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FORMATION OF THE MICROSTRUCTURE OF A COMPOSITE DURING ITS DYNAMIC LOADING

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It was shown in [1] that the use in hard alloys of a structurally unstable binding phase permits a significant improvement of their toughness and ductility while the strength level under quasi-static loading conditions is maintained. The studies make it possible to distinguish several basic features of the material's behavior under loading: in the composite, internal compressive stresses are formed, and the structure changes as a result of a phase transformation in the matrix under action of external loading; this provides for distortion in any small volume of the material and for its simultaneous strengthening.

The physical meaning of the use of a structurally unstable binding phase in composites consists in lowering the structural level of plastic deformation as a result of the formation of a microcrystalline state of the binding phase in the course of inhomogeneous loading. According to [1], such materials include the alloys TiC–TiNi, WC–NiAl, WC–G13 steel, etc. Apparently, these effects will also be preserved under dynamic loading conditions, providing for excellent mechanical properties of the composite. The object of the present work was to study the macro- and microstructure of the hard alloy, WC–G12 steel, with a stable and metastable state of the matrix after dynamic loading. The loading was carried out by the impact of a ball element made from this alloy against a plate of aluminum alloy at velocities of 700-2000 m/sec.

In the case of a stable state of the binding phase, impact loading results in fracture of the ball element whose fragments remain in the barrier and are partially removed from it. If the matrix is in metastable state, dynamic action does not impair the continuity of the material of the striker; at different speeds of collision, only the shape of the striking element changes. From the change in the geometrical dimensions of the striker, according to [2], one can calculate the dynamic yield strength

$$Y_0 = \frac{\rho V^2}{2} \cdot \frac{1}{\ln \dfrac{L_0 - h}{L_f - h}}, \tag{1}$$

where ρ is the density of the material; V is the velocity; L_0 and L_f are the initial and final lengths of the striker; h is the depth of the plastic front. As is evident from formula (1), to determine the dynamic yield strength, it is necessary to measure the initial and final lengths of the striker and the depth of the plastic front. Since the size of the structural elements of the composite is very small (for example, the size of the carbide particles is 1-2 μm), metallographic studies can not be used to determined the depth of the plastic front. On the other hand, since the geometrical shape of the specimen has changed, it may be assumed that the entire volume of the material will undergo plastic deformation. With this assumption, the dynamic yield strength was calculated (Fig. 1).

As the deformation rate increases, so does Y_0; for V = 1000 m/sec it amounts to 1700 MPa, and for V = 1800 m/sec, 3600 MPa, and exceeds the yield strength $\sigma_{0.2}$ by a factor of 3.5, as well as the ultimate strength (Fig. 1). As the velocity increases, a marked increase in Y_0 is observed, and when V = 1600 m/sec, it reaches the ultimate-strength values. A further increase in velocity causes fracture of the material. On the basis of these data, it may be stated that the velocity at which $Y_0 = \sigma_u$ is the limit of the ductile–brittle transition during dynamic loading of the material in the velocity range under consideration.

Analysis of the microstructure of the specimens following dynamic loading shows (Fig. 2) the presence in the material of a large number of microcracks concentrating in parallel bands or tracks running through the entire volume of the material or located near the surface of the striking element.

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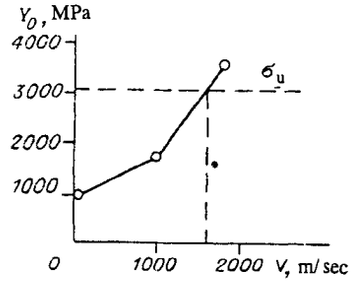

Fig. 1

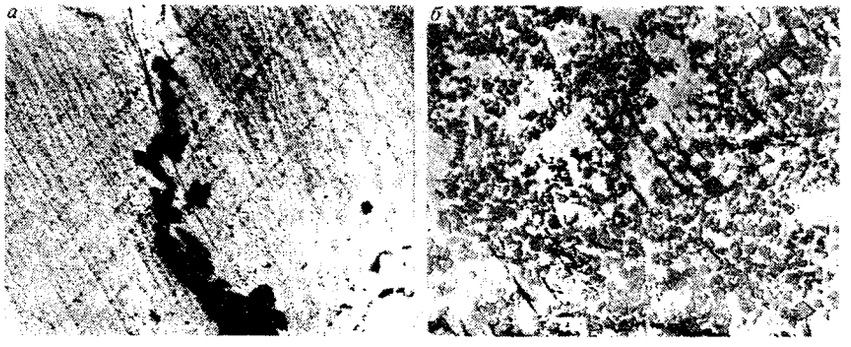

Fig. 2

The very fact of manifestation of such ordered microcracks in the form of tracks arrests to a special state of the structurally unstable binding phase. Moreover, it is obvious that they were formed in the earliest stages of collision, and spalling did not occur until later (Fig. 2a). This is apparently due to the presence of a dispersed carbide phase in this composite. To check this assumption, we plotted the distribution of the carbides with sizes and distances between the cracks formed, N(d) and N(h), for two alloys with different sizes of carbide grain (for d = 1 and 2.2 μm – Fig. 3a, b, respectively).

We note the regularity of the peaks in the distribution of the distances between microcracks and the drop in their intensity. Whereas for N(d), there is a maximum at d = 1 μm, there are at least three maxima at h = 2, 4, and 8 μm on the N(h) curve. A similar N(h) curve is seen for the material with a large size of carbide grain (d = 2.2 μm). This curve may be represented in the form

$$Y = \sum_{i=1}^{n} A_i \exp[-\alpha(x + 2^i d)], \tag{2}$$

where d is the average size of carbide particles (in our case, d = 1 and 2.2 μm); A_i is a function describing the change in the intensity of the maxima in the distribution of the distances between microcracks:

$$A_i = I_i(x + b_0)^2 \tag{3}$$

(I being the intensity of the maximum).

The only difference between these materials consists in the fact that in the first case, the macrocracks are ordered in the form of tracks running through the entire specimen, and in the second case, they are located near its surface. This difference may be due to the nature of the interaction between the impacting element and the barrier – a semi-infinite barrier in the first case and a fairly thin screen in the second.

The intensity of the harmonic (3) as a function of b_0 can be decreasing or increasing in the positive region of the argument: If the intensity of the harmonics decreases as their number increases, the value of b_0 is very large; otherwise it is very small.

The experimentally observed distribution of the intensities of the terms of Eq. (2) shows that for an unstable binder we have relatively sharp decrease in intensity as the number of the "harmonic" increases, i.e., b_0 is large. In the case of stable crystal, when the material breaks down and there is a small number of macrocracks (possibly, one), the parameter b_0 is small.

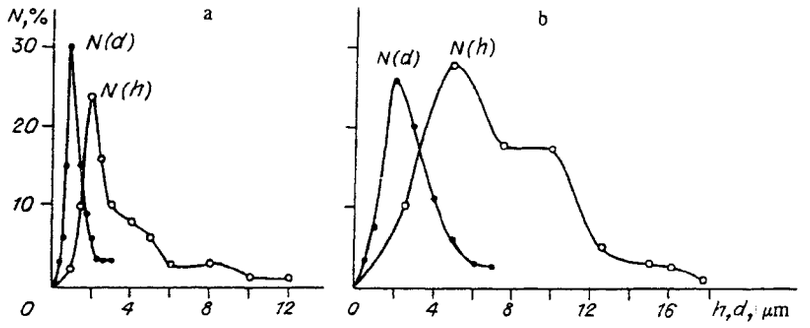

Fig. 3

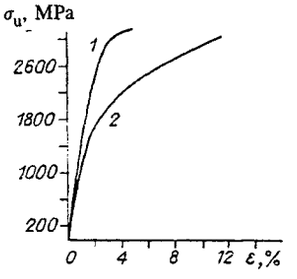

Fig. 4

It is well known [3] that stress and crack size are related as follows:

$$\sigma = (kE\gamma/C_{cr})^{1/2}, \tag{4}$$

where k is a coefficient equal to $2/\pi$, E is the elastic modulus of the composite, and γ is the energy, found from the expression

$$\gamma = (2\gamma_s + W_{pl}).$$

Here γ_s is the surface energy; W_{pl} is the work of plastic deformation during crack propagation. The yield point $\sigma_{0.2}$ of the material determines the critical crack below which the material will undergo only plastic deformation. It may be assumed, therefore, this quantity will also be determined by the parameter b_0 in formula (3): the greater C_{cr}, the lesser the tendency of the material to brittle fracture. This is also confirmed by the fact that the yield point for a structurally unstable state is substantially lower owing to the small stresses of martensite shear; this provides for a lower impact strength of the entire composite (Fig. 4, curves 1 and 2 – stable and metastable states).

Using formula (4), we can estimate C_{cr} for our materials. To do so, it is necessary to estimate W_{pl}. We have W_{pl} = 2RU, where U is the work of plastic deformation of the binder per unit volume, and R is the width of the zone covered by plastic deformation on one side of the crack, obtained from the stress–strain experimental curves for the binder [4]. It may be assumed that $R = l/2$ (l being the mean free distance between the carbide grains). In our case, for hard alloys with an average carbide grain size of 1 and 2.2 μm, l = 0.75 and 1.5 μm, respectively.

Substitution of these values into Eq. (4) shows that the critical crack size for an unstable binding phase is 40-50 μm and 80-90 μm for l = 0.75 μm and 1.5 μm, respectively, and for the stable state of the matrix C_{cr} = 15-20 μm and 20-25 μm.

Hence, in the case of a stable crystal, the size C_{cr} is smaller than the cracks forming in the composite. Therefore, only individual local cracks greater than 300 μm in size are observed in the material. For the unstable state, the size C_{cr} is appreciably greater than the cracks forming during loading of the composite. Therefore, the existing stress raisers in the composite lead only to the formation of multiple microcracks owing to the high ductility of the binding phase.

As already mentioned, the microstructure containing multiple microcracks is formed in the initial stage of loading, and this is followed by fracture of the composite. These microcracks themselves are an obstacle to a macrocrack. Therefore, the microcracks which has formed in the course of loading of the material may be regarded as a third phase of the composition material together with tungsten carbide and 110G13 steel.

Thus, the studies have shown that by using the structurally unstable state of the matrix in the composite, one can considerably lower the scale of the structural level of plastic deformation and of the fracture of such materials under conditions of not only quasistatic but also high-speed dynamic loading. At the same time, in the course of loading, a special microcrack structure is formed that provided an additional channel of dissipation of external energy, and as a consequence, an increase in toughness of the material.

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