EFFECT OF ALLOYING ON THE PROPERTIES AND HEAT TREATMENT OF LOW-ALLOY TUNGSTENLESS HIGH-SPEED STEEL 95Kh6M3F3T

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Saving of scarce alloying elements such as tungsten and molybdenum by creating tungstenless and less alloyed molybdenum high-speed steels is an important task for the national economy.

Among the alloying systems for steel on whose basis it is possible to solve this problem there is a priority for chromium—molybdenum—vanadium steels. In fact this alloying system relates to steels ÉI260 (100Kh4M3F2) and ÉI277 (110Kh4M3F3) created in the 1930s [1, 2], and also steel ÉP973 (65Kh6M3F3SB [3, 4]) developed in 1977.

It has been established [3, 5] that the high properties for steel ÉP973 are achieved with a content of 3% Mo and a ratio of $V:C \cong 4$. Although steel ÉP973 has been used industrially in a number of enterprises, its properties may be improved as a result of changing its composition, primarily by increasing the carbon content to 0.9-1.0%. However, it would also be necessary to define the optimum content of molybdenum and the V/C ratio in it. In order to solve this question several steels were melted with 0.95% C whose chemical compositions are given in Table 1. In these steels the ratio of vanadium to carbon was 3 and 4, and the molybdenum content was 3 and 5%. In addition, instead of niobium, which provided grain growth retardation in steel ÉP973, in the test steels titanium was added whose carbides and carbonitrides also do not dissolve well in austenite, as for the similar niobium carbides. This change in alloying system was also previously accomplished in steel 65Kh6M3F2T containing 0.2% Ti, which made it possible to increase its red hardness a little compared with steel ÉP973 [5].

Steels were melted in an open induction furnace. Ingots were forged at 1100-980°C. Bars were annealed at 820-850°C for 3 h, and cooled in the furnace. The hardness after this annealing was HB 190-220. Quenching and tempering of test steel specimens was carried out in salt electrode baths. Austenite grain size was determined according to the standard scale of GOST 5639-82. Carbide phase analysis was carried out by the method given in [6] and dilatometric analysis was carried out as in [7].

In the as-supplied condition (after annealing) the structure of the test steels was polygonized ferrite along whose grain boundaries and within subgrains MC type carbides based on VC with a crystal lattice parameter of 0.146-0.148 nm, type M_6C based on Fe_3Mo_3C, type $M_{23}C_6$ based on $Cr_{23}C_6$, and also carbonitrides M(C, N) based on Ti(C, N) with a crystal lattice parameter of 0.424 nm were observed. As the studies carried out have shown, distribution of alloying elements between solid solution (ferrite) and carbides was determined primarily by the V/C ratio.

TABLE 1

Melt	Nominal steel designation	Content of elements, %							(V+Ti)/C
		C	Cr	V	Mo	Ti	Mn	Si	
1	95Kh6M3F3T0.2	0,92	6,00	2,85	3,10	0,24	0,66	0,37	3,3
2	95Kh6M3F4T0.2	0,92	6,15	3,65	3,10	0,24	0,64	0,40	4,2
3	95Kh6M3F3T0.4	0,94	6,10	2,70	3,15	0,36	0,53	0,44	3,3
4	95Kh6M3F4T0.4	0,93	6,00	3,60	3,05	0,40	0,63	0,33	4,3
5	95Kh6M5F3T0.2	0,92	6,00	2,90	4,90	0,22	0,57	0,43	3,4
6	95Kh6M5F4T0.2	0,94	6,30	3,80	4,90	0,24	0,60	0,33	4,3
7	95Kh6M5F3T0.4	0,95	6,20	2,75	4.90	0,42	0,54	0,51	3,3

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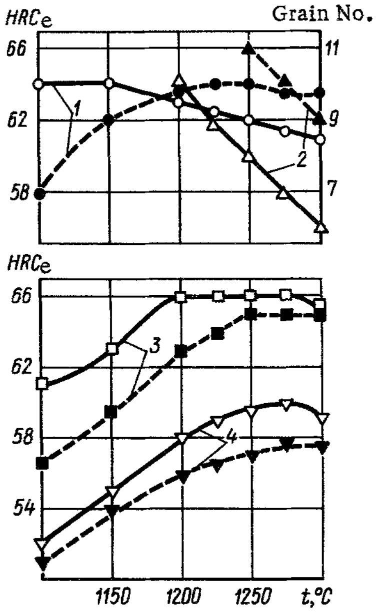

Fig. 1

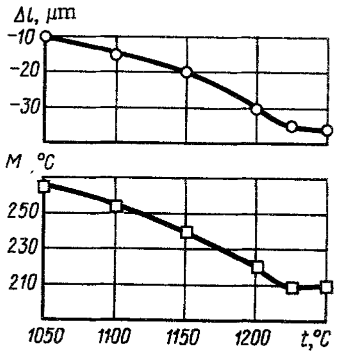

Fig. 2

Fig. 1. Effect of austenitizing temperature on hardness after quenching (1), austenite grain size (2), secondary hardness (3), and red hardness (4) for steels 95Kh6M3F3TO.4 (solid lines) and 95Kh6M3F4TO.4 (broken lines).

Fig. 2. Dependence of the change in specimen length Δl and martensite point M_s for steel 95Kh6M3F3TO.4 on austenitizing temperature.

Given in Table 2 [6, 8, 9] are the results of carbide phase analysis for annealed steels 95Kh6M3F3TO.4 with $V/C \cong 3$ and 95Kh6M3F4TO.4 with $V/C \cong 4$. Also given there for comparison are data for the distribution of alloying elements in steels type 65Kh6M3F2TO.2 with $V/C \cong 3$ [5] and 65Kh6M3FSB with $V/C \cong 4$ [6], and also in steel R6M5 [8]. It can be seen from Table 2 that the greatest amount of carbide phases is contained in steel R6M5 (about 20%), and the least in steel 65Kh6M3F3SB (about 7%).* Solid solution was more alloyed in steel with $V/C \cong 4$, when the total content of carbide-forming elements was about 7%, and an intermediate position is occupied by steel with $V/C \cong 3$ and total content of these elements of 5%, and in solid solution of annealed steel R6M5 there is significantly less chromium and vanadium, i.e., in total about 3% of the carbide-forming elements.

The austenitizing process was studied in detail with which there is dissolution of excess carbide phases. The completeness of their dissolution was determined with an increase in temperature from 1100 to 1300°C by measuring hardness after quenching, secondary hardness, red hardness (hardness of quenched and annealed steel after additional tempering at 620°C for 4 h), and dilatometrically with determination of austenite grain growth kinetics.

Results of determining the effect of austenitizing temperature on the properties of the test steels showed that independent of the V/C ratio titanium content for steel with 3 and 5% Mo they have practically the same hardness after quenching, secondary hardness, red hardness, and austenite grain size. Thus, an increase in molybdenum content in the steel from 3 to 5% does not improve its properties and therefore it is not rational (demonstrated previously in [1]).

Steels with 0.2 and 0.4% Ti independent of the concentration of the rest of the alloying elements have almost identical hardness after quenching, secondary hardness, and austenite grain size. A change in titanium content from 0.2 to 0.4% only affects red hardness of the test steels. With $V/C \cong 3$ the highest red hardness occurs for steel with 0.4% Ti,

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*Carbide phase content is indicated in weight fractions; in determining carbide phase content in volume fractions this difference would be insignificant. (Editor's note).

TABLE 2

Steel	Content of elements in solid solution, %							K*, %	Content of alloying elements in carbides, %								Phase composition of carbides
	C	W	Cr	V	Mo	Nb	Ti		C	W	Fe	Cr	V	Mo	Nb	Ti	
95Kh6M3F3T0.4	0,02	—	5,5	0,74	0,86	—	0,06	10,0	9,2	—	23,0	20,6	20,3	23,8	—	3,1	M_6C, $M_{23}C_6$, MC, M(C, N)
95Kh6M3F4T0.4	0,02	—	5,4	1,10	0,91	—	0,03	8,4	10,9	—	14,4	12,7	31,1	26,6	—	4,4	MC, $M_{23}C_6$, M_6C, M(C, N)
65Kh6M3F3SB	0,01	—	5,1	1,30	1,10	—	—	7,2	8,4	—	17,4	17,7	25,0	30,4	1,1	—	MC, $M_{23}C_6$, M_6C, M(C, N)
65Kh6M3F2T0.2	0,01	—	4,6	0,72	0,57	—	—	8,2	7,2	—	29,0	18,8	14,0	28,2	—	2,7	M_6C, MC, $M_{23}C_6$, M(C, N)
R6M5	0,03	0,4	2,9	0,20	0,70	—	—	21,3	3,9	27,5	29,7	8,8	8,1	22,0	—	—	M_6C, MC, $M_{23}C_6$

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*K is the amount of carbide phases.

TABLE 3

Steel	Content of elements in solid solution, %							K*, %	Content of alloying elements in carbides, %								Phase composition of carbides
	C	W	Cr	V	Mo	Nb	Ti		C	W	Fe	Cr	V	Mo	Nb	Ti	
95Kh6M3F3T0.4	0,63	—	6,0	1,8	2,7	—	0,06	2,6	12,7	—	7,7	10,0	36,9	21,2	—	11,5	MC, M(C, N)
95Kh6M3F4T0.4	0,45	—	5,9	1,9	2,3	—	0,04	3,9	12,9	—	3,9	7,5	45,2	21,2	—	9,3	MC, M(C, N)
65Kh6M3F3SB	0,43	—	5,9	2,4	2,3	0,03	—	1,7	12,0	—	9,6	10,2	43,7	21,6	3,0	—	MC, M(C, N)
65Kh6M3F2T0.2	0,46	—	5,7	1,5	3,0	—	0,02	1,3	11,5	—	18,3	9,9	21,4	23,7	—	15,3	MC, M(C, N)
R6M5	0,46	3,3	4,3	1,2	3,2	—	—	10,4	4,3	30,8	30,5	3,3	8,1	23,0	—	—	M_6C, MC

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*K is the amount of carbide phases %.

Note. Carbon content in tungstenless steels was calculated from consideration of carbide MC and carbonitride M(C, N) formation.

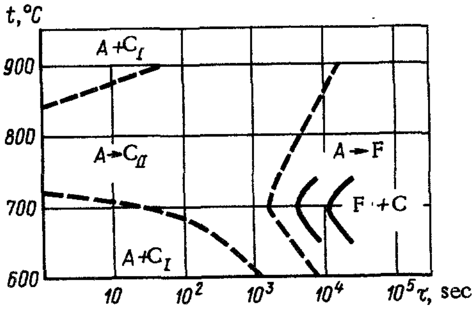

Fig. 3

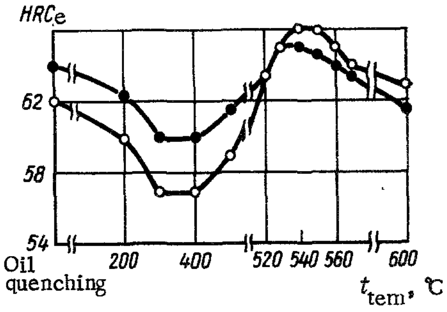

Fig. 4

Fig. 3. Isothermal decomposition diagram for supercooled austenite of steel 95Kh6M3F3T0.4 (austenitizing at 1225°C).

Fig. 4. Effect of three-stage tempering temperature on the hardness of steels 95Kh6M3F3T0.4 (○) and 95Kh6M3F4T0.4 (●).

and with $V/C \cong 4$ for steel with 0.2% Ti. A basic factor affecting hardness after quenching, austenite grain size, secondary hardness, and red hardness of the test steels is the V/C ratio (Fig. 1). Steel with $V/C \cong 3$ has higher secondary hardness and red hardness. Thus, on the basis of the results it has been established that the best properties are exhibited by steel 95Kh6M3F3T0.4 after oil quenching from 1225-1250°C.*

Dilatometric studies of carbide dissolution in this steel showed that during soaking at the austenitizing temperature, as also in steel 65Kh6M3F3SB [7], a reduction in specimen length (or volume) was observed (Fig. 2) caused by carbide disolution in austenite. The higher the austenitizing temperature, the greater the reduction in specimen length and the lower the temperature for the start of martensitic transformation (Fig. 2). From these data it also follows that the most complete carbide dissolution in steel 95Kh6M3F3T0.4 occurs at 1225-1250°C.

Given in Table 3 [6, 8, 9] is distribution of alloying elements between solid solution and carbides in tungstenless steels, and for comparison in steel R6M5. It can be seen that the total content of carbide-forming elements in solid solution (martensite) for all of the test steels, including R6M5, is almost the same and it is 10-11%.† The difference involves the fact that instead of ∿3% W (steel R6M5) the martensite of tungstenless steels additionally contains about 2% Cr and about 1% V.

In addition, as calculation shows, carbon concentration in martensite of steel 95Kh6M3F3T0.4 is one and a half times greater than in steels R6M5, 95Kh6M3F4T0.4, 65Kh6M3F3SB, and 65Kh6M3F2T0.2. In the test tungstenless steels there remain particles of MC carbides and M(C, N) carbonitrides insoluble in small amounts, and in steel R6M5 there is no transfer into solid solution of a much greater amount of M_6C carbides, which as for MC carbides retard grain growth during heating for hardening.

For a more complete idea about the austenite condition obtained after heating to the hardening temperature close to the optimum value, its stability was determined in steel 95Kh6M3F3T0.4 primarily with continuous cooling at rates from 3000 to 3°K/min. It was established that with this cooling austenite only undergoes martensitic transformation. The M_s temperature during cooling with rates from 300 to 50°K/min does not change (M_s = 210°C), but with a further reduction in cooling rate it gradually increases from 210°C at 50°K/min to 310°C with 3°K/min, which is connected with partial occurrence during cooling of impoverishment of supercooled austenite in carbon and alloying elements, and also as a result of its heterogenization.

A diagram was also plotted for isothermal decomposition of supercooled austenite of steel 95Kh6M3F3T0.4 and the region of intense carbide precipitation was determined (Fig. 3). It can be seen that at 900°C the start of carbide precipitation is recorded with a dilatometer after 1 min, and the end after 300 min. Exposure at 900°C for 360 min does not lead to polymorphic $\gamma \rightarrow \alpha$-transformation, and during subsequent cooling the steel experiences

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*Grain size is No. 8-9 (Fig. 2, curve 2), which is hardly admissible (Editor's note).

†Summing is permissible if the content of elements is given in atomic fractions. (Editor's note).

TABLE 4

Steel	HRC_e	Grain No.	σ_b, N/mm^2	a J/cm^2	Kp_{58}, °C
95Kh6M3F3T0.4	66	8 9	3100 3500	30. 40	620
65Kh6M3F3SB	63	7 .8	3800 4200	60. 70	590
65Kh6M3F2T	64,5	7 .8	2800 3200	20. 30	610
R6M5	64 . 66	10...11	3150. .3550	30...40	620

Note. Red hardness was determined after additional tempering for 4 h at different temperatures for different steels to a hardness not below HRC 58.

only martensitic transformation (M_s = 370°C). In the temperature range 840-720°C carbides commence to precipitate from austenite almost immediately as the specimen reaches the given temperature. At 720°C the end of the carbide precipitation process is recorded by the instrument after about 30 min, and the start of polymorphic $\gamma \to \alpha$-transformation is determined after 80 min. Exposure at this temperature for 300 min leads to complete austenite decomposition, which is indicated by absence of a martensitic point in the dilatogram during subsequent cooling. The least austenite stability for steel 95Kh6M3F3T0.4 is exhibited at 700°C. The start of carbide precipitation in this case occurs after about 1 min, and the end after about 30 min. Polymorphic $\gamma \to \alpha$-transformation commences after about 60 min and it is completed after about 200 min. At 650°C the process of carbide precipitation commences after about 10 min and it ends after about 40 min, but polymorphic $\gamma \to \alpha$-transformation develops during 180-400 min. At 600°C carbide precipitation commences after about 20 min and it ends after about 120 min. Isothermal soaking at this temperature for 400 min does not lead to austenite decomposition (during subsequent cooling there is martensitic transformation with M_s = 305 °C). Comparison of the isothermal decomposition diagrams for supercooled austenite of steel 95Kh6M3F3T0.4 (Fig. 3) and steel 65Kh6M3F3SB [7] showed that with an increase in carbon concentration from 0.65 to 0.95% there is a reduction in the temperature for least stability of supercooled austenite (from 800 to 700°C) and an increase in its stability towards polymorphic $\gamma \to \alpha$-transformation. On the basis of data for transformation of supercooled austenite with continuous cooling and under isothermal conditions it is possible to select a rational cooling schedule with normal or isothermal hardening for steel of this type.

After oil quenching from 1225-1250°C steel 95Kh6M3F3T0.4 contains 35-40% residual austenite, and steel 95Kh6M3F4T0.4 contains 25-30%. For sufficiently complete transformation of this amount of residual austenite, as studies have shown, it is necessary to carry out three-stage tempering. The effect of temperature of this tempering on the hardness of steels 95Kh6M3F3T0.4 and 95Kh6M3F4T0.4 is shown in Fig. 4. The reduction in hardness after tempering at 300 and 400°C is connected with a reduction in microstresses, the level of carbon concentration in martensite, and also with precipitation and coalescence of cementite-type carbides. It should be noted that the reduction in hardness in this temperature range for steel $V/C \cong 3$ is somewhat greater than for steel with $V/C \cong 4$. The maximum hardness was observed after tempering at 540-550°C for steel 95Kh6M3F3T0.4 and at 530-540°C for steel 95Kh6M3F4T0.4. After single- and two-stage tempering the hardness of steel 95Kh6M3F3T0.4 is HRC_e 63.5 and HRC_e 64.5, and for steel 95Kh6M3F4T0.4 it is HRC_e 64 and HRC_e 64.5 respectively. After a third temper the hardness increased to HRC_e 66 and HRC_e 65 respectively, and the amount of residual austenite in both steels did not exceed 5%.

The main properties of steel 95Kh6M3F3T0.4 after heat treatment (oil quenching from 1225-1250°C and three-stage tempering at 540-550°C for 1 h) are given in Table 4 [3, 5, 9]. Given for comparison are the properties of tungstenless steels similar in composition, and also stainless high-speed R6M5. It can be seen that steel 95Kh6M3F3T0.4 surpasses other tungstenless steels in hardness and red hardness, and it is no worse than steel R6M5 for basic properties. Preparation and sharpening of a tool of steel 95Kh6M3F3T0.4 was carried out by cutting and grinding schedules adopted for steel R6M5; correction of these schedules was not required.

Industrial proving of this steel in engineering enterprises of the country has shown that in the majority of cases it may substitute for steel R6M5.

CONCLUSIONS

1. The best properties of the tungstenless steels are exhibited by 95Kh6M3F3T0.4 (V/C≅3) With V/C≅4 there is a reduction in secondary hardness and red hardness for this steel, but with a reduction in titanium content from 0.4 to 0.2% secondary hardness does not change, but red hardness decreases. It is confirmed that an increase in molybdenum content from 3 to 5% does not improve steel properties and therefore it is not rational.

2. After oil quenching from 1225-1250°C and three-stage tempering at 540-550°C for 1 h each steel 95Kh6M3F3T0.4 is not surpassed by steel R6M5 in secondary hardness, red hardness, impact strength, and bending strength.

3. The high properties of steel 95Kh6M3F3T0.4 are governed by the almost identical degree of solid solution alloying as with steel R6M5. The total content of carbide-forming elements in martensite after quenching these steels is 10-11%.

4. With cooling at a rate from 3000 to 3°K/min steel 95Kh6M3F3T0.4 only experiences martensitic transformation. Isothermal decomposition of supercooled austenite for this steel commences with intense carbide precipitation, after which there is polymorphic $\gamma \to \alpha$-transformation.

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