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MILD-TO-SEVERE WEAR TRANSITION AND PLASTIC STRAIN LOCALIZATION

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ABSTRACT
Subsurface strain front propagation and localization in sliding wear has been investigated using laser speckle decorrelation technique. It was found that strain localization leads to intense plastic deformation and generation of nanocrystalline material, which is a precursor of mild-to-severe wear mechanism transition. SEM images of both worn and lateral surfaces have been analyzed using a procedure for calculating fractal dimension. Size scales of self-similar behavior have been found to characterize the nanocrystalline material deformation. Generation of large-scale nanocrystalline layer by severe sliding-induced deformation manifests itself as a new structural level of deformation, which may lead to catastrophic wear unless antiwear films are formed.
KEYWORDS:
speckle decorrelation, strain localization, nanostructuring, sliding, fractal dimension, scaling, wear

1. INTRODUCTION

It is well–known fact that crystal defects are first generated in the subsurface layer of metallic bodies during tensile deformation. A peculiarity of sliding wear tests is that all deformation is concentrated within such a layer and limiting deformation state is achieved here first. Subsurface deformation-induced nanostructures are real and well-known phenomenon in friction and wear generated due to enormous plastic strains and mixing [1]. One may say that people who studied friction and wear degradation of materials have always been involved in studying nanocrystalline microstructures. Thickness of such a nanocrystalline layer might be of negligible scale under definite loading conditions and thus would not interfere with the dominating wear mechanism. However, this would not be the case in transition from normal to catastrophic wear mode. The mechanism of nanocrystalline layer formation has been a subject of investigation for several decades and often related to quasistatic plastic strain accumulation, material transfer and mixing [2], effect of propagating elastic stress waves [3], and adiabatic shear [4]. In most cases, the nanocrystalline layer is not continuous but consists of contact spots where the real contacting between the counterbodies is performed. It is reasonable to stress that the deformation mechanism, thickness and real contact area of highly active nanocrystalline material are critical parameters for normal/catastrophic wear transition. In another words, scale transition determines wear mechanism transition. It is important for understanding normal and catastrophic wear of materials that both of them are determined by the scale level of adhesion interaction.

Sliding test sample/counterbody couple has several dimension scales amongst which three are the most important ones. Macroscopic scale *L* matches to the sample size, scale *l* to microstructure element, and λ is a crystalline structure scale. Strain localization and generation of nanocrystalline material at both microstructure and crystalline structure scales will not be critical for wear and friction, whereas macroscale localization will cause deterioration of load bearing ability of the whole sample. The objective of this paper is to gain understanding of strain localization, generation of nanocrystalline material and scale changes in transition from normal to catastrophic wear. Physical phenomenon responsible for scale increase is intensification of adhesion between counterbodies, which results in coarsening the wear debris. Adhesion is a material characteristic, which is inherent to surfaces at all possible size scales and therefore it may produce deformation structures of different size scales. Since these structures are obtained by severe deformation in the subsurface layer, they structurally consist of small fragments, size of which is smallest possible under given condition. Larger wear particles may consist of these units and look like fractal objects. Determining characteristic correlation function scales, which correspond to fractal areas, we may find a relation between the deformed area scale and the wear mode.

2. METHODOLOGY

Wear mechanism transition are usually observed by stopping the test and examining the worn surfaces. Meanwhile it is very important to observe the precursor stage, which determines how this transition is prepared and what happens to strain during the sliding. In connection with this, we used a method of laser speckle decorrelation [5] when sliding wear contact is modeled and deformation patterns are visualized on the lateral side of immobile specimen (Fig.1, *a*). Method of computational decorrelation involves illumination of the deformed surface by coherent light and successive performance of a set of operations (formation of a video frame of the surface, its digitalization and storing in a computer, computation of the map of sample coefficients of mutual decorrelation (SCMD) with another video frame shifted by a software-specified time interval, and finally, displaying the zones of strain localization by this map or digital representation of the SCMD in a file for subsequent representation of this strain by other technical means) [5].

The decorrelation is computed using the following formula [6, 7]:

$$D_{ijk}^{(m\times n,p)} = 1 - \left| \left\langle \left(S_{ijk} - < S_{ijk} >\right) / \sigma_{S_{ijk}} \times \left(S_{ij(k-p)} - < S_{ij(k-p)} >\right) / \sigma_{S_{ij(k-p)}} \right\rangle \right| \qquad (1)$$

where S_{ijk} , $S_{ij(k-p)}$ -) are the levels of the video signal in the given pixel of the video image of the first and second frames, respectively; $< S_{ijk} >$, $< S_{ij(k-p)} >$ are the levels of the video signal averaged over the correlation window; and $\sigma_{S_{ijk}}$, $\sigma_{S_{ij(k-p)}}$ -are standard deviations for the pixels characterized by levels S_{ijk} , $S_{ij(k-p)}$.

Since the intensities of speckles at the surface are changed as a function of both time and space during deformation it becomes possible to visualize the surface deformation pattern using video camera. Typically, the strain localization zones look like dark spots (bands), which either stay or travel across the field of vision. However, it is not easy both to encompass the whole picture of traveling zones and reveal quantitative parameters from this simple visualization. Therefore, we use specially developed software, which allows us to observe strain dynamics in a single pixel-wide line successively plotted at 1 s time interval in the form of time-scan.

Also we made an attempt to characterize worn surfaces by processing numerically the SEM images and determining their fractal properties. The fractal dimension d_{SEM} is determined from the slope of correlation function logarithm Lg(⟨ | J-J' | ⟩) vs. Lg(x), where correlation function [10] is determined as

$$\left\langle |J - J'| \right\rangle = \frac{\sum_i |J(x_i) + J(x_i + \Delta x)|}{N(\Delta x)} \sim \Delta x^{H} \qquad (2)$$

$J(x_i)$ intensity of SEM signal in point x_i, $J(x_i+\Delta x)$ intensity of SEM signal in point $x_i+\Delta x$, $N(\Delta x)$ – number of point pairs; Δx is a distance between points; H is the Hurst exponent.

The physical meaning of fractal dimension with our method is a characteristic of correlation between the pixel intensities *I*, and thus, it is a characteristic of correlation between the values of the scanned surface peak heights as a function of the distance between them. Appearance of the linear portions at the logarithmic plots indicates the scaling behavior at a definite scale level. Abscissa points, at which slope is changed correspond to the deformation scale limit, where scaling behavior is changed. Such a work has been carried out by computer treatment of SEM images of worn and lateral surfaces. It should be noted that fractal dimension itself was not of special interest. More interesting was length scales $l(x)$, which denoted the beginning and end of the scaling behavior intervals.

In this research we used samples machined of stainless alloy 36NkhTU, ferritic steel 15N3MA (Table), brass L70 (25% Zn, 74%Cu, Si+Fe 1%).

Table 1

Sample	Alloying elements, % wt.								
	C	Ni	Mo	Cr	Cu	Mn	Si	Ti	Al
15N3MA	0.13-0.18	2.75-3.15	0.2-0.3	0.30	0.2	-	-		-
36NKhTU	≤0.05	34.5-36.5	-	11.5-13.5	-	0.8-1.2	≤0.5	2.8-3.2	0.9-1.2

Wear tests have been carried out using 2169 UMT-1 tester in dry friction conditions. The counterbody disk was made of chromium-molybdenum 56HRc steel. Worn surfaces were characterized using TEM, optical microscopy and AFM.

3. RESULTS

During the test, we observed displacement of strain zones presented in Fig. 1, *c* in the form of a time-scan in A-A section. Three different types of strain fronts have been discovered. The horizontal bands (type I) correspond to stationary localized zones and the vertical bands (type II) correspond to the zones traveling in the A-A section with the slider velocity.

The sloped potions of dark bands (type III) correspond to their deceleration and strain localization. It is interesting that localized strain areas of types I and III may start travel again with the sliding velocity, i.e. became type II zones. This moment of time corresponds to fast deformation front propagated from strain localization zone.

Using the AFM SolverPro 47H instrument, we examined subsurface microstructure of test material (brass) resulted from the fast front propagation. Fig. 2. shows nanocrystalline grains generated by this front in the vicinity of sliding surface. The localized strain zone could be identified as a precursor for formation of new sliding wear participant – nanocrystalline material. Since deformation in sliding is concentrated in the subsurface layers, strain localization zones found on the lateral surface should be the result of deformation on real contact areas in the plane of sliding. We assume here that the worn surfaces produced by deformation under adhesion wear possess scaling behavior and therefore real contact areas should be distinguished from the neighboring undeformed areas.

The types of topography forming during friction and wear at the sliding surface and at the lateral surfaces vary in pattern with regard to testing conditions. The quantitative characteristics of the wear surface topography may be obtained from estimations of its fractal properties. Scanning electron microscopy (SEM) is one of the easiest ways to do it, however, it is necessary to take into account the peculiarity of SEM image formation by secondary electrons. We made an attempt here to evaluate the fractal properties of the worn and lateral surfaces with the use of SEM images obtained from the samples made from low carbon alloy steel 15N3MA. Prior to the wear tests the surfaces of samples have been polished. The lateral polished surface of the samples is of interest because it may give information on 3D behavior of the fragmented layer.

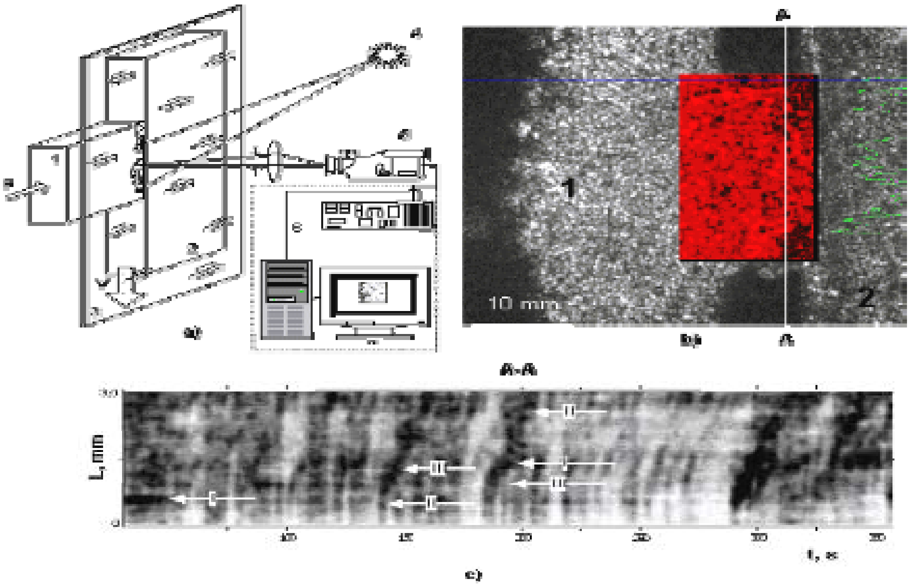


FIGURE 1. A) SCHEMATIC REPRESENTATION OF THE STRAIN VISUALIZATION METHOD IN SLIDING: 1) immobile specimen; 2) mobile specimen (slider); 3) glass; 4) source of light; 5) video camera; 6) interface; b) example of averaged distribution of decorrelation coefficient over a side surface of immobile specimen (the sliding velocity is 5 mm/min); the A-A line shows the section in which the chronogram of Figure 1c has been obtained; c) time scan (chronogram) of the distribution of processed digital speckle images in a longitudinal section (see Figure 2) over the direction of the sliding (brass specimen, sliding velocity is 5mm/min; the coverage interval is 1 s).

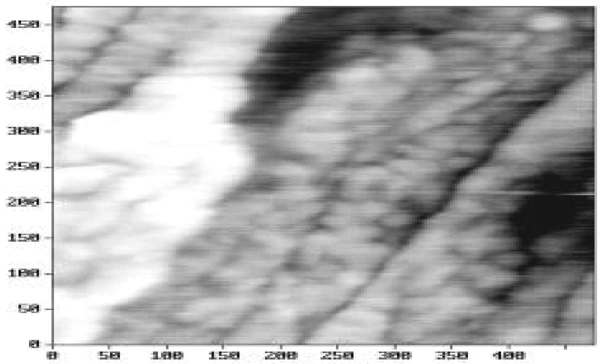


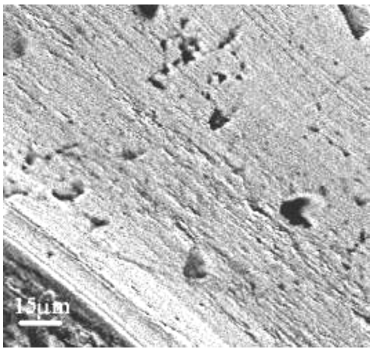
FIGURE 2. AFAM IMAGE OF NANOCRYSTALLINE MATERIAL GENERATED BY PLASTIC DEFORMATION IN SLIDING. SAMPLE SIDE PARALLEL TO THE SLIDING DIRECTION

The results of TEM examination of sliding deformed material are another confirmation of the AFM results with respect to nanocrystalline structures formed in subsurface layers in sliding [8]. In reality, the initial metal microstructure is transformed into ultrafine polycrystalline one by deformation. It is also has been shown that the wear debris have their both size and misorientations the same as those of fragments found in the severely deformed layer. It is worthwhile to note that such ultrafine structures are the results of very high strains and recovery processes. All materials tested show the fragment's size depends only on physical-mechanical properties of materials. As a rule, further plastic deformation of such structures does not result in smaller fragment size what enables a conclusion on a limiting deformation state achieved. The subsurface layer composed of such fragments shows new deformation mechanism, which allows enormous strain without fracturing. This mechanism is very much alike the mechanism developed in granulated media under deformation.

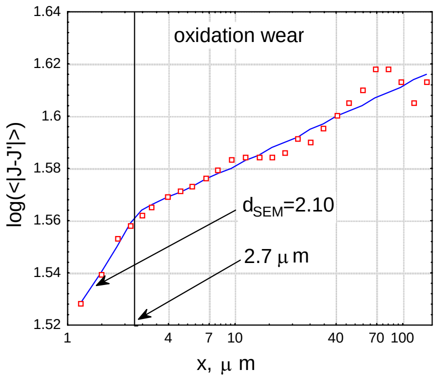
We made an assumption that the mass transfer in subsurface layers of metallic materials in sliding is by displacement of mesovolumes (fragments), which may be called the elementary plastic strain carriers. An argument in favor of this conclusion is the sliding-induced changes in topology of previously polished side, which is parallel to the sliding direction [9]. The topology has been examined using a method of determining the fractal dimension. The results of these studies give evidence of self-similar behavior, which is the result of self-consistent displacements of elementary structure units. It is natural that 10 to 100 nm fragments found in severely deformed layer and shown above may serve as such units.

Fig.3 shows the sliding speed and load dependencies of the coefficient of friction, temperature and wear intensity for 15N3MA steel samples. These dependencies are of complicated topology in the coordinate system. Let us note that variations in temperature and wear intensity reveal practically similar behavior. We may suggest that a dominating role here is played by only one factor; apparently it is plastic deformation that is responsible for formation of wear debris and temperature variations. At the same time, the coefficient of friction demonstrates other type of behavior with sliding speed and load. The decrease in friction that is observed at some values of load and speed may occur due to changes in the character of plastic deformation.

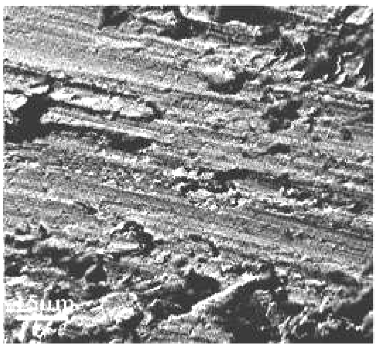
Metallographic SEM studies showed a correlation between the type of surface topology of the samples and speed-load dependence of the coefficient of friction. The sliding surface structures for different friction regimes are illustrated in Fig.3. The micrograph in Fig.3, *a* corresponds to a friction regime when low-temperature wear is by oxidation. Fig.3, *c* corresponds to the intense adhesive wear that is accompanied by mixing and transfer of metal in the friction zone. In Fig.3, *e*, lastly, shown is the surface of metal after high-temperature friction when there occurred a fall in friction due to high plasticity of the heated metal. Wear by flow (flow wear) is the case under these conditions. Visually, the roughest surface corresponds to the adhesive wear (Fig.3, *c*), whereas smoother surface without adhesion traces is obtained after high-temperature friction test (Fig.3, *e*). The wear dependence demonstrates how wear rate is changed with load and speed (Fig.4).



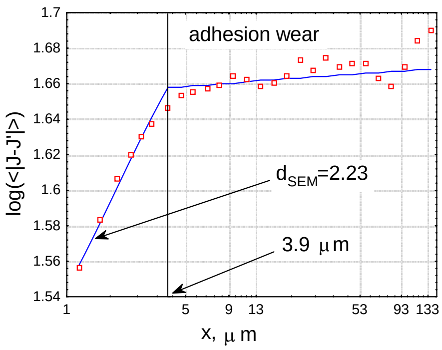
a) MILD (OXIDATION) WEAR



b)



c) SEVERE (ADHESION) WEAR



d)

FIGURE 3. WEAR MECHANISMS AND CORRELATION FUNCTION PLOTS SHOWING SELF-SIMILAR BEHAVIOR AND DEFORMATION SCALE LEVELS. SAMPLES WERE MADE OF 15N3MA STEEL.

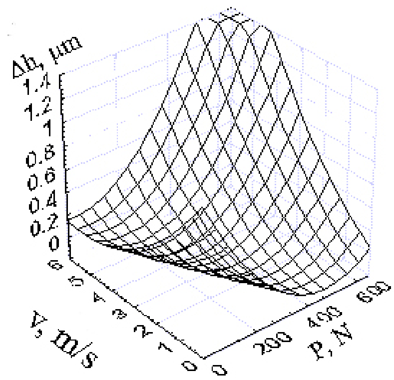


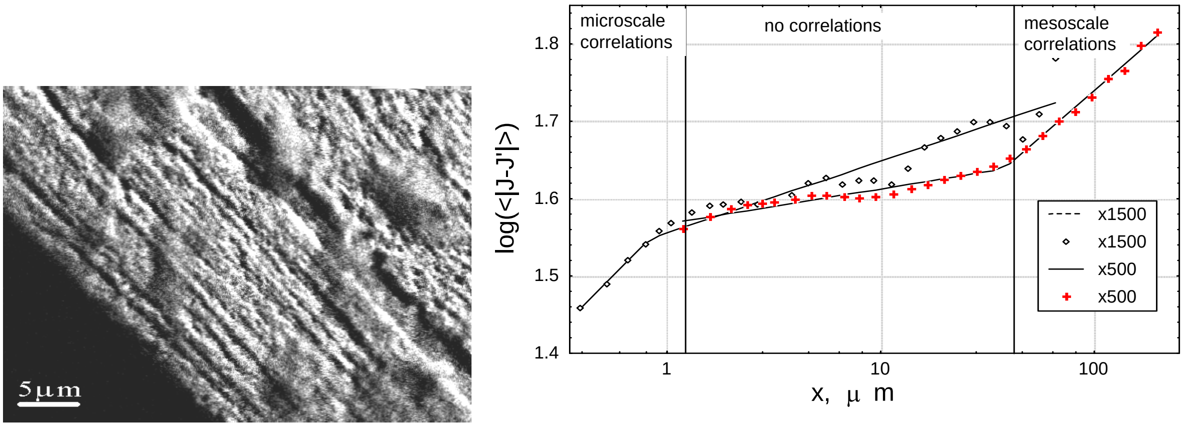
FIGURE 4. WEAR RATE (μm/min) vs. LOAD (N) AND SLIDING SPEED (M/S) FOR 15N3MA STEEL SAMPLE

Treating SEM wear surface images in Fig.3 *a,c,e* according to the above outlined procedure allowed us to obtain the results shown in Fig.3 *b,d,f*. The calculated data points are approximated by the lines, the slope of which to the **x**-axis gave us the value of the SEM image fractal dimension for a given scale of the SEM image. All plots clearly show the regions where scaling behavior is the case. The data points do not fit in the dependence that could characterize self-similarity of the surface above some critical value of the ***x***-coordinate. This means that the self-similarity may only exist within the limited zones of the surface and deformation in friction occurs within the local zones adjacent to the contact spots during some arbitrarily taken moment of time. The other sliding surface zones do not undergo deformation just now but may be deformed during the next moment.

With time the whole subsurface layer may be involved in deformation. Furthermore, the range where the scaling behavior may exist allows us to evaluate the dominating structure level involved in deformation by friction. In our case, an elementary structural unit, which is responsible for formation of the self-similar topology and deformation, is a crystalline structure fragment of 0.01 μm in size. The lower limit of the scaling behavior provides some support to this assumption. The position of the higher limit which is a dividing line between scaling behavior and the non-scaling one probably will depend upon the external factors such as load and sliding speed and will be determined by the size of the deformed zone adjacent to the contact spot.

The fractal characteristics (dimension and scale) reveal some correlation with the dependence of the coefficient of friction, whereas the wear intensity cannot be characterized by the SEM fractal behavior of the wear surface under these conditions, since wear mechanism varies from oxidative to adhesive and further to flow wear. Within the dominating of the only one wear mechanism, one can suggest the relatively constant topology of the wear surface and its characteristics. The oxidation wear does not result in formation of a thick enough fragmented subsurface layer. The thickness of the fragmented (fractal) layer is also negligible as compared to the surface size. The transition to the adhesive wear is accompanied not only by the sharp increase in the thickness of that layer up to 20-40 μm but also by intensification of metal transfer, diffusion, oxidation recovery and recrystallization processes at elevated temperatures. The ability of the layer to dissipate the friction energy by formation of the juvenile surfaces (fracture) should be defined by its fractal properties. The increase in the fractal SEM dimension of the wear surface may characterize the degree of the fracture energy dissipation, much alike it is done in case of the fracture surface analysis. The variations in the scaling correlation of the surface zones correspond then to the variations in structural level of the dominating friction energy dissipation process.

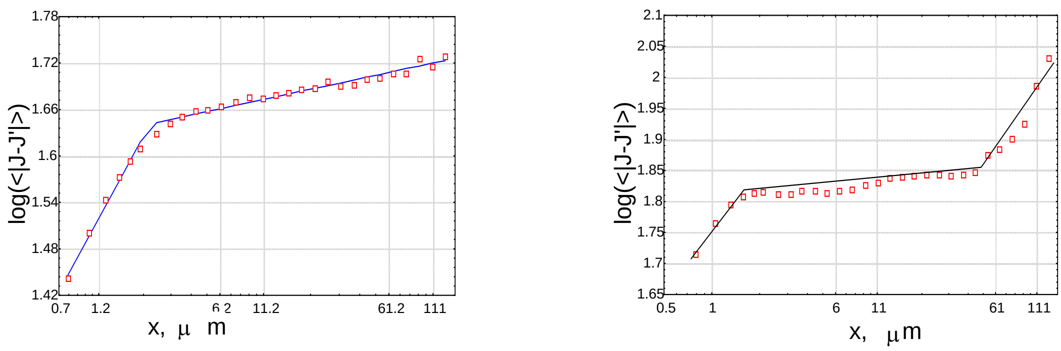
Fig.5 represents the SEM micrograph of the subsurface fragmented layer (lateral surface) and plotted results of SEM fractal dimension calculation obtained with different magnifications. Correlating experimental data points with the approximating lines calculated for different magnifications, one can see that there are two scaling ranges with the exponentially correlating pixel intensities. The scaling behavior of the fragmented layer lateral surface (magnification 1500) is observed for 1 μm and smaller zones (see the above discussion). The scaling behavior is not found within the range of 1 to 40 μm. Starting from 40 μm and higher, the correlation of the pixel intensities is again the case (both curves). This can only mean that the self-similar topology has been formed in the fragmented layer with the minimal structural unit of 40 μm. The coincidence between this value and thickness of the fragmented layer may prove not to be pure occasional. Apparently, this scale value is related to the development of deformation at higher mesoscopic level, which corresponds to the flow of the whole layer.



a) b)

FIGURE 5. SEM IMAGE OF NANOCRYSTALLINE LAYER (A) ON THE LATERAL SURFACE 15N3MA SAMPLE AND CORRELATION FUNCTION PLOT (B) SHOWING NEW MESOSCALE DEFORMATION LEVEL

Correlation function plot of a 36% nickel stainless specimen are shown in (Fig.6, *a*), which shows scaling behavior only within scale length of 0 to 4 micrometers. Fig. 6, *b* corresponds to formation of high scale nanocrystalline layer (40 to 50 micrometers) and correlation function plot also shows scale length L_2 along with length L_1. Since worn surfaces consist of real contact areas one may say that these areas demonstrate topology with scaling behavior and again these contact areas are nanostructured. Also brass specimens demonstrate reduction of scale during so-called "selective transfer" when smooth anti-wear films form on the surfaces due to preferential oxidation of zinc. However, it is only a special case of general mechanochemical wear and changing the load and sliding velocity one can transfer the system back to oxidation and even to catastrophic wear.



a) b)

FIGURE 6. CORRELATION FUNCTION PLOTS CORRESPONDING TO SMALL-SCALE AND LARGE-SCALE SLIDING-INDUCED DEFORMATION IN SUBSURFACE LAYERS OF STAINLESS ALLOY 36NKHTU.

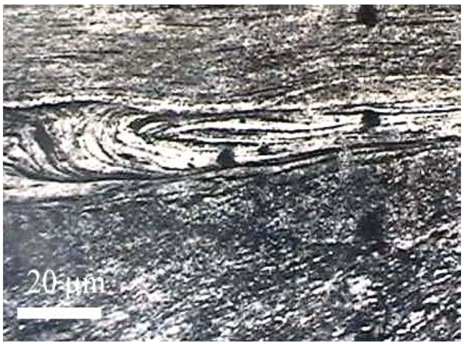


FIGURE 7. VORTICAL DEFORMATION MODE AT THE INTERFACE BETWEEN NANOCRYSTALLINE LAYER (TOP) AND PLASTICALLY DEFORMED ZONE

The above results and discussions serve to confirm an opinion that high strains and mass transfer on the worn surfaces are due to rotational mode of deformation with rotation of structure fragments around an axis normal to the sliding direction and parallel to the plane of sliding. Vertical mode of deformation induced by sliding is shown in Fig. 7 by the example of electrolytic copper. Such a mechanism is usually described in the theory of granulated structures to provide the high strains observed.

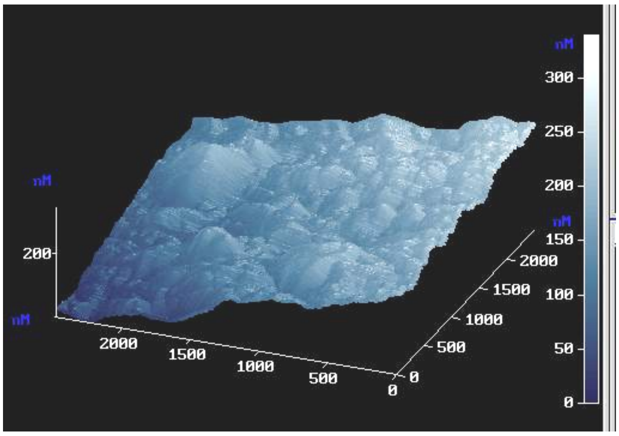


FIGURE 8. ANTIWEAR MICROCOMPOSITE FILM OF ZINC AND IRON OXIDES IN COPPER MATRIX ON THE WORN SURFACE OF MEDIUM CARBON STEEL AFTER SLIDING WEAR TEST IN MINERAL OIL ADDED WITH BRASS NANOPARTICLES

4. DISCUSSION

Formation of nanocrystalline material during severe sliding wear may be described in a following manner. Strain localization gives us a clue to understanding of deformation in subsurface layers. The first stage (running-in) is characterized by small contact areas at the sliding surface and strain fronts free traveling at sliding speed in the subsurface layers. At the second stage (normal wear), contact areas grow in size and strain fronts decelerate and even stop to result in strain localization. Since the sliding is not stopped, strain localization zone may start travel again as a result of plastic shear. The final stage of this process is formation of severely deformed material, a nanocrystalline layer, which will be new participant of sliding and a precursor of wear mechanism transition from normal to catastrophic wear, which determined both by thickness of this layer (scale) and its ability for form oxide films. The presence of microscale nanocrystalline structures is not critical since self-consisted rotational deformation is limited to the micrometer depth and the same size of contact spots. However, when the mesoscale level is achieved, i.e. contact spots overlap and nanocrystalline layer thickness is in order of tens of micrometers, it is the beginning of catastrophic wear. There should be a threshold behind which the scale will grow in irreversible manner and finally spread for nominal contact area. Morphology of this layer allows suggesting action of deformation mechanisms different of usual plastic flow. Actually, such a mechanism is well known as the grain boundary slip for superplastic deformation of nanocrystalline materials [11]. Length scale L_2 shows that new participant of sliding wear has been formed and its structure is self-similar at the given scale interval.

It is reported [12] that minimum size wear particle may exist in the system being the minimum possible structure scale from which larger wear particles may consist of. In our case that may be 10 to 100 nm fragment. Since resolution of SEM is not enough to see the fragments that small we cannot take them into account directly. Nevertheless, since the wear particles (as well as real contact areas) do really consist of these fragments we may consider them as fragment clusters and their deformation as a collective mode displacement of the fragments.

Second self-similar behavior scale is not found on the worn surfaces and it is because of intermixing deformed material. The lateral surface of the specimen is not subject to intermixing and therefore new scale level may be clearly seen.

The scale level growth means that larger volumes of subsurface metal are involved in deformation and wear. However, chemical composition and structural state of materials, environment determines further development of deformation and wear. Materials such as stainless steels and copper have no a mechanism which might stabilize or even reduce the deformation scale level and therefore the catastrophic wear is developed resulting in full seizure of sliding wear couple. Carbon steels are more active with respect to oxidation and once generated the severely deformed (nanocrystalline) material may form oxidation films, which are usually only few micrometers thick but provide good wear resistance. Physical meaning of this process may be expressed as minimizing the scale level of deformation within these films. The same process occurs at the surface of brass specimen with the only distinction that zinc is oxidized in the first case forming a composite with copper matrix reinforced by zinc oxide particles [13]. The scale level reduction may be achieved by a phase transformation as observed in testing titanium nickel alloys [14] or yttrium oxide stabilized zirconia [15].

Another route is to stabilize subsurface structure against sliding induced deformation. This could be achieved, for example, using hard coating or hard metal. The boronized coating may however form boron oxide during sliding wear, which also contributes to scale level reduction [17].

Sliding wear generated microcomposites like those found on brass specimens after sliding wear tests offer a way to reduce both wear and friction of steel specimens [13]. To obtain the microcomposite structures, we added nanosize brass powders to mineral oil and then tested medium carbon steel specimens for sliding wear in the suspension. The result was very smooth gradient nanocrystalline composite films consisting of zinc oxide, copper and iron oxide (Fig.8). Wear rate was order of magnitude lower as compared to specimens lubricated by pure oil. This film served to localize deformation within a micrometer-thick layer because of its low shear resistance during friction.

5. CONCLUSIONS

It was shown directly that strain fronts propagate in subsurface layers of metallic materials during sliding. The normal wear stage is characterized by fronts traveling at sliding speed, whereas strain localization caused intense plastic deformation in one place with formation of nanocrystalline layer in the vicinity of sliding surface. Structural units of this layer are 10 to 100 nm fragments and further deformation of this layer is by cooperative rotation of its structural units so that this layer moves as a whole formation to change the conditions of sliding. Thickness (deformation scale) of such a layer depends on the material characteristics and loading conditions. This stage is determined as a transition from normal to catastrophic wear for materials, which cannot form friction-reducing films such as stainless steels.

Nanocrystalline formations both at the worn and lateral surfaces show self-similar behavior at some length scales, which are the result of their self-consistent deformation in contact spots and layers. Catastrophic wear is characterized by nanocrystalline layer thickness (scale) in the order of 40-60 μm.

There is a correlation between behavior of strain fronts in subsurface layers and real contact areas of metallic materials during sliding wear. Since catastrophic wear by adhesion is just deformation at high scale level then reducing scale level will result in returning back to mild wear mode when adhesion and deformation are localized in micrometer-thick subsurface layer. There are two ways to do it. The first one is to stabilize subsurface material with respect to the friction-induced structural degradation while the second is, on the contrary, to make it softer as much as possible.

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