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EFFECTS OF STRAIN RATE ON WORK HARDENING OF HSLA AND Ti-IF STEELS

Notation:

A	– constant
b	– Burger's vector, nm, μm, mm, cm
B	– a coefficient dependent on the chemical composition and strain rate and directly correlates to the stage of austenite decomposition
$c_{1...5}$	– Zerilli – Armstrong model's constants
d	– grain size, μm, mm
D	– Cowper–Symonds coefficient
f	– the volume fraction of precipitates (for Nb-microalloyed steels: $f = 1.13 \cdot 10^{-4}[Nb]$)
G	– shear modulus, MPa
K	– microstructural stress intensity, MPa
k	– Boltzmann's constant, J/K
K_p	– dispersion strengthening contribution coefficient
k_s	– constant associated with subgrain boundary strength
Δl	– the distance between dislocation barriers
M	– orientation factor
$[M]$	– chemical composition
n	– hardening exponent
P	– constant dependent on ferrite grain size, MPa

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p – Cowper–Symonds coefficient
T – temperature, K
T_D – deformation temperature, ^{o}C
T_S – austenite-ferrite transformation start temperature, ^{o}C
v_o – the limiting dislocation velocity, cm/s
w – constant (for most of the Nb-steels: $w = 0.3$)
W_0 – energy characterizing the thermal activation process, J
X – volume of newly formed ferrite
ε – strain
ε_a – accumulated strain
$\dot{\varepsilon}$ – strain rate, s^{-1}
κ – the mean planar intercept diameter of the precipitated particles, nm
ν – Poisson's coefficient
ρ – total dislocation density, m^{-2}, cm^{-2}
ρ_0 – dislocation density of annealed material, m^{-2}, cm^{-2}
σ – stress
σ_a – athermal stress; MPa
σ_d – dynamic stress; MPa
σ_p – precipitation strengthening; MPa
σ_s – quasi-static stress; MPa
σ_{sg} – substructure strengthening; MPa
θ – misorientation angle, ^{o}

1. INTRODUCTION

The microalloyed HSLA and Ti-IF steels aimed for automotive industry have been of great interest for many years for the sake of their promising high strain rate performance with a high-energy absorption. The dynamic behavior of such materials is still the subject of extensive research. The steel hardening results from different strengthening mechanisms. The situation is more complex, whenever high strain rate of deformation is applied. It is difficult to describe material behavior during deformation at high strain rates using general, "universal" method of analysis. For example, mechanical twinning as a plastic deformation mechanism should be considered in addition to dislocation motion. The proper description of correlations among high strain rate and mechanical and microstructural behavior of the material are very important issues, not only because of the final products implementations, but also with respect to the manufacturing process. It has been determined that in many metal forming processes (drawing, rolling of long products, forging) the strain rates in the deformation zones are very high (up to 4000 s^{-1}). It is, therefore, clear that the influence of strain rate in modeling of such metal forming processes should be taken into account. There is a number of ideas that have been proposed and successfully employed to describe the mechanical behavior and microstructural evolution as a function of strain rate. The purpose of this paper is to present the mechanical and microstructural response of microalloyed steels subjected to dynamic (i.e. at very high strain rates, not considering mass forces) deformation in cold forming conditions.

2. EXPERIMENTAL PROCEDURES

Axisymmetrical compression tests were performed in order to study the effect of high strain rates on the material mechanical and microstructural behavior. The compression tests were carried out under various strains and strain rate conditions at room temperature. The evolution of the microstructure, as well as the mechanical response of the material affected by the strain rate, strain and generated heat, were observed and analyzed. The tests were performed using the tensile-compression testing machine (strain rate = 0.001 s^{-1}), drop-weight (strain rate = 200, 400, 600, 800 s^{-1}), Schenck servo-hydraulic compression testing machine (strain rate = 150 s^{-1}) and the Split Hopkinson Pressure Bar (strain rate = 2500 s^{-1}). The tests scheme and the basic chemical compositions of the investigated steels (in wt. %) are shown in Figure 1a and Table 1, respectively.

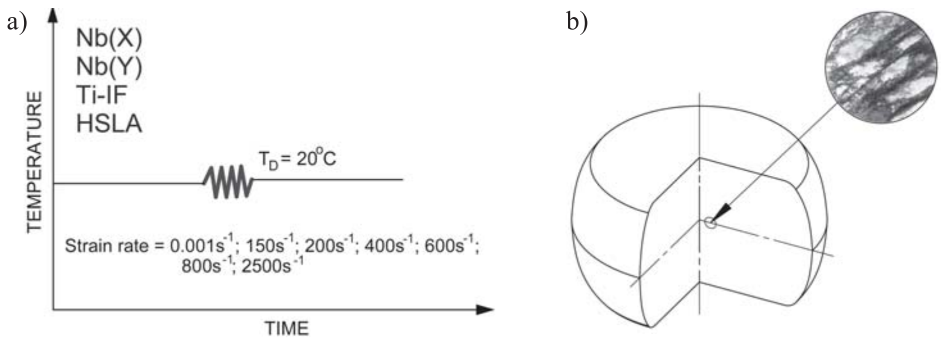

Fig. 1. *The scheme of compression tests (a), the location of TEM observations (b)*

Table 1. The chemical compositions (in wt. %) of investigated steels

Steel	C	Mn	Si	Al	Ti	S	N	P	Nb	B
Ti-IF	0.0022	0.112	0.009	0.037	0.073	0.006	0.0034	0.009	–	–
HSLA	0.067	1.3	0.34	0.037	0.024	0.006	0.0054	0.015	0.076	–
Nb(Y)	0.07	1.36	0.27	0.02	0.031	0.006	0.0098	0.015	0.067	0.003
Nb(X)	0.1	1.5	0.25	0.029	0.113	0.009	0.011	0.014	0.05	0.032

The transmission electron microscopy (TEM) was used to examine the microstructures of investigated steels after dynamic compression tests. The place of taking a specimen for TEM is shown in Figure 1b. Finally, hardness measurements (HV5) were carried out using Zwick hardness tester.

3. RESULTS

In the present analysis, the application of a wide range of strain rates makes possible to evaluate the strain rate sensitivity of studied materials and its influence on microstructure evolution and, in consequence, on mechanical properties. Based on the results of present study it can be stated that the change in loading conditions from quasi-static to dynamic causes significant discrepancies in microstructure evolution and in the hardening of final material. These effects are reflected in the final mechanical properties (e.g. hardness). The discrepancies between microstructures of materials deformed under quasi-static and dynamic conditions can be observed in all of investigated in present study steels.

a) b)

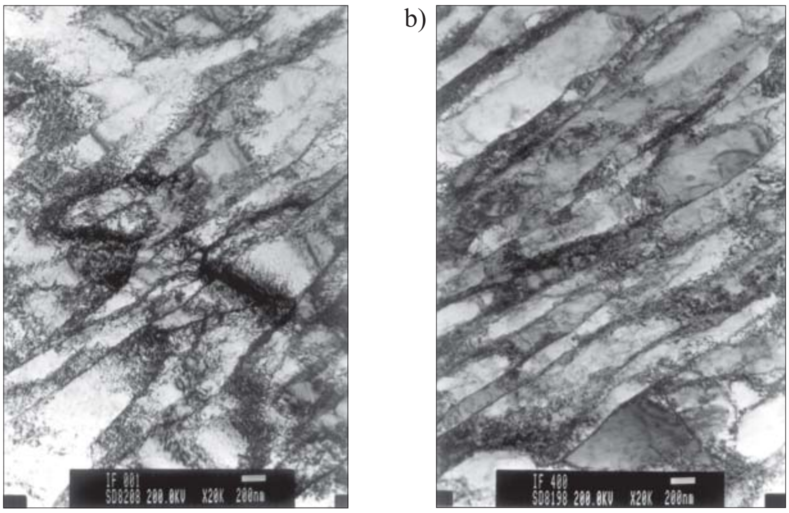

Fig. 2. *Transmission electron micrographs of Ti-IF steel after quasi-static (a) and dynamic (b) compression tests at room temperature. Strain rate 0.001 and 400 s^{-1}, respectively*

In the samples deformed under quasi-static conditions (Figs 2a and 3a), the tendency to form the cell structure is visible, although the cell walls are weakly outlined. The higher and lower dislocation density areas can also be seen. It can be noticed that microalloyed steel (Fig. 3a) shows more scattered dislocation structure. In the case of Ti-IF steel (Fig. 2b), increased strain rate and strain cause more refined and more clearly oriented structure. Observed structures are refined due to decreased spacing between the lamellar boundaries. Dislocation cells observed in the Nb(Y) steel (Figure 3b) are smaller and more clearly outlined comparing to the Ti-IF steel samples. However, it should be mentioned that in this case it is a result of coarsed initial structure of Ti-IF steel (100 μm comparing to 12 μm grain diameter of Nb(Y) steel).

a) b)

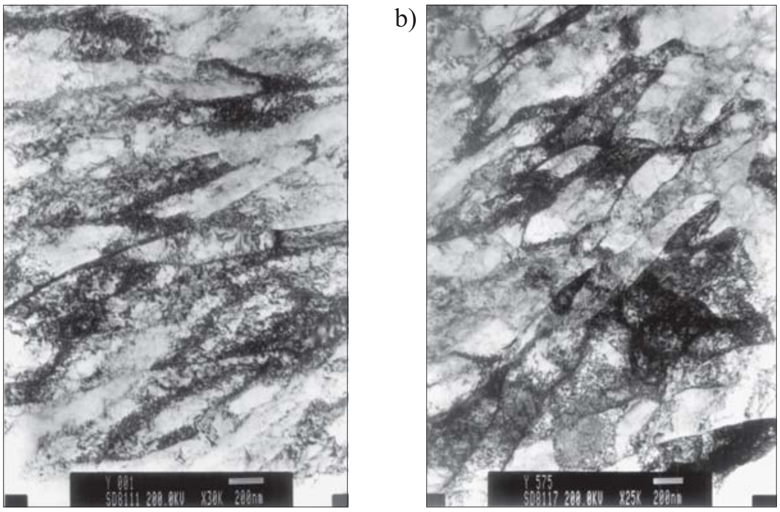

Fig. 3. *Transmission electron micrographs of Nb(Y) steel after quasi-static (a) and dynamic (b) compression tests at room temperature. Strain rate 0.001 and 600 s^{-1}, respectively*

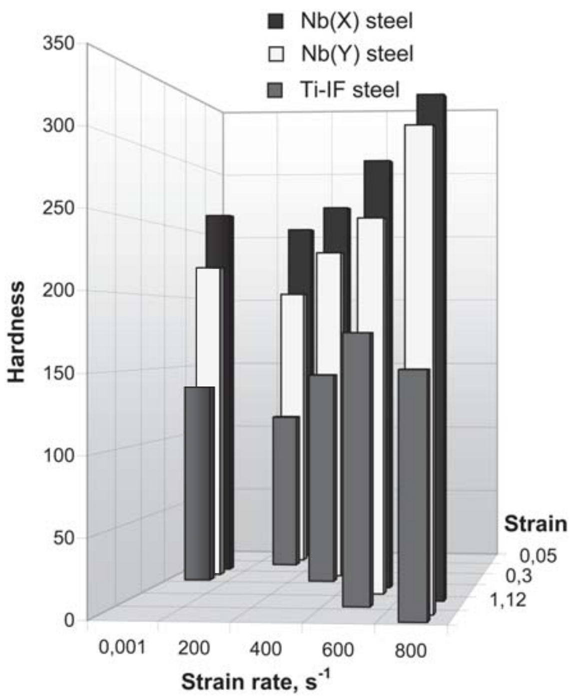

Fig. 4. *Variations of hardness with strain and strain rates at room temperature. Notice increased hardness for higher strain rates [1]*

The results shown in Figure 4 represent very interesting behavior of the investigated steels. It can be seen that influence of strain rate on mechanical properties is observed in each case of the tests, despite of history of deformation process. It can be observed that applied high strain rate directly influences the hardness. The chemical compositions of investigated steels do not change this correlation significantly. It can not be clearly stated, at present stage of study, that high strain rate is the only reason causing observed differences in hardness measurements. However, we can suppose that high strain rates affect the dislocation activity and finally influence on strengthening mechanisms.

4. MODELLING

One of the fundamental principles of thermomechanical processing is the simultaneous use and superposition of microstructural evolution and mechanical behavior of the deformed material. The degree of generated defects, cracks and phase transformation is decisive for the final microstructure to be continuous, partially fractured, or fully fractured. It is well known that the yield stress in the metallic materials increases steeply above a critical strain rate (4000 s^{-1} for mild steel) [2]. Following Meyers, the principle short–range barrier is the Peierls–Nabarro stress, which is especially important for b.c.c. structures. For f.c.c. and h.c.p. metals, dislocation forests are the primary short-range barriers at lower temperatures. The different nature of these barriers is responsible for the major differences in strain rate sensitivity between f.c.c. and b.c.c. phases.

The materials models are frequently used for extrapolation of the flow stress to very high strain rates. However, it is generally very difficult to determine mechanical response of the material due to the complex microstructural behavior that occurs during high strain rate loading. There are several models enabling reasonably good description of the flow stress which take into consideration strain rate effects.

The most frequently cited in the literature are:

- Cowper–Symonds [3]
- Johnson–Cook [4]
- Zerilli–Armstrong [5]
- MTS (mechanical threshold stress) [6]

The Cowper–Symonds equation is one of the simplest model of material behavior at different strain rates

$$\frac{\sigma_d}{\sigma_s} = 1 + \left(\frac{\dot{\varepsilon}}{D}\right)^{\frac{1}{p}} \tag{1}$$

where:

D and p – Cowper–Symonds coefficients,
$\dot{\varepsilon}$ – strain rate,
σ_d – dynamic stress,
σ_s – quasi-static stress.

One of the limitation of this model is that it does not include temperature and strain hardening effects. The calculated coefficients in Eq. (1) are presented in Figure 5b.

The results of calculated and measured flow stress (Fig. 5b) reflect the behavior similar to these observed by Campbell and Ferguson (Fig. 5a). For lower strain rate a linear stress dependence is observed. For higher strain rate (above 10^3 s^{-1}) significant increase in yield strength is exhibited.

The Johnson–Cook empirical constitutive equation is based on experimentally determined coefficients. This approach does not incorporate the structural response of deformed material and can be replaced by dislocation-based models such as the Zerilli–Armstrong and the mechanical threshold stress (MTS).

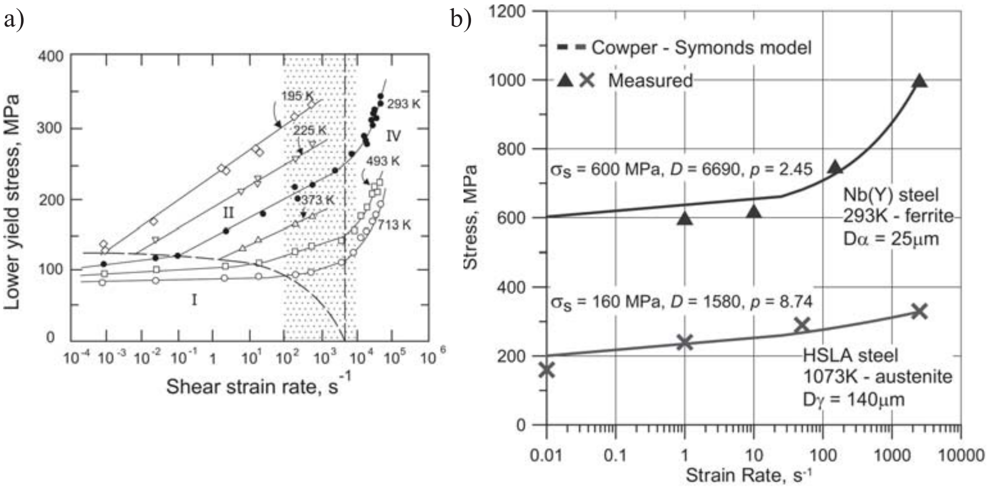

Fig. 5. *Experiment data by Campbell and Ferguson [2] (a), comparison of calculated (Cowper–Symonds model) and measured flow stress of microalloyed steels [1, 7] (b)*

In the present study, the Zerilli–Armstrong model has been chosen and modified. This model is one of the simplest dislocation mechanics formulations and was proposed in 1987 by Zerilli and Armstrong. Authors, based on the idea of the thermally activated motion of dislocations, proposed the microstructurally based constitutive model that is adequate for two-phase structure and dynamic loading. Armstrong also pointed out the importance of certain athermal effects due to grain size and related deformation twinning to be treated separately and explicitly.

The nature of the dislocation interactions leads to different forms of equation for the f.c.c. and b.c.c. structures. For b.c.c. metals, the motion of dislocations is governed by Peierls–Nabarro stress and it is resulting from the interaction produced by the overall lattice potential. This leads to a little increase in the flow stress with strain. For f.c.c. metals the motion of dislocations is constrained by their mutual intersections leading to substantial strain hardening.

Complete constitutive relation for f.c.c. metals is

$$\sigma = \sigma_a + c_2 \varepsilon^{1/2} \exp\left(-c_3 T + c_4 T \ln \dot{\varepsilon}\right) + K d^{-1/2} \tag{2}$$

For b.c.c. metals an additional empirical term of the form $c_5 \varepsilon^n$ is added to describe the strain dependence

$$\sigma = \sigma_a + c_1 \exp\left(-c_3 T + c_4 T \ln \dot{\varepsilon}\right) + c_5 \varepsilon^n + K d^{-1/2} \tag{3}$$

To increase an accuracy of the model for b.c.c. material, the precipitation and substructure strengthening description is incorporated into it. This is especially important in the case of microalloyed steels. Hence, Eq. (3) has a form

$$\sigma = \sigma_a + c_1 \exp\left(-c_3 T + c_4 T \ln \dot{\varepsilon}\right) + c_5 \varepsilon^n + K d^{-1/2} + \sigma_p + \sigma_{sg} \tag{4}$$

All constants and coefficients of modified Zerilli–Armstrong model of Ti-IF and microalloyed steels (equations (2) and (4)) are summarized in Table 2.

Table 2. Constants and coefficient of modified Zerilli–Armstrong model

Constant	Value		Literature
	Ferrite	Austenite	
σ_a, MPa	$\sigma_a = f([M],\rho)$		
	0 0	35 45	[1, 7]
c_1, MPa	1100 1080	– –	[1, 7]
c_2, MPa	– –	1356 1100	[1, 7]
c_3, K^{-1}	$c_3 = \frac{k}{W_0}\ln\left(\frac{b\rho v_0 \Delta l}{M}\right)$		[5]
	$\rho = \rho_o B \exp\ (\varepsilon X)\left(\frac{A}{T-273}\right)^s$		[8]
	$\left(\frac{A}{T-273}\right)^s$ represents an effect of annihilation process of dislocation activation during recovery process and depends on the parameters reflecting the quality of austenite and ferrite microstructure i.e. grain size, inhomogeneity of grain size distribution		
	0.00390 0.00558	0.00245 0.00185	[1, 7]
c_4, K^{-1}	0.000280 0.000225	0.000178 0.000150	[1, 7]
c_5, MPa	336 131	– –	[1, 7]
K, MPa·mm$^{0.5}$	5 18	23 5	[1, 7]
n	0.298 0.228	– –	[1, 7]
σ_p, MPa	$\sigma_p = \frac{0.538Gbf^{1/2}}{\kappa}\ln\left(\frac{\kappa}{2b}\right)$	–	[9]
σ_{sg}, MPa	$\sigma_{sg} = k_s l^m$		[10]
	$k_s \frac{1.6G\sqrt{b\theta}}{2\pi(1-\nu)} \cong 2.1\sqrt{\theta}$	–	[11]
	$\sigma_{sg} = \varepsilon_a P\left(\frac{T_S}{T_D}-1\right)^w$		[12]

Equation describing the precipitation strengthening (σ_p) by modified Ashby–Orowan equation (see Tab. 2) is difficult to manipulate because it is necessary to estimate the size of the particles. The example of calculations of the precipitation strengthening is shown in Figure 6. In most practical metal forming processes of microalloyed steels, the effect of precipitation can be described correctly by using the dispersion strengthening contribution coefficient K_p of the alloy [13]. However, the precipitation of niobium nitrides and carbides as well as complex titanium-niobium nitrides or carbonitrides can be significantly modified because of the change of the stored energy in dynamically deformed steel.

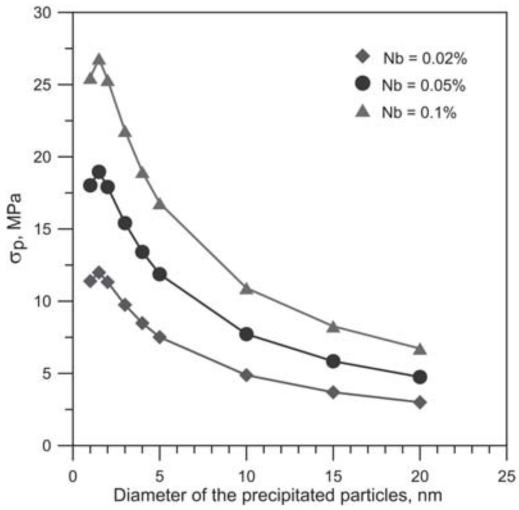

Fig. 6. *Effect of Nb content and particles diameter on the precipitation strengthening*

The limitation of dislocation motion and potentially increased significance of mechanical twinning lead to a better (from point of view of strengthening) dislocation arrangement in the final product. For example, the annihilation process of dislocations, typical for large plastic deformation, is reduced. As it was concluded by Leslie [14], in ferrite structure, strained at low temperature or at high strain rates, the edge components of dislocations can move at higher rates than screw components. Hence, elongated segments of the latter remain in the structure. These screw components are unable to cross-slip at low temperatures or high strain rates. Therefore, due to the lack of cross-slip, no cell structure is formed. This behavior can slow down the formation of the substructure, resulting in the increase of dislocation strengthening and finally increases hardness. Presented observations occur in wide range of strains (see Fig. 4).

The comparison of calculated and measured flow stresses is shown in Figure 7. Presented results reflect the ability of the modified Zerilli–Armstrong model to calculate the mechanical response of HSLA and Ti-IF steels under dynamic and quasi-static loading conditions. It can be noticed that the comparison indicates a good agreement between calculated and measured flow stresses. The presence of microalloyed elements and absence of solute C, N in fully stabilized steels leads to significant differences in the mechanical behavior between Ti-IF and HSLA.

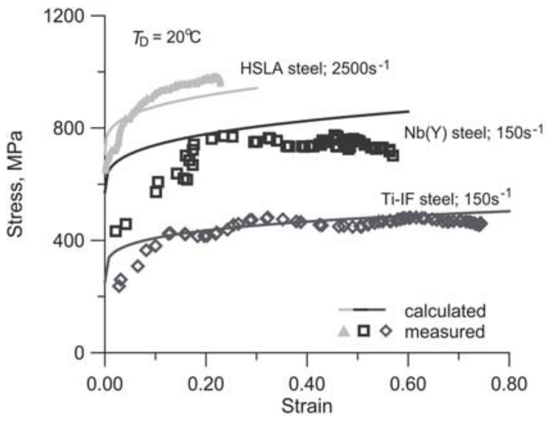

Fig. 7. *Flow stress as predicted by model and measured during compression tests [15]*

The various metallurgical phenomena that influence the microstructure evolution and mechanical response of microalloyed steels under dynamic loading are difficult to determine experimentally. However, the preliminary analysis of present study enables to plan a very precise matrix of experiments in support of future modeling of the mechanisms of plastic flow and microstructure evolution under the dynamic loading conditions.

5. CONCLUSIONS

Basing on the analysis of this research the following observations can be formulated:

– High strain rate has strong influence on mechanical properties (increased hardness) of all investigated materials.
– Combining the microstructural evolution and modified Zerilli–Armstrong models allows accounting for inhomogeneities of the material mechanical and microstructural behavior during deformation with very high strain rates.
– Present studies showed that application of high strain rate causes more intensive cell structure formation comparing to statically deformed structures.
– Effect of high strain rate on ultrafine structures is more complex comparing to the "typical" structures. However, applying dynamic deformation offers new possibilities of improving the mechanical properties and ductility of final product.

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