Influence of Lattice Curvature and Nanoscale Mesoscopic Structural States on the Wear Resistance and Fatigue Life of Austenitic Steel

Viktor E. Panin, Valery E. Egorushkin, and Natalya S. Surikova

Abstract The gauge dynamic theory of defects in a heterogeneous medium predicts the nonlinearity of plastic flow at low lattice curvature and structural turbulence with the formation of individual dynamic rotations at high curvature of the deformed medium. The present work is devoted to the experimental verification of the theoretical predictions. Experimentally studied are the influence of high-temperature radial shear rolling and subsequent cold rolling on the internal structure of metastable Fe–Cr–Mn austenitic stainless steel, formation of nonequilibrium ϵ - and α' -martensite phases, appearance of dynamic rotations on fracture surfaces, fatigue life in alternating bending, and wear resistance of the material. Scratch testing reveals a strong increase in the damping effect in the formed hierarchical mesosubstructure. The latter is responsible for a nanocrystalline grain structure in the material, hcp ϵ martensite and bcc α' martensite in grains, a vortical filamentary substructure on the fracture surface as well as for improved high-cycle fatigue and wear resistance of the material. This is related to a high concentration of nanoscale mesoscopic structural states, which arise in lattice curvature zones during high-temperature radial shear rolling combined with smooth-roll cold rolling. These effects are explained by the self-consistent mechanical behavior of hcp ϵ -martensite laths in fcc austenite grains and bcc α' -martensite laths that form during cold rolling of the steel subjected to high-temperature radial shear rolling.

Keywords Gauge dynamic theory of defects · Nanoscale mesoscopic structural states · Lattice curvature · Damping effect · Dynamic rotations · Fatigue failure · Wear resistance

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1 Introduction

Mechanical behavior of metals were subject of intensive theoretical and experimental studies over many decades. Conventionally, mechanisms of mechanical behavior under various loading conditions are associated with various-scale strain-induced defects, including cracks, in a translation-invariant crystal lattice. However, in most cases, the translational invariance of the crystal lattice in a deformed solid is strongly violated, which does not allow for a correct description of plastic deformation and fracture within the linear approach of Newtonian mechanics. In this work, we thoroughly study the mechanical behavior of Fe–Cr–Mn austenitic steel, in which complex structural phase transformations of the initial fcc lattice with the formation of hcp and bcc martensite phases occur in uniaxial tension. According to [1, 2], crystal lattice transformations are related to the appearance of nanoscale mesoscopic structural states at the interstices of lattice curvature zones. Thus, the account for lattice curvature provides a basis for nonlinear solid mesomechanics.

In this work, crystal lattice curvature throughout the Fe–Cr–Mn austenitic steel specimen is produced using the complex treatment by high-temperature radial shear rolling + cold rolling at room temperature. This complex treatment reveals important new regularities in the mechanical behavior of austenitic steel.

2 Gauge Dynamic Theory of Defects in the Heterogeneous Medium

2.1 Basic Equations of the Gauge Theory

A deformable solid as a multilevel hierarchically organized system requires a self-consistent description at the nano-, micro-, meso- and macroscopic levels. Gauge dynamic theory of defects in the heterogeneous medium provides a framework for such description. In this theory, fluxes of strain-induced defects J and their density α are described by the equations

$$\frac{\partial}{\partial x_\alpha} J_\mu^\alpha = - \frac{\partial \ln u_\alpha(x, t)}{\partial t}, \quad (1)$$

$$\varepsilon_{\mu\chi\delta} \frac{\partial J_\delta^\alpha}{\partial x_\chi} = - \frac{\partial \alpha_\mu^\alpha}{\partial t}, \quad (2)$$

$$\frac{\partial \alpha_\mu^\alpha}{\partial x_\alpha} = 0, \quad (3)$$

$$\varepsilon_{\mu\chi\delta} \frac{\partial \alpha_\delta^\alpha}{\partial x_\chi} = \frac{1}{c^2} \frac{\partial J_\mu^\alpha}{\partial t} + \frac{\partial \ln u_\beta(x, t)}{\partial x_\nu} \frac{C_{\alpha\beta}^{\mu\nu}}{E} - P_\nu^\beta \frac{C_{\alpha\beta}^{\mu\nu}}{E}, \quad (4)$$

$$\frac{1}{c^2} \frac{\partial^2 \ln u_\alpha(x, t)}{\partial t^2} + \frac{\partial^2 \ln u_\beta(x, t)}{\partial x_\mu \partial x_\nu} \frac{C_{\alpha\beta}^{\mu\nu}}{E} = -P_v^\beta \frac{C_{\alpha\beta}^{\mu\nu}}{E} \quad (5)$$

Equations (1)–(5) have the following sense and use the following notations:

- (1) is the continuity equation for a defected medium, which indicates that the source of plastic flow causes the defect flow rate;
- (2) is the condition of plastic-strain compatibility, the time variation of the medium density is determined in this case by the flux rotor, i.e. its heterogeneity, rather than by the divergence;
- (3) is the condition of continuity of defects, which corresponds to the absence of charges of the rotational component of the plastic strain field ($\alpha_\chi^\beta = \varepsilon_{\chi\mu\nu} \partial_\mu P_\nu^\beta$);
- (4) is the governing equation for the medium with plastic flow;
- (5) is the equation of quasi-elastic equilibrium, which presents the known continuum mechanics equation. Along with the elastic strain, it contains plastic distortions on the right-hand side. In fact, this summand corresponds to the nucleation of strain-induced defects in local zones of hydrostatic tension due to stress concentrators.

Expression (4) is applicable only to the medium with plastic flow. It relates the time variation in plastic flow to the anisotropic spatial variation in the defect density $\varepsilon_{\mu\chi\delta} \partial \alpha_\delta^\alpha / dx$ and sources $\left(\sigma_\mu^\alpha - P_v^\beta C_{\alpha\beta}^{\mu\nu} / E \right)$. Equations (4) and (5) differ from the corresponding elastic equations in that the time variation in plastic strain rate is determined by the stresses themselves rather than by $\partial \sigma_\mu^\alpha / \partial x$, as in the elastic case. In addition, the right-hand side of (4) includes plastic distortion $P_v^\beta(x, t)$ taken as sources, which indicates the duality of defects as field sources.

From the system of Eqs. (1)–(5), wave equations can be found for the dimensionless quantities of defect flux J and density α :

$$\frac{1}{c^2} \frac{\partial^2 J_\alpha^\mu}{\partial t^2} - \frac{\partial^2 J_\alpha^\mu}{\partial x^2} = \frac{\partial}{\partial t} \left\{ \frac{\partial \ln u_\alpha(x, t)}{\partial x_\mu} - \frac{1}{E} \frac{\partial \ln u_\beta}{\partial x_\nu} C_{\alpha\beta}^{\mu\nu} - \frac{1}{E} P_v^\beta C_{\alpha\beta}^{\mu\nu} \right\} \quad (6)$$

$$\frac{1}{c^2} \frac{\partial^2 \alpha_\alpha^\mu}{\partial t^2} - \frac{\partial^2 \alpha_\alpha^\mu}{\partial x_\nu^2} = \varepsilon_{\mu\chi\sigma} \left\{ \frac{\partial^2 \ln u_\beta(x, t)}{\partial x_\chi \partial x_\nu} C_{\alpha\beta}^{\mu\nu} - \frac{\partial P_v^\beta}{\partial x_\chi} C_{\alpha\beta}^{\mu\nu} \right\} \frac{1}{E}, \quad (7)$$

subject to the source compatibility condition

$$\frac{\partial N_\mu}{\partial t} + \varepsilon_{\mu\ell m} \frac{\partial M_\mu}{\partial x_\ell} = 0, \quad (8)$$

where M is the right-hand side of expression (6), N is the right-hand side of expression (7), and $u(x, t)$ is the inelastic displacements in the wave of inelastic localized strain.

The right-hand side of Eq. (6) characterizes defect flow sources. They are determined by the rate of quasi-elastic strain $\frac{\partial}{\partial t} \left(E_\mu^\alpha E - E_v^\beta C_{\alpha\beta}^{\mu\nu} \right) \frac{1}{E}$. Parenthesized is the difference between internal compressive (tensile) stresses and shear stresses

Fig. 1 Shape and velocity of plastic deformation depending on the curvature of the deformed region

associated with the stress distribution in the stress concentration zone. Relaxation processes of defect rearrangement (such as various atomic configurations or their conglomerates) are presented in (6) by the term $P_v^\beta C_{\alpha\beta}^{\mu\nu}/E$.

The right-hand side of Eq. (7) expresses a source of the strain-induced defect density. This is vorticity $\varepsilon_{\mu\chi\delta} \frac{\partial}{\partial x} (E_v^\beta - P_v^\beta) \frac{C_{\alpha\beta}^{\mu\nu}}{E}$ of shear strain induced by the shear stress relaxation in local zones of hydrostatic tension during the defect formation.

The wave pattern of strain-induced defect flows is defined by the right-hand side of Eqs. (6) and (7). Plastic distortion $P_v^\beta(x, t)$ plays a major role in the wave pattern of localized plastic flow.

Equation (6) for the strain-induced defect flux along the direction L (at $r < L$) is solved as

$$\vec{J} = \frac{b_1 - b_2}{4\pi} \chi(s, t) \vec{b}(s, t) \left(\ln \frac{2L}{r} - 1 \right) - \nabla f, \quad (9)$$

where $\vec{b}$ is the binormal vector in the local coordinate system, $\vec{n}$ is the normal, $\vec{t}$ is the tangent, χ is the curvature variation in a deformed region due to external loading, s is the current length of the region, b_1 and b_2 are the Burgers vector moduli of the bulk translational and subsurface rotational inconsistency, respectively, ∇f is the gradient part of the flux due to third-party sources.

The expression for the defect flux includes curvature χ of the deformed medium. Let us analyse the role of this factor in the behavior of the strain-induced defect flow.

2.2 Structural Turbulence at Severe Lattice Curvature

The wave pattern of localized plastic flow in Eqs. (6) and (7) is generally shown in Fig. 1. It demonstrates the dependence of the wave profile of localized plastic flow on the curvature of the deformed region $\chi(x, t) = 4\beta \text{sech}[2\beta(x + 4vt)]$.

In Fig. 1, it can be seen that an increase in the curvature of the deformed region greatly changes the shape and velocity of plastic deformation. The experiment confirms the theoretical prediction [1].

At a high curvature of the deformed medium, plastic distortion $P_v^\beta(x, t)$ strongly grows, too. This means that the right-hand sides of Eqs. (6) and (7) increase significantly. The wave pattern of plastic flow cannot be preserved in this case, and plastic deformation becomes turbulent, breaking up into individual dynamic rotations. Such a theoretical prediction of structural turbulence in a deformed solid was impossible under the conditions of translational invariance of the crystal. The violation of translational invariance and account of curvature in a deformed medium are fundamentally new. Their experimental confirmation was earlier obtained [2].

Severe curvature of plastic flow was previously achieved by introducing titanium carbonitride nanoparticles into a deformed solid [2]. Unlike carbides, which have a spherical configuration, carbonitrides have a cubic structure [3, 4]. Due to plastic flow in the presence of titanium carbonitride nanoparticles, a severe curvature is formed in the deformed medium, especially under shock loading. Figure 2 shows the fracture surface of low-carbon steel 09Mn2Si when measuring impact toughness with 0.15% titanium carbonitride nanoparticles introduced into the steel. The New View profilometer discovers pronounced dynamic rotations of turbulent flow on the surface in the form of individual vortical protrusions. The scratch test shows a significant decrease in the groove size in the presence of nanoparticles (Fig. 3b), in contrast to

Fig. 2 Structural turbulence on the fracture surface of low-carbon steel 09Mn2Si, with 0.15% TiCN nanoparticles introduced into it

Fig. 3 Cross profiles of the scratch groove in the 09Mn2Si specimens: initial (a) and with 0.15% TiCN nanoparticles (b)

a deep groove forming in the absence of nanoparticles (Fig. 3a).

Structural turbulence in a solid is characterized by relay-race transfer of momentum from particle to particle. This is what makes it different from classical turbulence characterized by the Avogadro number [5, 6].

Structural turbulence is even more pronounced on the fracture surface of a chevron-notched specimen shown in Fig. 4. Counter shear loads from the specimen notches induce couple stresses that form dynamic rotations (Fig. 5). A filamentary structure appears around the notches, which bears witness not to translational invariance but to the formation of special structural states in lattice curvature zones. These states also arise during the formation of dynamic rotations (Fig. 6). In other words, at the crystal structure curvature the electronic subsystem forms special structural states that are responsible for dynamic rotations by the mechanism of plastic distortion.

Fig. 4 Geometry of a chevron-notched specimen

Fig. 5 Formation of a filamentary structure in the vicinity of the edge-cut notches in the chevron-notched 09Mn2Si steel specimen in fracture

Fig. 6 Formation of noncrystallographic boundaries due to interstitial mesoscopic structural states in turbulent rotations on the fracture surface of a chevron-notched 09Mn2Si steel specimen

3 Role of Lattice Curvature in the Mechanical Behavior of Austenitic Steel

3.1 *Influence of High-Temperature Radial Shear Rolling and Subsequent Smooth-Roll Cold Longitudinal Rolling on the Austenitic Steel Microstructure*

High-temperature radial shear rolling exerts no effect on the phase composition of austenitic steel. It forms a layered structure with varying degrees of the crystal lattice curvature-torsion and refines the grain structure. In the near-surface layers, the material has a globular structure with the average grain size $d \approx 0.57 \mu\text{m}$; grains in the near-axial zone are elongated along the rolling direction. A small-angle substructure is well developed within the grains. The fraction of special twin boundaries with a misorientation of $\sim 60^\circ$ is significantly reduced. The yield stress $\sigma_{0.2}$ and ultimate strength σ_B increase, respectively, from 400 to 620 MPa and from 850 to 1050 MPa. The ductility decreases from 90 to 55% (the true strain $\varepsilon = 0.7$).

Radial shear and cold rolling to the cold strain $\varepsilon = 1.8$ changes significantly the austenitic steel microstructure. In austenite grains, a two-phase (ε martensite + α' martensite) nanocrystalline mesosubstructure is formed. Grains containing α' martensite with a distorted bcc structure prevail. According to X-ray structural estimates, the volume fraction of α martensite after rolling to $\varepsilon_{\text{true}} \sim 1.8$ comprises $\approx 85.6\%$, and the size of coherent scattering regions is 40 nm. Cold rolling forms a heterogeneous structure. In the specimen areas with a high plastic strain, the grain size is 40–100 nm. The specimen areas with a lower strain (Fig. 7) contain austenite grains 200–400 nm in size (Fig. 7b) and grains with ferrite zones 30–400 nm in size (Fig. 7d).

Fragmentation processes during cold rolling begin with the formation of fine stacking faults and ε -martensite nuclei (Fig. 8a) or α' -martensite plates (Fig. 8b) in various grains of the material.

The results presented in Figs. 7 and 8 indicate that ε -martensite and α' -martensite laths are important intermediate phases for the structural transformation of fcc austenite to bcc ferrite. Of special note is a strong texture of the intragranular structure. In austenite crystals, the direction and plane of rolling are close to $\langle 111 \rangle_\gamma$ and $\{110\}_\gamma$, respectively; in ferrite zones, to $\langle 110 \rangle_\alpha$ and $\{111\}_\alpha$.

Thus, within a nanostructured austenitic steel subjected to the complex treatment by radial shear rolling + smooth-roll cold rolling a multiscale hierarchically organized structure is formed that allows translational-rotational modes of plastic deformation from macro to nanoscale levels. Let us consider the mechanisms of such plastic deformation in fracture by uniaxial tension.

Fig. 7 Structure of steel 12Cr15Mn9NiCu after stepwise radial shear hot rolling to $\varepsilon \sim 65\%$ and subsequent cold rolling to $\varepsilon \sim 80\%$. **a** The bright field; **c** the selected-area diffraction image, shown are the two zones of reflections with azimuthal misorientations: $[110]$ austenite zone and $[111]$ ferrite zone; **b** the dark-field image in the 002_A austenite reflection; **d** the dark-field image in the 110_ϕ ferrite reflection

3.2 Fracture Surface in Uniaxial Tension of Austenitic Steel Specimens After Various Treatments

The fracture mode in tension of austenitic steel in the initial state and after the complex treatment is schematized in Fig. 9. In the initial steel, fracture starts with the propagation of an opening mode crack (Fig. 9a) and ends with the propagation of a tearing mode crack. Specimens subjected to radial shear rolling are fractured only by the propagation of a tearing mode crack (Fig. 9b).

The fracture surfaces of the steel specimens in the initial state (Figs. 10, 11 and 12) bears witness to the spatial stress state in fracture. After the treatment by radial shear rolling + cold rolling, the specimens under tension are fractured in the plane stress state (Fig. 13).

The fracture pattern of the initial steel specimens in the scale hierarchy of the grain structure is shown in Figs. 10, 11 and 12. Grain-boundary sliding of grain conglomerates (Fig. 10) is accompanied by stochastic microcracking and initiates individual accommodation rotations of individual grains with the formation of micropores in them (Fig. 11). A pronounced dimple relief on the opening mode surface (Fig. 12a) accompanied by the formation of microporosity is a sign of ductile fracture [7, 8].

Fig. 8 **a** The dark-field image of ε -martensite laths in the $0\bar{1}1_\varepsilon$ reflection. **c** The selected-area diffraction image for (a), shown are reflections in the two zones: $[001]$ austenite zone and $[\bar{1}11]$ ε -martensite zone; the arrangement and form of the reflections are given at higher magnification on the left and at the bottom of the image. **b** The dark-field of α' -martensite laths in the $10\bar{1}_\alpha$ reflection. **d** The selected-area diffraction image for (b), shown are reflections in the two parallel zones: $[110]$ austenite zone and $[111]$ α' -martensite zone

The micropore formation is associated with the coalescence of vacant lattice sites under the conditions of plastic distortion at the lattice curvature interstices [9, 10].

Quasi-elastic cleavages in the zones of tearing mode cracks (Figs. 11b and 12b) also exhibit microporosity. This shows the important role of accommodation processes of plastic distortion at the nanoscale level at various fracture mechanisms of the initial austenitic steel. In other words, the self-consistency of rotational deformation modes in a wide range of scales, from macro to nano, causes the high ductility $\delta = 90\%$ in uniaxial tension of the initial austenitic steel.

A high concentration of transition martensite phases in fcc austenite grains with ε martensite and in grains with bcc ferrite zones in α' martensite, which is due to radial shear plus cold rolling of the steel, provides a means for dynamic rotations as a mechanism of structural turbulence (Fig. 13). The phenomenon of structural turbulence in a deformed solid is detailed elsewhere [5, 6]. The presence of nanoscale mesoscopic structural states in the lattice curvature interstices makes it possible to synthesize vortical nanofilaments of a material under the structural turbulence condition. A hierarchy of dynamic rotations on fracture surfaces was numerously

Fig. 9 Schematic of the propagation of the main crack in fracture by uniaxial tension of the specimens of the initial austenite steel (a) and after the complex treatment by radial shear rolling + cold rolling (b)

Fig. 10 Stochastic cracks on the fracture surface in tension of the specimens made of the initial austenitic steel: a opening mode zone; b tearing mode zone

observed [9, 11–13]. However, the mechanism of their formation and their relation to structural turbulence of plastic flow of atoms in lattice curvature zones have not been discussed yet.

Fig. 11 Rotational mode of a grain conglomerate in the initial steel in the opening mode zone accommodated by rotations of individual grains with the formation of micropores in them (a); cleavages of grain conglomerates with the formation of micropores in the tearing mode zone (b)

Fig. 12 Ductile dimple fracture of the initial steel in the opening mode zone (a); quasi-elastic cleavages of grain conglomerates of the initial steel in the tearing mode zone (b)

Fig. 13 Dynamic rotations with the formation of a vortical filamentary structure on the fracture surface of the steel specimens after radial shear and cold rolling: initial (a) and final (b) zones of the main tearing mode crack

3.3 Damping Effect in the Structure of Austenitic Steel After the Treatment by Radial Shear Rolling + Cold Rolling

The damping effect in the heterogeneous internal structure of austenitic steel is studied by scratch testing at the indenter loads 50, 100, and 200 mN. The damping effect is distinct at all the studied loads. Figures 14 and 15 show groove profiles at the initial and treated surfaces of the steel specimens at the indenter load 100 mN.

As can be seen from Fig. 14, the highly plastic initial steel forms a pronounced smooth groove ~170 nm in depth. Bands of plastic shear are visible at the groove surface. A qualitatively different type of grooves is formed at the specimen surface treated by radial shear rolling + cold rolling (Fig. 15). The groove depth is only 15 nm, and its central zone is extruded ~15 nm above the initial surface. No traces of

Fig. 14 Smooth surface of a deep groove (a) and its profile (b) for the steel in the initial state

Fig. 15 Corrugated surface of a groove with the damping effect (a) and its profile (b) for the steel treated by radial shear rolling with subsequent cold rolling

Table 1 Nanohardness H , elastic modulus E^* , and shape recovery factor R at different states of the steel

State	Characteristics		
	H (GPa)	E^* (GPa)	R (%)
Initial	3.7 ± 0.2	229 ± 31	7
After radial shear and cold rolling	4.9 ± 0.3	190 ± 12	14

plastic shear are observed at the groove surface in the treated steel (Fig. 15a). Plastic extrusion is due to nanostructural transformations. The damping effect is very strong, while it is associated with a nonuniform stress distribution at the interface between the groove bottom and the substrate material.

Mechanical characteristics of the austenitic steel in different states are presented in Table 1.

From the tabulated data it is seen that complex radial shear and cold rolling increases the nanohardness H , decreases the elastic modulus E^* , and doubles the shape recovery factor R . The ductile mode of the material extrusion in scratch testing appears not only within the groove, but also in the material on its left and right. All these zones have no traces of plastic shear, but the stress distribution heterogeneity is clearly manifested geometrically at the interface between the surface layer plastically deformed by the moving indenter and the elastic substrate.

3.4 Influence of the Treatment by Radial Shear Rolling + Cold Rolling on the Development of Gigacycle Fatigue and Wear Resistance of Austenitic Steel

Nanostructuring of the austenitic steel and the strong damping effect of the material in scratch testing should increase the fatigue life of the crystal lattice [14, 15]. Moreover, it is known that nanostructuring of a material promotes the development of gigacycle fatigue. This is fully confirmed for the austenitic steel processed by radial shear rolling with subsequent smooth-roll cold rolling. The investigation results are shown in Table 2.

Table 2 Influence of the complex treatment of austenitic steel by radial shear rolling (RShR) + cold rolling (CR) on the characteristics of its fatigue life and wear resistance

Characteristics	State		
	Initial	After RShR	After RShR + CR
The number of cycles to fatigue failure	3 mln		65 mln
Wear rate coefficient, $10^5 \text{ mm}^3/\text{N m}$	8.58	8.25	1.30

Table 2 presents two fundamentally important results. First, the obtained data confirm high-cycle fatigue failure (3×10^6 cycles to failure) in the initial austenitic steel and gigacycle fatigue failure (more than 65×10^6 cycles to failure) after radial shear and cold rolling of this steel.¹ Gigacycle fatigue is usually realized at a significant reduction in external applied stresses [15, 16]; however, the tabulated results are derived at a high external stress. In other words, at high external stresses the transition from high-cycle to gigacycle fatigue can occur due to a specific internal substructure formed at the nanoscale structural level. Secondly, the wear resistance of austenitic steel does not vary after high-temperature radial shear rolling. To achieve this requires additional cold rolling that forms bcc ferrite zones in austenite grains of the steel. The mechanism of formation of wear particles is also associated with fatigue failure of the tribocontact. When the counterbody moves along the flat surface of the material, each mesovolume is first compressed and then stretched. Such cyclic deformation causes fatigue fragmentation of the material and the formation of wear particles.

A heterogeneous hierarchical structure formed in austenitic steel during radial shear and cold rolling effectively functions in tribological conditions. When the counterbody compresses mesovolumes of the heterogeneous austenitic steel, bcc ferrite grains are elastically compressed and hcp ϵ -martensite laths are embedded into the close-packed fcc austenite structure by the mechanism of forward + reverse martensitic transformation. In subsequent tension of this mesovolume, the hcp ϵ -martensite laths are recovered at the interstitial nanoscale structural states, and local stresses in the bcc ferrite grains relax. These processes are reversible and significantly retard plastic deformation, cracking, and the formation of wear particles. We emphasize that this effect is also associated with reversible structural transformations at the nanoscale structural level, where nanoscale mesoscopic structural states can exist at the lattice curvature interstices.

4 Structural Turbulence and Gigacycle-Fatigue Processes in a Solid with Lattice Curvature

4.1 *Structural Turbulence of Plastic Flow at Lattice Curvature and in the Presence of Nanoscale Mesoscopic Structural States at Its Curvature Interstices*

No turbulent plastic flow can exist in a translation-invariant crystal. However, the appearance of lattice curvature zones and of nanoscale mesoscopic structural states at the lattice curvature interstices radically changes the mechanisms of plastic deformation and fracture of solids. This concerns the effect of plastic distortion [11],

¹At the time of publication the cyclic loading experiment for specimens after radial shear and cold rolling is being continued.

formation of a vortical filamentary mesosubstructure [9, 12], structural turbulence, and dynamic rotations [2, 10, 13].

Structural turbulence of plastic flow was predicted when modeling grain-boundary sliding by the excitable cellular automaton method for grain boundaries that lack translational invariance [17]. In this case, consideration was given to lattice curvature at grain boundaries and in near-boundary regions.

Clusters of excess vacancies, that number in ~ 500 at the cluster size ~ 3.5 nm, in localized strain bands were viewed by Matsukawa and Zinkle [18] in the transmission electron microscope column in tension of a gold foil. Tetrahedra of stacking faults form in such vacancy clusters, which can move in the $\langle 110 \rangle$ direction at the migration energy $E_t = 0.19$ eV. Recall that the migration energy of a single vacancy in gold is $E_v = 0.85$ eV. This means that the migration of vacancy tetrahedra is not a diffusion process, but it is associated with structural transformations of nanostructured tetrahedra in the $\langle 110 \rangle$ direction.

As noted above, coherent scattering regions of the size ~ 40 nm appear in the structure of the martensite phase, whose volume amounts to as much as 85.6%. This is a very important nanostructural element, which contributes to the formation of nanostructured grains in austenite and ferrite grains, as demonstrated above in Figs. 1 and 2. Obviously, both in localized strain bands and in nonequilibrium martensite laths in Figs. 7 and 8, a variety of structural configurations can form: highly mobile stacking fault tetrahedra, slow-moving stacking fault octahedra, misoriented nanofragments, nanograins of various composition, including nano carbides, nanocarbonitrides, and others. The presence of the 85.6% nonequilibrium martensite phase in the metastable austenitic steel with coherent scattering regions of the size ~ 40 nm makes possible reversible structural-phase transformations in the steel specimens under cyclic loading. The mechanism of such transformations is discussed below.

4.2 *Influence of the Mechanism of Reversible Structural-Phase Transformations on Gigacycle Fatigue and Wear Resistance Increase in Austenitic Steel After Radial Shear and Cold Rolling*

Since individual volumes of a specimen periodically undergo alternating tension-compression under cyclic loading, this process can be reversible without cracking only subject to the condition of reversible structural-phase transformations. Nanostructured fcc austenite grains and bcc ferrite zones have different yield stresses and are surrounded by the martensite phase, which arises on the basis of interstitial nanoscale structural states in lattice curvature zones that lack translational invariance. Laths of the hcp ϵ martensite in compression can be embedded into the fcc austenite structure, transforming into its close-packed configuration. This governs inelastic compression deformation of the specimen. Laths of the bcc α' martensite in compression

will elastically change the spatial orientation of covalent d bonds and generate local stresses. In tension under cyclic loading, elastic stresses in the bcc ferrite zone will relax and cause a recovery of the hcp ϵ martensite in the austenite grains, implementing inelastic tensile deformation. Such processes of structural transformations are reversible in nanostructured materials [19–21], which determines the damping effect in their structure under cyclic loading, an increase in the gigacycle fatigue life and wear resistance.

A similar damping effect is revealed during scratch testing (Fig. 15). When the indenter moves during scratch testing, the martensite phase is first compressed. Laths of the hcp ϵ martensite transform their structure into the fcc lattice of close-packed austenite. Spatially oriented along the cube diagonals, structural elements of the bcc α' martensite associated with d electrons undergo quasi-elastic compression. After the indenter passes, the ϵ -martensite recovers its hcp structure, and the groove in Fig. 15 exhibits a damping effect. Relaxation of high local elastic stresses in the α' -martensite initiates the groove recovery after the indenter passes. Thus, structural transformations in austenitic steel after high-temperature radial shear and cold rolling are indeed reversible under cyclic external influences.

An increase in the fatigue life of austenitic steel, when loaded below the yield stress of a translation-invariant material, is explained by the nonequilibrium nanostructured martensite structure produced by radial shear and cold rolling and associated with lattice curvature. An important functional role is played by the spatial distribution of ϵ - and α' -martensite laths [22]. This distribution governs a complex spatial distribution of lattice curvature, different nanoscale mesoscopic structural states, the appearance of high local internal stresses, and the possibility of gigacycle fatigue without fatigue cracking at sufficiently high external stresses.

5 Conclusions

The description of a deformable solid as a multiscale hierarchically organized system is usually limited in the literature to a microscale structural level, where strain-induced defects of a translation-invariant crystal lattice are considered. An important role in the problem of the mechanical behavior of materials is played by curvature of the crystal lattice, in whose interstices nanoscale mesoscopic structural states arise [9, 10, 22]. In the present study, such nanoscale mesoscopic structural states were formed in Fe–Cr–Mn austenitic stainless steel using the complex treatment by multistage high-temperature radial shear rolling with subsequent smooth-roll cold rolling to the resulting plastic strain 1.8–2.0.

Such complex treatment causes the formation of nanostructured fcc austenite grains in the steel, bcc ferrite zones, and lattice curvature of nonequilibrium ϵ - and α' -martensite phases in the interstitial space based on nanoscale mesoscopic structural states. Under mechanical loading, the nonequilibrium heterogeneous martensitic structure of the specimens undergoes reversible structural-phase transformations, which are responsible for a nanocrystalline structure of the material, a vortical

filamentary structure and dynamic rotations on the fracture surface, an increased wear resistance, and the transition of high-cycle fatigue life of the initial material to gigacycle fatigue without reducing external applied stress.

This work was performed within the State contract for the Program of Fundamental Research of the State Academies of Sciences for 2013–2020 (project III.23.1.1), RFBR projects (No. 18-08-00221 and 17-01-00691), and Integration Project of the SB RAS No. II.1.

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