Reversible temper brittleness (RTB) — embrittlement of steels during tempering or slow cooling in the temperature range of 500-650°C — is considered to result from the formation of impurity segregates (P, Sb, Sn, As) in prior austenite grain boundaries [1-4]. However, there are data indicating [5] that not only these processes but other processes favoring embrittlement occur in quenched steel during tempering at 500-600°C.

This work concerns embrittlement during tempering of quenched chromium steels alloyed with molybdenum to prevent RTB in relation to the vanadium and phosphorus concentrations (see Table 1).

Ingots of heat 1 (0.35% V, 0.015% P) weighing 41 tons were forged to a size of 550 mm. Plates 15 mm thick were cut from them. The other heats were melted in a 100-kg induction furnace with use of ZhS-0 iron in the charge.

Vanadium was added to the steel during pouring of an ingot weighing 16 kg, which was forged and rolled to a plate 10 mm thick.

All heats in the form of plates 10-15 mm thick were oil quenched from 980°C (1 h). Samples were prepared from the plates after quenching and after tempering at 100-760°C for 10 h.

After quenching and low-temperature tempering, all heats had a typical martensitic structure, but after high-temperature tempering (above 600°C) a sorbite structure. The austenite grain size of the laboratory heats was smaller (grade 8-9) than in the commercial heat (grade 4), with greater dispersity and homogeneity of the structural components.

Embrittlement was determined from the variation of T50 with Ttemper, where Ttemper is the tempering temperature of the quenched steel and T50 is the ductile—brittle transition temperature, which most completely characterizes embrittlement.

T50 was determined on impact test samples 5 × 5 × 27.5 mm with a notch 1 mm deep (root radius 0.25 mm). T50 was taken as the testing temperature at which the fracture was 50% fibrous. The tensile strength was determined at 20°C on five samples 3 mm in diameter.

Figure 1 shows the variation of the mechanical properties of heat 1 with the tempering temperature. The mechanical properties of this heat are almost constant after tempering at 100-600°C. When the tempering temperature is raised from 600-760°C the strength characteristics decrease sharply, while the ductile characteristics increase.

Changing the vanadium concentration affects the stability of the steel during tempering. For the steel without vanadium the strength begins to decrease around 500°C (Fig. 2, curve 3).

### Table 1

<table>
<thead>
<tr>
<th>Heat No.</th>
<th>Composition, %</th>
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</thead>
<tbody>
<tr>
<td></td>
<td>C</td>
</tr>
<tr>
<td>1</td>
<td>0.15</td>
</tr>
<tr>
<td>2, 3, 4</td>
<td>0.17</td>
</tr>
<tr>
<td>5, 6, 7</td>
<td>0.18</td>
</tr>
</tbody>
</table>

Note. Heats 1, 3, 4, 6, and 7 contained 0.35, 0.29, 0.51, 0.32, and 0.55% V, respectively.

For heat 1, beginning with tempering at temperatures around 300°C, $T_{so}$ increases and reaches a maximum value at 500-600°C (Fig. 1). As compared with the quenched condition the increase of $T_{so}$ at the peak is 100°C. With further increase of the tempering temperature $T_{so}$ decreases, which coincides with the beginning of weakening.

For heats 2-7 the overall character of the variation of $T_{so}$ with $T_{temper}$ is the same, although the peaks occur at lower temperatures. For the steels in the quenched condition $T_{so}$ is ≈90°C lower than for heat 1, and amounts to −70 ± 10°C. After tempering at 730°C the value of $T_{so}$ is practically the same for heats 2-7 (−110 to −130°C).

The height of the peak and its position depends on the vanadium content of the steel. When the vanadium concentration is changed from 0 to 0.55% the peak rises ≈60°C and at the same time shifts ≈100°C (Fig. 2). Comparing heats 1 and 3, 6, differing in their metallurgical prehistory but similar in vanadium content (≈0.3%), one can see that the increase of $T_{so}$ in the region of the peak is almost the same as for the quenched condition.

In amounts of 0.005 and 0.022%, phosphorus has no effect on $T_{so}$ after the heat treatments tested.

The changes in the mechanical properties of the steel during tempering evidently depend on changes in fine structure. Figure 3 shows the microstructure of heat in the quenched condition, after tempering at 600°C, corresponding to the edge of the plateau of the strength characteristics and the peak of $T_{so}$, and after tempering at 760°C, where the strength and $T_{so}$ are lowest.

After quenching, the structure consists of lath martensite with well-developed dislocation arrays. The laths are slightly misoriented with respect to each other, with an average width of ≈0.3 μ and length ≈5 μ, and are grouped in colonies ≈5 × 5 μ. The laths are filled with evenly distributed dislocations with a density ≈10^{11} cm⁻². No carbide phase was observed in the quenched steel.

After tempering at 600°C for 10 h the overall character of the dislocation arrays and the fragmentation of the crystals are retained. The average size of the laths (width and length)