THE FINE STRUCTURE OF HADFIELD STEEL

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About 100 years ago an austenitic manganese steel (G13) was suggested by Hadfield [1] exhibiting a high capacity for strengthening during cold deformation and wear resistance, later called Hadfield steel. In numerous textbooks on metal science the high mechanical properties of Hadfield steel are explained by the breakup of mosaic blocks [2] and formation of deformation martensite [3]. In addition, among the reasons for significant strengthening of this steel during deformation, development of carbon segregation [4], reaction of carbon atoms with dislocations [5] and packing defects [6], features of the reaction of alloying element atoms [7], mechanical twinning (during creep resistance tests [8] and with high-rate deformation by impact [9]) are noted. Thus, there is no single idea about reasons for the high tendency of Hadfield steel toward strengthening.

In work carried out by us previously [10], a study was made of Fe-Mn-C steel containing 0.5% C and up to 35% Mn. A diagram was plotted for diffusionless metastable equilibrium phases, and it was shown that in austenite of manganese steel close to the metastable equilibrium boundary of γ- and ε-phases \( T^\gamma\epsilon \) the deformation mechanism is mechanical twinning providing high ductility for these steels with retention of good strength properties.

The aim of this work is a study of the fine structure of Hadfield steel after quenching and deformation by tension and comparison of the data obtained with those obtained previously.

The study was carried out on specimens of steel containing 13% Mn and 1.1% C melted in an open induction furnace. In addition, for comparison a study was made of steel containing the same amount of manganese and carbon, but melted in a Balzer-type vacuum furnace. The content of gas impurities in the vacuum melted steel was about half that in the openly melted steel (0.005 and 0.001% O; 0.018 and 0.034% N, respectively). Ingots were annealed at 1150°C and rolled into bar from which specimens were turned for study. Specimens were water quenched from 1150°C. Mechanical testing of specimens with a gauge length diameter of 3 mm was carried out in an MTS machine. Disks were cut in an electric-arc machine from quenched and also from the gauge length of deformed specimens, which were ground and electropolished in standard chromophosphoric electrolyte. Polished disks were studied in a Neofot-2 microscope and thin (foil) specimens were studied in an EM6G microscope with an acceleration voltage of 100 kV.

The results of metallographic study showed that the structure of quenched specimens consists of coarse austenite grains without signs of the presence of martensite or traces of plastic deformation.

Results of electron-microscope study of quenched and deformed specimens of open-melted steel are presented in Fig. 1. A characteristic feature of quenched steel specimens is the clearly defined dislocation structure. Dislocation loops, accumulations, and interweaving of dislocations are observed in microphotographs (Fig. 1a). In the electron-diffraction patterns for these regions only reflections from the austenite fcc-lattice are observed. In individual areas of foil (as a rule, at crack-type defects) on a background of the dislocation structure, numerous extended rectilinear bands are observed (Fig. 1b). In the electron-diffraction patterns for these regions with an orientation of \( (110)^A \) (Fig. 1c), apart from reflections from the fcc-lattice, additional reflections located along the \( (111) \) direction at a distance equal to one third of that between neighboring matrix reflections are observed. This indicates that additional reflections apply to twins of the system \( (111) \langle 112 \rangle \). Continuous diffuse traces in electron-diffraction patterns along the \( (111) \) direction point to
the small thickness of twins, which is 4-20 nm. Twins are lit up on dark field images in these reflections. In the often encountered foil orientations of the type (100)*A (Fig. 1d) additional twin reflections are located not along the <111> direction, as would be for twins of the system {111}<112> in an fcc structure, but along directions close to <110>. In considering the reverse lattice of the fcc structure twinned by the system {111}, it has been established [11] that this "incorrect" location of twin reflections may be caused by small local bending of the foil.

The structure of quenched vacuum-melted steel is similar to the structure of quenched open-melted steel.

As a result of deformation by tension, specimens of Hadfield steel experience uniform elongation and they break without forming a neck. The mechanical properties of open-melted steel are \( \sigma_F = 1050 \text{ MPa}, \sigma_0.2 = 340 \text{ MPa}, \delta = 52\% \), and for vacuum steel \( \sigma_F = 1100 \text{ MPa}, \sigma_0.2 = 350 \text{ MPa}, \) and \( \delta = 65\% \). Metallographic study of microsections of deformed steel revealed a considerable number of thin intersecting slip bands in all austenite grains, for whose identification the method of electron-diffraction microscopy was used. The results of this study are presented in Fig. 2.

In light-field electron microphotographs of specimens deformed with \( \varepsilon = 20\% \) (Fig. 2a, b) the amount of deformation twins compared with the amount of them in the condition after quenching or deformation to a small degree (structure of steel after deformation with \( \varepsilon = 10\% \) was also studied) is much greater. Twins form in an austenite matrix having a cellular dislocation structure only by one system \{111\}<112>, and they contain dislocations and stacking defects. A characteristic feature of the twin structure in vacuum-melted steel is strong curvature of twins (Fig. 2b). As a result of steel deformation to failure (breakage), austenite grains are almost completely filled with deformation twins (Fig. 2c, f). Twinning proceeds by several systems \{111\}<112> (as a rule, by two, which follows from analysis of the electron-diffraction pattern presented in Fig. 2d). Twins forming in the first instance (by the first system) are more extensive than secondary twins, which are located between primary twins (Fig. 2e) and sometimes intersect them, as with formation of crystals of lamellar athermal martensite. In vacuum-melted steel deformation twins are curved (Fig. 2f). No indications of \( \alpha^- \) or \( \epsilon^- \)-martensite formation in the deformed steels were detected.