High-speed P6M5 steel (0.8–0.88% C; 5.5–6.5% W; 5–5.5% Mo; 3.8–4.4% Cr; 1.7–2.1% V) is used extensively in Russian industry. It is employed in various cutting tools and stamps that differ considerably in design and dimensions and hence in operating conditions and in the basic sources of faults.

In the present work, we study the influence of hardening heat treatment on the properties of P6M5 steel and select the optimal heat treatment for different operating conditions.

We also investigate the influence of the quenching temperature on the structure and mechanical properties of the steel. The structure is characterized specifically by the austenite grain size, determined by quantitative microscopic analysis (the secant method at 500-fold magnification [1]), and the quantity of residual austenite, determined by the X-ray structural method [2]. The mechanical properties determined after triple tempering at 560°C are the hardness, the flexural strength (σ_{fl}), the yield point in compression (σ_{0.2 co}), the impact strength, the destructive (fracturing) torque for drill bits, and the red hardness (RH 620°C), which is the hardness after additional 4-h heating of the quenched and tempered steel at 620°C.

Endurance tests of cutting tools are conducted in drilling by ground bits of 8-mm diameter, with small spread of the working life on account of the high geometric precision of the bit. The tests continue until complete wear of the bit (its fracture); the wear is measured at the vertex and rear surface. (In the figures, only the wear at the vertex is shown, since the wear at the rear surface is qualitatively the same.)

Endurance tests of a die are conducted in producing a hexahedral internal depression in M10 screws (State Standard GOST 11738–74).

The change in properties of P6M5 steel as a function of the quenching temperature is determined by its phase composition in the annealed state: ferrite alloyed with chromium; and carbides based on chromium (of type M_{23}C_{6}; about 9%), tungsten and molybdenum (of type M_{6}C; about 17%), and vanadium (of type MC; about 1.5%) [3]. The carbides have a complex composition, and therefore we denote the number of metal atoms by M; for example, a carbide of type M_{6}C may correspond to the composition (W,Mo)_{4}Fe_{2}C.

The structural changes (grain size, quantity of residual austenite) on quenching over a broad temperature range (1100–1250°C) reveal three characteristic intervals of the quenching temperature T_{qu} (Fig. 1).

(1) At T_{qu} = 1100–1160°C, the carbides based on chromium (M_{23}C_{6}) dissolve; with increase in T_{qu}, the quantity of residual austenite markedly rises (on account of the increase in carbon concentration in the solid solution); the chromium content also grows. At relatively low T_{qu} (for high-speed steel), the mobility of the boundaries is small, and therefore very small grains are retained. Moreover, grain growth is slowed by the undissolved carbides based on tungsten and molybdenum (M_{6}C).

(2) At T_{qu} = 1160–1200°C, there is slight solution of the carbide M_{6}C and hence slight change in the residual-austenite content and slow grain growth.

(3) At T_{qu} > 1200°C, there is active solution of the carbides M_{6}C. This leads to pronounced change in structure, i.e., to increase in the size and number of austenite grains.

In accordance with the change in phase composition on quenching at different temperatures, the mechanical properties of P6M5 steel also change.
The hardness and red hardness of the tempered steel increase continuously with increase in $T_{\text{qu}}$ (Fig. 2). This is due to the increased alloying of the solid solution on account of the large quantity of carbides dissolved in the austenite on heating before quenching and the more intense dispersional hardening on tempering. However, these characteristics increase at different rates in different temperature ranges: the rate is markedly lower at $T_{\text{qu}} = 1160–1200^\circ\text{C}$ (Figs. 1 and 2).

The impact strength declines monotonically with increase in hardness. This is consistent with numerous studies (for example, [3]). However, the rate of decline is markedly lower in the second interval; the sharp drop begins at $T_{\text{qu}} > 1180–1190^\circ\text{C}$, when the grain size exceeds $B = 11–12$ (Figs. 1 and 2).

The variation in strength $\sigma_\text{fl}$ of the tempered steel is extremal. Increase in $\sigma_\text{fl}$ is associated with the hardening influence of disperse carbides deposited on tempering. (The increase is greater when more carbides dissolve in the austenite on quenching, i.e., at higher $T_{\text{qu}}$.) Decrease in $\sigma_\text{fl}$ is associated with marked austenite grain growth at $T_{\text{qu}} > 1200^\circ\text{C}$. The maximum $\sigma_\text{fl}$ corresponds to $T_{\text{qu}} = 1160–1180^\circ\text{C}$ (Fig. 2). On quenching, satisfactory impact strength (0.4 MJ/m$^2$) and hardness (62–63 HRC) is retained at maximum strength.

As shown by the test data, however, heat treatment ensuring maximum strength cannot be used for tools operating at a cutting speed $\nu = 30–40$ m/min on account of the reduced red hardness. (Such rates are used in the treatment of very workable steels—for example, in the machining of annealed steel 45 by high-speed steel tools.) At $T_{\text{qu}} = 1160–1200^\circ\text{C}$, the red hardness RH 620$^\circ\text{C}$ is less than the value RH 620$^\circ\text{C}$ > 58 specified by State Standard GOST 19265–73 (Fig. 2).

Thus, to obtain the required red hardness, we require $T_{\text{qu}} > 1200^\circ\text{C}$. With increase in $T_{\text{qu}}$, the hardness and red hardness will rise. The increase in $T_{\text{qu}}$ is limited by the decrease in strength (Figs. 1 and 2).

This means that $T_{\text{qu}}$ must be specified on the basis of the principle of sufficient strength: we select the maximum possible $T_{\text{qu}}$ corresponding to the minimum strength sufficient to prevent tool fracture or chipping of its cutting edge. Correspondingly, we will obtain the maximum possible hardness and red hardness and the maximum tool life. At below-optimal $T_{\text{qu}}$, accelerated tool wear occurs on account of insufficient red hardness and hardness of the high-speed steel tool. At above-optimal $T_{\text{qu}}$, chipping and fracture of the steel tool is observed on account of its inadequate strength.

Drilling tests confirm these conclusions. The maximum bit life and minimum unit wear (wear per hole drilled) correspond to $T_{\text{qu}} \geq 1240^\circ\text{C}$ (Fig. 3; Table 1).

Note the difference in the wear of tools quenched at different $T_{\text{qu}}$. Typical $J_{\text{un.v}}$ curves for bits quenched at different $T_{\text{qu}}$ are shown in Fig. 4. (The other curves for the bits are practically the same as those in Fig. 4.)

At $T_{\text{qu}} = 1200$ and 1250$^\circ\text{C}$ (Figs. 4a and 4d), the bit is withdrawn on account of catastrophic wear. For some bits, no normal-wear section is seen. However, the causes of failure are different for bits with $T_{\text{qu}} = 1200^\circ\text{C}$ and 1250$^\circ\text{C}$: for $T_{\text{qu}} = 1200^\circ\text{C}$, failure is due to accelerated wear on account of inadequate red hardness; for $T_{\text{qu}} = 1250^\circ\text{C}$, chipping and even frac-