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Motion and deceleration of explosively accelerated solid particles in a metallic target

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Abstract—The process of hardening materials with different types of crystalline lattices by means of explosive doping is considered as multifaceted. The likeness of, and the difference between, the mechanisms underlying the hardening of materials having body-centered cubic (bcc) and face-centered cubic (fcc) lattices are discussed. The role of long-range fields of stresses in the formation of dislocation structure of metals subjected to explosive doping has been investigated. An intricate nature of the space configuration of stress fields near retarded particles and the role of these fields in hardening the treated metallic materials have been demonstrated.

INTRODUCTION

THE METHOD of bulk doping of metallic materials is effected by superdeep penetration of disperse particles as a consequence of dynamic interaction with a barrier [1]. Such a treatment has considerable potential, since it increases the number of physico-mechanical and service properties of metals and alloys. The mechanism underlying the interaction of solid disperse particles with a metal matrix has been investigated by several authors [1, 2]. However, the problem of structural changes taking place in the matrix near particles and the trajectories of their motion has yet to be adequately studied. It would provide the possibility of understanding the mechanism of mass transfer of particles into metals with different compositions and structures.

MATERIALS AND PROCEDURES

Experiments were carried out with samples of armco-iron, molybdenum, silicon iron, and stainless steel (08X18H10T). To prepare the specimens for treatment, the first two were subjected to annealing and forging, and the latter two were put through annealing. Doping materials for the first two were made of particles of silicon nitride and titanium diboride powders, and for the latter two of a mixture of titanium carbonitride with nickel. The sizes of the particles were assigned within 40–125 μm. Their acceleration was achieved by using the energy of explosion [3]. The acceleration velocity attained 2×10^3 m s^{-1}, the pressure in the metallic barrier on collision of particles amounted to 15 GPa.

The microstructure of samples before and after the treatment was studied with the aid of a scanning electron microscope 'Nanolab-7' and a transmission electron microscope EMV-100B. Foil specimens of the studied materials 0.3–0.4 mm thick were spark-cutted transversely to, and in parallel with, the axis of the samples at a depth of 5–7 mm from the surface. The specimens were lapped with the aid of a 'Minimet' instrument, polished with diamond pastes and then polished electrochemically in an electrolyte, consisting of ortho-phosphoric acid (640 ml) and chromic anhydride (120 g), at a temperature of 70–80°C and voltage of 6–8 V.

The residual long-range fields of stresses were measured microscopically by bending extinction contour parameters [4, 5]. The fine structure of the materials was studied before and after explosive doping. The identification of particles implanted into the metallic barrier with those flung by explosion was conducted. To measure the parameters of the bending extinction contour, indexing of the latter was preliminarily carried out, and the inclination axis of the meter (goniometer) was oriented normally to the foil reflection. The presence of extinction contours near implanted

NOMENCLATURE

b	Burgers vector	γ	iron phase
G	shear modulus	ε	iron phase (martensite)
Δl	displacement of extinction bending contour	κ	component of plastic bending–twisting
R	radius of lattice curvature	ν	Poisson factor
β	tensor of excessive density of dislocations	$\rho^{\pm}$	local excessive density of dislocations
		σ_g	stress of long-range fields.

particles testified to the bending of a thin foil or to the corresponding curvature of the metallic crystalline lattice if the foil preserves the form of a plate [5]. The studied portions of the structure did not contain the boundaries of grains or subboundaries.

The measurement of long-range fields is based on the measurement of $\partial\varphi/\partial l$ [4, 5]. For this purpose, the goniometer measured the displacement of the bending extinction contour with a controlled variation in the foil inclination angle. The long-range stress fields were determined by combining the following formulae, since, as will be shown below, there was elastico-plastic bending:

$$\sigma_g = Gt\frac{1-\nu}{1-2\nu}\frac{\partial\varphi}{\partial l} \quad \text{and} \quad \sigma_g = Gb\sqrt{\rho_{\pm}}, \qquad (1)$$

where [6]:

$$\rho_{\pm} = \rho_{+} - \rho_{-} = \frac{1}{b}\frac{\partial\varphi}{\partial l}.$$

RESULTS AND DISCUSSION

Variation of the metallic target structure on high-velocity injection of disperse particles

Explosively-accelerated particles interact with the target material where local pressures of the order of dozens of kilobars are realized [1]. The particles may penetrate a metallic matrix in consequence of overcoming the forces of intermolecular bonds and due to the transition of local regions of the matrix into a quasi-liquid state. As shown in ref. [1], the collisional energy is spent for dynamic losses amounting to 46% of the dynamic yield stress of steel. While moving in a matrix, the particles excite shock-plastic waves in the front and displace the high-pressure region. Along the paths of particles, the zones with a strongly disordered state are being formed in the metal. These zones represent the sections of energetically resolved excited states of metals with atoms displaced from the crystalline lattice nodes (atomic-vacant state) [7]. The state is transmitted to the zones of metallic matrix on the path of the particles. This allows one to obtain an exceptionally high mobility of atoms and explains the high velocity of particles and also the fast complete or partial collapse of the channels along their paths. After the passage of particles, the pressure in the region behind them decreases due to unloading via the reverse jet being formed on compression of the channel.

In the case of incomplete closing of the channel, the particles find themselves within the cavities of the damage spots (craters). In longitudinal specimens the channels are identified in the form of thin cavities [Figs. 1(a) and 2(a)]. In transverse specimens the channels have the form of oval holes [Fig. 2(b)]. The channels and the particles are often surrounded by spidery extinction contours [Figs. 1(b) and 2(b)].

Analysis of a thin layer which directly adjoins the channel walls in all of the materials has shown that the region experienced transition to an amorphous state [Fig. 1(a)]. This is confirmed by examination of electrographs with blurred halos and wide diffuse rings as well as by the presence, on the surface of the channels, of a distinctive "veiny" pattern characteristic of the fracture of amorphous material [Fig. 2(a)]. The reasons for such a transition can be traced to the origin, on the path of the particles, of a highly excited state with a disordered atomic structure being fixed due to the smallness of the time of the process (10^{-7}–10^{-8} s [1]) and good heat removal, to the friction heat-induced melting of a thin metal layer, and to the great plastic deformation developing at a high rate [8]. The X-ray microspectral examination has also revealed the saturation of the surfaces of the channel walls with the elements entering into the composition of the material of the particles, indicating its partial dissociation.

The great plastic deformation occurring at a high rate is evidenced by the presence, near the particles and their paths, of a gamut of dissipative structures changing from one to the other and testifying to the processes of self-organization in the metallic matrix. Since moving particles excite shock-plastic waves ahead of them, the concentrators of stresses originate inevitably on their paths, and this entails deformation-type relaxation processes [9]. Based on the relaxation approach theory, the region of the matrix in front of a moving particle and the trajectory of its motion may be considered as a set of relaxation oscillators coming into action as soon as a certain stress is attained on them. As a result, the region subjected to compression on all sides splits into zones with relaxed stresses. The process of plastic deformation near the particles and their paths shows up in the form of relaxation waves involving shearing and rotational components and

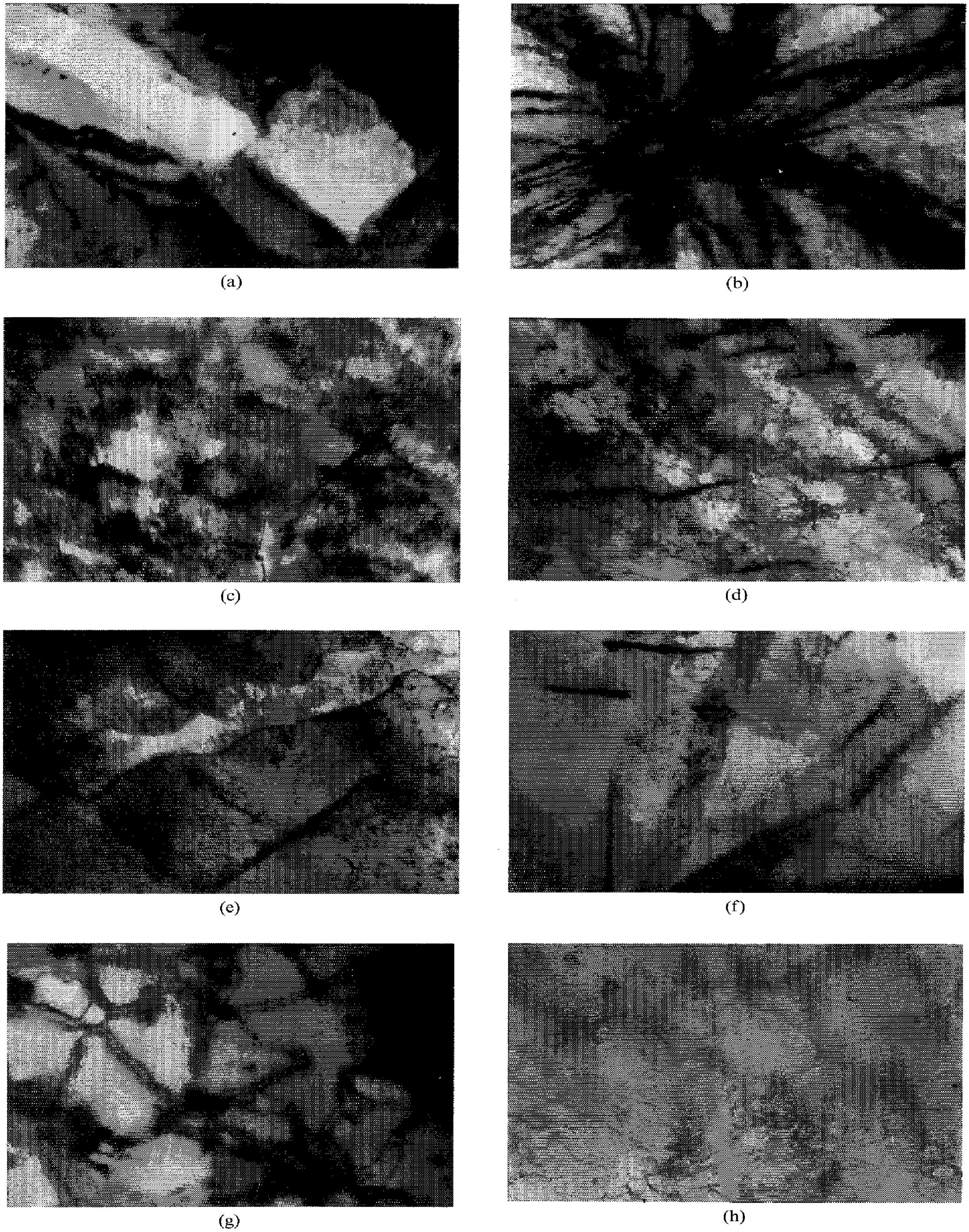

FIG. 1. Microstructure of armco-iron (a)–(f), (h) and molybdenum (g) following explosive doping: (a), (g) ×60 000; (b)–(f) ×20 000; (g) ×100 000.

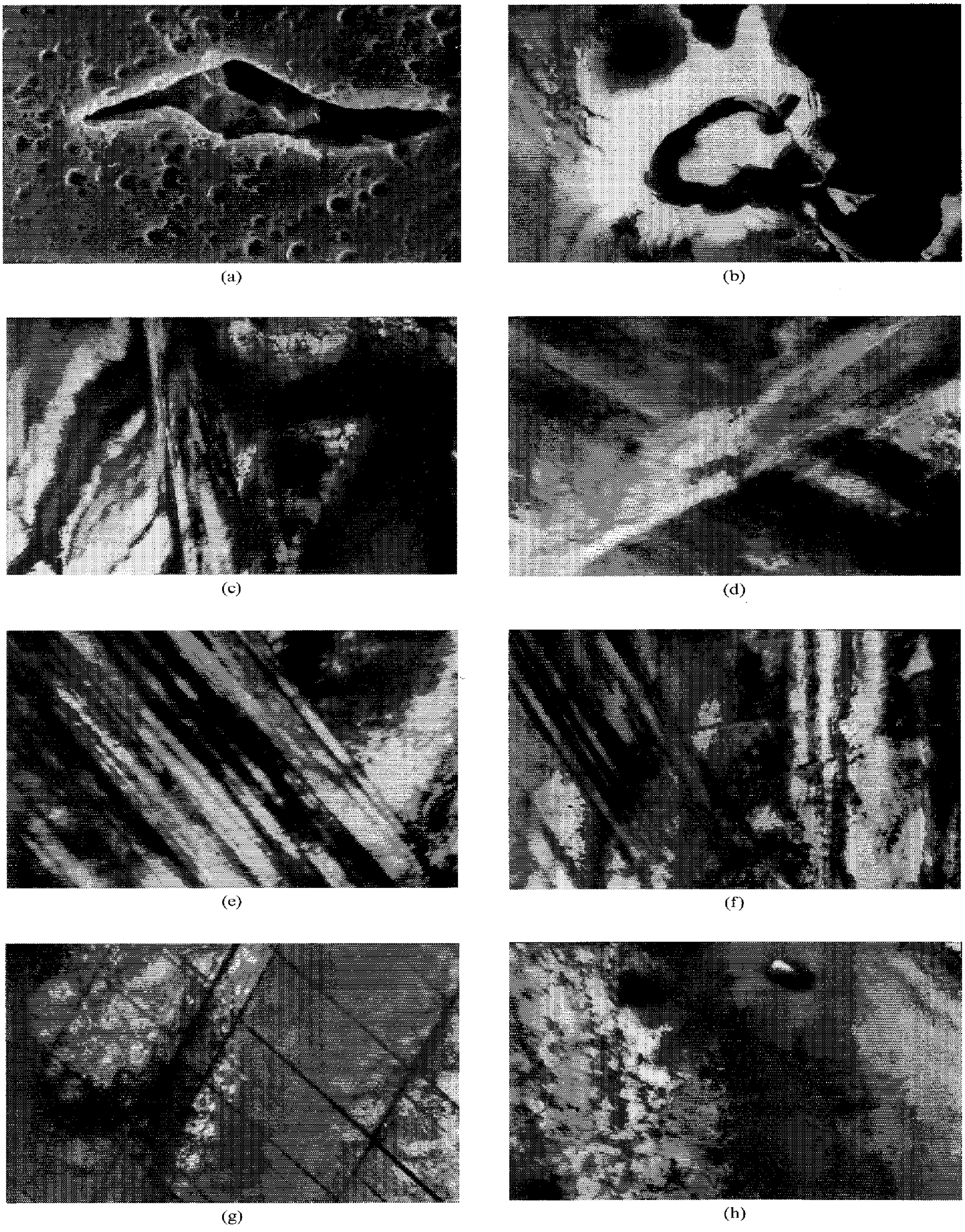

FIG. 2. Microstructure of steel 08X18H10T after explosive doping: (a) ×5000; (b)–(h) ×20 000.

propagating with a high velocity [10]. The particle trajectory comprising the zone of the collapsed channel emerges as a highly distorted structure in the form of a localized plastic deformation band [Fig. 2(c)].

A strongly stressed state near the trajectory of the particles is responsible for the origination of deformation-type relaxation processes the nature of which depends on the crystalline lattice of the material, its chemical composition, and on the energy of the packing defect. In materials with bcc lattices, the thin layer of the channel surface with an amorphous structure borders directly on the zone with a strongly fragmented structure characterized by high-density dislocations, 10^{11}–10^{12} cm^{-2} [Fig. 1(c)]. The fragmentary structure gradually goes over into a blurred cellular dislocation structure in which new subboundaries are being formed [Fig. 1(d)]. The boundaries of the fragments and cells are not very clearly defined and perfect, i.e. torn and loop-like patterns are seen. The new boundaries make up dipole configurations. They have the form of closed loops, or terminate representing the regions that in space confine the front of the plastic deflection of one portion of the grain with respect to the other.

Against the background of the fragmentary and cellular structures, thin black bands can be seen which are aligned in a strictly parallel direction within the grain, the so-called zones of localized deformation [Fig. 1(d)]. The electron diffraction patterns of these sections with arc-shaped reflexes reveal appreciable distortions in the crystalline lattice and great disorientation of separate regions. The formation of such reorientation bands is indicative of the presence of a considerable rotational component of plastic deformation. At large magnifications, these bands resemble flat assemblies of packing defects, which represent the onset of the pre-twinned state.

In the zones with fragmentary or cellular structure, one can see a great number of strongly distorted twins [Fig. 1(e)] having a prismatic shape in the lateral cross-section. At the places of bending of the twins, the boundaries of their fragments intersect and this purely geometrically reflects the turn associated with the passage of disclinations. In these regions the shears of one portion of the pair with respect to the other are possible. They lead to a full or partial loss of the twin conjugation with the matrix due to the interaction of the ensemble of dislocations with the twin boundary.

In the zone of strong deformation, thin perfect boundaries were also observed which formed dipole configurations and which were called "knife boundaries" in ref. [11]. Such boundaries produce large, of the order of several degrees, disorientations of the adjoining regions of the crystal. With the treatment, the boundaries of the grains showed a tendency to split [Fig. 1(f)], since they probably originated in the layers corresponding to the structure of special boundaries with a smaller energy [12] which have moved rapidly to the positions having a high density of coinciding nodes.

The specific features of the molibdenum structure are split dislocations [Fig. 1(h)], which were not absent in the initial state. They form pentagonal configurations with a loop at the centre, with the neighbouring "stars" having split dislocation. The split dislocations were observed in the zones corresponding to great and small deformation degrees. It should be noted that while static (quasi-static) deformations of bcc-metals, having a high energy of the packing defect, exhibit light rearrangement of non-split dislocations with the formation of dislocated ensembles, the rearrangement of the dislocational structure under high-rate deformation conditions can be accompanied by the splitting of dislocations.

The zone with the cellular structure goes over gradually into the region of weak deformation for which the specific feature is the positioning of dislocations in intercepting glide planes with the formation of a grid [Fig. 1(h)]. At the places of the interception of the dislocations of two systems, dislocation loops are often seen. Finally, in the next zone a chaotic distribution of dislocations is observed.

In the austenite structure of steel 08X18H10T, which has a low packing defect energy, a strongly fragmentary structure originates near the channels [Fig. 2(d) and (e)] going over into a cellular one [Fig. 2(f)]. A strongly stressed state near the paths of the particles in the given material causes the origination of relaxation processes leading to the appearance of a great amount of the assemblies of thin microtwins with strictly parallel boundaries and packing defects against the background of fragmentary and cellular structures [Fig. 2(d)–(f)]. In the case of a high-rate deformation in austenite, the deformation microtwins are formed easily by displacements of twinned defects in the plane {III} and of the segments of such defects [10]. The density of dislocations in the zone of fragmentation amounts to about 10^{11}–10^{-12} cm^{-2} and in the zone with the cellular structure to about 10^{11} cm^{-2}. In these portions, as well as in metals with bcc lattices, the scraps of different types of boundaries are detected, bearing witness to local rotations of microregions.

The analysis of the electrondiffraction patterns taken over the sections with a great density of packing defects has revealed the existence of a new phase ε against the background of a strongly deformed structure [Fig. 2(d)–(f)]. This phase represents thin twinned plates with a face-centered lattice which were formed after the merging of twinned packing defects or of their segments. A similar structure was also observed by the authors of ref. [13]. Packing defects easily originate in austenite, and this process, provoked by a high-rate treatment, has led to phase transition due to the shear rearrangement of the crystalline lattice. The deformation microtwins and plates of the ε-martensite exist simultaneously in the austenite matrix. They are located in the identical austenite planes {III}, therefore it would be logical to assume that they can grow from identical nuclei. As follows from ref. [14], both the twinning and the phase transition $j \rightarrow \varepsilon$ may take place via the formation of packing defects on full dislocation splitting into partial ones. In this case, the twin grows due to the movement of the twinning dislocation into the parallel layers of the lattice in the section with the packing defect, whereas the growth of the face-centered interlayer in austenite (ε-phase) takes place because of the formation of "bonds" between the packing defects.

Gradually, the densities of the twins, packing defects and of the ε-phase in the cellular structure decrease. Their clusters originate more rarely; more often they form a system of intersecting defects [Fig. 2(h)]. In going over to the basic structure with a chaotic distribution of dislocations, their density gradually decreases [Fig. 2(h)] and attains the level 10^9 cm^{-2}. It should be noted that the split dislocations were observed in all of the zones; this is typical of metals with a low energy of the packing defect [10].

THE LONG-RANGE STRESS FIELDS NEAR RETARDED PARTICLES

It is known that the long-range fields of stresses originating near various defects (dislocations, disclinations, twins, disperse inclusions, boundaries of grains and subgrains, including those torn at the junctions of boundaries) play an important part in evolving the defect structure and developing deformation strengthening. This role strengthens as the inhomogeneity of the structure of metals and alloys increases [4].

The process of strengthening a metallic target involves several factors (reinforcement with channels and channel zones, local transition to amorphous state and microalloying, deformational strengthening as a result of local plastic events, dispersion strengthening with particles). We shall consider in more detail the latter factor which is associated with the action of the particles that came to a halt. Moving particles, which experience the resistance on the side of the matrix, slow down, retard and come to a stop. At the places of their retardation, a long-range field of stresses appears which generates relaxation processes having an obviously wavy nature [Fig. 3(a)–(c)]. The shock waves excite collision pulses; the stress waves generated by these processes superimpose and form the superposition of pulses. Relaxation processes near the particles are realized by means of successive drops in stresses because of the defects being emitted from the boundaries 'particle-matrix' (dislocations and disclinations). Due to the non-isotropic properties of translational fluxes, the relaxation shears inevitably generate the field of rotating moments acting on the particle and on the adjacent zone of the matrix. High-frequency vibrations of particles and grains of the matrix in shock waves account for the origination of a turbulent plastic flow. The fluxes of defects emerging from the particles form new sources of force fields.

The origination of the long-range fields of stresses near retarded particles is evidenced by the presence of wave-type spidery extinction contours the character of which indicates their viscoelastic origin [Fig. 3(d)–(f)]. In all of the cases the fields are screened by the field of dislocations originating around the particles. The excitation contours result from the localization of vortical deformation at the place of the retardation of particles. This is a fixed state of plastic deformation taking place at a high rate in the course of explosion and leading to the appearance of long-range fields of stresses of viscoelastic character around the particles. It seems that from the start the originating stress fields are elastic. The partial relaxation of stresses has made them elastico-plastic. The extinction contours have the branches of several orders which often intersect with one another [Fig. 3(e)]. In addition to the spidery profile contours, the formations in the form of a "comet tail" were observed, i.e. spherical, tetragonal or rhombic distortions originating around the particles. The extinction bands have different thicknesses; there are clearly defined and smeared branches, and sometimes the so-called "shagginess" of the branches [Fig. 3(f)] is observed which is conditioned by the presence of dislocation knots.

Near some particles, extinction contours were observed the character of which is similar to the fields near coherent inclusions [5], though coherence is out of the question in the present investigation. Sometimes, the pattern of the contours pointed to purely elastic distortions. For example, in steel 08X18H10T in its symmetric position {III} one observes the intersection of dark bands which corresponds to the lines of systematic reflections $2\bar{2}0$, $20\bar{2}$, and $02\bar{2}$. Or, in the case of the grain orientation [100], two dark bands corresponding to systematic reflections 020 and 002 intersect at right angles, while weaker bands (second-order branches), which intersect with them at an angle of 45°, appear due to systematic reflections 022 and $02\bar{2}$. In the armco-iron with the orientation near [111] in the symmetric position, there is an intersection of the dark bands which corresponds to the lines of systematic reflections $1\bar{1}0$, $0\bar{1}1$, $\bar{1}01$, and $\bar{1}10$.

In order to study the character of variation in the long-range fields of stresses near retarded particles, measurements were made of the crystalline and lattice bending-twisting tensors available due to various dislocation–disclination formations in the structure.

The fields of stresses near the particles gave rise to a considerable deflection of atomic planes $\partial\varphi/\partial x$ which is related to $\boldsymbol{\kappa}$, β and R [4]:

$$\frac{\partial\varphi}{\partial x} = \boldsymbol{\kappa}_{ij} = \beta_{ij} = \frac{1}{R} = b\rho_{\pm}\,. \tag{2}$$

In the case of elastico-plastic bending $\kappa > \beta = b\rho_{\pm}$, whereas the bending–twisting tensor is characterized by two components which were measured at a distance of 1 μm from the particle (dots on the lines, Fig. 4):

$$\boldsymbol{\kappa} = \begin{vmatrix} 0 & 0 & 0 \\ 860 & 0 & 860 \\ 0 & 0 & 0 \end{vmatrix} \quad (\text{rad cm}^{-1}). \tag{3}$$

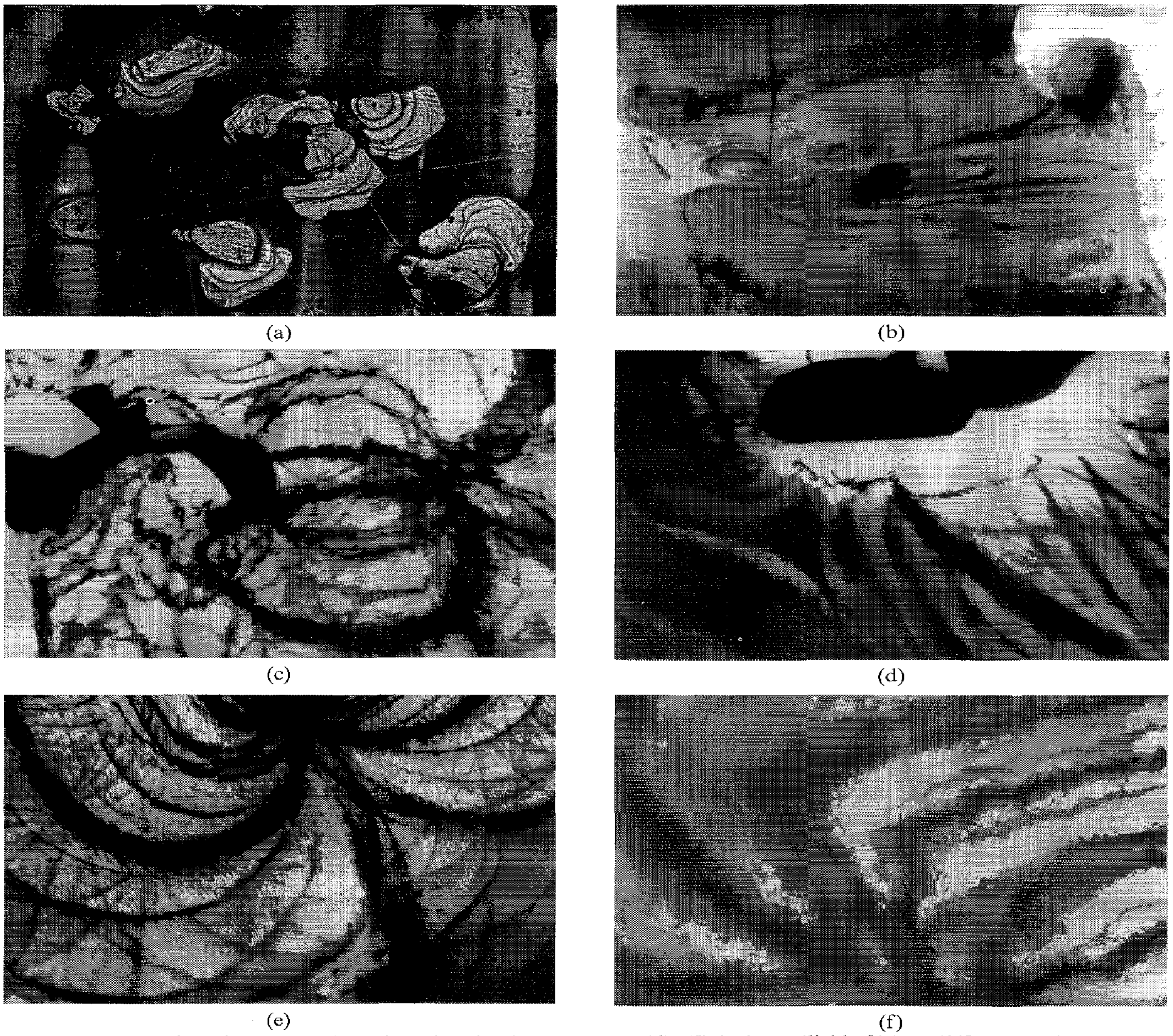

FIG. 3. Relaxation waves (a)–(c) and extinction contours (d)–(f) in iron silicide [(a) $\times 500$], armco-iron [(b) $\times 20\,000$]; steel 08X18H10T [(b),(f) $\times 20\,000$; (d) $\times 60\,000$] and molybdenum [(e) $\times 60\,000$).

The results of the determination of stress fields testify to the decrease in these fields with distance from the place of particle retardation (Fig. 4). The stresses decrease along the y axis, i.e. the gradient $\partial\boldsymbol{\kappa}/\partial y$ takes place. The larger the particle, the higher the stress σ_1. The fields of the stresses of neighboring particles can interact. In such a case the stress σ_1 along the y axis varies in a more complex fashion, and the variations in the components of $\boldsymbol{\kappa}$ turn out to be out of proportion to each other.

The fields of stresses near retarded particles interact with other defects of the metallic matrix structure which were present in the matrix before treatment and which originated on explosive doping. The wavy relaxation zones and the long-range fields of stresses develop near the surfaces of contact and disorientation, non-metallic inclusions, and near dislocation clusters. The fields from various sources sum up and form a three-dimensional system of long-range fields of stresses characterized by different wavelengths, dispersion, spatially localized maxima and minima.

The stress waves, originating near a retarded particle, which came to a stop not far from the junction between the boundaries of the grains, generate new waves of stresses on the other side of the boundary (in the adjacent grain) and can cause such a strong stress at the place of junction which leads to the generation of a microcrack. The stress fields near the particles interact with the stress fields caused by the formed fragmentary or cellular dislocation structure and also with the twins, twin interlayers of the ε-martenside of deformation origin, with the bands of

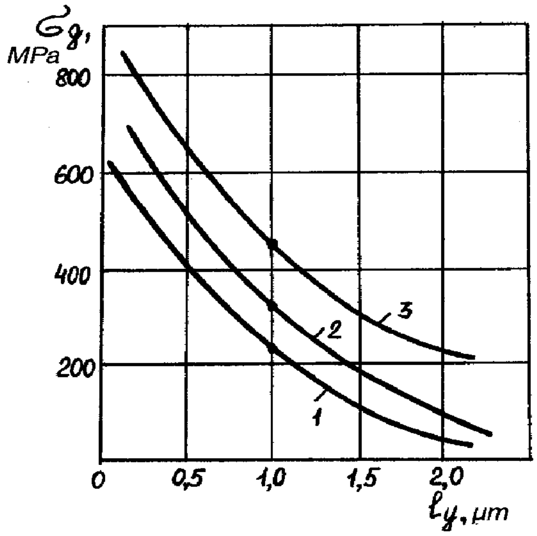

FIG. 4. Variation of the stress field in steel 08X18H10T near particles of size 20(1), 40(2), and 60(3) μm.

localized deformation in which one can also see the traces of wave plastic relaxation. These defects do not violate the continuity of extinction contour if they appear earlier or simultaneously, or, occurring later, they violate the continuity of the foil bending and complicate the profile of the contours. At the places of intersection additional stresses originate and the appearance of microcracks is possible.

CONCLUSIONS

A comparative analysis of the microstructure of material matrices with bcc and fcc lattices has shown that the superdeep penetration of disperse particles accelerated by the energy of an explosion in both of them is governed by similar laws which include the motion and retardation of particles, formation and collapse of channels along their path, as well as the origination of complex dissipative structures indicative of the occurrence of relaxation processes of deformational character at which local plastic deformation takes place resulting in the self-organization of the material structure. The common feature of the relaxation processes taking place in the materials studied is the existence of substantial plastic shears and turns near the particles and their paths typical of a strongly deformed state when collective forms of motion arise in the assemblies of strongly interacting dislocations [11]. In the regions with a fragmented or cellular structure, large-scale crystallographic non-uniformities develop that give rise to considerable disorientations.

The difference in the character of relaxation processes taking place in a metallic matrix near particles and their paths is attributable in the first place to the type of the crystalline lattice and to a different energy of the packing defect, to an abnormally great mobility of the edge components of dislocation loops in the bcc-metals and to the predominant slip of dislocations along octahedral systems in fcc-metals, to the difference in the shape and in the behaviour of dislocation loops and to different behaviour of twinning dislocations in bcc- and fcc-metals [10]. If the shear-turning type of deformation takes place in the studied materials similarly to the formation of fragmented and cellular structures, of the zone with a system of intersecting dislocations and their random distribution, then the difference is observed in the processes of twinning and formation of packing defects. In armco-iron and molybdenum, on the one hand, and in 08X18H10T steel, on the other, the mechanisms of twinning and formation of packing defects differed greatly. This is responsible for the difference in the morphology of twins (see Figs. 1(e) and 2(d)–(j)) and in the outcomes of the formation of packing defects (the pretwinning state in bcc-metals and the formation of the new ε-martensite phase in the fcc-material). The difference was also observed in the behavior of dislocations and boundaries of the grains (their abnormal splitting in the bcc-metals and traditional splitting in the fcc-material). The long-range fields of stresses originating in the places of retardation of disperse particles accelerated by the explosion energy produce wave relaxation processes in a metallic matrix which influence the fine structure and contribute to the strengthening of the material processed.

In metallic materials, the process of strengthening by disperse particles accelerated by the energy of explosion is multifaceted and consists of the reinforcement of the material by channels and channel zones, transition to amorphous state and microalloying, dispersion strengthening by particles and deformation strengthening as a result of local plastic events of relaxation type. The nature of the latter is greatly determined by the type of the matrix crystalline lattice.

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