4. Retained austenite formed in steels 03Kh11N10M2T and 03Kh12N10MT during tempering and nitriding is stable at negative temperatures as low as $-196^\circ$ and also when subjected to tensile stress as high as the proof stress at temperatures as low as $-196^\circ$.

5. After nitriding, $\sigma_0 = 100-120$ kgf/mm$^2$ for steel 03Kh11N10M2T; with a case depth of 0.20-0.25 mm, the ductility and toughness are superior to those of widely used nitriding steels 30Kh3VA and 38KhMYuA with a case depth of 0.45 mm and it is recommended for nitrided machine parts operating at temperatures from $-70$ to $+450^\circ$.

6. Nitrided stainless steel 03Kh12N10MT has high toughness and ductility in combination with high strength at cryogenic temperatures, and it can be recommended for nitrided machine parts operating at temperatures from $-233$ to $+350^\circ$.

LITERATURE CITED

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THERMAL EMBRITTLEMENT OF STAINLESS MARAGING STEELS

V. P. Il'ina, L. N. Kuz'minskaya, P. G. Lapin, V. B. Spiridonov, and V. S. Fridman

It is known [1] that thermal embrittlement, manifest in a reduction of fracture toughness with slow cooling of hot-worked semifinished products in the temperature range of 1000-800°C and associated with precipitation of titanium carbides in austenite grain boundaries, is characteristic of stainless maraging steels to the same extent as for steels of the 03N18K9M5T type [2-6].

The purpose of this work was to determine the maximum heating temperature where slow cooling from this temperature does not induce thermal embrittlement and to determine the minimal temperature and heating time to remove thermal embrittlement (where it cannot be avoided in the manufacturing process).

The investigation was conducted with cold-rolled plates 3 mm thick of steel 03Kh11N10M2T2-V (EP679) produced at the Novosibirsk Metallurgical Factory; the chemical composition of the heats is given in Table 1.

Fig. 1. Effect of heating temperature and cooling rate on the mechanical properties of steel 03Kh11N10M2T2-VD after cold rolling (1, 2) and after preliminary heating to 1200°C for 1 h (3). 1, 3) Water quenched; 2) furnace cooled in a container at a rate of 50 deg/h.

Table 1

<table>
<thead>
<tr>
<th>Heat No.</th>
<th>Composition, %</th>
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<tr>
<td></td>
<td>U</td>
<td>Z</td>
</tr>
<tr>
<td>1</td>
<td>0.02</td>
<td>1.0</td>
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<td>2</td>
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Changes in the microstructure, mechanical properties (σb, σn, σp), and phase composition of anodic residues were determined.

The microstructure (boundaries of prior austenite grains and precipitation of excess phase) was revealed by electrolytic etching in a reagent consisting of 50% HNO3 + 50% glacial acetic acid at a current density ~0.6 A/cm² (preliminary electropolishing was conducted in the same reagent at a current density ~5 A/cm²).

The anodic residues for physicochemical phase analysis were separated in an electrolyte consisting of 50 ml HCl (sp. gr. 1.19) + 5 g citric acid + 1000 ml methanol at 0° and current density 0.04 A/cm². X-ray analysis of the anodic residues was made in dual RKU-114M cameras with KαCu and Co radiation, with asymmetrical loading of the film.

We determined the effect of heating temperatures at 800-1200°C (holding 1 h) on the thermal embrittlement of the steel after furnace cooling in a container at a rate of ~50 deg/h and, for comparison, in water. The steel was tested as-received (hot deformation + 40% cold deformation) and after high-temperature water quench (1200°C for 1 h).

The changes in the mechanical properties after these treatments are shown in Fig. 1, and the changes in microstructure in Fig. 2. The values of σn and σp decrease considerably at 700-850°C, which is due to the precipitation of intermetallic compounds – χ phase (Fe3Crl2Mo10) and Fe2Mo [1].

The lower values of σn and σp after slow cooling from temperatures above 1100°C are due directly to so-called thermal embrittlement and define the maximum temperature at which slow cooling does not induce thermal embrittlement (Fig. 1, curve 2). Precipitation of titanium carbides in the grain boundaries (Fig. 2) is observed at these same temperatures, particularly after heating at 1200°C. A slight reduction of σp (~10%) is observed even after slow cooling from 1100°C. After cooling from 1150°C, the value of σp decreases 20-25%, and 75% after cooling from 1200°C.

It should be noted that preliminary high-temperature treatment at 1200°C with rapid cooling in water has no noticeable effect on the microstructure or properties of the steel (Fig. 1, curve 3).

Thus, slow cooling of steel 03Kh11N10M2T2-VD from 1100-1200°C is unsuitable. The most intensive precipitation of carbide phase and susceptibility to brittle fracture are observed after slow cooling from 1200°C. The results obtained are explained by the fact that with heating to 1100°C or higher (holding 1 h) the titanium carbides that are fairly evenly distributed after hot deformation go into solution and are then precipitated in austenite grain boundaries during slow cooling in the range of 1000-800°C.

It can be assumed that the solution of carbides also occurs with longer holding times at temperatures below 1100-1200°C. This is necessary in cases where thermal embrittlement occurs in some stage of production of semifinished products (especially large sections) but reheating at high temperatures is either impossible (due to lack of facilities for rapid cooling in the dangerous temperature range) or undesirable (because of considerable grain growth at high temperatures – from grade 5 at 1100°C to grade 1-2 at 1200°C).