Unstable plastic deformation in bimetal

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Abstract. The present work aims the characterization of different types of waves generated upon the plastic flow in single-crystal metals and alloys. The propagation of new types of waves is highlighted during the linear work hardening and easy glide stages of flow. As found, the motion velocity of waves is the inverse function of the work hardening coefficient.

1. Introduction

As shown in [1-3], plastic flow inhomogeneities is a natural process that accompanies all stages of deforming solid. The specific pattern of plastic deformation macrodomains distribution depends on the work hardening coefficient as a function of strain dependence, i.e., $\theta = (1/G) \cdot \frac{d \tau}{d \epsilon}$ ($G$ is the shear modulus, $\tau$ is the shear stress, and $\epsilon$ is the strain), and is attributed to the appropriate stages in the flow curve $\tau(\epsilon)$. The local strain distribution, arising upon the linear work hardening stage, looks like a wave spreading with a velocity $V$ along the extension axis of the deformed sample. Thus, elucidating the nature and properties of such waves is a relevant task.

It is well known that the deformation of metals whether exposed to shock excitation or cyclic loading is induced by the wave processes. A study of elastic waves occurring in elastically deformed solid as well as plastic waves, whose difference is that the latter are associated with plastic flow front propagation in the deforming material [4], is a relevant task. In the last decades, many investigators have paid attention to the generation of the above two classes of waves in deforming solids. The physical properties of these waves together with obstacles leading to their emergency have thoroughly been studied in [5-7]. Nonetheless our recent investigation implies the feasible generation of yet another kind of waves in a solid undergoing the tensile loading at a constant rate [1-3]. In this connection, the essential parameters of localized deformation waves, such as motion velocity and wavelength, are within the scope of this work.

2. Experiments and materials

The plastic flow processes were studied on a series of samples with different structural features and mechanisms related to plastic deformation, using the following face-centered cubic (FCC) single crystals:

1. Pure metal single crystals (cooper and nickel).
2. Cu+10%Ni+6% Sn solid solution single crystals in a quenched state.
3. Alloyed $\gamma$-Fe single crystals (Mn austenitic steels).
4. NiTi equiatomic alloy single crystals.
Anticorrosive bimetal composed of dissimilar metals, such as A 283 Grade C low-carbon steel (BCC) and 301 AISI (FCC) austenitic stainless steel.

The first two samples are subjected to the deformation via the dislocation slip, the third one is within the twinning, and the fourth one undergoes the B2→B19′ phase transformation.

The tensile tests were implemented at a constant straining rate, exploring samples that were cut out of single crystals whose extension axes were oriented in certain directions. The deformation curve of Cu exhibits three ranges that correspond to easy glide (I), linear work hardening (II) and parabolic work hardening (III). Each stage was described by means of the work hardening coefficient. For steps I and II, the acting shear stress $\tau$ is a linear function of plastic deformation $\varepsilon$, i.e. $\tau=\theta\varepsilon$, with a coefficient $\theta$ varying with respect to each stage of flow ($\theta_I<\theta_{II}$). The flow curves of another crystals (Cu + 10%Ni + 6% Sn solid solution and NiTi) show the presence of a sharp yield point and a yield plateau with $\theta=0$. The tension of Hadfield steel (alloyed $\gamma$-Fe) single crystals with a carbon content of ~1% along the [377] axis gives the ordinary twinning in the [211](111) system [8], which manifests itself as the sharp yield point and long yield plateau in the deformation curve. The intermediate [355] orientation occupies the same side of a standard crystallographic triangle. The flow curve of these crystals is thus similar to that of crystals oriented along the [377] axis, giving rise to a sharp yield point and a yield plateau. However, a slight (~13°) deviation of the extension axis from the [111] pole leads to the stage I, where $\theta_I$ is close to zero. Furthermore, the emergence of complementary twinning systems and their interplay with each other [5] go up to $\theta_{II}$ at the stage II in the $\tau(\varepsilon)$ curve.

The comprehensive inspection of each stage from the loading diagrams of bimetals using the work hardening coefficient $\theta=d\sigma/d\varepsilon$ and the constant value $n$ (here it is the exponent of deformation hardening in the Lüdwik equation) evidences the features of the deformation curves for our samples, as below. The transition from the elastic range to the plastic flow is succeeded by the yield plateau followed by the linear-hardening. Then, it observes the Taylor parabolic work-hardening stage that is supervened by the pre-fracture.

A holographic approach elaborated to test severe plastic strains [9] allows the high-precision experimental study of plastically deformed single crystals and polycrystals of metals and alloys. The method enables the instantaneous detection of plastic flow macrodomains. Holograms were collected every 30 s, and a series of $\tau(\varepsilon)$ flow diagrams was acquired contemporaneously for tensile samples, providing abundant information on the deformation kinetics.

A digital image correlation technique [10] enabled displacement vector fields to be measured on a tested surface. The experiments were made through the whole process of plastic flow process from the tensile yield point to the fracture. The approach is suitable for evaluation of the main characteristics of spreading localized plastic deformation zone, among them are spatial $\lambda$ and temporal $T$ periods of the process.

### 3. Results

Plastic deformation is prone to deployment within its stages, and each of them is accompanied by the birth of strain localization [1-3], predominately at certain sites of sample, which are known as local strain macrodomains (with a width of up to 3 mm), alternating with non-deforming slabs of the sample. These strain nuclei generate mobile or immobile sets that are regularly distributed within a sample. The types of strain localization are distinguished with respect to the corresponding stages of flow.

The present work is mainly aimed at considering the spread of local strain macrodomains that arise upon the yield plateau, easy glide (I) and linear work hardening (II). It is worth noting that, a single front moving in the yield plateau is associated with Cu+10%Ni+6%Sn, NiTi and Hadfield steel crystals. In stages I (easy glide) and II (linear work hardening), a sequence of similar macrodomains
propagates along Cu and Ni samples in an orderly manner. The local strain distributions were investigated by plotting the coordinates of the local strain peaks versus the tension time $t$ of Hadfield steel sample with the [355] stretching axis (at $\dot{\varepsilon} =$ const $\approx \varepsilon$) (figure 1). At the yield plateau stage, a single wide deformation macrodomain, separating strained and unstrained parts of the sample, begins spreading from a grip of the testing apparatus. The strain is localized at exactly this macrodomain, whereas other parts of sample remain unstrained. At the onset of the stage I, another strain front secedes from this macrodomain and moves in the opposite direction within the strained region of the sample. Prior to the separation, there is a drop in velocity of the main macrodomain, propagating through the unstrained area of the sample. The situation repeats twice during the stage I (figure 1). A transition from the stage I to the stage II manifests itself by several local elongation peaks as soon as the main strain macrodomain goes towards the grip (figure 1). The stage II exhibits the synchronous movement of equispaced local-strain macrodomains with a constant velocity. Here, the local-strain macrodomain distributions are more complicated as compared to those observed in previous descriptions. In the stage III (parabolic work hardening), there are various stationary localized strain macrodomains in the sample, which imposes further restrictions on consideration of any types of strain localization for the moving macrodomain. The stage II (linear work hardening) evidences the emergence of a typical wave picture in the material. Furthermore, the main wave parameters measured for all the tested materials are found to be equal in magnitude, giving the wavelength of $5 \leq \lambda \leq 10$ mm and the wave propagation rate of $10^{-3} < V < 10^{-4}$ m/s. To give adequate interpretation of established relationships, the low rate of wave propagation as well as the shape of the $V(\theta)$ dependence has to be taken into consideration.

![Figure 1](image-url)

**Figure 1.** Space-time locations of the localized deformation macrodomains in the alloyed $\gamma$-Fe single crystals extended along the [355] axis.
Figure 2 displays the measured propagation rates of local strain macrodomains versus angle $\theta$ for all samples (stages I and II). It appears that

$$V_{aw} = V_{aw}^0 + \Omega / \theta,$$

where $\theta$ is the above specified quantity. Here the constants $V_{aw}^0$ and $\Omega$ are evaluated for the afore described two stages and are found to be different.

For this type of waves, the propagation rate versus the dynamic process variable (work hardening coefficient $\theta$) deviates from those of other waves. It is evident from the experimental data in figure 2 that $V \sim \theta^{-1}$, whereas for plastic waves $[4]$ $V_{pw} \sim (\theta/\rho)^{1/2} \sim \theta^{1/2}$ ($\rho$ is the density of the material). It is therefore obvious that the wave propagation rates found as a function of $\theta$ for studied wave processes diverge from values of plastic waves $[4]$. Furthermore, the motion velocity of waves is inferior to those of other types of waves, being in a range of $10^{-5} < V < 10^{-4}$ m/s, which is 4-5 orders of magnitude lower than those for plastic waves.

Based on the aforesaid, one can definitely affirm that a new type of waves is distinguished in a strained material. In what follows, their origin will be thoroughly discussed. Using the dimensional work hardening coefficient, $\theta_d = d\tau / d\varepsilon$ (for samples made of the same material, e.g. $\gamma$-Fe single crystals), (1) can be rewritten as

$$V_{aw} = V_{aw}^0 + J / \theta_d.$$  

Here the coefficient $J$ is measured in units of $\text{Pa} \cdot (\text{m/s}) \equiv \text{W/m}^2$, which is attributed to the energy flux through the sample from the loading machine. It appears from the data measured for the same single crystals that $J = 3 \times 10^4 \text{ W/m}^2$. The same value is found from the outward relation of $J = \tau \cdot V_m$ ($V_m = 10^3 \text{ m/s}$ is the motion rate of the grip of the testing device and $\tau$ is the acting stress). It is thus established that the calculated $J$ values are close to measured ones in an increasing order of magnitude. Therefore, the rate of the process depends on the energy flux passing through the sample. The easy glide with a minimum shear stress results in the lowest coefficient $J$.

Plastic deformation of bimetal arises from the Lüders band interface. As found, a stationary system of waves with a distance $\lambda = 4$ mm between them is observed upon the parabolic stage at the motion rates $V_1 = 0.8 \times 10^{-4}$ m/s and $V_2 = 2.3 \times 10^{-4}$ m/s of the above observed waves (figure 3), the linear deformation strengthening with a spatial period $\lambda = 4$ mm and their propagation rate $V_{aw} = 6 \times 10^{-5}$ m/s.

**Figure 2.** Propagation rates of localized strain nuclei versus work hardening coefficient in stages I (a) and II (b).

**Figure 3.** Visualization of $X(t)$ kinetic plot of strain localization zone along sample axis as functions of time at the yield plateau.
4. Conclusion
The examination of instantaneous strain distribution patterns suggests that the deforming medium [11, 12] will separate spontaneously into active nuclei of localized plasticity alternating with regions where no plastic deformation occurs at a given instant of time. Plastic flow macrolocalization has the following qualitative and quantitative characteristics:
- Regions of inhomogeneous plastic flow behavior have a macro-scale 5…10 mm, which is commensurate with the size of samples used for mechanical testing.
- The local strain nucleus can be viewed as a meso-defect responsible for plastic flow development on the macro-scale level; with growing deformation level, within the nucleus there occur regular increments in the local elongations, shears and rotations.
- The sum of contributions of all the localized plasticity nuclei determines increment in the total deformation per loading step.

The matching of experimental data obtained for a wide range of materials (single-crystals and BCC, FCC polycrystals of metals and alloys) assigns a specific meaning to our contention that the occurrence in the deforming solid of macrolocalization nuclei evolving in a regular manner is a universal feature. It is found that the main regular features of local strains distributions are due to a changeover in the deformation stages with increasing extent of loading. The same scenario is observed for all test samples in both single-crystal and polycrystalline state, no matter what crystal lattice type, chemical composition or micro-mechanism involved in the plastic deformation. A one-to-one correspondence is established between the emergent local strains pattern on the one hand and the work hardening law acting at the given flow stage on the other.

The regularities highlighted within the evolution of plastic flow macrodomains that take place in strained solid coincide with concepts put forward in [13]. As found, these are related to the trend of a deforming system to self-organization. According to [11, 12], self-organization is a process where a system is capable of attaining spatial, temporal or functional inhomogeneity, without needing control by any external agent. These regular features must be considered for the adequate interpretation of experimental data and the development of the theory of solid plasticity.

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