

Letter to the Editor

RELATION BETWEEN MICROSTRUCTURE AND COERCIVE FORCE OF PLASTICALLY DEFORMED Cr-Mo-V STEEL ¹⁾

СВЯЗЬ МЕЖДУ МИКРОСТРУКТУРОЙ И КОЭРЦИТИВНОЙ СИЛОЙ ПЛАСТИЧЕСКИ ДЕФОРМИРОВАННОЙ СТАЛИ Cr-Mo-V

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The effects of magnetic hysteresis are strongly influenced by dislocations, precipitates and other lattice imperfections. In our previous studies of 2.5 Cr-Mo-Nb steels the density and arrangement of dislocations were found to be decisive factors for the coercive force and the initial permeability [1, 2]. The question arises whether similar low-alloy steels exist in which also precipitates would play a substantial role. Therefore, the Cr-Mo-V steel was chosen because vanadium is known to form very disperse carbides and a high density of precipitates can be expected.

The samples (rods, diameter 10 mm, length 100 mm) used for the study had the composition (wt. %):

C	Mn	Si	P	S	Cu	Ni	Cr	Mo	V	Al	N
0.14	0.58	0.31	0.013	0.01	0.29	0.34	0.66	0.46	0.30	0.012	0.01

and were subjected to the following heat treatment:

960 °C/1 h/oil/720 °C/7 h/air.

They were deformed in tension at room temperature up to plastic strains of 1%—10%. The deformation was macroscopically non-homogeneous along the length of the samples, especially for low strains. Thus, an undeformed part of the sample with a 1% strain could be used for electron-microscopical studies to define the initial state of the undeformed samples.

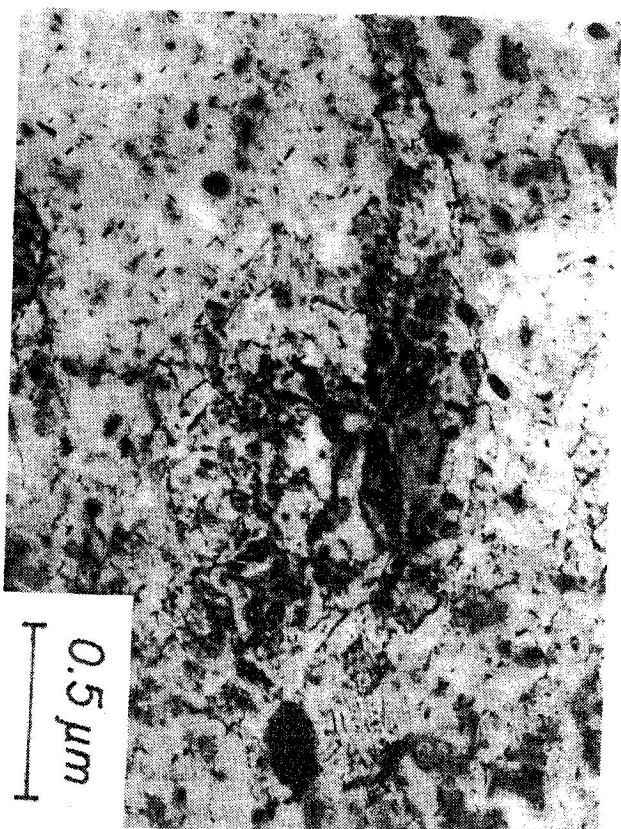


Fig. 1. Minimum dislocation density in V-steel, initial state.



Fig. 2. Maximum dislocation density in V-steel, initial state.

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Fig. 3. Dislocation configuration at 8% strain, V-steel.

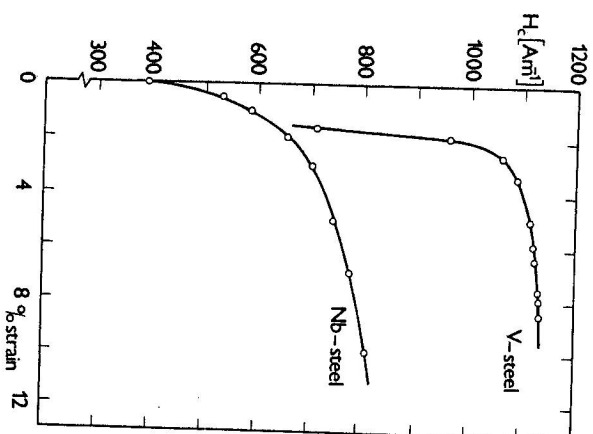


Fig. 4. Coercive force H_c vs. strain for V and Nb steels.

The details of the preparation of the samples, the determination of the microstructure by the electron microscopy and magnetic measurements were the same as described in [1, 2].

The character of the dislocation structure of the undeformed state is shown in Figs. 1, 2. Even in this case the dislocation density ρ is non-homogeneous, varying from the minimum value $\rho = 280 \times 10^{12} \text{ m}^{-2}$ (Fig. 1) to the maximum value $\rho = 430 \times 10^{12} \text{ m}^{-2}$ (Fig. 2), the average value being $\rho = 349 \times 10^{12} \text{ m}^{-2}$. It is seen that the dislocations do not form tangles but there are only regions with higher dislocation densities, where single dislocations can easily be distinguished.

At an 8% strain the dislocation density reaches the average value of $640 \times 10^{12} \text{ m}^{-2}$ (Fig. 3). In this case we can already notice a certain degree of the tangling of dislocations so that inside the grains there are tangles of many dislocation dipoles separated by regions of low dislocation densities.

The dependence of coercive force vs. strain is given in Fig. 4. The upper curve refers to the present studies on V-steel, the lower concerns the previous measurements on Cr-Mo-Nb steel [3]. To compare the microstructure of both systems of steels in the undeformed state, we shall consider the sample of Nb-steel treated at $1050^\circ\text{C}/2\text{h}/\text{air}$ (i.e. sample No. 62 in [1, 2]) and the sample of V-steel in the initial state as described above.

Table 1
Coercive forces, dislocation and precipitate densities in Nb and V steel samples

	Coercive force H_c in Am^{-1}	Average dislocation density in 10^{12} m^{-2}	Average precipitate density in 10^{19} m^{-3}
Nb-steel [2]	885	$\sim 340^*$	$\sim 7^*$
V-steel	707	349	115

* The values of the dislocation and precipitate densities given in [2] for the Nb toroidal samples had to be corrected as in [2] the conventional foil thickness was considered for density calculations instead of the real one.

The results of our interest are presented in Tab. 1. From the table we can see that the values of coercive forces and dislocation densities are comparable, whereas the precipitate densities differ by more than one order of magnitude. Already these facts confirm that in both systems the coercive force is controlled mainly by dislocations. As to the role of precipitates we can say that the high density of precipitates in V-steel samples causes blocking of dislocations. This follows from the fact that, in spite of annealing, the recovery of V-steel samples is the same as that of Nb-steel air-cooled samples.

Regarding the magnetic properties, the blocking of dislocations in V-steel samples can also explain a higher increase of the coercive force with strain (Fig. 4, upper curve) in comparison with Nb-steel samples (Fig. 4, lower curve), because the precipitates hinder dislocation tangling. This is in agreement with our previous observation that, in general, the tangling of dislocations lowers the coercive force [2].

The higher values of the coercive force for the V-steel samples (Fig. 4) are obviously caused by higher dislocation densities in the samples, the average dislocation densities varying from 349 to $640 \times 10^{12} \text{ m}^{-2}$ as compared to the corresponding values 13.6 to $371 \times 10^{12} \text{ m}^{-2}$ in Nb-steel (see Tab. 3 in [2]).

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