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IMPROVING THE PROPERTIES OF SPRING AND SPRING STEELS BY CHANGING THEIR STRUCTURE

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Abstract- A large number of various spring alloys are used in industry, since depending on the service conditions of the elastic elements, as well as their shape and size, they must differ in mechanical, physicochemical and technological properties. Therefore, spring alloys have different chemical composition and structure and are subjected to various types of heat treatment. Classification of spring alloys can be carried out according to a number of features. Firstly, alloys can be divided by the main methods of strengthening. Steels and alloys, strengthened by cold plastic deformation and subsequent low-temperature heating (tempering or annealing). This group includes steels with an increased carbon content (0.4-1.3%), subjected to strengthening by cold plastic deformation (after preliminary heat treatment), and after cold working - low tempering. In this case, preliminary heat treatment makes a certain, and in some cases, a very significant contribution to the overall strengthening. Depending on the structure obtained during heat treatment, the absolute degree of strengthening in the process of cold plastic deformation also changes. During the cold working process, the phase state of the steel remains practically unchanged, but only the fine structure changes due to an increase in the dislocation density. As a result of additional tempering (aging), residual stresses are removed, dislocations are redistributed, segregations are formed on them, which causes a strong increase in the elastic limit, relaxation resistance and other properties of the steel.

Keywords: Steel, Iron, Nickel, Plastic, Deformation, Structure.

1. INTRODUCTION

In addition to carbon and alloy steels of the pearlitic class, this group also includes austenitic steels that do not undergo phase transformations of the $\gamma \rightarrow \alpha$ or $\gamma \rightarrow \varepsilon$ type during cold plastic deformation. During subsequent tempering, essentially the same structural processes occur in them as in pearlitic steel, with the only difference being that during tempering, segregations of other types may form and the dislocation structure may be different.

1. Steels and alloys strengthened by work hardening (strain hardening).

In addition to steels, this group also includes alloys of copper (single-phase brass and bronze), molybdenum and rhenium, niobium, etc. As a result of cold plastic deformation, not only does the density of dislocations increase in them, which are arranged as a cellular substructure or flat clusters, but the original distribution of atoms is also disrupted. During subsequent pre-recrystallization annealing, residual stresses are removed, dislocations are redistributed according to the polygonization type, and changes in the distribution of component atoms occur.

A characteristic feature of all alloys of the class under consideration is the anisotropy of elastic properties, which is sharply expressed in the deformed state and decreases after tempering (or during pre-recrystallization heating) as a result of the redistribution of stresses and dislocations.

2. Steels and alloys strengthened as a result of martensitic transformation.

This class includes carbon and alloy steels (including martensitic-aging steels), strengthened as a result of martensitic transformation during quenching, as well as stainless steels of the transition austenitic-martensitic class, undergoing martensitic transformation either during quenching or cold treatment, or during strain hardening. As a result of martensitic transformation, crystallites are refined, the dislocation density sharply increases, changing substructure, segregations of carbon atoms are formed on dislocations, and possibly excess phases are released. Then, martensite quenching creates high hardening of steel.

If martensitic transformation occurs during deformation work hardening of austenite, then with a general increase in the density of dislocations, their uneven distribution occurs. The consequence of this is anisotropy of strength and plastic properties, but especially the elastic limit. All these steels with a martensitic structure acquire

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maximum resistance to small plastic deformations after additional tempering, during which residual stresses are removed, redistribution of dislocations occurs, possibly by the type of polygonization, movement of carbon atoms to dislocations, as well as the possible release of particles (areas) of excess phases.

3. Steels and alloys hardened by dispersion.

This group includes alloys based on iron hardening (aging) systems. nickel, iron-nickel-chromium, nickel-chromium, etc., with additives mainly of titanium and aluminum, which create hardening phases that dissolve at the hardening temperature and are released in dispersed form during subsequent aging. This group also includes one of the main spring starters, beryllium bronze alloys, precipitation hardening brasses, and some alloys based on precious metals (platinum-silver, palladium-silver, etc.). The increase in resistance to small plastic deformations in these alloys is associated with a change in the fine structure of the matrix and with the blocking effect of particles (areas) of the excess phase. An additional increase in the strength properties of these alloys, accompanied, however, by a decrease in the plasticity properties, can be achieved as a result of plastic deformation of these alloys after quenching and final tempering (aging).

4. Alloys hardened as a result of internal oxidation.

This group includes silver-based alloys, for example, silver-magnesium-nickel, in which, when heated in an oxygen-containing atmosphere, the solid solution is saturated with oxygen, and particles of magnesium oxides are formed. These changes in the phase state of the alloy cause significant hardening. This classification highlights the main methods of strengthening. However, as indicated above, the most promising direction for obtaining high strength properties in existing alloys and for creating new high-strength spring alloys is the combination of several structural strengthening methods (thermomechanical treatment, etc.) in each alloy. In this case, the classification, even by the main strengthening methods for each group of alloys, loses its certainty and becomes too complex and at the same time not clear enough. Secondly, spring alloys can be classified by purpose.

5. General-purpose spring alloys (steels).

General-purpose spring alloys used as structural materials should mainly have high resistance to small plastic deformations, high tensile strength with sufficient viscosity and plasticity, as well as increased fatigue strength and relaxation resistance. Special-purpose Spring alloys. These alloys, distinguished by their improved mechanical properties, as well as high resistance to fatigue and stress relaxation, must also have certain physicochemical or physical properties, the requirements for which vary depending on the service conditions of the corresponding elastic elements. In particular, these alloys may be required to have specific corrosion resistance, non-magnetism, low electrical resistance, independence of the modulus [4].

In what follows, we adhere to the classification by purpose dislocations formed during straightening create stresses of the erratic sign and previously existing dislocations annihilate. In the process of warm deformation (straightening), the Bauschinger effect is almost not manifested, since the dislocations arising in this case are redistributed and fixed by carbon atoms. Therefore, annihilation of previously existing dislocations does not occur. These results show that cold deformation operations (upsetting, straightening, etc.) in the manufacture of springs from patented wire and their stabilization are dangerous, since they can cause softening. It is better to use warm deformation, especially during stabilization (forced deformation, etc.).

With increasing wire diameter, due to slower cooling and local increase in bath temperature at the point of wire entry, there is an ever-increasing risk of austenite decomposition during cooling in the temperature range corresponding to the formation of coarsely differentiated structures. This leads to deterioration of the entire complex of mechanical properties after patenting and further drawing. Therefore, the larger the wire diameter, the lower the temperatures of the patenting baths. The duration of holding during tempering is determined by the rate of heat removal from the wire into the bath and the time of decomposition of supercooled austenite [2].

2. INFLUENCE OF COLD PLASTIC DEFORMATION AND TEMPERING ON THE STRUCTURE AND PROPERTIES OF NORMALIZED SPRING STEEL

Deformation hardening of spring steel provides a high complex of mechanical properties not only after preliminary patenting, which was discussed above, but also after normalization. This is explained by the fact that even when cooling in air during the normalization of semi-finished products or parts of a relatively small cross-section made of steel with an increased carbon content and especially low- or medium-alloyed steel, a structure of thin-plate sorbite is formed, which differs little from that obtained during patenting. The use of normalization instead of patenting is more cost-effective. In addition, patenting is not applied to the processing of a number of profiles from which elastic elements are made - wires of increased cross-section, rods and strips.

The method of strain hardening of normalized steel grades S-65A (0.65-0.70% C; 0.45-0.55% Mn; 0.15-0.20% S; $\leqslant$0.025% S; $\leqslant$0.025% P), 65G and 50 KhG turned out to be very effective for the manufacture of many very important types of springs Figure 1 shows the effect of plastic deformation the strength properties of steels S-65A and 65G after normalization 830-850 °C and at 910-930 °C, respectively. It follows from these data that the deformation of normalized steels sharply increases their strength properties - strength and proportionality limits. Further improvement in strength properties, although with some reduction in plasticity, is achieved after additional tempering, which causes the same substructural changes as tempering of deformed steel after preliminary. The most important fact is that as a result of tempering the deformed steel its fatigue strength increases significantly [3].

M.A. Kotkis explains this effect by a decrease in the mobility of dislocations due to the formation of Cottrell and Snook atmospheres and the counteraction of other dislocations. It is important that leaf springs made of normalized steel 65 G after deformation and tempering have a significantly higher fatigue strength than after conventional heat treatment (Figure 2).

The same effect was found when testing springs made of C-65A steel (Figure 3) and samples made of 50 X G spring steel (Figure 4). It is interesting that tests of spring sheets (compression at 30% deformation) confirmed better resistance after deformation and aging (106,000-255,000 cycles before failure) than after quenching and tempering (70.000-150.000 cycles before failure). Spring tests were also favorable.

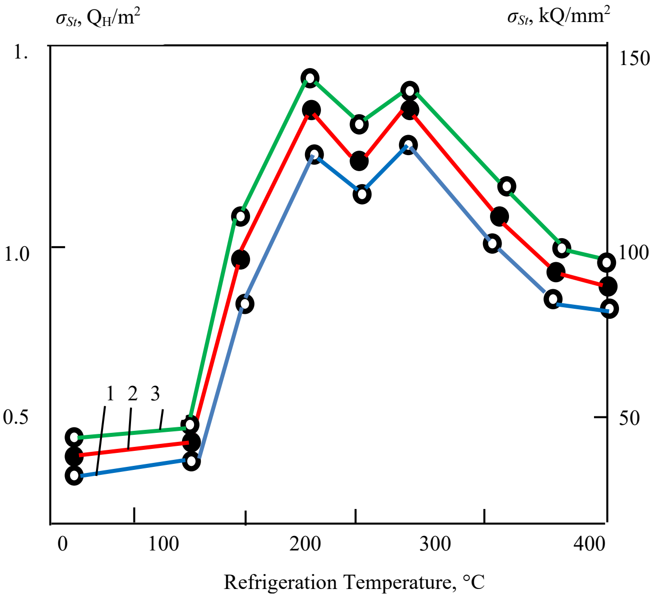

Figure 1. Dependence of mechanical properties of steels 65G (top) and S-65 A (bottom) on the degree of deformation (numbers on the curves) and tempering temperature

Currently, the first option is being implemented at the Azerbaijan Pipe Rolling Plant for casing pipes with a diameter of 114-168 mm, combining the TMO process with the established hot deformation scheme on installations with an automatic mill: For this purpose, hardening devices and a furnace for tempering pipes are introduced into the gap between the rolling and calibrating mills. After hot rolling on the automatic mill, the pipe blanks are fed to the rolling mills, where they are rolled to the required size. To increase the effect of HTMO, it is envisaged to install more powerful rolling mills with reductions in wall thickness up to 20% (instead of 3-5% on existing mills). After rolling, the pipes are subjected to hardening in spray devices combined with the equipment on the output side of the rolling mills. Heat hardening is applied at the Sumgait Pipe Rolling Plant to spiral-seam gas and oil pipeline pipes with a diameter of 820-1200 mm and a wall thickness of 8-12.5 mm.

The main differences in the process flow chart and equipment are as follows: an additional sprayer is installed behind the tempering furnace to reduce heat emission into the shop and reduce distortion of the geometric dimensions of the pipes; the designs of the devices transporting the pipe through the furnaces and the sprayer have been changed, tangential-axial, flat-jet sprayers are used. A heating mode is also used, allowing for a small distortion of the geometry of the pipes before hardening [1].

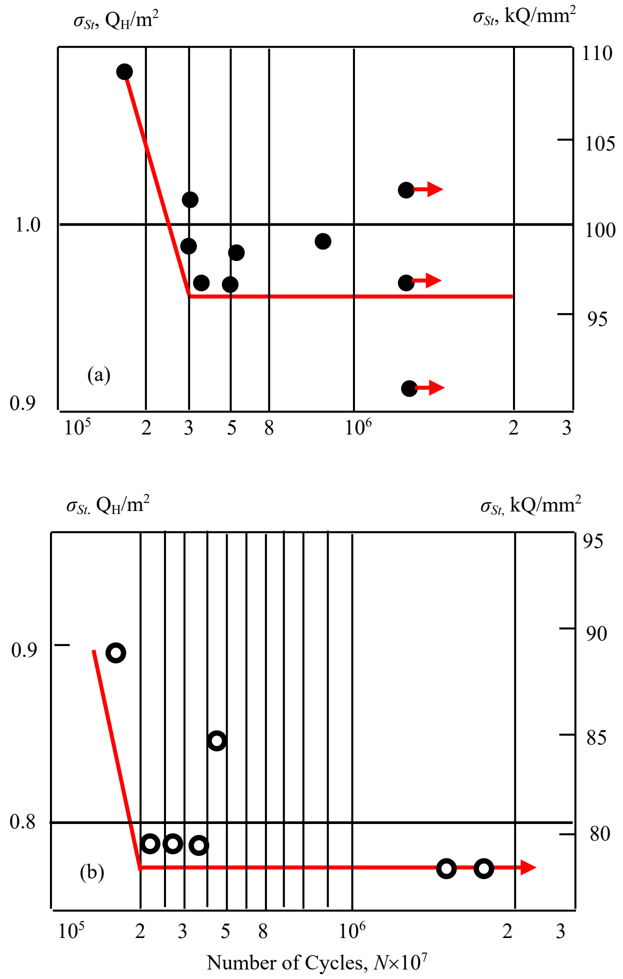

Figure 2. Dependence of the elastic limit of steel 65G after normalization and deformation with compression of 50% from the tempering temperature (M.A. Kotkis): 1. $\sigma_{0.002}$; 2. $\sigma_{0.005}$; 3. $\sigma_{0.001}$

As a result of the conducted research, which showed a certain improvement in the properties of springs and leaf springs as a result of the use of normalization, cold plastic deformation and tempering, a technological process called plastic-thermal treatment was developed at the Gorky Automobile Plant. In this case, the normalized rolled stock or wire is subjected to cold plastic deformation (rolling or drawing) with a compression of 40-60%. Various elastic elements are made from the strip, tape or wire obtained in this way by cutting, molding or winding. These operations are followed by final processing - tempering or aging at 280-300 °C for 20-40 minutes. In this case, not only high mechanical properties and high durability of the springs under service conditions are achieved.

It is known that isothermal treatment of lower bainite provides a combination of high strength in steel along with increased plasticity and toughness. Taking into account these latter properties, it is possible to perform strain hardening of steel with a lower bainite structure.

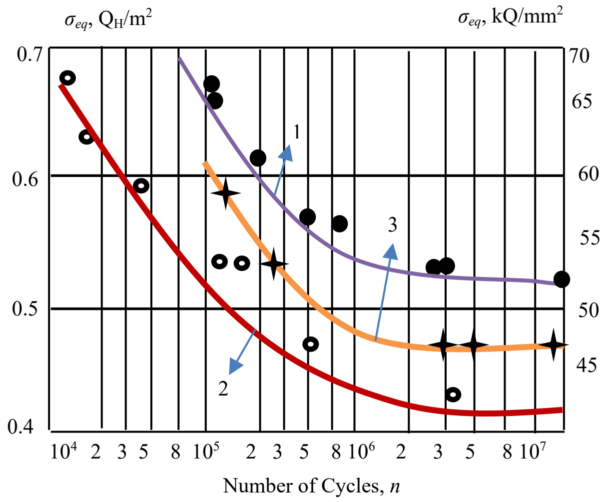

Figure 3. Fatigue strength of valve springs: 1. steel C-65A after deformation aging (normalization, deformation and tempering); 2. the same, after hardening and tempering; 3. steel 70 after patenting, deformation and tempering Q$_{MH}$ /m^2

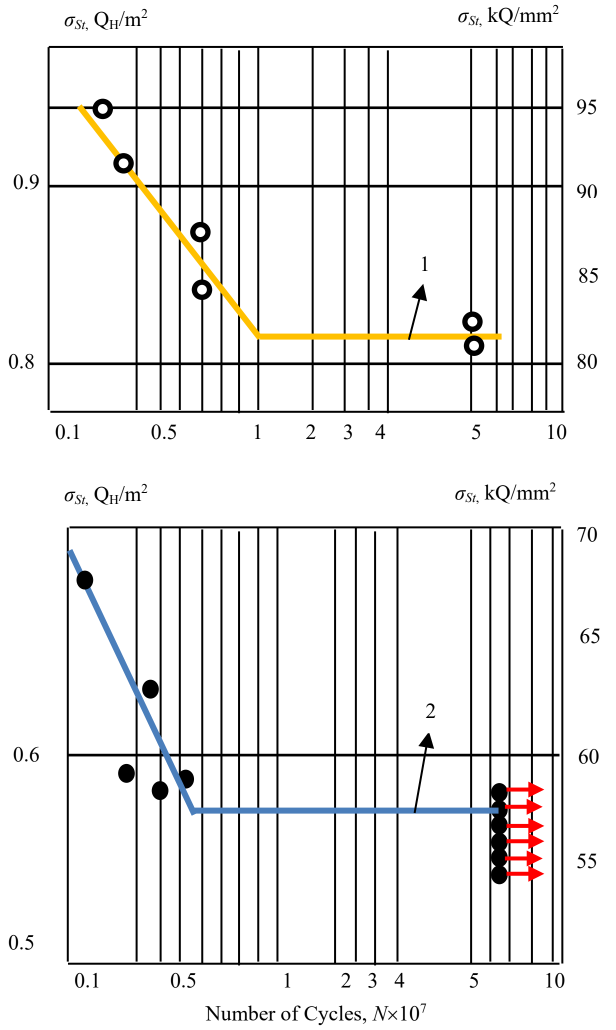

Figure 4. Fatigue strength of steel 50 KhG hardening, 1. strain hardening and tempering; 2. hardening with tempering ensures precision and simplifies and reduces the cost of the technological process of hardening) preparation of springs

This was first shown by A. A. Sazonova [176] using the example of U7A steel, which was subjected to isothermal treatment in the region of pearlite (at 430-500 °C and intermediate transformation (at 300 °C) and subsequent drawing with a reduction of 92%. It was shown that pre-treatment to lower bainite (transformation at 300M$_H$) provides the highest strengthening [σ_B =3030 M$_H$/mm^2 °C (309 kQ/mm^2)] while maintaining almost the same ductility (ψ=35%) as after conventional patenting (transformation at 500 MH/mm^2, 2 min), although the strength in the latter case is significantly lower [σ_B=1910 M$_H$/mm^2 (195 kQ/mm^2)]. showed that after patenting 50XFA steel for sorbite and lower bainite, its strength (σ_B) is practically the same: 1070 M$_H$/mm^2 (109.7 kQ/mm^2) and 1050 M$_H$/mm^2 (107.5 kQ/mm^2), respectively. In the same way, the strength of steel in the specified initial structural states and after deformation differs little. Thus, after drawing with a reduction of 75%, the strength limits are 1655 M$_H$ /mm^2 (169.2 kQ/mm^2) in the first case and 1675 M$_H$/mm^2 (170.7 kQ/mm^2) in the second. At the same time, it is technologically more convenient and economical to carry out patenting at 420 °C on lower bainite in saltpeter baths, which are widely used in industry, than at 600 °C on sorbitol in lead.

Since the duration of the transformation of supercooled austenite at 420 °C requires a fairly long time, proposed two new compositions of chrome-vanadium steel: 60 Kh05F and 67 Kh05F, in which, due to a decrease in the chromium content to 0.5%, the decomposition period of supercooled austenite was reduced by about three times. As showed, patenting on lower bainite with subsequent deformation and hardening using electric contact heating ensures a high level of mechanical properties of spring wire made of chrome-vanadium steel. In this case, the significance of patenting on lower bainite also lies in the fact that in the process of austenite transformation, there is no redistribution of alloying elements between the present phases. Therefore, even with subsequent high-speed heating to the austenitic state for hardening, the composition of the austenite quickly levels out, which is also facilitated by the previous deformation. Thus, there are certain advantages of using patenting of alloy steel for lower bainite.

In the work [3], a comparison of strain hardening of spring steel grades U8A and 80S2KhA was carried out practically. If the transformation temperature for lower bainite is reduced, its strain hardening will increase significantly and will exceed the hardening of steel with sorbite structure. in all structural states: quenched martensite, tempered martensite (tempering at 100 and 200 °C), tempered sorbite (tempering at 600 °C), isothermal quenching to lower bainite (transformation conditions: 300 °C, 1 h), patenting of the sorbite structure (for steel U8: 520 °C, 15 min; for 80C2XA: (600 °C, 20 min). After processing the above types, the steel was deformed with a large reduction (samples with a martensitic structure could hardly be deformed). The greatest strengthening as a result of drawing with a reduction of ~80% is achieved after preliminary isothermal treatment on lower bainite, although the plasticity of the wire turned out to be lower than after conventional patenting on sorbite. The lowest properties were obtained for steel with a tempered sorbite structure (Table 1). The data in Table 3 show that patenting on lower bainite is very effective.

Table 1. Mechanical properties of some spring and spring steels

Dependence of properties of steel 8 a and after cold plastic deformation with reduction of 80% of the original structural state			
		Heat treatment	Drawing with 80% compression
Steel	Processing mode	σ_{el}, M_H/m^2 kQ/mm^2	σ_{el}, M_H/m^2 kQ/mm^2
U8A	Patenting (520 °C, 15 min)	1080 (110)	1470 (150)
80C2XA	Patenting (600 °C, 20 min)	1190 (122)	1760 (180)
U8A	Quenching and tempering (600 °C, 1 h)	1030 (105)	1420 (145)
80C2XA	Quenching and tempering (600 °C, 1 h)	1130 (115)	1760 (180)
U8A	Patenting of lower bainite (300 °C,1 h)	1420 (145)	1960 (200)
80C2XA	Patenting of lower bainite (300 °C, 1 h)	1470 (150)	1900 (296)

The deformation of steel with a bainitic structure is characterized by a different pattern of strengthening than after conventional patenting - a linear increase in strength with an increase in the degree of compression (Figure 5). Equal strengths of steel with a lower bainite and sorbite structure are achieved after different degrees of compression - lower for the first structural state. Thus, for steel of the 70 K 4 Cl type, the ultimate strength of 2650 M_H/mm^2 (270 kQ/mm^2) is achieved in the case of a bainitic structure after a reduction of 43% (ψ=51%), and under conditions of conventional patenting for sorbite - after a reduction of 95% (ψ=35%). At the same time, after high degrees of reduction (90%), steels treated for lower bainite have significantly greater strength, but withstand a significantly smaller number of twists [5]. Thus, steel 65G after this treatment and deformation has a tensile strength of 2730 M_H /mm^2 with a twisting number of 32, whereas after patenting for sorbitol, the values of these properties are equal to 2420 M_H/mm^2 and 45, respectively.

However, according to the same work, the change in strength properties as a result of aging at 200 °C of steel with a lower bainite structure is less (5.9%) than after preliminary patenting (7.3%). This lower tendency to aging-induced transformation of supercooled austenite at 380-490 °C from the degree of compression is an advantage of treatment on lower bainite. Another advantage of this preliminary treatment is better relaxation resistance of steel during heating [107, p. 270], which is important for springs.

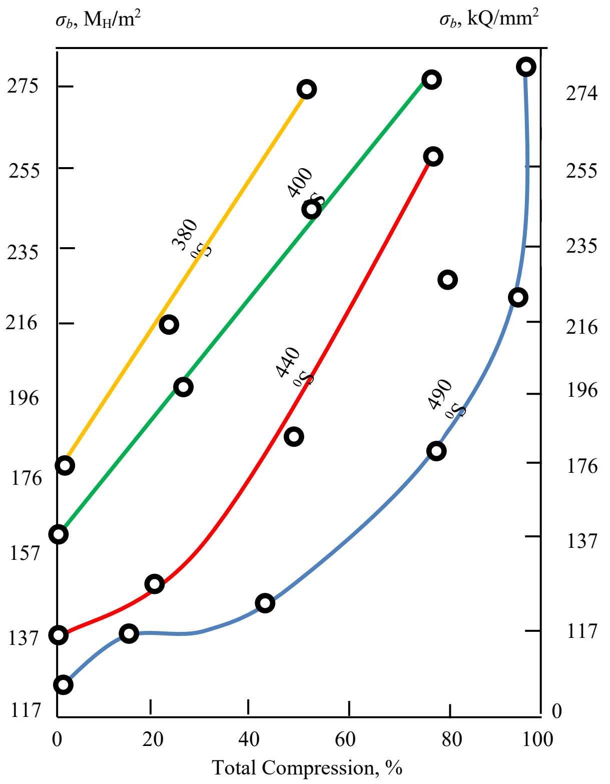

Figure 5. Dependence of the ultimate strength of 70 K 4 Cl steel after isothermal transformation of supercooled austenite at 380-490 °C on the degree of compression

3. STRUCTURE AND PROPERTIES OF SPRING STEEL AFTER HARDENING WITH SUBSEQUENT HIGH-SPEED ELECTRIC TEMPERING AND STRAIN HARDENING

The works of V.N. Gridnev, et al. [4, 6] showed that the lamellar structures formed during the transformation of austenite in the pearlite region can be even more dispersed if the martensite is subjected to high-speed electric tempering to a temperature corresponding to the third stage of the transformation 550-620 °C). In this case, along with highly dispersed plates of the α-solid solution with a crushed substructure, thin, filiform plates of carbides are formed. It is possible that this form of the precipitated carbide particles is associated with the presence of twins in the martensite crystals. For steel 70, after quenching from 1000 °C in oil and electric tempering with heating to 550 °C at a rate of 10 to 1000 deg/sec and subsequent rapid cooling, the tensile strength changes from 1350 to 1750 MH/m^2 (140-180 kQ/m^2) while maintaining high ductility.

During deformation of steel with the structure of martensite tempered under the specified conditions, practically the same processes occur as during deformation of eutectoid-type structures, but strengthening is more intense due to the high dispersion of carbides and, to a lesser extent, the features of the substructure According to it has the same character as during deformation of lower bainite, and reaches high values even after small reductions (Figure 6). The magnitude of strengthening during deformation, as well as plasticity (transverse narrowing and the number of bends with overbending) are greater, the higher the heating rate during electric tempering, since the hardness of the original structure is higher.

On the wire made of steel 70 after this treatment and deformation with a compression of 90%, the ultimate strength was 2950-3050 M_H/mm^2 (300-310 kQ/mm^2) at ψ=48-50% and n=15-16. The true strength reached 6500 M_H/m^2 (670 kQ/mm^2), i.e. it was about half of the theoretical strength of iron, which is amazing in itself. The combination of high strength and plastic properties obtained as a result of deformation of steel after high-speed electric tempering indicates that this process will be one of the most effective methods for producing high-strength spring wire from both carbon and alloy steel.

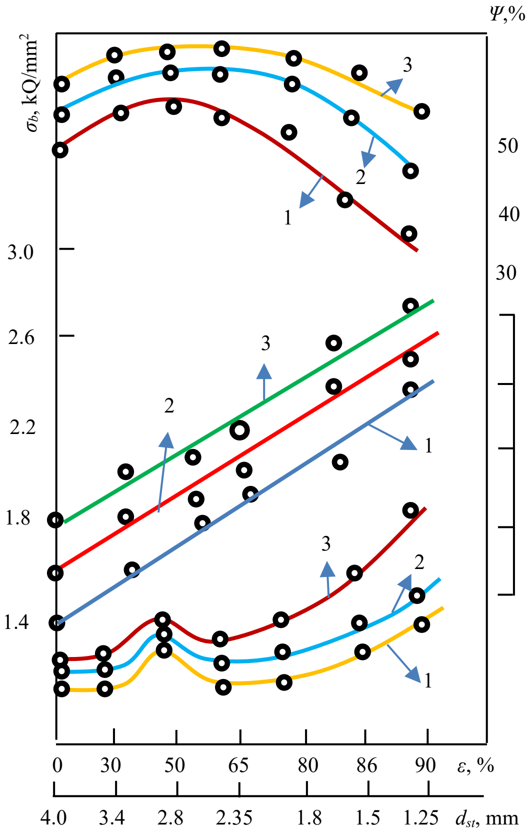

Figure 6. Dependence of mechanical properties of bends) the greater, than (n is the number of bends) on the degree of deformation of wire made of steel 70 after electric tempering with different heating rates: 1. 10 deg/sec; 2. 100 deg/sec; 3. 1000 deg/sec

4. INFLUENCE OF COLD PLASTIC DEFORMATION ON PROPERTIES OF SPRING STEEL WITH MARTENSITIC STRUCTURE

The process of strain hardening of steel with martensitic structure, or marforming, sharply increases the strength limits and especially the yield strength of steel: up to 2740-2990M_H /m^2 (280-300 kQ/mm^2) and up to 2550-2740 M_H/m^2 (260-280 kQ/mm^2) respectively, while maintaining a certain plasticity, and an increased resistance to brittle fracture is noted. It is significant that simultaneously with the strength properties under static uniaxial tension, the strength under biaxial tension also increases, which is a measure of the tendency of steel to brittle softening.

Thus, as a result of mar forming, a set of strength properties increases and therefore this process can be used to process critical parts. However, the question of using mar forming to strengthen springs has not yet been fully resolved. This is explained by the insufficient study of the effect of plastic deformation on those properties of steel with a martensitic structure that determine the main characteristics of springs. As follows from many experimental studies conducted under conditions that exclude the diffusion of carbon atoms and other impurities during hardening and subsequent testing, "fresh" martensite of hardened steel is characterized by very low resistance to small plastic deformations [6, 7] with reduced hardness and in this sense is "soft".

If during quenching, as usually happens in practical conditions, processes of diffusion redistribution of carbon atoms have time to occur, leading to the formation of segregations and even carbide particles, then the resistance to small plastic deformations, although it increases, remains at a fairly low level. This feature of the properties of freshly quenched steel is associated with the fact that the dislocation systems that arise during the martensitic transformation and create strong fields of macro- and microstresses are weakly fixed, and therefore unstable and can be reconstructed - spontaneously even before the decomposition of martensite as a result of relaxation of internal stresses and even more so under the influence of even very small external forces. Apparently, during quenching, even under conditions when relaxation of internal stresses and diffusion processes of redistribution of carbon atoms occur, the latter still do not lead to complete saturation of segregations on dislocations or on the boundaries of twins. Therefore, the resistance to microplastic deformations ($\varepsilon_{rec.}$ $\approx 10^{-4} \div 10^{-6}$ in this case, according to [8], is very low and does not depend on the content of carbon (Figures 7 and 8) and alloying elements [7, 8], if the diffusion of these atoms to dislocations was excluded during mechanical tests at low temperatures.

However, as shown in [9], the very low resistance to microplastic deformation of hardened steel tested under the specified conditions may be to some extent the result of the fact that the residual deformation that occurs under loading is a consequence not only of the movement of dislocations, but also of the isothermal transformation of austenite into martensite. According to calculations [11], it is 0.09% per 1% of transformed austenite, which agrees with direct measurements. Therefore, even with low contents of residual austenite, its decomposition under load can make a large contribution to microplastic deformation. The dependence of the resistance to small plastic deformations ($\varepsilon_{rec.}$ $\approx 10^{-4}$) on the carbon content presented by the authors [10] is also justified because (Figure 33) it is impossible to consider the studied steels with a variable carbon content (the martensite point for all steels remained the same, the amount of residual austenite in these steels was equal to -35 °C), the degree of its transformation under the action of the applied stresses, the contribution made only by the movement of dislocates to the total deformation, and this determines the value of the resistance to deformations.

Microplastic considerations, based on the data given in 3 and 34, it is impossible to quantitatively estimate the actual resistance of "fresh" martensite to the development of microplastic deformations. However, this characteristic of freshly quenched martensite should indeed have a low value, since a large number of weakly pinned dislocations arise during martensitic transformation.

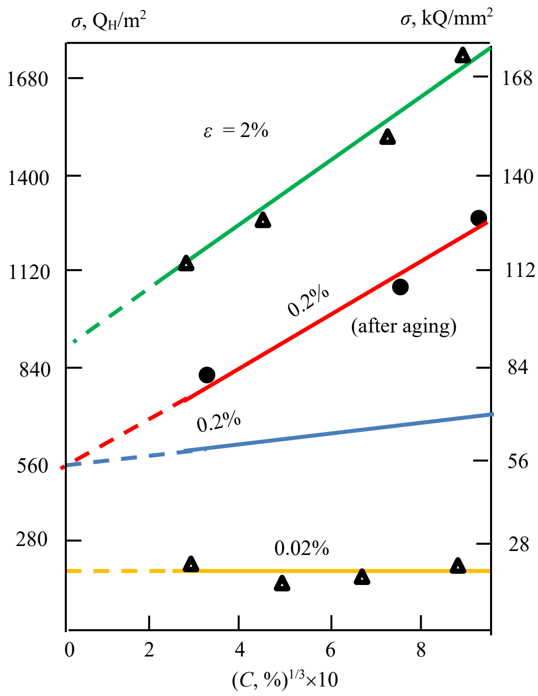

Figure 7. Flow stresses of freshly quenched martensite after various deformations depending on the carbon content (the curve for the alloy after 0.2% deformation and aging is given for comparison) [7]

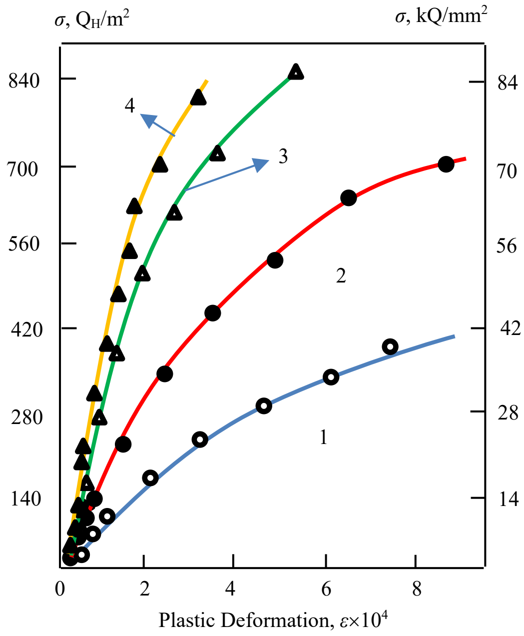

Figure 8. Comparison of the development of microplastic deformations in martensite with 0.39% C and 22.7% Ni in the unaged state and after aging at 20 °C during testing under tensile and compressive conditions at -186 °C (A - properties of steel after LTMO at 20 °C): 1. tensile tests without aging; 2. compression test without aging; 3. tensile test after aging at 20 °C, 4 h; 4. compression test after aging at 20 °C, 1 h

This difference in the degree of increase of $\tau_{0.0002}$ and $\tau_{0.02}$ is a result of the fact that the hardening coefficient increases sharply during aging (Figure 9), especially in steel with a high carbon content. The higher the carbon content in the steel, the smaller the distance between the interstitial atoms in the martensite lattice and the defects

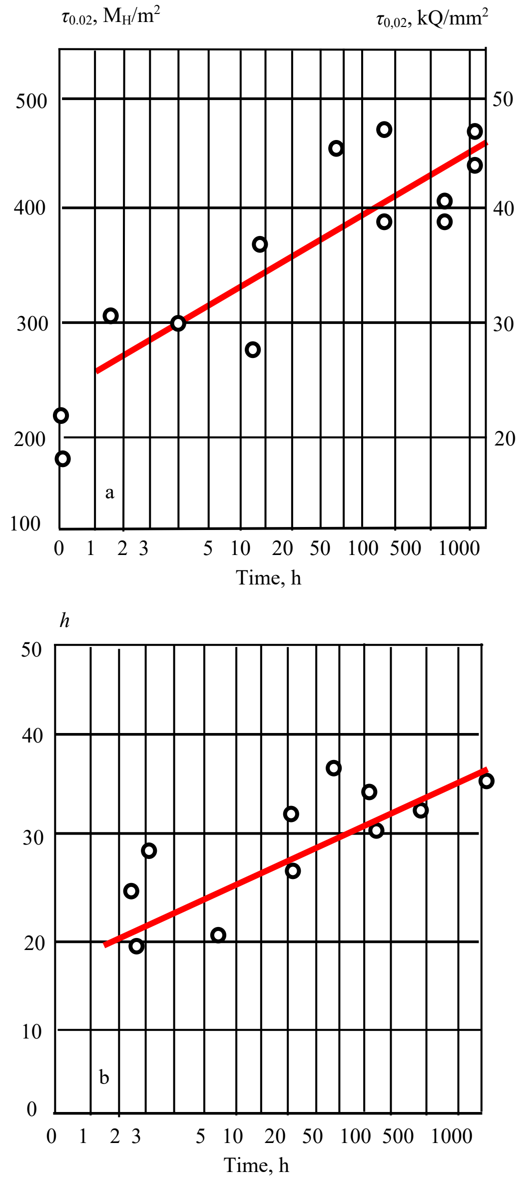

Figure 9. Effect of aging duration at 20 °C on the elastic limit, a) and hardening coefficient h, b) of hardened steel U7A

Apparently, the Bauschinger effect also plays a certain role. A manifestation of this effect is partly the difference between the resistance to microplastic deformations determined under tension and compression conditions (Figure 8). This difference, however, can also be due to the different intensity of development of the isothermal transformation of austenite into martensite under these conditions and, therefore, the process of pinning dislocations and twin boundaries occurs more quickly and efficiently, increasing the resistance to the development of microplastic deformations (Figure 10).

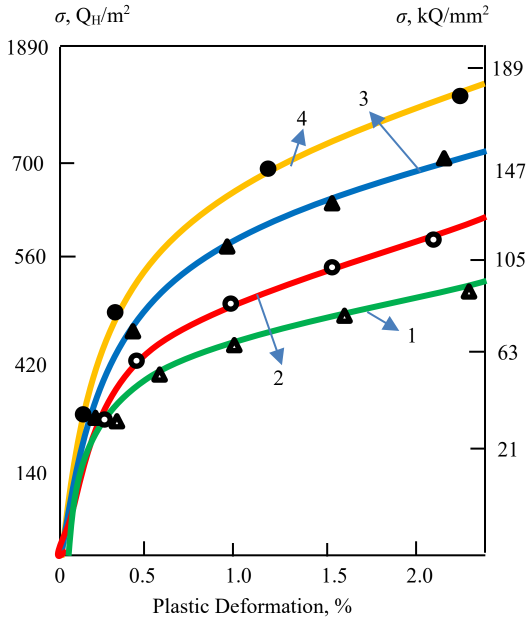

Figure 10. Strain diagram of steel at –186 °C (freshly quenched martensite) [182]. Steel compositions: 1. 0.017% C; 29.3% Ni; 2. 0.11% C; 29.4% Ni; 3. 0.39% C; 22.7%Ni; 4. 0.57% C; 19.0% Ni

Thus, from the point of view of resistance to small plastic deformations, fresh martensite is indeed “soft”, as S.T. Kishkin assumed. This position is confirmed by the data of work [12], which shows that untampered martensite cooled to -65 °C, during testing at the same temperature (without heating), i.e., under conditions that practically exclude aging, exhibits increased plasticity, a low tendency to form microcracks with high resistance to destruction. During further holding of hardened steel at normal temperatures, i.e., during aging, saturation of segregations on structural defects with carbon atoms occurs and therefore resistance to small and microplastic deformations increases. According to [13], the value of $\sigma\_{0.0002}$ obtained at –186 °C increases fourfold after aging of freshly quenched martensite at 20 °C.

Steel with a martensitic structure is characterized by rapid and effective strain hardening, which ensures high values of yield strength and tensile strength. High values of strength properties of steel with a martensitic structure are achieved as a result of deformation with a very small reduction (usually up to 5%), whereas strengthening of steel with a lower bainite or sorbite structure to the same level of properties requires significantly greater reductions. The strengthening of steel with a martensitic structure as a result of deformation is greater, the higher the content of carbon and alloying elements. The effect of carbon content on strain hardening of martensite is well illustrated by the data of work [11] (Figures 7 and 10). The tests were carried out at deformations of ~10^{-3}-10^{-2}, significantly exceeding those that arise due to the transformation of retained austenite into mar-forming.

As shown in [15], the aging effect is noticeably expressed in nickel steel with twinned martensite, whereas in steel of the same type, but with lamellar or strip martensite, this effect is not observed. However, after a small preliminary deformation (~1%) at low temperatures, aging causes an equal increase in the hardness of these steels, regardless of the initial morphology of the martensite. Therefore, at deformations of ~10^{-3}-10^{-2}, the movement of dislocations is undoubtedly of primary importance.

The mechanisms of the influence of carbon on the strengthening of martensite and supersaturated ferrite [12] are similar in many ways. Ferrite after quenching at 725 °C (before testing at 25 °C, the quenched samples were kept at -80 °C) is characterized by a very weak dependence of the flow stress [an increase of 24.1 M_H/mm^2 (2.46 kQ/mm^2)] on the increase in carbon content within the range from 0.004 to 0.02% with plastic deformation of ~10^{-6}, but with its increase to 10^{-3} (Figure 11) the effect of carbon sharply increases [an increase in stress of 144 M_H/mm^2 (14.7 kQ/mm^2)]. However, the dependence shown in Figure 11, should be considered only as a qualitative characteristic of the effect of carbon on hardening, since it cannot be asserted that all the carbon in the iron was in solid solution.

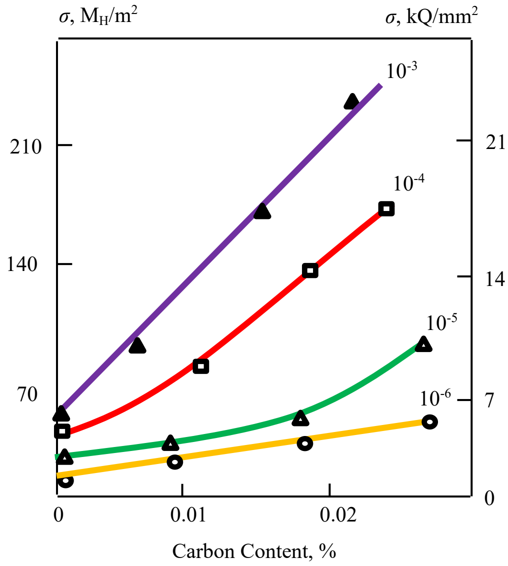

Figure 11. Flow stress of iron-carbon alloys as a function of carbon content at different values of plastic deformation

The data in Figure 12 indicate that in the case of freshly quenched ferrite, as well as freshly quenched martensite, carbon has a relatively weak effect on the resistance to microplastic deformations, i.e. on the initiation of dislocation motion, but sharply increases the hardening coefficient due to a change in the substructure. In particular, the ferrite substructure changes from cellular to random dislocation arrangement (after 6% deformation) with increasing carbon content under the action of deformation, since carbon inhibits transverse slip and dislocation multiplication. The high strain hardening of martensite is explained by the fact that dislocations arising from the so-called "seed" ones during deformation are linked into more stable systems and are effectively inhibited by numerous embedded carbon atoms and other obstacles - segregations and carbide particles. In this case, the effective length of dislocation loops continuously decreases as plastic deformation develops and the dislocation density increases, and the faster the higher the carbon content in the steel. Therefore, the stresses required to advance dislocations also continuously increase, which is manifested in the growth of strength properties. A certain contribution to the strain hardening of steel is made by the process of decomposition of residual austenite.

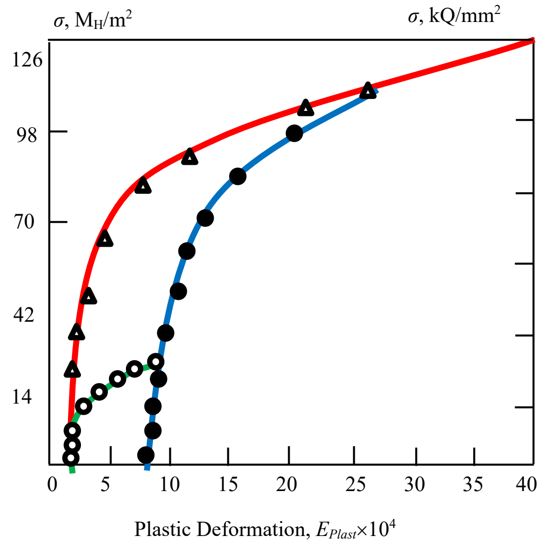

Figure 12. Effect of preliminary deformation at –186 °C and aging on the deformation diagram of steel with 0.39% C and 22.7% Ni with martensite structure [15]: 1. after quenching; 2. after aging at 20 °C, 4 h; 3. after quenching, 0.06% deformation and aging at 20 °C, 4 h

As was indicated above, the deformation of steel with a martensitic structure under conditions that exclude aging processes (deformation and testing at very low temperatures) increases the resistance to increased plastic deformations the more, the higher the carbon concentration (Figure 7). It follows that this increase in the strength properties of steel with a martensitic structure under the specified conditions is a consequence of strain hardening itself, during which a substructure is formed that depends on the carbon concentration in the steel.

Believes that the effect of carbon content on the increase in hardening is associated with a more intensive development of aging processes occurring during deformation and leading to the formation of segregations and the release of carbide particles. Although strain aging does indeed occur, but only this process, judging by the results of work [18], does not determine the high intensity of strain hardening of steel with a martensitic structure. At the same time, it has not been established what contribution to hardening is made by the process of martensite disordering, which, as is assumed, occurs under the action of plastic deformation.

The coefficient of strain hardening of hardened steel increases as a result of natural aging (Figure 12) and it is greater, the higher the carbon content (Figure 13). This increase in hardening is already evident in the region of small plastic deformations. If steel with a martensitic structure is subjected to very small plastic deformation and then aged, the strain hardening coefficient increases sharply (Figure 12).

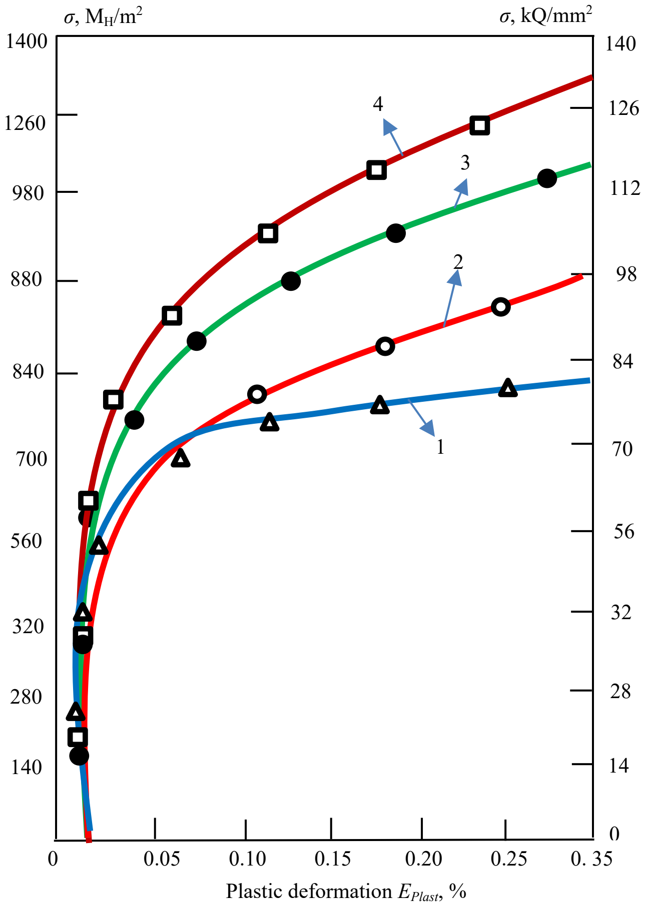

Figure 13. Deformation diagram at –186 °C of steel with martensite structure after aging at 20 °C (for steel compositions as Figure 10)

A characteristic feature of steel with a martensitic structure is the high intensity of strain hardening, and therefore, to increase strength properties while maintaining a certain plasticity, deformation with small degrees of compression is used, usually not exceeding several percent. In this case, as well as after additional tempering, the tendency of steel to brittle fracture does not increase, which is of great practical importance. It is precisely on this high tendency of tempered martensite to strain hardening that the mar forming process is based. The study of steels subjected to strain hardening after quenching and after additional tempering shows that untampered martensite is strengthened, or rather, strengthened, values of the strength and yield strength, greater than tempered, and moreover, the higher the carbon content in the steel, the stronger. Thus, according to data from [20], steel containing 0.38% C and 1.9% Cr, after quenching and deformation with a reduction of 2.5% (I) and after quenching followed by tempering at 180 °C, 1.5 h and deformation by 2.5% (II) had the following properties:

Table 2. Changes in fluidity properties in different situations

The following properties	I	II
σ_B/m^2 (kQ/mm^2)	2240 (228)	2050 (210)
σ_T/m^2 (kQ/mm^2)	2020 (204)	1940 (198)
σ_{pl}/μ^2 (kT/mμ^2)	1940 (198)	1180 (192)

The data of the same work [19] indicate the influence of carbon content on the growth of strain hardening of steel with a martensitic structure. Thus, as a result of deformation with a 2.5% reduction of quenched and low-tempered steel with 0.18% C and 1.7% Cr, the ultimate strength increases by 145 M_H/m^2 (15 k/m^2) and the proportionality limit by 190 M_H/m^2; for steel with 0.67% C and 1.8% Cr under the same processing conditions, these properties increase by 275 M_H/mm^2 (28 kQ/mm^2) and 490 M_H/mm^2 (50 kQ/mm^2), respectively, reaching σ_B = 2720 M_B/m^2 (278 kQ/mm^2) and σ_{Pt} = 2540 M_H/mm^2 (260 kQ/mm^2). At the same time, unlike carbon, alloying elements - manganese, nickel and chromium in an amount of up to 3-5% do not affect the effect of strain hardening of steel with 0.37% C (after quenching, low tempering at 170-180 °C) [16]. Unlike the influence of the indicated alloying elements, silicon increases the effect of strain hardening of hardened steel. The plasticity of alloy steel changes with increasing degree of deformation in approximately the same way as carbon steel with the same carbon content. The highest absolute values of plasticity are noted for nickel steel, and the lowest - for silicon steel, which exhibits brittleness after deformation by drawing by more than 5%.

The strengthening of quenched steel compared to tempered steel is greater, the higher the temperature of this preliminary tempering. Despite the greater strengthening of steels with the structure of "fresh" martensite, they are usually not subjected to strain hardening due to their high tendency to brittle fracture and, more often, low tempering is performed after quenching. In steel with a tempered martensite structure, during deformation, in addition to those changes in structure and properties that are characteristic of quenched steel, dissolution of metastable and coherently bound carbides with the matrix can occur, as well as the precipitation, and moreover, in larger quantities (true, only at increased degrees of deformation), of new carbide particles due to the increased decomposition of the supersaturated solid solution. With small deformations of steel with a tempered martensite structure, for the reasons stated above (unblocking of existing dislocations and mainly generation of new ones), the elastic limit should be greatly reduced while maintaining high hardness, which is proven in the work [17].

It is interesting that a similar strong decrease in the elastic limit under the action of a small additional deformation was observed on patented and cold-formed high-carbon steel and on many other alloys - solid substitution solutions, pre-hardened by significant plastic deformation and low-temperature annealing (see p. 446). Additional deformation of steel quenched for martensite can lead to a decrease in resistance not only to small but even to large plastic deformations and, in particular, to a decrease in hardness. Thus, as a result of cold rolling with reduction of 5% steel (0.44% C and 27.56% Ni), pre-hardened for martensite, the hardness decreased from 380 to 340 HV. At high compressions the hardness of such steel increased, although not very intensively [20].

The effect of softening was explained by these authors on the basis of the results of direct electron microscopic observations. It is believed that the twins of the transformation of austenite into martensite are partially detwinned as a result of deformation, which is also carried out by twinning. It is significant that the deformation twins are formed along the same crystallographic planes and with the same orientation as the transformation twins, but their magnitude is much greater (60 rm instead of 10) [18].

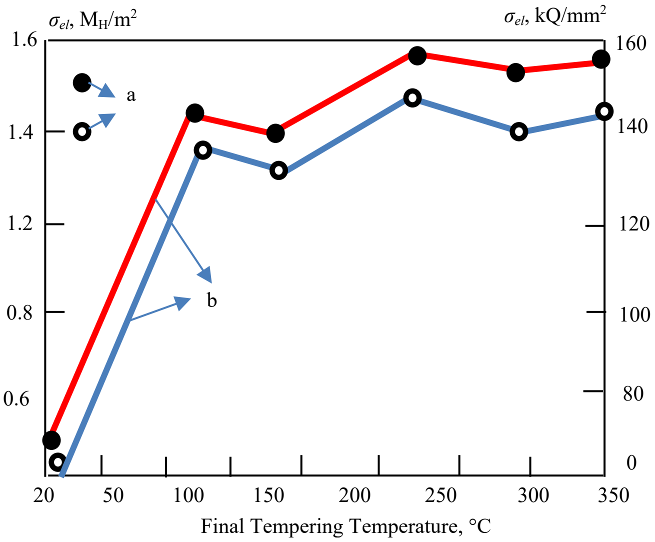

Figure 14. Dependence of the elastic limit of steel 55 KhGRTS on the final tempering temperature [193, p. 65]: a. quenching from 920 °C and tempering at 350 °C; b. the same, tensile deformation by 0.7-1%

5. CONCLUSION

1. Spring and spring steels are mainly required to have an elastic limit ($\sigma_{0.005}$).
2. As a result of the study, it was determined that the large elastic deformation of spring steels is determined by its construction, that is, the number of coils, diameter and length.
3. In order to increase elastic limit of the spring and spring steels studied in article, we subjected them to quenching at a temperature of 830-900 °C, depending on composition of the steels, first in water and then in oil, followed by medium-temperature quenching (350-500 °C).
4. Since the quenching depth was high, the structure of the quenched spring steel over the entire cross section was determined to consist of troostite after the quenching operation.
5. It was determined that in order to improve the properties of spring and spring steels, the amount of silicon in the composition should be increased to 1.8
6. The fact that the structure consists of troostite allowed us to achieve the requirements imposed on spring and spring steels.

REFERENCES

[1] V.F. Gahramanov, A.I. Hasanov, N.S. Rzayev, R.K. Hasanli, “Use of Advanced Metallurgical Technologies to Improve Properties of Structural Elements for Various Purposes”, International Journal on Technical and Physical Problems of Engineering (IJTPE), Issue 63, Vol. 17, No. 2, pp. 21-29, June 2025.

[2] V.F. Gahramanov, N.S. Rzayev, O.Sh. Salmanov, I.H. Mammadova, "Study of Increasing Properties of Pearlite Structure Steels by Cold Plastic Deformation", International Journal on Technical and Physical Problems of Engineering (IJTPE), Issue 63, Vol. 17, No. 2, pp. 103-110, June 2025.
[3] V.F. Gahramanov, E.A. Aslanov, "Influence of Preliminary Oxidation on the Process of Nitriding of Iron Alloys", Mechanics of Machines, Mechanisms and Materials, Scientific and Technical Journal, No. 2, Vol. 55, pp. 33-38, Baku, Azerbaijan, 2021.
[4] S.N. Namazov, V.F. Gahramanov, A.A. Dzhafarova, "Structural Characteristics of Powdered Iron Graphite with Subsequent Oxidation", International Scientific Practical Conference, pp. 91-94, Yurga, Russia, 2016.
[5] V.F. Gahramanov, "Diffusion Chroming and Carbochroming of Stamping Steels", Oil and Gas Business, Scientific and Technical Journal, Founder: FGBOU VO Ufa State Petroleum Technical University, Vol. 3, No. 19, pp. 121-127, 2021.
[6] A.V. Velikanov, "Effect of Chemical Composition on the Properties of Heat-Treated Rail Steel", Rails of Increased Operational Durability: Works of VNIIZhT, Transport, Vol. 196, pp. 138-152, Moscow, Russia, 2018.
[7] V.V. Polyakov, "Fundamentals of Technology for the Production of Railway Rails", Metallurgy, p. 416, Moscow, Russia, 2017.
[8] Masaru Ueda, "Characteristics of Heat-Strengthened Rails and the Latest Developments of Nippon Steel", Engineering Solutions, No. 1, pp. 9-11, 2012.
[9] V.A. Rezanov, "Development of a Method for Flashing Contact Welding of Alloyed Rail Steels", FGBOU VPO MATI - Russian State Technological University Named After K.E. Tsiolkovsky, p. 20, Moscow, Russia, 2013.
[10] A.B. Yuryev, "Development and Implementation of the First Technology in Russia for the Production of Differentially Heat-Strengthened Rails Using the Heat of Rolling Heating", Steel, No. 11, pp. 33-35, 2016.
[11] A.V. Kuklev, L.K. Fedorov, V.V. Tinya-Kov, "Improving the Technology of Continuous Casting of Rail Steel", Steel, No. 10, pp. 54-56, 2016.
[12] R.A. Iskanderov, J.M. Tabatabaei, "Vibrations of Fluid-Filled Inhomogeneous Cylindrical Shells Strengthened with Lateral Ribs", International Journal on Technical and Physical Problems of Engineering (IJTPE), Issue 42, Vol. 12, No. 1, pp. 121-125, March 2020.
[13] V.F. Gahramanov, E.A. Aslanov, N.S. Rzayev, J.V. Asgarov, "Effect of Carbide Coatings on Corrosion of Iron-Based Composite Materials", International Journal on Technical and Physical Problems of Engineering (IJTPE), Issue 60, Vol. 16, No. 3, pp. 244-250, September 2024.
[14] R.A. Iskanderov, J.S. Mammadov, D.A. Mahmudov, "Substantiation of Initial Provisions of Construction of a Methodology for Modeling Corrosion Degradation", International Journal on Technical and Physical Problems of Engineering (IJTPE), Issue 60, Vol. 16, No. 3, pp. 274-280, September 2024.
[15] A.P. Dzyuba, R.A. Iskanderov, Y.M. Selivanov, S.A. Valiyeva, "Holographic Study of Non-Stationary States of Thin-Walled Structure Elements Under Thermal and Thermo Mechanical Loading", International Journal on Technical and Physical Problems of Engineering (IJTPE), Issue 59 Vol. 16, No. 2, pp. 342-348, June 2024.
[16] V.F. Gahramanov, E.A. Aslanov, "Improving Properties of Pipes for Oil and Gas Transportation by Heat Treatment", International Journal on Technical and Physical Problems of Engineering (IJTPE), Issue 59, Vol. 16, No. 2, pp. 162-166, June 2024.
[17] V.F. Gahramanov, A.H. Guliyev, N.S. Rzayev, Z.S. Musayev, "Effect of Increasing Electrical Resistance of Alloys of Various Compositions", International Journal on Technical and Physical Problems of Engineering (IJTPE), Issue 61, Vol. 16, No. 4, pp. 355-361, December 2024.
[18] I.A. Ismayil, "The Structural Analysis for The Connecting Rod of Piston Machine Using Ansys Software", International Scientific Journal of IFToMM Problems of Mechanics, No. 2, Vol. 83, pp. 19-24, Tbilisi, Georgia, 2021.
[19] V.I. Bakhshaliev, I.A. Ismayil, A. Bakirova, "Optimization the Wear Process and Clearances in Rotating Nanosurfaces of the Kinematic Couples of Reciprocating Machines", International Scientific Journal of IFToMM Problems of Mechanics, No. 1, Vol. 74, pp. 17-23, Tbilisi, Georgia, 2019.
[20] A.P. Dziuba, R.A. Iskenderov, V. Sirenko, L. Levitina, "Optimization of Two-Directional Variable Wall Thickness Shells of Rotation by Methods of the Theory of Optimal Processes", Proceedings of the 9th International Conference Control and Optimization with Industrial Applications, pp. 556-562, Istanbul, Turkey, August 2024.

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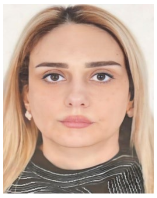

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