

## OPTIMIZATION OF MECHANICAL PROPERTIES IN ENGINEERING STEELS VIA HIGH-TEMPERATURE THERMOMECHANICAL TREATMENT

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**Abstract-** Influence of the deformation scheme. Up to now, there are practically no more or less systematic data on the use of stamping as a method for deforming austenite in the HTMT (high-temperature thermomechanical treatment) cycle. Apparently, this is due to the peculiarities of the stamping process itself, primarily with the strong non-uniformity of deformation in different sections of the forging, which determines comparatively less strengthening as a result of HTMT than in rolling or drawing. Meanwhile, the widespread use of this method of manufacturing parts in industry leads to the need to determine the possibilities of their strengthening in the case of using HTMT with stamping deformation. Such work was carried out on 50KhFA steel; the increased carbon content in this steel, as well as the presence of chromium and vanadium, which delay the decomposition of martensite during tempering and the recrystallization process, determine the feasibility of using this steel for HTMT strengthening. To determine the optimal conditions for HTMT with stamping deformation, hardening was studied under a wide range of process conditions. Various stamps and blanks of different shapes and sizes were tested. Forgings with diameters of 10 and 5 mm were stamped at 900 °C in a shaped stamp on a 1000 -t high-speed crank press, after which they were immediately (within 1.5 s) transferred to an oil quenching tank. In this case, one batch of samples was deformed with one blow, the other - with two. When deforming with two blows, the blank was first squeezed between flat dies, and then quickly tilted at 90 °C and the sample was stamped in a shaped stamp. After quenching, low tempering was carried out.

**Keywords:** Austenite, Stamping, Rolling, Drawing, Hardening, Chromium, Vanadium, Decay.

### 1. INTRODUCTION

The issue of improving the mechanical properties of structural steels using high-temperature thermomechanical

treatment has been relevant at all times. This is because it is possible to increase the hardness of steels with the same chemical composition by thermomechanical treatment, which will lead to the transformation of these steels into tool steels. Heavy-scale pneumatic (hydraulic) hammers carry out high-temperature thermomechanical processing in the factory; the geometric dimensions of the ends of the hammers are selected in accordance with the requirements of the fields of application of forged steels.

The reason why steels are subjected to high-temperature thermomechanical treatment is to increase hardness by removing and bringing together bonds between molecules by interfering with the molecular structure of steel and dynamic cooling. On the other hand, the small size of the tips when hammered with pneumatic hammers leads to an increase in hardness, which is explained by the fact that since the cross-sectional area of the tip is small, it receives a large deformation because of the same load (the thickness of the material being knocked out decreases).

The thicknesses of the sheets selected during the manufacturing process or experiment range from 5-6 mm to 200 mm. When the thickness is large, it is stamped at high temperature. Cold stamping of sheets is usually performed under the workshop temperature conditions. One of the difficulties of this process is mainly that the stamped product is forged. In this case, a lot of force is required. The cold sheet stamping process is usually performed on roller presses when stamping sheets of various designs, the force of which can create a force equal to 4.5-6 tons.

Stampa Zamani metal lichenin-stained pneumatic hammer marks (of various sizes and depths) quilt. Various methods are used to polish these plastic marks: milling or, depending on the depth, the melting jet method. To obtain comparative results under the same conditions (on forgings for making tensile specimens), conventional quenching with low tempering was carried out.

To determine the optimal conditions for the HTMT process, the effect of the forging diameter on hardening was initially investigated. In the first series of experiments, samples with a working part diameter of 10 mm were tested; the increase in strength was small, and a spread of properties was observed. It was found that a decrease in the forging diameter leads to better and more uniform results, so subsequent experiments were carried out on samples with a diameter of 5 mm.

As follows from the data in Table 1, after HTMT, the tensile strength increased by approximately 100 MPa, and the yield strength by 200 MPa with an insignificant decrease in relative narrowing to the permissible limits established for the high-strength condition.

Table 1. The effect of HTMT with deformation by stamping on the properties of 50KhFA steel

Processing mode	$\sigma_B$ , MPa	$\sigma_{0.2}$ , MPa	$\sigma_{0.02}$ , MPa	$\psi$ , %
Standard heat treatment: turned samples	2162	1777	1527	41.7
stamped samples	2145	1750	1462	38.6
Round blanks with a diameter of 7.3 mm ( $\epsilon=32\%$ )	2233	1943	1580	33.3
Rectangular blanks measuring 4.5x10 mm ( $\epsilon=87\%$ )	2233	1960	1590	27.2

Microstructural study of the size and shape of austenite grains confirmed that the deformation is very uneven across the cross-section. To obtain a greater strengthening effect as a result of thermomechanical treatment, it is necessary to have a more complete and uniform deformation across the forging cross-section. For this purpose, HTMT was performed during stamping with deformation in two blows with an interval of 5-6 s between them. 2-3 s after the second blow, the forging was immersed in oil.

The initial dimensions of round and rectangular blanks were selected with the expectation of obtaining different degrees of deformation (Table 2). Deformation was carried out in such a way that the metal completely filled the groove. Mechanical properties after different deformation options are given in Table 3. As shown by microstructural analysis, when stamping in two blows, deformation occurs more uniformly and undeformed zones are significantly reduced (compared to stamping in one blow). In areas 1 and 2, the grains do not differ significantly in shape from the grains in similar areas when stamping in one blow [1].

## 2. PROBLEM SOLVING

In area 3, the grains are deformed along the height of the sample (the effect of the first impact). There is a small area of hindered deformation with equiaxed undeformed grains on the surface of the sample. Such a picture with small deviations can be observed on all forgings deformed according to the variant specified in Table 2. It follows from the data provided that the effectiveness of hardening during HTMT is determined not only by the degree of compression, but also by uniformity. Higher properties are

found in those samples obtained from the narrowest blanks. This is obviously explained by the fact that with wide blanks, the effect of friction is stronger and the length of the hindered deformation zones increases.

It should be noted that during stamping it is very difficult to maintain a constant deformation mode, since it is affected by many process factors: the accuracy of the stamp installation, the uniformity of lubrication (graphite dissolved in oil was used to reduce friction), the temperature of the stamp, the accuracy of the blank installation in the center of the groove, the position of the blank during the second installation (during stamping in two strokes), etc. It was possible to avoid large deviations from the established parameters only after accumulating a certain amount of experience. Uneven deformation across the cross-section was also detected during the study of microhardness and physical broadening of the lines (110)  $\alpha$  and (211)  $\alpha$ .

It was shown that the main strengthening as a result of HTMT with stamping deformation occurs already at relatively small degrees of deformation (up to 20-30%); a further increase in the degree of deformation (up to 90%) has little effect. An important factor determining the degree of strengthening during HTMT with stamping is the size of the zones of difficult deformation.

Table 2. Deformation modes during HTMT of 50KhFA steel under stamping conditions

Processing index	Size or diameter of the original workpiece, mm	One hit		Two strikes $\epsilon$ (%)
		bxh, mm	$\epsilon$ (%)	
A	8	4.2x12	40	105
B	7.1	4.2x9.5	20	82
C	10x6	4.2x14	87	120
D	10x4.5	4.2x10	87	96
E	7x6	4.2x10	51	87
F	8x6	6x8	35	64
G	10.5x5.3	7x8	51	64
H	10.5x6	8x8	64	64
I	9.4x6	7x7	51	51
K	9.4x6	6x7	64	51
Final diameter of the sample $d=5.4$ mm				

Table 3. Mechanical properties of 50KhFA steel after VTMO under stamping conditions

Processing index	$\sigma_B$ , MPa	$\sigma_{B 0.2}$ , MPa	$\sigma_{0.02}$ , MPa	$\psi$ , %
A	2293	1958	1511	34.6
B	2313	1999	1597	33.7
C	2320	2030	1643	32.5
D	2296	2005	1630	30.2
E	2327	2003	1588	31.8
F	2250	1966	1635	35.9
G	2250	1991	1624	30.1
H	2205	1910	1632	36.0
I	2260	1948	1590	31.5
K	2240	1910	1602	28.2
Deformation modes see Table 2				

According to the nature of deformation, the cross-section of the forging can be divided into four areas (1, 2, 3, 4 in Figure 1b).

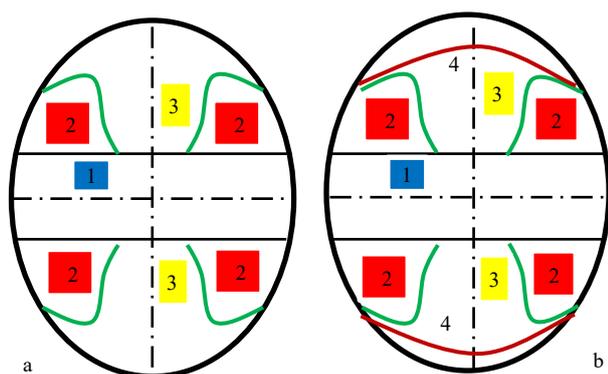


Figure 1. Distribution of deformation across the section of a 5 mm diameter tensile specimen forging after VTMO: 50KhFA steel: a- stamping in one blow; b- stamping in two blows

After the optimal mode of HTMT and tempering at 200 °C for 2 hours,  $\sigma_B=2300$  MPa (150-200 MPa more than after conventional treatment),  $\sigma_T=2000$  MPa (250-300 MPa more),  $\sigma_{0.02}=1650$  MPa,  $\delta=6\%$ ,  $\psi=30\%$ ,  $KCU=350$  kJ/m<sup>2</sup> were obtained. The fatigue limit after HTMT increased in the considered variant (by ~10%) compared to conventional heat treatment and amounted to 980 MPa. The resistance of steel in the region of limited fatigue increased significantly (5-8 times).

The strengthening effect obtained as a result of HTMT is stable up to relatively high tempering temperatures (up to 400 °C). In production conditions, high-temperature thermomechanical treatment with deformation by forging of a number of automobile parts and samples made of steels 45 and 40X was carried out. For steels of both grades at a high temperature of the onset of deformation of 1000 °C, an increase in the degree of deformation to 75% and a decrease in the duration after deformation holding increases the ultimate strength and yield strength and its ratio to ultimate strength, while the plasticity parameters remain virtually unchanged, and the impact toughness increases by 30-50% compared to conventional heat treatment). The maximum increase in strength was ~ 15% for steel 45 and about 20% for steel 40X. The increase in the mechanical properties of steels 45 and 40X after HTMT is greater than the homogeneous deformation during forging.

A study of the structure of thermomechanical Ly hardened steels 45 and 40X showed that after HTMT martensite is characterized by an increased dislocation density. In addition, the proportion of martensite packets with complex boundaries and smaller crystal sizes increases than in the case of conventional quenching. The features of carbide formation after tempering at 500 °C for steel 45 and at 550 °C for steel 40X were studied. It was found that the number of precipitated carbide particles after HTMT is much greater than after conventional heat treatment, and their size is smaller. The number of carbide particles increases from  $7.5 \times 10^6$  to  $10 \times 10^6$  mm<sup>-2</sup> for steel 45 and from  $6.2 \times 10^6$  to  $10 \times 10^6$  mm<sup>-2</sup> for steel 40X. The average diameter of carbides decreased from  $0.1 \times 10^{-3}$  to  $0.06 \times 10^{-3}$  mm for steel 45 and from  $0.09 \times 10^{-3}$  to  $0.06 \times 10^{-3}$  mm for steel 40X.

The effect of the deformation scheme - rolling, pressing, stamping - during HTMT was studied on widely used structural steels of the 40X, 40XN, 40XNM, 40XNMV, 40XNMVF types. To perform pressing with different degrees of deformation, blanks of different diameters were made and the press tool was changed; the diameter of the die eye was constant, equal to 15 mm. Pressing was carried out on a hydraulic press with a force of 4 MN without press residue using graphite lubricant and with container heating at a temperature of 250 °C. The pressing speed was about 60-80 mm/s. The time from the end of deformation to cooling in oil did not exceed 0.5-0.6 s. The heating temperature of the blanks was selected taking into account the heating of the metal during the pressing process, and it was 870 °C in the case of deformation by 25 and 45% and 830 °C for deformation by 70 and 100% (the degree of deformation  $\varepsilon$  was calculated using the logarithmic formula).

To obtain comparable results in studies conducted on HTMT with different deformation methods, the degrees of compression during rolling  $\varepsilon$  were as follows: 15; 25; 35 and 50%; the last two values correspond to  $\varepsilon = 45$  and 70% according to the logarithmic formula. Rolling deformation was performed in one pass at a speed of 0.23 m/s. The blanks in the form of plates were rolled along the height to a final thickness of 12 mm.

Taking into account the small heating during rolling, some of the samples ( $\varepsilon = 50\%$ ) were heated at 860 °C. The heating temperature for the remaining blanks was 880 °C. The break time between rolling and quenching in oil was about 1-1.5 s.

HTMT during stamping was carried out in a specially designed die with a heater. A blank of rectangular cross-section was fed into the die groove and deformed along the height in one blow with the degrees of compression  $\varepsilon=15$ ; 25; 45 and 70%. A flash-free cylindrical sample with a diameter of 14.5 mm and a length of about 110 mm was obtained. The study of the carbide formation process at low tempering was carried out on 40X steel after HTMT with different deformation methods and on 40XNM and 40XNMVF steels - with deformation by pressing. For comparison, these processes were also studied after conventional heat treatment.

After HTMT of 40X and 40XNM steels with deformation by pressing and tempering at a temperature of 150-250 °C, some grinding and a more equiaxed shape of carbides are detected compared to conventional quenching. With an increase in the tempering temperature to 350 °C, carbide clusters appear. Unlike conventional heat treatment, they are located not only along the boundaries of martensite crystals, but also inside them; this nature of the precipitation of the carbide phase is associated with increased fragmentation of martensite formed from hot-deformed austenite. An even greater decrease in the size of carbide particles is observed upon tempering at 150 °C of 40KhNMVF steel subjected to BTMO with deformation by pressing.

A further increase in the tempering temperature leads not only to the coarsening of plate-shaped carbides, but also to the precipitation of new small equiaxed carbide

particles, most likely carbides enriched in tungsten and vanadium. The plate-shaped carbides precipitated at a tempering temperature of 400 °C have a certain direction; It is generally accepted that this direction aligns with the dominant axis of plastic deformation and is governed by the underlying crystallographic texture, particularly the orientation of slip systems and lattice anisotropy. The results of quantitative analysis of the data obtained show that BTMO, regardless of the deformation method, slows down the precipitation of the carbide phase during martensite tempering. Thus, in steel 40 X, compared to conventional heat treatment during tempering at 150 °C, the total area occupied by carbide after HTMT is two times smaller. With an increase in the tempering temperature to 350 °C, the same situation continues to be maintained, and a more significant decrease in the amount of the precipitated carbide phase is observed in the case of deformation by rolling (by 1.5 times) and to a lesser extent - in the case of deformation by pressing (1-1.2 times).

The slowdown in the precipitation of the carbide phase during low tempering of thermomechanical Ly hardened steels can be explained by a decrease in the total cross-linking of the solid solution, caused by the interaction of carbon with dislocation structures (polygonal sub boundaries) created during hot deformation in the HTMT process. The increased density of structural defects in martensite formed from hot-deformed austenite determines a greater number of nucleation sites than in the case of quenching and, as a result, a greater dispersion of the carbide forming elements during tempering.

As a result, after HTMT using pressing of 40 X and 40 XHMB steels at deformation degrees  $\varepsilon=70$  and 25%, an axial texture of martensite with the  $\langle 110 \rangle$  direction, coinciding with the pressing direction, is formed.

On the pole shapes of the central part of the rods subjected to HTMT rolling ( $\varepsilon=50\%$ ), a significant pole density in the transverse direction was found. This is the result of a more intense flow of metal of the inner zone in the transverse direction during rolling of rods of rectangular cross-section. The most intense orientation is [111]. In addition, weaker orientations are present in the martensite texture (the latter only in 40XHMB $\Phi$  steel). When HTMT is carried out under stamping conditions, the nature of the metal flow in the transverse direction is very close to the deformation by rolling. Thus, in steels subjected to HTMT with pressing deformation, an axial texture  $\langle 110 \rangle$  was found. Of the many wind-induced orientations in martensite after HTMT under rolling and stamping conditions, the most common and intense are [2] and [3]. The change in texture intensity occurs mainly with a decrease in the degree of deformation and depends little on the composition of the steel. The maximum strengthening effect for 40 KhNM, 40 KhNMV and 40 KhNMVF steels during HTMT with pressing deformation is achieved with an increase in the degree of compression to 70%. The tensile strength of 40 KhNM steel samples increases from 2150 MPa (without deformation) to 2350 MPa, and the yield strength - from 1500 MPa to ~1700 MPa. For 40 KhNMV and 40KhNMVF steels, these characteristics increase even more:  $\sigma_B=2380\div 2450$  MPa,

and  $\sigma_{0.2}=1760\div 1830$  MPa. Increasing the degree of deformation to 100% (when recrystallization begins) leads to some decrease in the strength characteristics of these steels, more significant for the yield strength. A distinctive feature of hardening during HTMT under pressing conditions is obtaining high ductility for a high-strength state (after low-temperature tempering). For 40 KhNM and 40 KhNMV steels, increasing the degree of deformation over 45% significantly increases ductility:  $\varphi=41\div 37\%$  and  $\delta=11\%$ . After conventional heat treatment, 40XHM steel has  $\varphi=29\%$  and  $\delta=8.5\%$ ; steel 40XHMB shows a tendency to brittle fracture. Deformation by pressing with a degree of compression of 25% or more also eliminates brittle fracture of steel 40 X after tempering at 150 °C and increases ductility to  $\varphi=32\%$ ,  $\delta=10\%$ .

Deformation by pressing during HTMT has a particularly strong effect on the resistance of steels to fracture. Analysis of the change in the criterion  $K_{te}$  with an increase in the degree of deformation shows greater resistance to fracture in the presence of stress concentrators. Maximum values are obtained at  $\varepsilon=70\div 100\%$ ; the increase in the value of the criterion  $K_r$  in this case is 40-50%. On steels 40X, 40 XH and 40 XHM, an increase in the value of  $K_{ro}$  is observed even under the conditions of the initial stage of recrystallization (when the yield strength drops). The results of tensile tests of samples strengthened by HTMT with rolling deformation show a higher increase in the ultimate strength and yield strength than after HTMT under pressing conditions. The curves of the change in ultimate strength and yield strength of steels 40 X and 40 XH have an extreme character with a maximum at a degree of reduction  $\varepsilon=35\%$  (for steel 40 XH closer to  $\varepsilon=25\%$ , and for steels 40 XHM, 40XHMB and 40 XHMBF - at 50%). Under HTMT conditions with rolling deformation at comparable degrees of compression, recrystallization processes are less intense than during pressing.

The maximum increase in the ductility of the studied steels with a martensite tempering structure after HTMT under rolling conditions is observed at a degree of compression of about 25%. At the same time, the ductility of steels subjected to deformation by pressing during HTMT is higher compared to deformation by rolling. This result is not unexpected if we take into account the nature of metal flow under three-dimensional compression during pressing (a simpler texture in the case of pressing and a more complex one in the case of rolling).

The results of mechanical tensile tests of samples made of 40X, 40XH, 40XHM, 40XHMB and 40 XHMB F steels after HTMT with stamping deformation also show an extreme nature of the change in ultimate strength and yield strength depending on the degree of deformation. The maximum increase in these properties compared to conventional heat treatment is observed at  $\varepsilon=45\%$ . As a result of HTMT under stamping conditions with subsequent low tempering,  $K_I$  increases by approximately 30-40%, which is lower compared to HTMT under pressing conditions.

The results of determining the impact toughness of steel 40X after HTMT with deformation with pressing and

tempering at 150 °C show its maximum increase compared to conventional heat treatment at a compression ratio of 45%;  $KCU=250$  kJ /m<sup>2</sup>. With an increase in the degree of deformation, the KCU values decrease due to structural changes occurring during the development of recrystallization of the treatment. For other steels, high impact toughness values are retained even at a deformation ratio of more than 45%. The maximum increase in this characteristic is observed at  $\epsilon=70\%$ . Apparently, the preservation of grain elongation with a simultaneous increase in plasticity as a result of pressing straight HTMO has a favorable effect on increasing the resistance to impact fracture of steel in the transverse direction.

The impact toughness after HTMO under rolling and stamping conditions increases, and the maximum increase in KCU is achieved at compression ratios of 15-25%. This is the result of high anisotropy of impact toughness depending on the direction of metal flow. The maximum values of ultimate strength and yield strength in torsion tests are achieved on samples deformed during HTMO rolling. It should be noted that for almost all steels, the increase in yield strength is higher than the increase in ultimate strength.

### 3. RECOMMENDATION TO ENGINEERING PRACTICE

When torsion of samples textured as a result of rolling under HTMT, the planes along which the shear occurs remain in the zone of maximum shear stress until failure, i.e. they coincide with the main directions during rolling. Therefore, hardening during torsion testing reaches a greater value than during tension, where the zone of action of the greatest shear stresses corresponds to an angle of 45° to the rolling direction. Based on these data, it is possible to explain the increase in the yield strength of samples deformed by stamping under HTMT. In the case of deformation by pressing, an axial texture is formed and shear during torsion is carried out more easily than with more complex textures. Therefore, no significant change in strength is observed compared to conventional hardening during torsion testing. Thus, HTMT with the considered deformation methods leads to different steel strengthening efficiencies:

- a) under conditions of limited recrystallization during pressing, it is possible to obtain a significant increase in the ultimate strength (~200 MPa) and yield strength (180-200 MPa) with high ductility after tempering at a temperature of 150 °C ( $\delta \approx 11\%$ ;  $\psi \approx 40\%$ ). In this case, an increase in impact toughness and crack propagation resistance is also observed;
- b) the maximum increase in steel strength is observed under rolling conditions: 200-400 MPa for  $\sigma_p$  and 250-300 MPa for  $\sigma_{0.2}$ ; ductility and impact toughness are lower compared to the results during pressing ( $\delta \approx 7\%$ ;  $\psi \approx 30\%$ );
- c) deformation by stamping under the selected conditions leads to lower strengthening efficiency as a result of non-uniformity of the structure and insufficient ductility; in accordance with this, the resistance to crack propagation also decreases.

HTMT under hot rolling conditions were carried out on pipes used in the oil industry made of steels D (0.44% C; 1.10% Mn; 0.32% Si), 36T<sub>2</sub> C (0.38% C; 1.65% Mn; 0.58% Si; 0.48% Mo) (0.37% C; 0.83% Mn; Mo). Hot rolling; was carried out on a pipe rolling plant at the Nikopol South Pipe Plant. The rolling conditions were adopted in Table 4.

Table 4. HTMT of pipe steels

	D	36G2S	38HNM
Workpiece diameter, mm	130	90	130
Pipe size, mm	114×9	76×6	133×9
Compression degree, mm	77.6	78	73.5
Temperature, °C			
Start of rolling	1250	1200	1200
End of rolling	800-840	800-840	760-800
The degree of compression was calculated using the formula $\epsilon = 1 - \mu^{-1}$ 100, where $\mu$ is the overall drawing coefficient.			

The rolling duration was 100-120 s, and the time it took for the pipe to pass from the rolling mill to the sizing mill was 15-36 s. Since the temperature of the metal after the rolling mill was 900-950 °C, part of the recrystallization processes occurred. Therefore, the degree of compression, judging by the hardening, was less than calculated. Immediately after rolling, the end of the pipe was cut off and quenched in water (processing I). The rest of the hot-rolled pipe was cooled on a table (roller table), and then the pipes cut from it were subjected to comparison of mechanical properties by processing according to the following modes:

- II - heating at 840-850 °C, 30 min + quenching + tempering;
  - Sha - annealing at 880 °C + heating to 840-850 °C, 30 min + quenching + tempering;
  - 1116 - normalizing with 880 °C + heating to 840-850 °C, 30 min + 3 quenching + tempering;
  - IV - normalizing with 1100 °C + heating to 840-850 °C, 30 min + quenching + tempering.
- Tempering was carried out in the temperature range from 100 to 600 °C every 100 °C, holding at the tempering temperature was 1%.

The treatment variant using the hereditary influence of HTMT was carried out according to the following regime: unmilled quenching after rolling, intermediate tempering at 400 or 500 °C, rapid heating in a salt bath (3 min at 850 °C) - quenching - tempering. The BTMO variant was also studied using compressed air under high pressure as a cooling medium after rolling. After such HTMT, tempering was carried out only at 400-600 °C.

The mechanical properties of the pipes obtained after HTMT were compared with the properties of pipes cooled on a roller table and subjected to heat treatment, which provides for intermediate annealing for 1 h at 880 °C to remove the effect of hot deformation, reheating for quenching at 840-850 °C, 30 min, quenching and tempering under similar conditions. The results of mechanical tests are given in Tables 3-5. At all tempering temperatures, higher properties are obtained after HTMT than after conventional heat treatment.

The maximum difference in the values of ultimate strength was observed after tempering at 100 °C. With an increase in tempering temperature to 400 °C, the difference between the ultimate strength after HTMT and conventional quenching decreases, but it continues to be preserved even after tempering at 500-600 °C. This indicates high stability of the hardened state of the pipes created by HTMT. The decrease in the difference in the values of ultimate strength after HTMT and conventional quenching with an increase in the tempering temperature from 100 to 200 °C is explained by the reduced value of this characteristic in steels D and 36G2S usually quenched and tempered at 100 °C due to their premature brittle fracture. When the fracture was relatively more ductile after conventional quenching and tempering at 100 °C (steel 38 XHM), the difference in the values of ultimate strength after HTMT and after conventional quenching and the same tempering was smaller and decreased insignificantly with an increase in the tempering temperature. The difference in the yield strength and proportionality limits of D and 36G2S steels after

increases ductility, but significantly reduces the yield strength. In the case of pipes made of 38 XHM steel, which differs from the previous ones in its increased content of alloying elements and a lower end rolling temperature (760 °C), the recrystallization of the treatment during HTMT was less. The yield strength for pipes made of 38XHM steel after HTMT was 100-250 MPa higher than after conventional quenching, and the proportionality limit was 100-350 MPa higher.

A strong strengthening effect of HTMT is manifested after tempering at low and medium temperatures. Maximum plasticity indices in the tempering temperature range of 200-400 °C are achieved after HTMT with reductions of 15 and 50% (Figure 2). However, in all cases, HTMT of 55 XPP steel provides an increase in plasticity compared to conventional quenching and tempering under comparable technological conditions.

The favorable effect of hot deformation in the HTMT cycle is very stable. Heat treatment of roller table-cooled pipes according to TO<sub>1</sub> scheme (heating to 840-850 °C, 30 min, water quenching, tempering at 100-600 °C) did result in lower mechanical properties than after HTMT, but they were still significantly higher than after conventional heat treatment in similar modes. As a result of intermediate annealing at 880 °C (TO<sub>3</sub>) or intermediate normalization (TO<sub>2</sub>) to remove the effect of preliminary hot deformation, the properties decreased compared to those obtained after HT (Tables 1-3), but still remained at a sufficiently high level. After heat treatment according to TO<sub>1</sub> and even TO<sub>2</sub> and TO<sub>3</sub> options, the strength characteristics are significantly higher than those indicated in the reference literature for similar steels after conventional heat treatment; At the same time, an increase in plasticity properties is also observed. It is appropriate to recall the well-known fact that with the same heat treatment, the properties of samples cut from pipes are always higher than those of samples made of the same steel, but cut from rods. This is possible only in the case of a stable influence of the previous method of complex stress loading during hot deformation on the structure of the metal of the pipes.

The fracture of the specimens of pipes of steels D and 36G2S after BTMO, as well as after TO, with tempering at 100 °C was mainly ductile, whereas after TO<sub>2</sub> and TO<sub>3</sub>, as well as in the case of specimens 113 of bars of steels 35 GS and 35 XI 2 (similar to steel 36G2S) it was completely brittle after tempering at 100 °C. Specimens of steel D after TO<sub>4</sub>, as well as specimens of similar steels, were brittle even after tempering at 200 °C.

This indicates that the effect of hot deformation during pipe production on the properties of steel will be preserved to a certain extent after reheating for quenching. Even after TO<sub>4</sub> treatment and subsequent quenching (again completely refining the grain) with tempering, the mechanical properties are not completely comparable with the properties of similar steels given in reference books. Comparing the mechanical properties after different processing options, it can be seen that after TO<sub>4</sub> varnaite, they are lower than after TO<sub>2</sub> with intermediate heating at 880 °C (although the effect of high heating TO<sub>4</sub> is removed by subsequent recrystallization). In the latter case, the

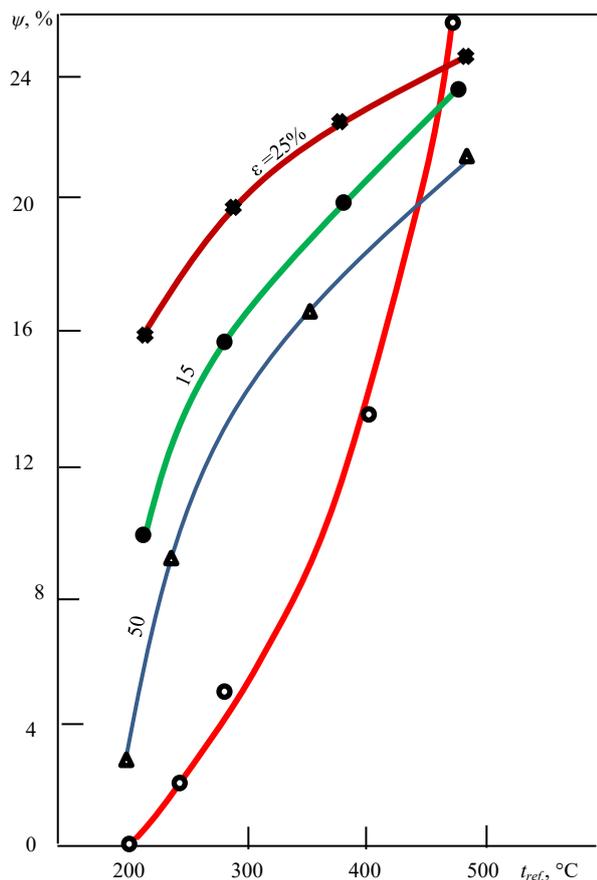


Figure 2. Change in transverse contraction of 55XTP steel depending on the tempering temperature after conventional quenching (0%) and HTMT with reductions of 15, 25, 50%

In the case of pipes made of 38 XHM steel, the annealing holding time was 3 hours, HTMT and conventional heat treatment under these experimental conditions was low. This is explained by the partial recrystallization of the treatment during HTMT, which

properties are lower than after TO<sub>1</sub>. After processing according to the TO<sub>2</sub> or TO<sub>3</sub> mode, the strength characteristics are approximately the same, whereas after intermediate normalization, the plasticity characteristics are higher.

Although after TO<sub>3</sub> the grain was refined and therefore the steel properties should improve somewhat, they still turned out to be lower than after TO<sub>1</sub>, when there was no intermediate normalization. This indicates that in the case of TO, the positive effect of previous hot deformation was not completely eliminated. The preservation of the hot deformation effect after slow cooling of rolled pipes is confirmed by the data obtained by B.P. Kolesnikov also on pipes made of 36G2S steel. After heating for quenching at rates up to 8 deg/s, but without holding, i.e. under conditions when the effect of preliminary hot deformation could be preserved, the yield strength was 200-250 MPa higher than after conventional heat treatment, when the holding time at the austenitizing temperature was 30 min. This increase in the yield strength can be explained by the effect of increased stability of the hardened state of steel after hot deformation, which is typical for pipes, when each elementary volume is subjected to a complex shape change: compression, tension and torsion. It should be noted that although the level of mechanical properties after TO treatment is higher than after TO<sub>2</sub>, TO<sub>3</sub> and TO<sub>4</sub>, the size of austenite grains in the first case was 3-3.8 (steel D) or 2.2-3 times larger (steel 36G2S).

Consequently, there is no direct relationship between the level of mechanical properties of steel and the size of austenite grains in the thermomechanically strengthened state. The mechanical properties of steel in the strengthened state are more significantly influenced by the degree of development of subboundaries, the intragranular fragmented structure, than the grain boundaries themselves, the nature of which, naturally, cannot be established by the usual method of metallographic analysis. At the same time, a significant effect on the properties is exerted by an increased dislocation density, as well as concentration heterogeneity in the places of former grain boundaries, which is manifested in a double network of boundaries (detected by S.Z. Bokshsteyn by the autoradiography method) after recrystallization or recrystallization, i.e., the boundaries of grains that existed before recrystallization and newly formed grains. This means that the number of boundaries serving as barriers and resisting plastic deformation and destruction, as it were, doubles.

Steel D was also subjected to impact testing at -196 °C after HTMT and treatment according to variants TO<sub>1</sub> and TO<sub>3</sub> and tempering at 550 °C, 2 h with slow cooling with a furnace, i.e. under conditions of development of reversible temper brittleness. After HTMT, fracture occurred along the grain body. During treatment TO<sub>3</sub>, the crack developed mainly along the grain boundaries, and during treatment TO<sub>1</sub>, fracture clearly occurred along the grain body, which is determined by the preservation of the effect of hot deformation after repeated heating for quenching. HTMT of pipes with cooling with compressed air.

The study of mechanical properties obtained after various treatment options showed that in the case of pipe rolling, the strengthening effect of hot deformation is very stable. When cooling rolled pipes on a roller table, recrystallization of the treatment does not occur completely, and collective recrystallization does not even begin. The fragmented structure obtained as a result of high-temperature plastic deformation is preserved to a significant extent even after reheating the pipes. Therefore, a type of processing that provides for accelerated cooling of the pipes immediately after rolling with compressed air is effective. Due to the preservation of the strengthening effect of the preceding hot deformation, higher properties are achieved than after conventional normalization.

The structure formed during accelerated cooling of pipes made of steel D (hardening troostite) corresponds to that obtained as a result of the decomposition of austenite at 500-550 °C; the structure of pipes made of steels 36G2S and 38 XHM after accelerated cooling is a mixture of troostite and bainite formed at 400-500 °C. The data from mechanical tests carried out under factory conditions are given in Table. 3-5 (the average values obtained as a result of testing 12 pipes for each processing option), and the mechanical properties that oil pipes must have according to the standard depending on the category.

As Tables 5 and 6 from the data provided, after HTMO with compressed air cooling and tempering at 400-500 °C, pipes made of steel D meet the requirements for category L pipes. Pipes made of steel 36G2S after tempering at 600 °C also meet these requirements for strength properties, and their ductility and impact toughness properties even higher the mechanical properties obtained on pipes made of 38 XHM steel after the same treatment with tempering at 600 °C meet the requirements for pipes of category M.

Table 5. Mechanical properties of stainless-steel pipes after HTMO with compressed air cooling and after normal normalization

$T_{sub}, °C$	$\sigma_B, MPa$	$\sigma_{0.2}, MPa$	$\delta, \%$	$\psi, \%$	$KCU, kJ/m^2$	HRC
VMTO with compressed air cooling						
400	930	690	18	53	750	25
500	930	690	17	51	720	23
600	865	640	20	55	740	19
Normalization						
400	800	480	22	59.5	780	16
500	815	495	21	57	780	16
600	800	470	22	60	810	16
No vacation	815	490	20	58	820	16

Table 6. Mechanical properties of stainless-steel pipes 36G2S after VTMO with compressed air cooling and after normal normalization

$T_{sub}, °C$	$\sigma_B, MPa$	$\sigma_{0.2}, MPa$	$\delta, \%$	$\psi, \%$	$KCU, kJ/m^2$	HRC
VMTO with compressed air cooling						
400	1430	1045	9	24	200	37
500	1100	840	13	40	640	20
600	855	680	20	57	980	23
Normalization						
400	860	525	20	58	970	19
500	850	525	22	58	940	19
No vacation	86	535	22	56	920	19

Thus, the use of the BTMO variety with compressed air cooling is very promising, since it allows obtaining improved mechanical properties, which makes it possible to replace alloyed M rolling of carbon steels (D) or low-alloy steels (36G2S) does not require significant re-equipment in workshops and does not reduce the rolling rate (Table 7).

Table 7. Mechanical properties of stainless-steel pipes 36G2S after veto with compressed air cooling

$T_{sub}$ , °C	$\sigma_B$ , MPa	$\sigma_{0.2}$ , MPa	$\delta$ , %	$\psi$ , %	KCU, kJ/m <sup>2</sup>	HRC
400	1285	980	11.6	44.5	630	36
500	1125	865	12.4	55	970	29.5
100	1000	820	13.6	58	1040	28.5

#### 4. CONCLUSIONS

This study set out to demonstrate how high-temperature and low-temperature thermomechanical treatment (HTMT and LTMT) can be systematically tailored -through rolling, pressing, and stamping-based deformation schemes- to enhance the mechanical performance of widely used engineering steels. The following consolidated findings emerge:

1. Superior Strength-Ductility Synergy
  - Under optimized HTMT schedules, ultimate tensile strength rose by 150-400 MPa and yield strength by 180-300 MPa over conventional quenching-and-tempering, while elongation and reduction in area remained within-or even exceeded-industrial acceptance limits.
  - Pressing with 70 % total reduction produced the best all-round balance ( $\sigma_{sub>B</sub>} \approx 2350$  MPa,  $\sigma_{sub>0.2</sub>} \approx 1830$  MPa,  $\delta \approx 11$  %,  $\psi \approx 40$  %).
2. Process-Structure-Property Links
  - Hot deformation in the austenitic range generated a highly fragmented martensitic sub-structure and elevated dislocation density; both delayed carbide precipitation during tempering, slowed recovery/recrystallisation, and proved decisive for strength retention up to 400 °C.
  - Axial <110> textures after pressing, and complex transverse textures after rolling/stamping, explained the direction-selective improvements observed in torsion and impact testing.
3. Fatigue and Fracture Resistance
  - The fatigue limit of 50KhFA steel increased by  $\approx 10$  %, and limited-fatigue life by a factor of 5-8, relative to conventional heat treatment.
  - Crack-propagation resistance ( $K_{sub>IC</sub>}$  and  $K_{sub>Tp</sub>}$ ) improved by 30-50 % when deformation schemes suppressed large, equiaxed austenite grains and promoted fine packet martensite.
4. Industrial Applicability
  - For oil-and-gas pipes (steels D, 36G2S, 38KhNM) rolled to high reductions then quenched or air-blasted, HTMT delivered L- and M-category properties *without* alloy upgrades or major mill retrofits, cutting material cost and furnace time.
  - Automobile forgings and pressings fabricated under two-blow stamping achieved  $\sigma_{sub>B</sub>} \approx 2.30$  GPa with reliable impact toughness ( $> 350$  kJ m<sup>-2</sup>), meeting modern lightweight-safety targets.

The process selection guideline is defined in Table 8.

Table 8. Recommended deformation modes and parameters for targeted mechanical properties

Desired outcome	Recommended deformation mode	Key parameters	Typical tempering window
Maximum strength in thin sections	Rolling ( $\epsilon \approx 35-50$ %)	880 °C re-austerities; $\leq 1$ s quench lag	150-300 °C
Best strength-ductility trade-off	Pressing ( $\epsilon \approx 45-70$ %)	830-870 °C entry; container $\approx 250$ °C	150-350 °C
Cost-effective pipe production	Air-blast cool from final pass	$\epsilon > 75$ %, start 1200 °C, finish $\approx 800$ °C	400-600 °C
Shape-critical forgings	Two-blow stamping ( $\epsilon_{sub>1</sub>} \approx 20$ %, $\epsilon_{sub>2</sub>} \approx 40$ %)	Interval $\leq 6$ °C; oil quench $\leq 3$ °C	200 °C for 2 h

#### 5. LIMITATIONS AND FUTURE WORK

- Uniformity challenge: Stamping still suffers from localized under-deformation; industrial dies and lubricant delivery should be refined to shrink “hindered” zones.
- High-alloy steels: The present database covers C-Mn and Cr-Mo-V grades; Ni-Cr-Mo and advanced bainitic steels warrant investigation under identical HTMT routes.
- Thermo-mechanical modelling: Coupled FE-based microstructure models could shorten optimization loops, linking strain-rate fields to final packet size and carbide dispersion.

Thermomechanical treatment -when engineered with a clear understanding of deformation uniformity, quench delay, and tempering response- provides a practical, scalable pathway to properties formerly attainable only in expensive tool steels or via complex surface alloying. By embedding these process maps into plant practice, manufacturers can unlock lighter, tougher, and more fatigue-resistant components for transport, energy, and infrastructure applications-with minimal capital expense and demonstrable sustainability benefits.

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