dispersed carbides in the volume of the grain leads to a significant wear resistance of the steel. An increase in austenitizing temperature and time causes coagulation of the carbides. The embrittlement of the steel as the result of carbide formation is compensated by austenitic grain refinement. This agrees with the results of investigation of the change in impact strength and the character of failure of the steel after aging and austenitizing using similar cycles.

Therefore, the complex heat-treatment cycle for 110G13L steel developed by us (harden from 1100°C + age at 460-480°C for 4-6 h + harden from 850°C with a hold of 3-6 h) makes it possible to effectively increase its wear resistance with maintenance of a sufficiently high level of impact strength without additional alloying. Combined solution of the interpolation equations obtained opens possibilities for optimization of the wear resistance and resistance of the steel to brittle fracture in relation to its specific service conditions.

IMPROVING THE HEAT RESISTANCE OF AUSTENITIC
STEELS BY THE THERMOCHEMICAL TREATMENT METHOD

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The heat resistance of austenitic steels containing volatile components becomes an important factor when parts made of such steels are used in various systems and devices where these parts are subjected to heat, either during service or under vacuum. Some undesirable processes (such as the austenite—ferrite phase transformations [1]) affecting physicochemical properties of the steel might occur due to heating. The heat resistance of austenitic steels can also be impaired due to blistering which occurs when such steels are used under radiation conditions [2]. This makes the improvement of austenitic steels' heat resistance especially important. The authors consider that one way to achieve this improvement is to use thermochemical treatment (TCT) for fabrication of protective coatings consisting of chemical elements with a low atomic number, such as Al and Ti (on the basis of minimum sz^2 value, where s is the material erosion coefficient and z is the atomic number).

The present work* studies the following Cr—Mn and Cr—Ni austenitic steels: ÉP838 (12% Cr, 14% Mn, 4% Ni, 1.4% Al, <1% Mo; <0.03% C) and 316 (16-18% Cr, 10-14% Ni, 2% Mn; 2% Mo; <1% Si; <0.03% C; <0.03% S; <0.045% P).

A microbalance was used for studying the material evaporation process under high, oil-free vacuum conditions (about 5 µPa). Evaporation rates from an open surface to vacuum were measured by the weight loss of 8 × 16 × (0.1-0.2) mm polished plate specimens suspended from a quartz microbalance with $4 \cdot 10^{-7}$ g/µm sensitivity.

The protective coatings were deposited on the steel surfaces by TCT methods. Two methods were used: 1) saturation in a powder medium; and 2) electrolytic aluminizing in molten salts (see Table 1).

As was established previously [3] on the basis of kinematic curves obtained for vacuum evaporation of ÉP838 and 316 steels at different temperatures, the evaporation rate of their volatile components is of parobolic form. Phase analysis of the condensate carried out by the electrographic transillumination method showed that the main component of the vapor phase formed above the ÉP838 steel is Mn, while above 316 it is Cr. Evaporation of the above elements is controlled by their diffusion to the surface, which explains the parabolic form of their evaporation rate curves. Evaporation rates of the Cr—Ni steels are considerably lower than those of the Cr—Mn steels in similar temperature ranges.

ALUMINIZING

It is necessary to prevent or slow down the evaporation of volatile elements from the steel surface during aluminizing. The efficiency of that protective coating is proven by

TABLE 1

Method	TCT method	Saturating medium composition	TCT conditions	
			t, °C	τ, min
I	Al-coating	(10% Al+90% Al$_2$O$_3$)+ +1,5% NH$_4$Cl	1100	300
II		35% NaCl+35% KCl+ +15% AlF$_3$+15% NaF	750	5
III	Ti-coating	(50% Ti+50% Al$_2$O$_3$)+ +1% NH$_4$Cl	1000	180
IV		(50% Ti+50% Al$_2$O$_3$)+ +5% ZnCl$_2$+ +14% NH$_4$Cl	1000	180

Footnotes: 1) Methods I, III, and IV correspond to saturation from a powder medium; while method II corresponds to electrolytic aluminizing; 2) Electrolytic aluminizing current density was 0.3 A/cm^2.

an insignificant specimen weight loss due to evaporation of the volatile element. The evaporation rate of the coating material as such (in this case, aluminium) at 835°C (the experimental temperature) is very low, since, during aluminization, layers of heat-resistant intermetallic compounds (such as Fe$_3$Al, etc), rather than pure aluminium layers, are formed on the steel surface. The evaporation rate of aluminium from its intermetallic compounds with iron at 835°C is lower than the sensitivity level of the equipment used. Because of this, the weight loss of the ÉP 838 steel specimens at that temperature can possibly be considered as the product of diffusion and evaporation of Mn from both uncoated specimens (Fig. 1a, curve 1) and specimens aluminized by the powder saturation method (Fig. 1a, curve 3).

During the process of electrolytic aluminizing (Fig. 1a, curve 2) even a very thin, 10-15 μm, layer (the layers formed by powder saturation aluminizing are approximately 100 μm thick) considerably slows down the Mn evaporation from ÉP838 surfaces. The primary reason for this is the increased resistance to diffusion of Mn through the thin aluminized layer. It can be concluded that there was no flaking and crumbling of aluminized layers during the vacuum tests in both cases, because such processes would inevitably be registered by the microbalance and revealed by affecting the smoothness of the evaporation kinetic curve. We can therefore infer that a good bond exists between the aluminium coatings and the matrix within the limits of a single heating—cooling cycle (20°C → 835°C → 20°C).

The aluminizing of Cr—Ni steel 316 by the powder saturation method decreases the evaporation of Cr and Ni at 950°C to such low values that they could not be registered by the equipment used. Aluminizing virtually eliminates the evaporation of volatile elements from 316 steel at the above temperature.

COATING WITH TITANIUM

Unlike the diffusion layer formed by aluminizing, coating with titanium usually produces a pure titanium layer on the surface, as well as a thin diffusion layer between the steel and the titanium. Such a layer provides a very effective passivation method for evaporation from the steel ÉP838 specimens. It was not possible to register any weight changes of the Cr—Mn steel specimens at 835°C (Fig. 1a, curve 4) using the microbalance method. Apparently, the titanium layer prevents Mn diffusion to the surface. Because the tests at 835°C showed that coating with titanium completely eliminated the evaporation of volatile components from the surfaces of ÉP838 steel specimens over a relatively long period of time (8 h), it was decided to carry out additional tests on titanium-coated specimens at 950°C. Kinetic curves for evaporation of ÉP838 at 950°C with and without a titanium layer (formed by method III) are shown in Fig. 1b (curves 4 and 1 respectively). Curve 1 is parabolic, which is characteristic for diffusion processes in this case for the diffusion and evaporation of Mn. However,
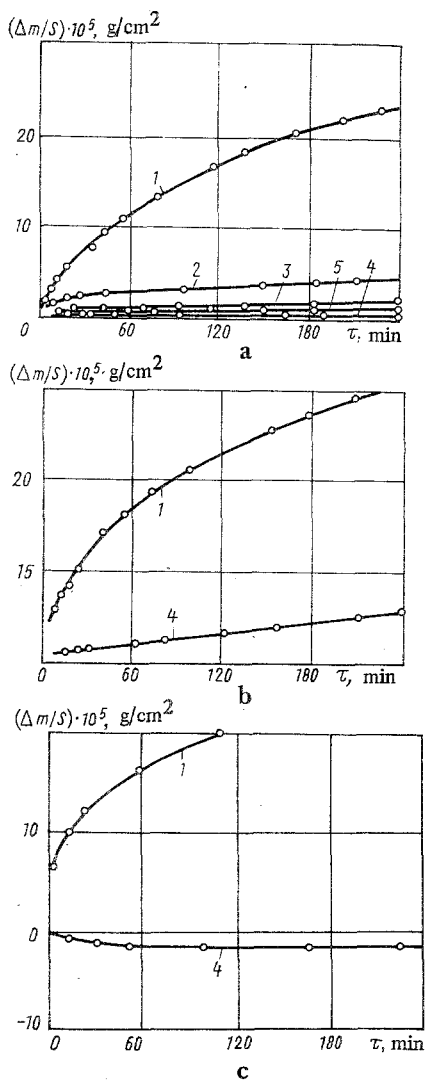
Fig. 1. Kinetic curves for the weight changes of Cr—Mn steel ÉP838 specimens in a 5 µPa vacuum at 835°C (a) and 950°C (b) and of Cr—Ni steel 316 specimens in low vacuum (100 µPa) at 835°C (c). 1) Un= coated; 2, 3) aluminized by the molten salt electrolysis method (I) and by the powder saturation method (II); 4, 5) titanium-coating using the powder saturation method (III and IV).

for the titanium-coated specimen the weight loss was a linear function of time at constant temperature (curve 4), which suggests evaporation of Ti from the specimen's surface. Of course there is Mn diffusion in the titanium layer; however even at 950°C the rate of that process is low and does not distort the linear character of the kinetic curve, which means that Mn does not reach the evaporation surface. The magnitude of the experimentally measured Ti evaporation rate from the surface of the steel coated with titanium by method III, $w_{950}^{III} = 4 \cdot 10^{-9}$ g/cm$^2 \cdot$sec, agrees with the tabulated data on Ti evaporation rate at 950°C.

The titanium coating formed on ÉP838 specimens using method IV is apparently not as dense as that produced by method III. Because of this a low rate of evaporation of Ti is observed even at 835°C (Fig. 1a, curve 5). The titanium evaporation rate from the surface of Cr—Mn steel coated by method IV is: $w_{950}^{IV} = 7 \cdot 10^{-10}$ g/cm$^2 \cdot$sec.

Tests of the method IV titanium-coated ÉP838 specimens at 950°C showed that, in that case, Ti evaporation follows a linear law with a somewhat higher rate than in the case of

Ti-coating by method III: $w_{950}^{IV} = 5\cdot10^{-9}$ g/cm^2·sec. Experiments with evaporation of Cr–Ni steel 316 at 835°C under high vacuum (10 µPa) showed that, similar to the Cr–Mn steel EP838 the specimen weights did not change. Because of this, the vacuum conditions were deliberately lowered. Tests of the titanium-coated steel 316 in 100 µPa vacuum at 835°C are shown in Fig. 1c. The Ti layer absorbs gas molecules from the residual atmosphere in the working chamber, which causes a weight gain of the specimens at 835°C. That weight increase is insignificant in absolute terms, however it must be remembered that the molecular weights of the absorbed gases (oxygen, carbon dioxide, water vapor, etc.) are low. Therefore the absorption even of a large amount of gases does not result in a significant weight gain of the specimens. Gas absorption of titanium coatings on austenitic steels at high temperatures can be a very important factor for their application in various vacuum systems.

It is concluded therefore that the titanium coating of austenitic Cr–Mn steel EP838 and Cr–Ni steel 316 can be recommended as a protective coating for high-temperature applications. It is also worthwhile to test the heat resistance of titanium coatings in thermocyclic conditions, as well as to experiment with other coating methods, for instance, plasma coating, etc.

The application of TCT-formed aluminium and titanium protective layers on Cr–Mn (EP838) and Cr–Ni (316) austenitic steels decreases the evaporation rates of volatile components from these steels in vacuum at 800-950°C. Titanium coating has an additional advantage: high absorption of gas molecules which can be used as a method of improving the vacuum conditions.

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METHOD OF IMPROVING THE HEAT RESISTANCE OF PARTS
WORKING UNDER CONDITIONS OF DISCONTINUOUS LUBRICATION

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In the present work the possibilities are studied for using thermochemical treatment methods (TCT) to improve the heat resistance of parts working under conditions of discontinuous lubricant supply. The main aim of the study is to produce rubbing surfaces, the microstructure of which would possess high antifriction as well as antigalling properties.

Boriding produces a friction-resistant layer. The borided layer retains a high hardness (H1400-2000 at a load of 0.5 N) up to 850-900°C. However, because of low galling resistance, such layers cannot be used without lubrication.

Sulfiding is the most promising method for improving galling resistance. However, the hardness of sulfide layers is so low that even a quenched steel abrades it, and consequently, that process cannot be used as a treatment for working parts [1].

Simultaneous boron and sulfur saturation of steels failed to improve their galling resistance. This can be explained by the fact that such processes do not produce, on the steel surface, dispersed phases of iron sulfides which would at high temperatures form, due to their melting point, sulfide films between the rubbing surfaces.

Boron and silicon impregnation was used to produce microstructures at the steel surface containing boride and sulfide areas [2]. The studies were carried out using steel 20. Im-