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HETEROGENEITY OF PLASTIC FLOW OF ZIRCONIUM ALLOYS WITH A PARABOLIC LAW OF STRAIN HARDENING

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The specific features of plastic-strain macrolocalization at the stage of the parabolic law of strain hardening in samples from industrial zirconium-based alloys are considered. It is shown that in predeformed blanks, zones with a different character of plastic-strain localization are formed. It is also shown that the strain-localization macropattern can be used as a characteristic of the susceptibility of a material to further plastic form-changing, for example, upon tube rolling. The sign of fracture of alloys upon plastic deformation is revealed. The scale effect in the formation of localized plastic-flow zones is shown and studied.

Introduction. In the production of articles from industrial alloys by the methods of plastic metalworking, the problem of estimation and prediction of the plasticity margin that ensures continuation of the process without fracturing a semifinished or finished product arises. This is especially important upon cold rolling, when recrystallization processes are practically impossible. The traditional methods of prediction on the basis of the plasticity indices determined from stress–strain (σ–ε) diagrams do not reflect the general tendency to strain localization during the entire process of form-changing [1]. As is shown by Zuev et al. [1–3], the form of the dependence of the strain-hardening coefficient $\theta \equiv d\sigma/d\varepsilon$ on the strain $\theta(\varepsilon)$ can be used to determine the specific features of the distribution of localized tensile-strain zones.

Among the possible forms of the dependences $\theta(\varepsilon)$, the cases $\theta \approx 0$ (easy sliding or yield sites) and $\theta = \text{const}$ (the stage of linear hardening) are observed rather seldomly, mainly, in single crystals [4]. In polycrystalline materials, the parabolic law of strain hardening $\sigma \sim \varepsilon^{1/2}$ is realized more often, i.e., $\theta \sim \varepsilon^n$ ($n < 0$). In this case (see [1–3]), in the stretched sample, an ordered system of equidistant immobile sites of plastic strain is formed, i.e., the spontaneous separation of the material into deformable and undeformable layers occurs. (In the case $\theta = \text{const}$, such layers are mobile [1–3].) Nicolis and Prigogine [5] noted that lamination of the medium is the result of self-organization in open thermodynamic nonequilibrium systems. In the present work, the strain-localization patterns that make it possible to determine the further plastic form-changing and estimate the technological-plasticity margin of zirconium (Zr)-based alloys, which are widely used in atomic power engineering, are investigated.

1. The Technique and Materials. The industrial zirconium-based alloys É125 and É635, which are utilized for manufacturing the units of the active zone of nuclear reactors, were used as the material for our study [6]. The mass fraction of the elements entering the composition of Zr-based alloys is given in Table 1. The alloy É635 is intended for thin-walled, cold-rolled shell pipes for heat-releasing units; large-diameter tubes and other units are manufactured by welding from the alloy É125, which is produced mainly in the form of sheet rolling. Cold rolling of tubes is accompanied by large strains, and the necessary technological

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TABLE 1

Mass Fraction of the Elements
in Zr-Based Alloys [6]

Alloy	Nb	Sn	Fe
É635	0.95–1.05	1.20–1.30	0.34–0.40
É125	2.45–2.55	—	0.006–0.012

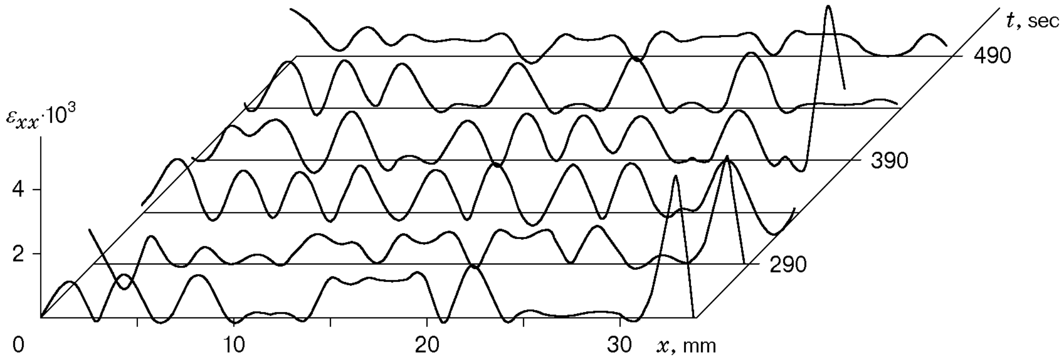

Fig. 1

plasticity is reached by intermediate thermal processing, including quenching and tempering [6]. The choice of the optimum temperature–time regime of heat treatment is determined by stringent strain conditions and is complicated by the fact that Zr and its industrial alloys undergo a low-temperature, hexagonal, close-packed, α-phase–high-temperature, body-centered, cubic, β-phase polymorphic transformation.

The method of speckle interferometry, which allows one to obtain the quantitative strain characteristics, i.e., to determine the displacement-vector field of a flat sample upon tension and calculate the plastic-distortion tensor components (local lengthening ε_{xx}, shear ε_{xy}, and rotation ω_z [3]), is convenient for revealing localized-strain zones [7]. The use of this technique in plasticity analysis under loading of the sample makes it possible to determine the complete pattern of strain evolution and the character of its localization in the deformable sample at different stages of the process [1–3].

2. Strain Localization in Zr-Based Alloys. Figure 1 shows the evolution of the component ε_{xx} upon constant-velocity tension of the sample fabricated from an É635 quenched sleeve tempered after stitching (the blank from which a tube is rolled). In this case, the curve σ–ε contains a site of linear strain hardening, and the localization pattern is wavy. The plastic-strain localization zones displace along the sample axis with a velocity of $V_w \approx 3.5 \cdot 10^{-5}$ m/sec [8] for a wavelength of $\lambda \approx (5.5 \pm 1)$ mm. The successive involvement of the new volumes of material into the process results in quite a highly plastic state of the alloy.

To study the mechanical properties and localizations of plastic flow, samples cut out from a blank rolled from a tube from an initially cylindrical sleeve, whose deformability has been described above, were used. The conic shape of the blank in this case is determined by stronger compression of the metal in the center of plastic strain on the cold-rolling mill for tubes [6]. Figure 2 shows the profiles of the blank and of the place where the samples for mechanical tests (1–5) and speckle interferometry (6–8) were cut out. In mechanical tests, after reaching the yield point, the samples were hardened according to the parabolic law, so that a stationary equidistant system of localized-strain sites was formed in them.

It has been found that such mechanical characteristics as the yield point $\sigma_{0.2}$, the strength $\sigma_{\rm t}$, the elongation up to rupture δ, and the narrowing of the transverse cross section ψ, which were determined from the strain diagrams for five samples, and the hardness HV are almost the same and, therefore, they do not allow one to judge the properties of the blank metal (Table 2). However, the differences between the plastic-strain localization patterns shown in Fig. 3a–c for three samples cut out from different parts of the

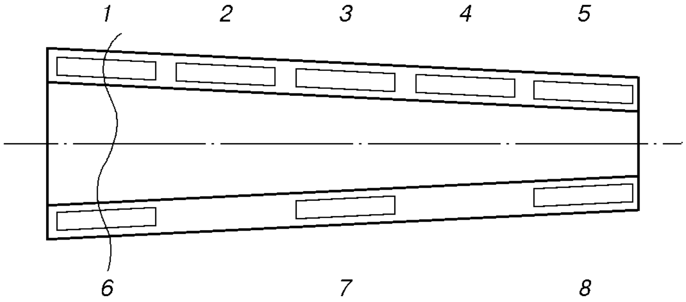

Fig. 2

TABLE 2

Mechanical Properties of the Samples Cut Out from a Tubular Blank (Fig. 2)

Sample No.	$\sigma_{0.2}$, MPa	σ_{t}, MPa	δ, %	ψ, %	HV, MPa
1	331	556	8.6	34.0	2100
2	441	659	4.1	26.0	2200
3	398	706	4.0	25.0	2250
4	352	738	4.0	22.5	2270
5	360	683	3.5	22.5	2300

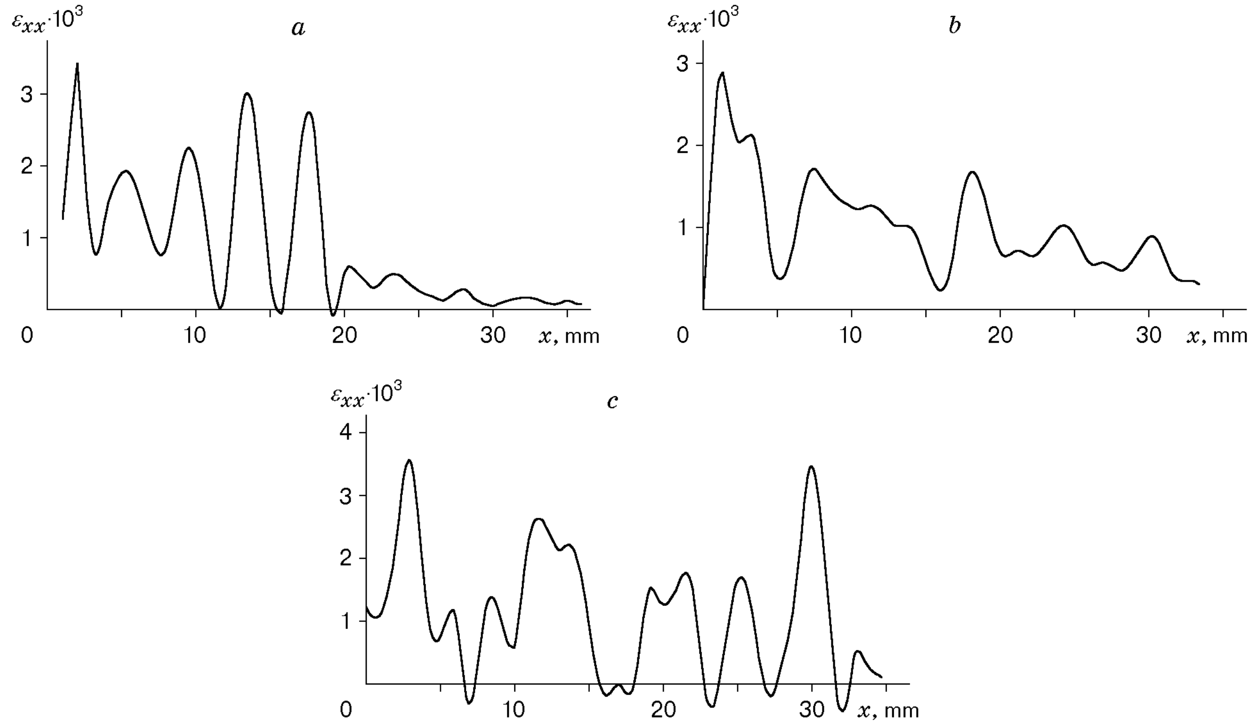

Fig. 3

blank are so considerable that they are suitable for control of the plastic properties of alloy É635 at different stages of the technological process. At the thick end of the blank (small strain upon rolling), a zone in which plastic deformation practically does not develop is distinctly seen (Fig. 3a). In this case, in the blank, there is a dangerous cross section (shown by the wavy curve in Fig. 2) over which fracture upon rolling occurs in some cases. One can assume that at the initial stage of rolling of a tubular blank, a sharp jump in the mechanical properties of the material arises, which is manifested in plastic-flow localization patterns.

As the compression increases, i.e., the blank diameter decreases, the distribution of the localization zones becomes more uniform (Fig. 3b and c). As a result of the plastic deformation of the metal upon cold rolling, the mechanical properties become the same over the volume. This preserves a satisfactory level of technological plasticity of alloy É635 with a noticeable growth of its strength owing to cold working [9, 10].

3. The Sign of Fracture upon Uniaxial Tension of the Samples. In the tension of the samples to the limiting degrees of deformation, the distribution of the localization zones acquires a specific form for this stage. The system of stationary localized-strain sites, which is characteristic of the stage of parabolic hardening, is replaced by one, gradually growing stationary maximum (peak) of the local-elongation component ε_{xx} of the plastic-distortion tensor which indicates the site of subsequent viscous fracture (Fig. 4). From the moment at which the peak appears, almost the entire deformation is localized in this narrow zone of

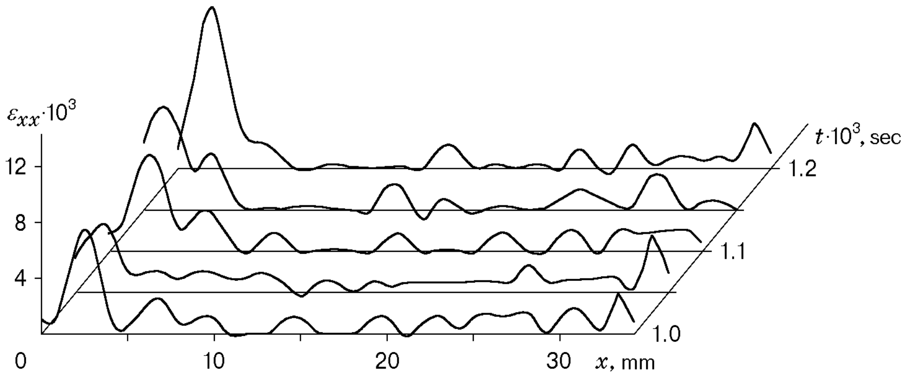

Fig. 4

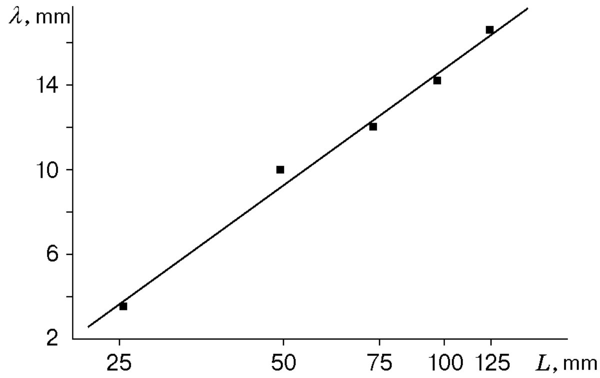

Fig. 5

the sample. The maximum arises at the stress $\sigma < \sigma_t$ and is followed by the formation and visual identification of a deformation neck. This phenomenon is characteristic of all cases of viscous fracture, practically does not depend on the nature of the material [1–3], and can serve as a sign of subsequent fracture.

4. Scale Effect upon Strain Localizations. The problem of the possible effect of the dimensions of the deformable system on the plastic-strain localization parameters (the scale effect) is important for understanding the nature of the deformation-localization phenomenon. The dependence between the sample length L and the spatial localization period λ (the average distance between the localized-strain sites in the distributions similar to those shown in Figs. 1 and 3) allows one to obtain additional information on the nature of such localization zones and the kinetics of their onset. The dependence $\lambda(L)$ was studied on samples of length 25, 50, 75, 100, and 125 mm and of the same 5-mm width; the samples were made from a 1.6-mm-thick uniform sheet from alloy É125. Because the mechanical characteristics of Zr–Nb alloys are sensitive to changes in the structure and phase composition [11, 12], the use of sheet material in this case allowed us to maintain sufficiently homogeneous properties in the experiment. The length of the spatial localized-strain period was determined under the same common deformation $\varepsilon_{\rm com} \approx 2.5\%$ for all the dimensions of the samples.

It has been found that, for the same tension velocity, sample width, and temperature, we have

$$\lambda(L) = \lambda_0 + \alpha \ln L. \tag{1}$$

We note that $\lambda_0 = -21.4$ mm, $\alpha = 7.8$ mm, and the correlation coefficient $\rho = 0.99$ (Fig. 5). The resulting dependence can be interpreted as follows. We assume that the derivative $d\lambda/dL > 0$ is inversely proportional to the probability of onset of the localization site w; in turn, the latter is proportional to the sample length: $d\lambda/dL \sim w^{-1} \sim L^{-1}$. Thus, $d\lambda = \alpha dL/L$, which gives the dependence (1). The proportionality coefficient α has the meaning of the scale unit of the spatial uniformity of plastic deformation in the sample.

It follows from (1) that $\lambda = 0$ for $L_0 \approx 2\alpha \approx 15.5$ mm (Fig. 5), and the quantity L_0 can be regarded as the minimum dimension of the sample in which the plastic-flow localization patterns can still occur periodically. In the samples of length $L \leqslant L_0$, one should expect a uniform distribution of deformation. In particular, this, probably, explains the fact that in the literature devoted to deformation-localization problems, the onset of periodic plastic-strain localization patterns is mentioned rarely (see, e.g, [13]). The possibility of deformation localization in large samples probably causes the scale effect (the dependence of strength on dimensions [14], which plays an important role in engineering).

Conclusions. Thus, the mechanical properties of alloys obtained in standard mechanical tests do not completely describe the properties of materials, especially upon the nonuniform predeformation that is characteristic of technological processes, and should be supplemented by information on the evolution of the localized-strain zones. Here the important role is played by the sample dimension, because, as has been shown above, it is connected with the spatial scale of localization. This circumstance should be taken into account, for example, in estimating the mechanical properties by means of small-sized samples. In laboratory practice, in many cases (especially in studies concerning single crystals), in order to save materials, samples whose working part does not exceed 15–20 mm are used. It is clear that in experiments similar to those described above, on a small-sized sample cut out from the same conic blank, the localization sites could remain unnoticed. At the same time, the plastic-strain site in technological plastic-working processes usually has significant dimensions, and one must to expect the development of significant flow heterogeneity, which determines the deformability of the metal and the degree of readiness of blanks to fracture.

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