy of the crystal lattice [4]. The strength of the surface friction layer depends primarily on the grain size of the phases in the friction zone and its structure [5]. Such

research results are particularly important for assessing the wear resistance of applied coatings.

Traditional coating technologies do not meet modern operational requirements for friction units in mechanisms operating under dynamic loads, high temperatures, exposure to corrosive environments, and abrasive wear. An effective solution that meets these requirements is wear-resistant heterogeneous coatings with a significant difference between the filler and the metal base in terms of hardness, plastic deformation capacity, and fracture behavior [6-9].

Given the large amount of information on the study of deformation processes in the surface friction layers of composite materials, their stress-strain state, tribotechnical and tribometric characteristics depending on the ratio of the physical and mechanical characteristics of their components, the specifics of defectiveness and discreteness of the process of damage accumulation in the transition zone between the filler and the matrix due to deformation effects remain insufficiently studied. Since intensive plastic deformation occurs in thin layers of the

* Correspondence e-mail: stebeletska@ukr.net



metal matrix during friction, the assessment of structural defects must be carried out taking into account point and linear defects (dislocations), which are important in the processes of metal strengthening and weakening, mass transfer, and wear.

The results of the analysis of the anti-friction characteristics of composite materials showed the importance of the size of filler inclusions and their volume content [10; 11]. The stress state that arises as a result of contact interaction with the mating surface also has a significant impact on the wear resistance of heterogeneous materials [12]. In [13], which was based on the calculation of the stress-strain state of composite materials with uniformly oriented columnar formations, a relationship was established between the stress concentration depending on the ratio of the mechanical properties of the matrix and inclusions, as well as the density of their location. Areas with increased stress levels, which are most prone to the initiation of destruction processes, were identified in the material structure. The work [14] shows that the stress at the onset of plastic deformation $\langle \sigma_1 \rangle$ of a composite material under friction can be qualitatively predicted, taking into account the nature of the components, their volume ratio, and the friction coefficient f , according to analytical studies of the stress-strain state using the formula:

$$\langle \sigma_1 \rangle = \frac{\sqrt{2}(\sigma_T)_m}{\sqrt{A+f^2B}}, \quad (1)$$

where $(\sigma_T)_m$ is the yield strength of the matrix material; f is the coefficient of conversion of normal forces into tangential forces during friction, which, with certain assumptions, can be considered the coefficient of friction; A , B are coefficients that depend on the mechanical properties and volume content of the components.

The relative minimum value of the average longitudinal compression stress $\langle \sigma_1 \rangle / \langle \sigma_T \rangle_m$, at which the process of plastic deformation begins in the material, can be taken as the plastic compression coefficient, which characterizes the degree of strengthening. It has been shown that at a volume content of filler $\xi < 10\%$ vol., strengthening is insignificant, $\langle \sigma_1 \rangle / \langle \sigma_T \rangle_m$ tends to 1. At a denser placement of $20 < \xi < 40\%$ vol., there is a significant limitation of plastic deformation. However, at $f > 0,3$ increasing the volume content of the strengthening phase to 40% vol. is ineffective, as it does not affect the average normal stresses at the onset of plastic deformation. Due to the decrease in the distance between particles at $\xi > 40\%$ vol., there is a significant local strengthening of the matrix, which reduces the strength in the weakened areas of the matrix between the particles due to the formation of cracks.

There is an assumption that using alloy phase diagrams, it is possible to predict the result of transition layer formation during diffusion interaction [15]. In this case, it is assumed that the layer on the diffusion isotherm is formed after an increase in the concentration of the diffusion element in the alloy system, and the process of its growth is governed by the laws of phase transformations. In this case, from the point of view of atomic diffusion theory, phases of lower

composition should appear first, then medium composition, and finally, higher composition. According to the theory of reaction diffusion, the formation of the diffusion layer can begin with the formation of phases of higher or medium composition without the stage of formation of the phase of lower composition under more favorable thermodynamic conditions. In both cases, knowledge of phase transformations in metals and alloys, according to the phase diagram of the systems under study, is crucial for understanding which phases can form in the transition zone, allowing us to predict the sequence of their occurrence and the qualitative composition as a result of the diffusion process.

Despite the absence of generalized laws in modern materials science linking phase equilibrium with the mechanical and physical properties of alloys, the results of experimental studies show that metal systems with similar phase diagrams are similar in structure and properties, and the direction of their change is determined by the corresponding phase equilibrium diagrams.

The assumption of a linear dependence hypothesis regarding the strength properties of a solid solution: they either decrease with a decrease in the concentration of the diffusing element, or exceed the properties of the matrix and filler material, remaining practically unchanged within the newly formed phase, makes it possible to simulate any stress-strain state of a composite material loaded with friction forces and containing a transition zone with mechanical properties that change according to a given law together with the filler in the matrix [13].

As an example, Fig. 1a shows the results of the influence of the law of change in mechanical properties in the transition zone on the stress state in the elementary volume of components at given average stresses of longitudinal compression $\langle \sigma_1 \rangle$ and longitudinal shear $\langle \sigma_{12} \rangle$, characteristic of the load of sliding friction forces of the material with inclusions located normally to the friction surface. The dependence of the relative stress intensity $\sigma_i / \langle \sigma_1 \rangle$ ($\sigma_{12} / \langle \sigma_1 \rangle$ at $f = 1$), at a point on the x_2 axis, on the nature of the change in mechanical characteristics in the transition zone is also illustrated (Fig. 1, b).

The numbered diagrams correspond to cases of longitudinal compression with shear, and the curves with hat symbols are for comparison of similar dependencies, valid only when the average longitudinal shear stresses $\langle \sigma_{12} \rangle$ are applied. All dependencies are constructed based on the assumption that the elastic modulus of the inclusion significantly exceeds the elastic modulus of the matrix:

$$E_f / E_m = 5 - 6 \quad (2)$$

Local stress fields significantly depend on the law of change in mechanical properties in the transition zone.

If there is no transition zone between the inclusion and the matrix, the stresses in the matrix are concentrated at the phase boundary, while in the fiber their concentration is maximum (curves 1 and 1'). The presence of a transition zone in the form of a solid solution of one of the components, which is characterized by a linear change in strength properties, contributes to a decrease in the concentration of

stresses in the fiber and ensures a uniform, smooth transition to the matrix to the level of its stresses (curves 2 and 2').

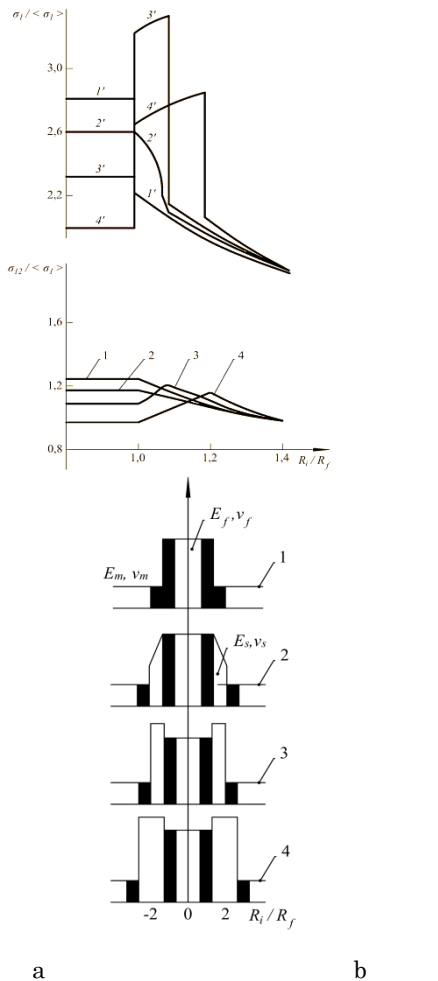


Fig. 1 – Distribution of relative stress intensity $\sigma_i, \langle \sigma_i \rangle$ in the transition zone at different distances from the surface of inclusion in the matrix volume (a) and depending on the nature of the change in the normal elastic modulus in the cross-section of the elementary volume (b): 1 – no transition zone; 2 – the normal elastic modulus changes linearly; 3 – mechanical properties change abruptly; 4 – the thickness of the abrupt zone of property change exceeds the previous one; 1'– 4' – the same, under friction loading [13]

Thus, local stresses in the composition under various types of loading by friction forces of an elementary volume can be predicted and calculated, taking into account the mechanical properties, type, and structure of the transition zone formed as a result of the diffusion interaction of the matrix and inclusion according to the phase diagram. The reduction in stress concentration at the “matrix-inclusion” boundary is achieved by the formation of a transition zone with enhanced physical and mechanical properties and a higher threshold for the onset of plastic deformation, as well as by reducing the intensity of strain hardening and dislocation density.

2. RESEARCH METHODS AND DISCUSSION OF THEIR RESULTS

The influence of the physical and mechanical characteristics of the composite material matrix and the dispersibility of inclusions on the dislocation density, deformation processes, and fracture kinetics under friction conditions was studied theoretically and experimentally on composite electrolytic coatings (CEC) that underwent chemical-thermal treatment (chromium plating).

Composite electrolytic coatings were obtained by co-deposition from an electrolyte of nickel and a strengthening phase of TiC of varying dispersibility onto samples measuring $10 \times 10 \times 5$ mm, which were obtained in a bath with a PH of 3 – 4, at a temperature of 25°C and a current density of 10 A/dm^2 .

Wear resistance tests were carried out on the M-22M installation in friction modes without lubrication at a load of 20N and a sliding speed of 0.5 m/s according to the coupling scheme – shaft-plane, and the friction path was 1 km.

Ni-TiC composite with various TiC fractions ranging from 5 to $100 \mu\text{m}$ was subjected to diffusion chromium plating in a powder mixture: 65% Cr, 25% Al_2O_3 , 1.5% NH_4Cl in metal containers with a fusible lock for 3–5 hours at a temperature of 1100°C .

Fig. 2a shows an image of the microstructure of Ni-TiC CEC with TiC filler with a dispersion of $30 \mu\text{m}$, where a uniform distribution of filler particles across the cross-section of the polish is observed. After diffusion chromium plating, the image of the CEC microstructure is shown in Fig. 2b. A low concentration of TiC particles is observed in the surface layer of the coating, while its carbide phase has accumulated slightly at depth and remained fairly evenly distributed across the micro-section, which did not lead to significant changes in the structure across the depth of the layer, except for the structure of the chromium-doped surface. The diffusion zone formed at the “coating-base” boundary provided increased adhesion strength of the CEC to the base and improved the mechanical properties of the nickel-doped nickel matrix.

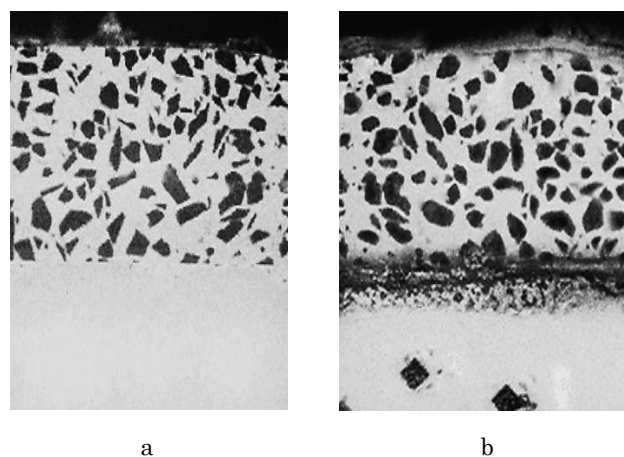


Fig. 2 – Microstructure of Ni-TiC CEC: a – with TiC filler $\times 300$; b – after diffusion chromium plating $\times 300$

Fig. 3 shows the dependence of the wear resistance of CEC with different TiC fractions at the same volume filling of the sediment, equal to 24% by volume, and after its diffusion chromium plating for 3 hours at a

temperature of 1100°C in Fig. 4. The research was carried out under sliding conditions at different loads and friction path lengths.

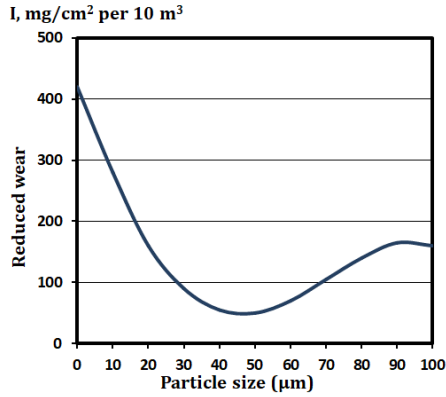


Fig. 3 – Dependence of the wear resistance of Ni-TiC CEC on the size of TiC particles at a load of 20N and a friction speed of 0.5 m/s. The volume content of the filler is 24% [4]

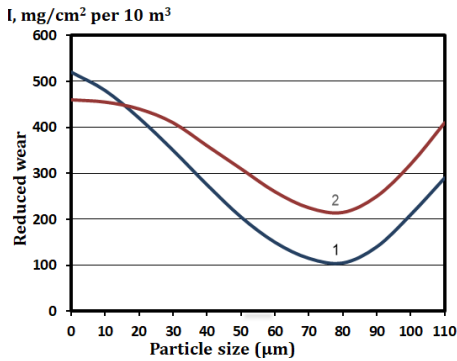


Fig. 4 – Dependence of wear resistance of Ni-TiC system after diffusion chromium plating under loads of 10N (1) and 20N (2)

Fig. 3 and Fig. 4 allowed us to conclude that the wear resistance of samples that were not subjected to chemical-thermal treatment and were subjected to diffusion chromium plating depends on the size of the filler particles, which differs in the threshold value of wear resistance reduction, at which its reduction occurs in the first case from a carbide inclusion size of about 50 μm, while in the second case – from 70 μm under a load of 20N and 80 μm under a load of 10N. This difference is due to the increase in the adhesion forces of the particles in the matrix, which underwent diffusion chromium plating, from the adhesion forces of the carbide phase in the second case, since in both cases elastic forces are present and, accordingly, the energy of plastic deformation is different.

Fig. 5 shows a diagram of a friction plane with a solid phase located in one of them. Individual particles are localized at the tops of micro-protrusions of irregularities due to the action of friction forces, and their retention in the matrix is determined by the interaction energy, which depends on the nature of the material and the size of the particles. The absence of plastic deformation in the contact zone, which is caused by friction forces, leads to the accumulation of elastic deformations and is characterized by the dissipation of part of the energy expended by the work of friction forces.

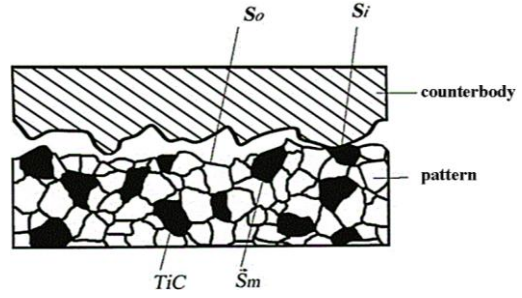


Fig. 5 – Schematic representation of TiC particle distribution in the surface layer of CEC

Since these processes are thermodynamically unbalanced, they can be described using a dissipative function [11]:

$$T \frac{dS}{dt} = \bar{\Psi} = \sum \frac{dW_{dis}^i}{dt} + \bar{I}_c A_c \quad (3)$$

where T – system temperature in °K; $\frac{dS_i}{dt} = \frac{dS_i}{S_k dt}$ – rate of change of system energy; $\frac{dW^i}{dt} = \frac{dW_{dis}^i}{S_m dt}$ – energy dissipation rate; S_i – temporary contact surface area; $\bar{I}_c A_c$ – influence of concomitant factors.

The dissipative functions of plastic deformation processes and the formation of free interaction surfaces of different phases, in which the dissipative forces of energy and the emergence of defects in the crystal structure – dislocations, preceded by plastic deformation – are concentrated, can be determined on the basis of the following considerations. Dislocations accumulate at phase boundaries, leading to the destruction of the forces of connection between them and, as a result, to the spalling of one of the phases – in this case, the TiC phase, which contributes to the occurrence of abrasive wear.

For clarity, let us introduce the concept of chemical potential of a dislocation as the work required to form a single dislocation:

$$\varphi_d = \frac{\tau}{\beta}, \quad A_d = \Delta\varphi_d = \frac{\Delta\tau}{\Delta\beta} \quad (4)$$

where φ_d is the chemical potential of the dislocation; β is the number of dislocations per unit volume of deformed material; τ is the stress acting in metals; A_d is the work required to form and move dislocations; $\Delta\varphi_d$ is the difference in chemical potentials for two stressed states of the material; $\Delta\tau$ is the strengthening of the material during its plastic deformation.

For its part, the amount of plastic deformation work expended on the occurrence of dislocation will be recorded as follows:

$$dW_{pl} = d\bar{D} A_d \quad (5)$$

where $d\bar{D} = \frac{dD}{S_i}$ is the rate of dislocation accumulation in the system.

Dividing both parts of equation (5) by dt , we obtain an expression for the dissipative function of plastic deformation of the material:

$$\varphi_{dis} = \frac{dW_{pl}}{dt} = \frac{d\bar{D}}{dt} A_d = \bar{I}_d A_d \quad (6)$$

where $\tilde{I}_d = \frac{d\bar{D}}{dt}$ is the dislocation accumulation rate calculated per unit of contact surface area and determined by the plastic deformation energy in the contact zone.

Considering the mechanism of destruction in the friction zone as the chipping off of particles at the boundaries of their volume as a result of dislocation accumulation along them, we write:

$$\sum \sigma_i dS_i = \sigma_m d\tilde{S}_m + \sigma_0 dS_0 - \sigma_d d\tilde{S}_d \quad (7)$$

where σ_0 – specific surface energy in the contact zone; dS_0 – specific contact surface area (plastic deformation zone); σ_m – specific surface energy when the contact area changes $d\tilde{S}_m$; σ_d and $d\tilde{S}_d$ – specific surface energy and change in the surface area of dislocation boundaries, respectively.

Thus, taking into account the dissipative energy of plastically deformed friction surfaces in the contact zone, determined by the rate of change of their area (friction rate), we obtained the expression for the dissipative function:

$$\Psi_s = \frac{d\tilde{S}_0}{dt} (-\Delta\sigma_0) + \frac{d\tilde{S}_m}{dt} (-\Delta\sigma_m) \quad (8)$$

$$\text{where } \Delta\sigma_0 = \sigma_0 - \frac{\sigma_0^0}{2}, \Delta\sigma_m = \sigma_m - \frac{\sigma_m^m}{2}.$$

The expressions $\frac{d\tilde{S}_0}{dt}$ and $\frac{d\tilde{S}_m}{dt}$ are indicators of the rate of change of dissipative energy of plastic deformation at contact areas during friction and depend on many factors, primarily on the load and speed of the friction process. This is confirmed by the works of many researchers [16], which show that friction energy is transformed into dislocation energy, leading to the brittleness of surface layers and chipping of individual phases in the friction zone.

The inclusion in the dissipative function expression of coefficients characterizing friction factors (load, speed, cyclicity, temperature, etc.), as well as the structural composition of the substance, makes it possible to assess the nature and degree of material wear during friction.

Based on the results obtained, an explanation of the extreme nature of the dependence of wear resistance on the size of filler particles in CEP is justified. It has been established that with an increase in filler particle size, dissipative energy at the grain boundaries is more rapidly transformed into elastic energy, which leads to crumbling – the loss of bond strength with the matrix and, as a result, its spalling. Subsequently, these carbides cause abrasive wear. The increase in wear resistance in the presence of carbide inclusions with a grain size up to certain values is probably due to the regularities of the friction pair working process and the undestroyed forces of grain bonding with the matrix due to the insufficient amount of elastic energy required for brittle delamination of the carbide grain.

Considering the dependence of the internal energy of the system on both interatomic interactions in the crystal lattice and the influence of external forces, in [17] it is substantiated that under the action of external mechanical loads in friction pairs, the internal energy of the crystal increases over time according to a

law that is close to linear. When the material reaches a certain critical value of internal energy (activation energy), it leads to its destruction. Thus, the destruction process is a consequence of the loss of stability of a certain volume of material caused by reaching a critical value of its internal energy density.

By accepting a physically infinitely small volume of a solid body as an open thermodynamic system, which under constant external conditions is in a state of local thermodynamic equilibrium, the condition for material destruction can be presented as follows:

$$\Delta u = u_0 \quad (9)$$

where Δu and u_0 are, respectively, the change in the molar internal energy of the system and its molar activation energy of destruction. The latter of these quantities is a fundamental energy characteristic of material strength. The former accumulates during the friction process. When it reaches a value equal to u_0 , the material is destroyed [16].

The change in the internal energy of the system is described by Gibbs' fundamental equation [17]:

$$\Delta u = T\Delta S - P\Delta V + \sum \mu_i \Delta w_i \quad (10)$$

where T is the absolute temperature, P is the pressure, V is the volume, S is the entropy, μ_i is the chemical potential of the i -th component, whose molar fraction in the system is equal to w_i .

In general, the internal energy of a system changes under the combined influence of various thermodynamic processes: mechanical, electromagnetic, thermal, diffusion, chemical, etc. Taking these factors into account, the condition for material destruction in the surface layer of friction can be written [17] as:

$$TR \ln\left(\frac{t}{\tau_0}\right) + V_m 10^{-6} \left(\frac{\Delta\sigma^2}{2E} + \sigma \bar{\epsilon} t\right) + \Delta u_n + \Delta gB \geq u_0 \quad (11)$$

where R is the universal gas constant, V_m is the molar volume, σ is the stress, E is the elastic modulus, $\bar{\epsilon}$ is the average rate of inelastic deformation, t is the time the system remains in a loaded state, and τ_0 is the period of thermal oscillations of atoms.

In expression (11), the first term determines the increase in the internal energy of one mole of the surface layer substance due to an increase in entropy, the second term determines the increase due to an increase in the energy of elastic and plastic deformations of the crystal lattice (correspondingly, the first and second terms of the expression in parentheses), the third is due to technological processing of the surface; the fourth is due to physicochemical interactions with the environment.

Research conducted in [16] on the influence of each of the terms on the left side of inequality (11) showed that the main factor in the process of material destruction is the deformation component, the value of which increases monotonically over time in accordance with a practically linear law.

Since the energy of elastic deformations is a function of the instantaneous stress value, it does not accumulate in the material over time. To estimate it in the presence of dynamic influences, such as periodic loads, the

maximum stress value σ_{max} is taken as the effective value $\Delta\sigma$.

During friction and wear, the act of destruction is localized in a small volume of material, which, upon reaching critical stresses, flakes off and is transferred to the surface of the counterbody in the form of transfer particles. Their accumulation on the surface of the counterbody due to adhesive transfer leads to the formation of wear particles, which subsequently act as an abrasive, transforming the sliding friction mechanism into low-abrasive wear, increasing its intensity. Critical stresses are determined by the conditions of destruction. Thus, the formation of transfer particles occurs when an inequality similar to (11) is satisfied in the volume of the exfoliated fragment.

To estimate the size of the transport particle, it is assumed that it is limited by a hemisphere with radius $d/2$, where d is the diameter of the contact spot [18]. Therefore, its volume is taken to be equal to $\pi d^3/12$. If ν is the number of moles of substance in a given volume, then the condition for the formation of a transport particle is the fulfillment of the inequality:

$$E_{def} \geq E_s \quad (12)$$

where $E_{def} = \frac{\pi d^3}{12} 10^{-6} \left(\frac{\Delta\sigma_{max}^2}{2E} + \sigma_{max} \bar{\epsilon} t \right)$ is the deformation energy, $E_s = \nu \left(u_0 - TR \ln \left(\frac{t}{\tau_0} \right) - \Delta u_n - \Delta g \right)$ is the activation energy of destruction (crushing of a fragment of a given volume). The value of the latter is determined both by the cohesive properties of the material (the first term) and by the change in its internal energy due to an increase in entropy, technological processing of the surface, and its physicochemical interaction with the environment.

Given that the formation of new surfaces is accompanied by particle spalling, it is logical to assume that the activation energy of destruction is proportional to the surface area of the transport particle $S = \pi d^2/2$. The coefficient of proportionality is determined by the cohesive characteristics of the material in the near-surface zone:

$$E_s = 2\gamma \frac{\pi d^2}{2} \quad (13)$$

where γ is the specific cohesion energy. Assuming $\sigma_{max} = E \epsilon_{max}$, where ϵ_{max} is the maximum deformation, we conclude that the deformation energy is proportional to the modulus of elasticity E :

$$E_{def} = \frac{\pi d^3}{12} 10^{-6} \left(\frac{\epsilon_{max}}{2} + \bar{\epsilon} t \right) \epsilon_{max} E \quad (14)$$

Therefore

$$\frac{E_{def}}{E_s} \sim \frac{E}{\gamma} d \quad (15)$$

Fig. 6 illustrates the dependence of the size of transport particles on the physical and mechanical characteristics of the material.

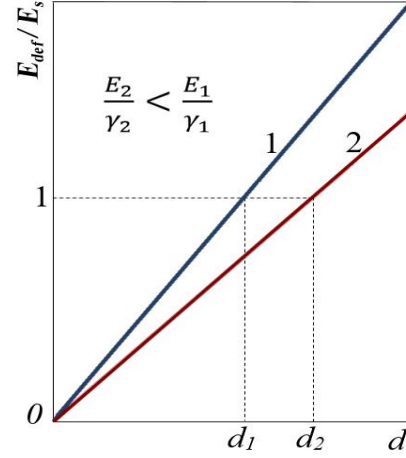


Fig. 6 – Dependence of the transport particle size (d) on the physical and mechanical characteristics of the material (elastic modulus E and specific cohesion energy γ)

It has been established that the size of the particles of abrasion is influenced by the elastic modulus of the substance, and its penetration into the friction zone is influenced by the specific cohesive energy.

3. CONCLUSIONS

The wear resistance of the Ni-TiC system CEC is primarily determined by local stress fields, which are significantly influenced by the law of change in the mechanical properties of the transition zone between the matrix and the filler. Diffusion chromium plating promotes the formation of a transition zone in the form of a solid solution, which reduces the difference between the mechanical properties of the filler and the matrix and increases the yield strength of the material. In the filler, the stress concentration decreases and is smoothly redistributed into the matrix, monotonically decreasing to the stress level in its volume. This reduces the likelihood of carbide phase chipping, which is caused by dissipative energy arising from plastic deformation in the friction zone, transforming into elastic energy and localizing at the phase boundaries, which leads to an increase in brittle carbide grain chipping.

The conducted analysis showed that under the condition of friction, the destruction of the material occurs faster for larger transfer particles. In addition, the size of the transfer particles is larger in the material with higher strength (large values of γ), and in the material with higher hardness (large values of E), the size of the transfer particles is smaller.

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Вплив будови і властивостей перехідної зони між матрицею і наповнювачем на руйнування композиційних покриттів при терті

М.В. Кіндрачук¹, Ю.І. Богомол², Т.С. Черепова³, Н.М. Стебелецька⁴, В.В. Харченко¹, А.Л. Гловин⁵

¹ Національний університет «Київський авіаційний інститут», 03058 Київ, Україна

² Національний технічний університет України «Київський політехнічний інститут імені Ігоря Сікорського», 03056 Київ, Україна

³ Інститут металофізики ім. Курдюмова Національної академії наук України, 03142 Київ, Україна

⁴ Відокремлений підрозділ Національного університету біоресурсів і природокористування України «Бережанський агротехнічний інститут», 47501, Бережани, Україна

⁵ Відокремлений структурний підрозділ «Бережанський фаховий коледж Національного університету біоресурсів і природокористування України», 47501 Бережани, Україна

Проведено дослідження щодо процесів зношування композиційних електролітичних покриттів (КЕП), основою яких є нікель, в умовах тертя без мащення з урахуванням зміни механічних властивостей у перехідній зоні, впливу дифузійного хромовання й дисперсності карбідної фази TiC. Сформована в поверхневому шарі на межі «покриття-основа» дифузійна зона забезпечила зростання адгезійної міцності КЕП з основою та сприяла підвищенню механічних характеристик нікелевої матриці й легованої нікелевої матриці. Механізми зношування проаналізовано, виходячи з енергетичної теорії, згідно якої в зоні контакту енергія пластичної деформації перетворюється в енергію дислокацій в області тертя, що є причиною викришування твердої фази. Виконано оцінку механізму руйнування поверхневих шарів пари тертя на основі аналізу теоретичних та експериментальних даних дослідження у межах енергетичної концепції формування частинок зносу у їх приповерхневих зонах. Показано вплив механічних властивостей матеріалу на розмір частинок зносу. Встановлено, що розмір частинок викришування визначається модулем пружності речовини, а їх здатність до проникнення в зону тертя обумовлюється питомою енергією когезії.

Ключові слова: Композиційні електролітичні покриття, Тертя, Частинки зносу, Перехідна зона, Зносостійкість, Дифузійне хромовання, Структура.