A HEAT TREATMENT FOR OBTAINING A CONTROLLED FERRITIC–MARTENSITIC STRUCTURE IN STEEL

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In recent years practical interest has been aroused by so-called two-phase ferritic–martensitic steels, which combine increased strength and high production plasticity [1, 2]. Requirements for specific parameters of the structure of these steels are determined by their purpose and necessary mechanical properties. Areas of martensite create the strengthening. Each 1% of martensite increases the tensile strength of a two-phase ferritic–martensitic steel by about 10 MPa regardless of the strength and geometry of the martensite phase. At the same time, the morphology of the martensite exerts an influence on the plastic properties and intensity of strain hardening. The optimum is obtaining islands of dispersed martensite at the triple joints of ferrite grains [3-4]. To obtain high plasticity it is important to provide the minimum degree of impregnation of ferrite with interstitial atoms, and to increase the fracture resistance a reduction in the carbon content in the martensite (for details see [2-4]).

In order to correctly select the heat-treatment cycle for obtaining the specified ferritic–martensitic structure it is necessary to know the influence on it of various factors such as the temperature and length of heating, the cooling rate, and the original structure of the steel.*

The influence of composition on the structure and properties of two-phase ferritic–martensitic steels is a subject for separate discussion.

The basic variation in heat treatment for obtaining a ferritic–martensitic structure is heating to temperatures in the intercritical temperature range Ac_1-Ac_3 with subsequent cooling.

Processes Occurring in Heating to Temperatures in the Intercritical Temperature Range. The quantity of martensite in steel hardened from intercritical temperature range temperatures is determined by the quantity of austenite formed in heating. Since the transformation products reproduce the geometry of the former austenite, to obtain a two-phase ferritic–martensitic structure of a certain type it is necessary to take into consideration the points of preferable origin and the morphology of the γ-phase formed in heating, which are determined by the type of original structure [5]. For example, to obtain a two-phase ferritic martensitic steel with dispersed islands of martensite an original structure formed as the result of deformation or hardening and tempering is most favorable. To obtain a two-phase ferritic–martensitic steel with acicular martensite it is necessary to heat steel first hardened to martensite to intercritical range temperatures.

The quasibinary diagram of equilibrium type shown schematically in Fig. 1 may be used only as a qualitative reference point for determination of the possible ratio of the phases in a two-phase ferritic–martensitic steel. In actuality under conditions of a real heating time the ratio of the phases (α and γ) does not conform to this plan, as was shown in [6]. The highest rate of austenitization is observed in the first 4 or 5 min of hold in the intercritical range, after which the rate of occurrence of the process steadily decreases [7, 8]. The γ-phase grains are formed preferentially at the ferrite grain boundaries, which is the result of the difference in the diffusion coefficients parallel and perpendicular to the grain boundaries. As the result the increase in the quantity of austenite during the isothermal hold in the intercritical range develops in an undesirable direction, by envelopment of the ferritic grain boundaries (Fig. 2).

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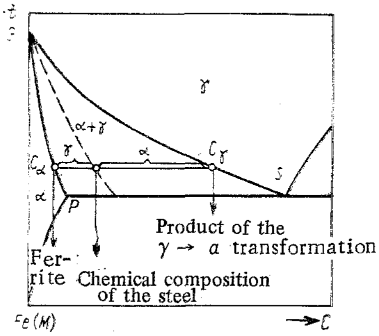

Fig. 1. Quasibinary equilibrium diagram of the Fe(M)—C system (plan). The broken line corresponds to obtaining approximately 20% γ-phase.

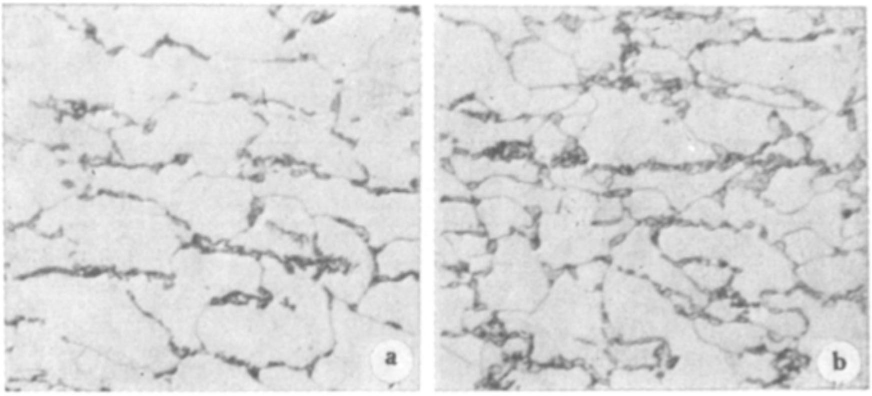

Fig. 2. Structure of type 08G2S steel hardened from 740°C after a hold of 1 (a) and 10 (b) min. 1000×.

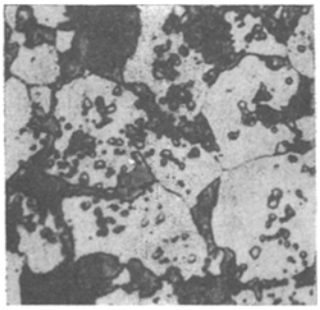

Fig. 3. Structure of 20 steel hardened from 760°C after holding for 10 min. 1000×.

The type of original structure has a significant influence on the kinetics of austenitization [9]. With an increase in the specific area of ferrite—carbide boundaries, that is, with an increase in the degree of dispersion of the original structure, the rate of formation of γ-phase increases. Investigation of the kinetics of austenitization of cold-worked steel (rolled with a reduction of 60%) given preliminary heat treatments with different cycles showed that with short (2.5-5 min) holds the influence of the original structure on the kinetics of austenitization is preserved after cold working [10].

An increase in heating temperature in the intercritical range leads to an increase in the rate of austenite formation and in the maximum quantity of it. This is accompanied by a reduction in the carbon content in the γ-phase (C_γ). The rate of increase in the area of

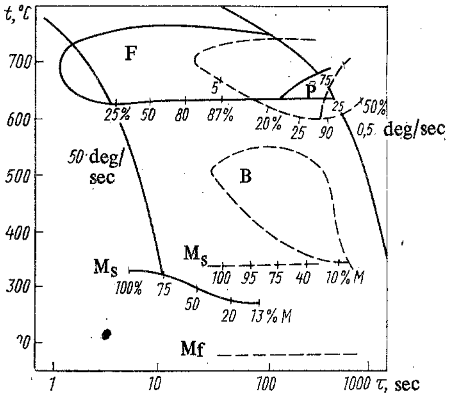

Fig. 4. Thermokinetic diagrams of 06KhGS steel after austenitization at 780°C (solid lines) and of completely austenitized 30KhGS steel (broken lines) with the same carbon content in the γ-phase (isothermal 10-min hold).

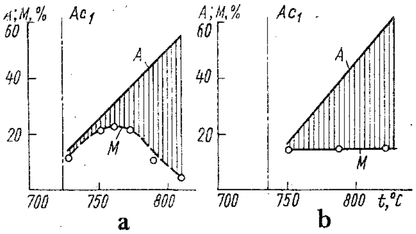

Fig. 5. Influence of heating temperature (10-min hold) on the quantity of austenite and martensite formed in cooling at a rate of 70 deg/sec in types 06KhGM (a) and 06KhGSR (b) steels. Shaded area is the quantity of "new" ferrite (dilatometric measurements).

the austenite phase is proportional to the rate of carbon diffusion and the difference in its contents at the austenite boundaries with ferrite and carbides and inversely proportional to the gradients of the contents at the austenite—ferrite and austenite—carbide boundaries [11]. With an increase in temperature all of these factors change in the direction of an increase in the rate of growth of γ-phase but the gradient of the contents at the boundaries of it with ferrite decreases more sharply than at the boundaries with cementite and therefore the rate of growth of austenite is greater in the direction of ferrite than in the direction of the carbide. As the result in heating to intercritical range temperatures in steels with stable carbides or a high carbon content in the formation of a specified quantity of austenite a large quantity of undissolved carbides is maintained (Fig. 3), which leads to an additional decrease in C_γ.

An increase in strength properties of two-phase ferritic—martensitic steels may be obtained as the result of aging at 170-200°C. Therefore, the heat treatment must provide a certain content of interstitial impurities in the ferrite (C_α). The values of C_α depend upon the heating temperature. In accordance with the equilibrium diagram, an increase in heating temperature leads to a decrease in C_α (Fig. 1, line PG).

Transformations of Austenite in Cooling from the α + γ-Area. The basis for an understanding of the influence of the parameters of "intercritical" heat treatment on the structure of two-phase ferritic—martensitic steels is thermokinetic curves of the transformation of austenite in cooling of a two-phase ferrite—austenite mixture. The features of the phase transformations after incomplete austenitization (if it is compared with homogeneous austenite of the same steel) are caused primarily by the increased carbon content

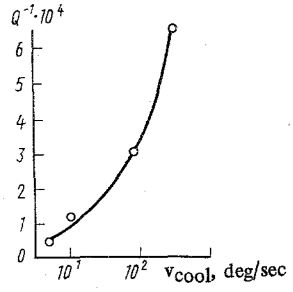

Fig. 6. Influence of cooling rate on the height of the Snoek peak. Type 08G2S steel heated at 760°C.

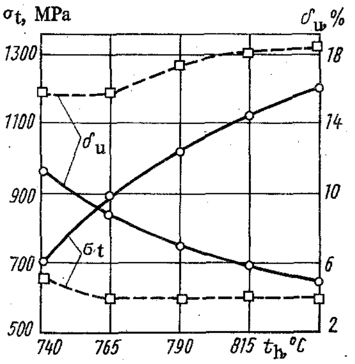

Fig. 7. Influence of heating temperature in the intercritical range on the mechanical properties of a two-phase ferritic—martensitic steel obtained in water hardening (solid lines) and in cooling at a rate of about 30 deg/sec (broken lines). Steel type 08G2SF.

in the austenite (Fig. 1, line GS). On the other hand, the presence of a developed interphase surface eases development of diffusion transformations in cooling of steel heated to intercritical range temperatures. This is revealed in a shift in the area of precipitation of ferrite in the direction of higher cooling rates in comparison with completely austenitized steels with the same carbon content in the austenite (Fig. 4). Easing of the conditions of precipitation of ferrite and the corresponding enrichment with carbon of the remaining austenite lead to thinning of the bainitic area on the thermokinetic diagram and a reduction in the M_s point.

With an increase in the heating temperature in the two-phase area there is an increase in the quantity of austenite formed (Fig. 5) and its stability changes simultaneously as the result of the change in C_γ. The maximum heating temperature in the intercritical range in obtaining a two-phase ferritic—martensitic steel is limited by the plasticity, which drops with an increase in the quantity of the hardening phase, and also by the danger of an extreme reduction in the stability of the austenite to a level at which the specific cooling conditions may lead to the pearlite transformation. The selection of the heating temperature in the intercritical range is determined primarily by the specified strength (necessary quantity of martensite).

The cooling rate has a decisive influence on the share of transformation of austenite according to the diffusion-free mechanism and also on the strength and plasticity of the ferrite as the result of the change in the content of the interstitial atoms dissolved in it (the change in the height of the Snoek peak on the curves of the temperature relationship of internal friction, Fig. 6).

To obtain the specified ratio of ferrite and martensite in the structure of the steel, a certain combination of heating temperature in the intercritical range and cooling rate from this temperature is important. The lower the heating temperature in the two-phase area, the more rapid the cooling rate necessary to obtain the specified structure. The higher the heating temperature, the stronger the relationship of the quantity of martensite in the structure to the cooling rate. With comparatively slow cooling (at a rate of 5-30 deg/sec) precipitation of ferrite is observed and the lower the stability of austenite, that is, the greater the original quantity of it, the greater the quantity of it. Such an effect of self-regulation of the structure makes it desirable to obtain a two-phase ferritic—martensitic steel with the minimum possible cooling rates excluding the pearlite reaction. At the same time, there is a quite wide range of heating temperatures (40-80°C) which provide approximately constant structure and strength properties of two-phase ferritic—martensitic steels (Fig. 7, broken lines). Within the limits of this range with an increase in heating temperature the plasticity of the steel increases.

An additional advantage of cooling at relatively low rates is a reduction in the interstitial impurity content and also refinement of the martensite islands as the result of precipitation of new ferrite [12]. In slow cooling of steels from intercritical range temperatures the maximum quantity of residual austenite is observed, which is caused by flow of carbon into the last portions of the γ-phase and also by the appearance of stabilization. In heating to temperatures providing an equal carbon content in the austenite the minimum cooling rate making it possible to obtain a ferritic—martensitic structure depends upon the composition of the steel and for type 09G2 steel may reach 2-5 deg/sec [12].

The use of high cooling rates from the intercritical range makes it possible to obtain a ferritic—martensitic structure and the necessary strength for steels with a low carbon content or with a significant reduction in alloy element content.

Obtaining Two-Phase Ferritic—Martensitic Steels in Controlled Cooling. The most favorable variation for obtaining two-phase ferritic—martensitic steels with a specified ratio of phases is controlled cooling from the heating temperature in the intercritical range (T_1) first with a rate of 5-30 deg/sec and then with a rate of more than 100 deg/sec [2, 12]. In the stage of comparatively slow cooling to the temperature T_2 the conditions are created for precipitation of new ferrite and a reduction in the interstitial impurity content in the old ferrite, which provides an increase in plasticity of the steel and promotes the obtaining of dispersed islands of martensite in the structure. In the stage of rapid cooling from the temperature T_2 supersaturation of the ferrite occurs, providing precipitation of the impurities during the subsequent short temper. In the general case the temperature T_2 must be higher than the temperature of the pearlite transformation at the corresponding cooling rate in the T_1-T_2 area.

Tempering of Two-Phase Ferritic—Martensitic Steels. In obtaining two-phase ferritic—martensitic steels by rapid hardening from intercritical range temperatures it is desirable to give a final short temper. The basic purpose of this temper is to decrease the content of dissolved impurities for the purpose of increasing the plasticity of the ferrite and preventing overaging of the two-phase ferritic—martensitic steel in heating of deformed parts. There is a simultaneous increase in plasticity (toughness) of the martensite and a partial relief of residual stresses occurring as the result of the local increase in the volumes of the areas of former austenite in the martensite transformation.

The tempering temperature and time of two-phase ferritic—martensitic steels intended for forming or working by other methods of cold plastic deformation must provide preservation of their basic substructure features, which cause the absence of the yield plateau on the tensile curve of these steels, a low nominal yield strength, and a high rate of strain hardening. These features are the result of an increased dislocation density close to the martensite areas and a certain level of residual tensile stresses in the ferrite. A sign of too high a temperature or too long a tempering time (optimum 5-10 min at 200-250°C) is the appearance of a yield plateau on the tensile curves.

It has been established that low temperature tempering of two-phase ferritic—martensitic steels, including after working, makes it possible to increase their impact strength and to reduce the T_{50} temperature by about 60-80°C, that is, it promotes an increase in the service reliability of these steels [13]. Under production conditions tempering of finished parts to increase the strength properties by the strain aging mechanism or to increase the fracture resistance is combined with the paint or lacquer coating drying operation or (for fastener parts) with dehydrogenation baking.

Conclusion. The specific heat-treatment cycle for obtaining a two-phase ferritic–martensitic steel with specified properties depends upon the composition of the steel and the parameters of the heat-treatment equipment (heating rate and time, means of cooling the heated strip, capability of reheating for tempering, etc.). At the same time, based on the concepts presented above and the experimental data obtained, it is possible to recommend certain optimum conditions of treatment with heating to intercritical range temperatures which must be determined more accurately experimentally on the heat-treatment equipment available in the plant.

1. The heating temperature must correspond to Ac_1 + (60-90)°C in order to obtain 50-60% austenite in the structure with the given heating time. Subsequent cooling must provide the obtaining of 30-45% new ferrite with refinement of the martensite islands. As the result the structure of the steel must consist of about 20% martensite and 80% ferrite.

2. It is preferable to cool from intercritical range temperatures at moderate rates (5-30 deg/sec) if only in the area from the heating temperature to 400-500°C, which promotes an increase in plasticity of the steel and provides stability in obtaining the necessary quantity of martensite (15-25%) within limits of ±3-5%.

3. The original structure of the steel must be as highly dispersed as possible. This promotes acceleration of austenitization and an increase in the number of centers of origin of austenite.

4. Tempering for increasing the plasticity of two-phase ferritic–martensitic steels intended for subsequent cold working must be done at 200-250°C with a time of not more than 10 min.

5. The criterion of correct choice of the basic heat-treatment parameters in obtaining a two-phase ferritic–martensitic steel must be assumed to be obtaining of a structure of 15-25% martensite and the remainder ferrite, the absence of a yield plateau on the tensile curve, and a $\sigma_{0.2}/\sigma_t$ ratio of 0.5-0.6.

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FEATURES OF THE STRUCTURE AND PROPERTIES OF HIGH-SPEED STEELS AFTER LASER TREATMENT

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For development of scientifically based laser heat-treatment cycles for high-speed steel tools it is necessary to establish the relationship between the irradiation conditions and the structures and properties of the steels formed in the zone of action of the light beam. Existing information on this question refers primarily to R6M5 and R18 steels. For steels with an intermediate tungsten content and additionally alloyed with vanadium, cobalt, and molybdenum such data is limited although such steels are widely used for machining of difficult to machine materials. In connection with this an increase in the life of cutting tools of these steels is especially important.

This article presents the results of an investigation of the influence of pulsed laser treatment on the structure of the irradiated layers and the structure and properties of previously hardened R9K5, R9M4K8F, R10K5F5, R12F5M, and R12F2K8M3 high-speed steels. The specimens were laser treated on a Kvant-16 machine in air. After irradiation transverse microspecimens were prepared, etched in a 10% ethyl alcohol solution of HNO_3 using the method described in [1], and then investigated by metallographic, electron microscopic (carbon replicas tinted with tungsten oxide), x-ray diffraction, and microhardness methods. The electron microscopic and x-ray methods of investigation were used after layer-by-layer grinding off of the surface layers of the irradiated areas, which made it possible to clearly compare the data obtained with the results of metallographic investigation.

The carbon content in the α-solid solution was determined by the method presented in [2]. The microhardness was measured on the transverse specimens in the center portion of the irradiated volume on a line coinciding with its axis of symmetry (Fig. 1a). The life tests of the laser treated tools were made using the method described in [3, 4]

The irradiated portions formed under the action of the pulses, the power density of which exceeds 320 MW/m^2, consist of three characteristic zones with different structures (Fig. 1b), which are responsible for the character of change in microhardness (Fig. 1c).

In the original condition (zone I) the microhardness of R9M4K8F steel is 880H. In laser treatment the hardness of the steel in zone II (tempering zone) drops as the result of decomposition of the martensite. The minimum microhardness (760H) is observed at the boundary with zone III (heating temperature corresponds to the Ac_1 point) [5]. The results of electron-microscopic investigations made is possible to establish that in the zone considered even after treatment at temperatures close to the Ac_1 point the size of the carbide particles does not exceed the size of the carbides formed in standard tempering in through heat treatment (Fig. 2a, b). Consequently, in laser treatment in the subcritical temperature range coagulation of the carbide particles does not occur since the heating time is short.

On the basis of the results presented it may be concluded that in contrast to normal tempering in the 560°C-Ac_1 temperature range, in which the reduction in hardness is caused both by decomposition of the martensite and by coagulation of the carbide phases [6], in laser treatment softening occurs only as the result of decomposition of the solid solution. This may be explained by the fact that after laser treatment in the temperature range considered the hardness of the steel is significantly higher than after normal tempering (for R9M4K8F steel after tempering at temperatures close to Ac_1 760H and 400H, respectively).

The maximum microhardness is reached at the boundary with zone IV (heating temperature corresonds to T_S, the solidus temperature) [5]. In external appearance the structure of

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