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FOREIGN TECHNOLOGY DIVISION

FORGINGS FROM SPECIAL STEELS (SELECTED CHAPTERS)

by

I. G. Generson

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FORGINGS FROM SPECIAL STEELS (SELECTED CHAPTERS)

By: I. G. Generson

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This book considers problems involved in the production of special-steel forgings used in power-plant construction and other branches of industry, describes the technological procedures for forging low-deformability high-alloy steels, presents and analyzes the results of research and industrial experience, presents data characterizing forging quality in relation to fabrication conditions, and makes practical recommendations regarding the selection of optimum technological processes.

This book is intended for engineering and technical workers at appropriate plants, as well as at scientific research and technological planning institutes.

FOREWORD

Of the diversity of special-steel forgings used in the fabrication of various machine and equipment components, this book considers principally those produced from high-hot-strength and corrosion-resistant steels, which are coming into wider and wider use in modern technology in connection with the development of powerplants and chemical machinery.

In many cases, the specific properties and technological characteristics of these steels necessitate special smelting, forging, and heat-treatment procedures, which can in large measure be extended to certain typical forgings produced from other groups of special steels. The information in this book can therefore be applied to a broad range of forgings from alloy and high-alloy steels employed in various branches of industry.

Special attention is paid to analysis of metallurgical factors adn the characteristics of hot machining and heat treatment, which affect the quality, mechanical and special properties, and operational reliability of forgings.

This book is based on data obtained in numerous investigations conducted at the Neva Machine-Building Plant imeni V.I. Lenin [NPL] (НЗЛ) when production of large special-steel forgings was begun and on many years of industrial experience. The experience of other plants is also taken into account.

The industrial-research results, the critical evaluation of production observations, and the systemmatization of individual data on current production revealed a number of technological characteristics and patterns that can be successfully utilized in the search for optimum technological solutions and in bringing new types of forgings from alloy and high-alloy steels into production.

The success of much of the experimental work that forms the basis for this book was furthered by the participation of metallurgists at the NPL, including Engineers V.N. Tokarev, P.M. Libman, Ye.V. Babayeva, N.I. Belan, and others, to whom the author wishes to express his gratitude.

Chapter 1

CHARACTERISTICS AND PROPERTIES OF CERTAIN HIGH-HOT-STRENGTH AND CORROSION-RESISTANT STEELS EMPLOYED FOR LARGE FORGINGS

High-hot-strength and corrosion-resistant steels differ from ordinary steels in their special properties, which are produced by a complex of physicochemical and technological factors. Such steels usually have a high content of alloying elements (chromium, nickel, molybdenum, tungsten, vanadium, etc.), although their chemical composition is not the sole factor responsible for their special properties. Technological factors, including melting and casting methods, thermomechanical deformation conditions, and heat-treatment regimes, also have a strong influence on the properties of steel. The best results with respect to special hot strength and corrosion resistance are yielded by a favorable combination of chemical composition and optimum technological conditions.

The most important requirements imposed on the properties of high-hot-strength steels are a guarantee against component failure under given operating conditions and of minimum deformation over the entire service period (within the limits established for a given machine design). These requirements are evaluated from the main criteria of hot strength: yield strength and long-term ultimate strength. Other important indices of steel durability are also taken into account, particularly long-term plasticity and notch sensitivity. The plasticity margin governs the working deformation during component service. This criterion is therefore one of the main factors evaluated in selecting a metal. The tendency of a steel toward premature failure under the action of notches (stress concentrators) is directly related to its deformability under creep. Research has shown that most turbine-component failures are due to the presence of stress concentrators whose action has been intensified by low steel plasticity [1].

The physical constants of a steel (coefficient of linear expansion α and coefficient of thermal conductivity λ) are also of great importance from the standpoint of high-temperature structural applications and technological applications.

The temperature stresses in components are directly related to the coefficient of linear expansion, increasing with the value of α. Designers sometimes give preference to a metal with a lower coefficient of linear expansion, even when it has a somewhat lower hot strength. Austenite steels have the highest coefficients of linear expansion.

TABLE 1

Chemical Composition of Certain High-Hot-Strength and Corrosion-Resistant Steels, %

	Марка стали [1]	C	Si	Mn	Cr	Ni	Ti	Mo	W	S	P	Прочие элементы [3]
										не более [2]		
	1. Стали перлитного класса [4]											
5	ЭИ415	0,16—0,24	≤0,4	0,25—0,60	2,40—3,30	—	—	0,35—0,55	0,30—0,50	0,030	0,035	0,60—0,85 V
6	Р2	0,22—0,32	0,30—0,50	≤0,60	1,5—1,7	—	—	0,60—0,80	—	0,025	0,030	0,20—0,30 V
7	15Х1М1Ф	0,10—0,17	0,15—0,35	0,40—0,70	1,1—1,4	—	—	0,90—1,10	—	0,030	0,030	0,20—0,35 V
	2. Высокохромистые стали мартенситного класса [8]											
9	2Х13	0,16—0,24	≤0,6	≤0,6	12—14	≤0,6	—	—	—	0,030	0,035	—
10	1Х17Н2(ЭИ268)	0,11—0,17	≤0,8	≤0,8	16—18	1,5—2,5	—	—	—	0,030	0,035	—
11	1Х12ВНМФ	0,12—0,18	≤0,4	0,5—0,9	11—13	0,4—0,8	—	0,5—0,7	0,7—1,1	0,030	0,030	0,15—0,30 V
12	15Х11МФБ	0,12—0,18	0,20—0,55	0,6—1,0	10,0—12,0	0,5—0,9	—	0,80—1,05	—	0,025	0,030	0,20—0,30 V; 0,20—0,30 Nb
	3. Стали аустенитного класса [13]											
14	Х18Н9Т	≤0,12	≤0,8	≤2,0	17—20	8—11	До [24] 0,8 *	—	—	0,030	0,035	—
15	16-13-3 с Ti	0,08—0,12	≤0,8	≤2,0	15—17	12—15	0,3—0,6 **	2,5—3,25	—	0,030	0,030	—
16	Х18Н12М2Т	≤0,12	≤0,8	≤2,0	16—19	11—14	0,3—0,6 **	2—3	—	0,030	0,035	—
17	Х18Н22В2Т2	≤0,10	≤0,7	1,0—1,6	17—20	21—24	1,15—1,65	—	1,3—2,2	0,020	0,035	—
18	ЭИ405	≤0,12	0,8—1,0	≤0,5	16—17	12,5—14,5	—	1,5—2,0	—	0,020	0,030	0,8—1,2 Nb
19	ЭИ395	≤0,12	0,5—1,0	1,0—2,0	15,0—17,5	24—27	—	5,5—7,0	—	0,020	0,030	0,1—0,2 N
20	ЭИ572	0,26—0,35	0,3—0,8	0,75—1,5	18—20	8—10	0,2—0,5	1,0—1,5	1,0—1,5	0,030	0,030	0,2—0,5 Nb
21	ЭИ726	0,08—0,12	≤0,6	1,0—2,0	13—15	18—20	—	—	2,0—2,75	0,020	0,020	0,9—1,3 Nb; ≤0,025 B ***; ≤0,020 Ce ***
22	ЭИ612	≤0,12	≤0,5	1,0—2,0	14—16	34—38	1,1—1,5	—	2,8—3,5	0,020	0,020	—
23	ЭИ612К	≤0,12	≤0,5	1,0—2,0	14—16	34—38	1,2—1,6	—	2,8—3,5	0,020	0,020	3,5—4,5 Co; 0,01 B ***

*Minimum titanium content equal to C%·5.

**Minimum titanium content equal to (C% −0.03)·5, but no less than 0.3%.

***Calculated content.

1) Steel designation; 2) no more than; 3) other elements; 4) steels of perlite class; 5) EI415; 6) R2; 7) 15Kh1M1F; 8) high-chromium steels of martensite class; 9) 2Kh13; 10) 1Kh17N2(EI268); 11) 1Kh12VNMF; 12) 15Kh11MFB; 13) steels of austenite class; 14) Kh18N9T; 15) 16-13-3 containing Ti; 16) Kh18N12M2T; 17) Kh18N22V2T2; 18) EI405; 19) EI395; 20) EI572; 21) EI726; 22) EI612; 23) EI612K; 24) up to.

The coefficient of thermal conductivity affects principally the temperature field of a component and hence the strength and operating behavior of the assembly of which it is a part. A decrease in this coefficient causes an increase in the temperature gradient and thermal stresses. This important factor is always taken into account in evaluating whether or not a steel of a given structural class should be selected for a component intended to operate at high temperatures.

The technological significance of a low coefficient of thermal conductivity lies in the fact that metals with such conductivities require prolonged preforging heating and heat treatment in order to avoid large thermal stresses and development of cracks. Low thermal conductivity also causes stress concentration during local heating, which leads to warping of the component. The thermal conductivity of steel is strongly affected by the extent to which it is alloyed: carbon steels have the highest thermal conductivity, while high-alloy austenite steels have the lowest. The difference in the thermal conductivities of steels of different classes is greatest at normal temperatures.

Powerplants (steam and gas turbines), whose critical components operate under the action of high temperatures, substantial stresses, and corrosive media, give rise to a number of problems relating to the search for and selection of steels with the requisite combination of hot strength and appropriate physical properties, as well as satisfactory resistance to chemical agents at elevated temperatures.

Complex problems also arise in selecting chemically stable steels intended to operate in aggressive media at ordinary temperatures, e.g., in compressors intended to compress and deliver various gases. Despite the absence of creep, i.e., irreversible residual deformation, such applications often require high-alloy special steels combining good corrosion resistance with high structural strength. The latter requirement is especially important in connection with the continuous rise in the working parameters of machines using corrosion-resistant steels. As requirements for hot strength and anticorrosion properties are raised, the chemical composition of steels usually becomes more complex and their workability in all stages of metallurgical production becomes poorer.

Table 1 shows the chemical composition of some of the corrosion-resistant steels of the perlite, martensite, and austenite classes most widely employed in construction of powerplants; these steels are briefly described below.

1. STEELS OF THE PERLITE CLASS

The advantages of high-hot-strength perlite steels include a low alloying-element content, good thermal conductivity, a low coefficient of linear expansion, good workability, and relatively low forging cost.

The common alloying feature of this group of steels is the fact that they contain 1-3% chromium, 0.5-1.0 molybdenum (the principal solid-solution hardening element), and vanadium (as the carbide-forming element). Some types of steels contain additional components (e.g., tungsten) that have a favorable influence on the mechanical strength of the metal at elevated temperatures.

The perliet steel with the highest hot strength is ЭИ415, which is widely used in turbine building for fabricating wheels, rotors, and other components. In terms of hot strength, this steel is intended for operation at temperatures of up to 550°C for prolonged periods (100,000 h) and at 550-580°C for short periods. Type ЭИ415 steel has a high capacity for thermal improvement and good temperability.

When used in forged wheels with hubs 300 mm high and no central hole, ЭИ415 steel is characterized by the following mechanical properties (in tangential specimens): yield strength $\sigma_{0.2}$ = 70-78 kgf/mm², ultimate strength $\sigma_{в}$ = 80-86 kgf/mm², relative elongation δ = 16-18%, relative reduction in cross-sectional area ψ = 50-60%, and impact strength $a_{н}$ = 8-10 kg-m/cm². Research has shown that such wheels can be treated to give them higher strength while retaining complete satisfactory plasticity indices.

A great deal of experience in the production of forged wheels and other components of various sizes and shapes from ЭИ415 steel at the Neva Machine-Building Plant imeni V.I. Lenin (NPL) has confirmed that it is technologically suitable for hot working and provides reliable mechanical properties.

High mechanical properties and good homogeneity of properties over the forging cross-section have also been noted in a large rotor with a body diameter of about 900 mm [1]. At $\sigma_{0.2}$ = 60-65 kgf/mm² and $\sigma_{в}$ = 70-78 kgf/mm², the plasticity indices were δ = 16-17%, ψ = 50-60%, $a_{н}$ = 10-12 kgf.m/cm². The difference in the yield strengths at the periphery of the body and in the region close to the central hole (100-120 mm in diameter) was about 5%.

When subjected to long-term tensile loading, ЭИ415 steel treated to $\sigma_{0.2} \geq 65$ kgf/mm² has high plasticity. Steel intended for components to serve for 100,000 h can be deformed by 1%. As a result of its lower plasticity, the permissible deformation of steel treated to $\sigma_{0.2} \geq 75$ kgf/mm² is limited to 0.5%.

In view of its combination of hot strength and stability of properties, P2 rotor steel, which was developed by the laboratory of te Leningrad Metals Plant [LMP](ЛМЗ), can be used for turbine components intended to operate at temperatures of up to 535°C. The principal advantages of this steel is that it has no tendency toward thermal embrittlement and has satisfactory plasticity at elevated temperatures.

Heat-treated P2 steel is characterized by high mechanical properties at 20°C. A series of 6 wheels with hubs 100 mm high exhibited the following mechanical properties: $\sigma_{0.2}$ = 80.3-88.0 kgf/mm², $\sigma_{в}$ = 89.8-97.0 kgf/mm²; δ = 15.2-17.6%; ψ = 46.7-70.0%; $a_{н}$ = 6.0-13.7 kgf·m/cm².

In accordance with the recommendations made by the LMP laboratory, large rotor forgings fabricated from P2 steel are subjected not to refining but to single or double normalization followed by high tempering. Rotor forgings in this state have the following mechanical properties: $\sigma_{0.2}$ = 50-55 kgf/mm², $\sigma_в$ = 70-75 kgf/mm², δ = 15-18%, ψ = 40-60%, $a_н$ = 4-10 kgf·m/cm². When cut, a rotor more than 800 mm in diameter was found to have high uniformity of properties throughout the forging volume. The yield strength $\sigma_{0.2}$ holds constant at 41-45 kgf/mm² over the temperature range 450-550°C.

The satisfactory plasticity of P2 steel when subjected to long-term tensile testing at 500-550°C permits a deformation of 1%.

Type P2M rotor steel with an elevated molybdenum content (up to 0.8-1.0%) has a higher hot strength. According to the data of the LMP, the increase in molybdenum content has a favorable effect on mechanical properties at room temperature and ensures satisfactory deformability under long-term tensile loading. Type P2M steel is used for large rotors with steam parameters of up to 580°C and 240 atm (abs).

Type 15X1M1Ф steel is employed for forged flanges, pipes, T-joints, and other steam-turbine fittings intended to operate at temperatures of up to 565°C. After normalization and high tempering, forged flanges of the wheel type with a height of 170 mm have the following mechanical properties (tangential specimens): $\sigma_{0.2}$ = 33-36 kgf/mm², $\sigma_в$ = 52-56 kgf/mm², δ = 26-31%, ψ = 75-78%, $a_н$ = 18-29 kgf·m/cm². Increasing the forging height to 340 mm causes the yield strength $\sigma_{0.2}$ to decrease to 30-34 kgf/mm².

Thorough investigations of experimental forgings from 15X1M1Ф steel, involving mechanical tests of different zones conducted at the LMP laboratory, showed their properties to be sufficiently uniform over their cross-section.

According to the data of the LMP, the mechanical properties of quenched and tempered forgings have the following values: $\sigma_{0.2}$ = 40-50 kgf/mm², $\sigma_в$ = 57-65 kgf/mm², δ = 20-23%, ψ = 65-70%, $a_н$ = 7-15 kgf·m/cm².

This group of high-hot-strength perlite steels has a very favorable combination of technological properties, which permits fabrication of high-quality forgings without material metallurgical difficulties.

2. HIGH-CHROMIUM STAINLESS AND HIGH-HOT-STRENGTH STEELS OF THE MARTENSITE CLASS

Chromium martensite steels have high corrosion resistance and hot strength combined with good mechanical properties, which has resulted in their finding broad industrial application. Large forgings from these steels are used in chemical machine building and in other production areas involving exposure to corrosive media. In turbine building, stainless chromium steels are employed principally for blades intended to operate at temperatures of up to 450-500°C and sealing bushes. Additional alloying of chromium martensite steels permits them to be used as high-hot-strength materials, replacing austenite steels.

Type 2X13 chromium steel has relatively high strength and good plasticity. Forged shafts 150 mm in diameter, subjected to normalization and tempering, have the following mechanical properties: $\sigma_{0,2} = 50-55$ kgf/mm^2, $\sigma_e = 68-70$ kgf/mm^2, $\delta = 21-25\%$, $\psi = 60-65\%$, $a_H = 8-12$ kgf·m/cm^2. When tested in the quenched and tempered state, forged wheels have higher strength indices: $\sigma_{0,2} = 60-65$ kgf/mm^2 and $\sigma_e = 80-85$ kgf/mm^2 with $\delta = 14-18\%$, $\psi = 57-65\%$ and $a_H = 5-7$ kgf·m/cm^2. In a forged rotor with a body diameter of 720 mm subjected to the same heat treatment, $\sigma_{0,2}$ and σ_e are reduced to 42-45 and 61-63 kgf/mm^2 respectively with $\delta = 22-28\%$, $\psi = 45-55\%$ and $a_H = 6-10$ kgf·m/cm^2; the mechanical properties remain very uniform over the forging cross-section.

Technologically the most complex of the martensite steels is 1X17H2 (ЭИ268). Its principal advantage over other stainless steels is its combination of high mechanical properties (a yield strength of up to 60-75 kgf/mm^2) with extremely good corrosion resistance in aggressive media. In view of its chemical and mechanical properties, this steel is regarded as the most suitable for shafts and wheels for nitrous-gas compressors and certain other machines.

High-hot-strength steels based on 11-14% chromium and subjected to additional alloying with tungsten, molybdenum, vanadium, and certain other hardening elements have come into wide industrial use in recent years, especially in powerplant construction. The effectiveness of such alloying lies in the fact that it hardens and increases the hot strength of chromium stainless steels, bringing the operating-temperature limit up to 580-600°C.

While having high mechanical properties at normal temperatures and a hot strength intermediate between those of perlite and austenite steels, hardened high-chromium steels are also distinguished by a number of other characteristics (scale resistance equivalent to that of austenite steels, high thermal conductivity, and a low coefficient of linear expansion), whose practical utilization in powerplant construction is very important. These properties result in smaller thermal stresses and permit naked components of high-chromium martensite and perlite steels to be used in machine assemblies.

Economically, chromium steels differ from austenite steels in their greater technological workability and relative cheapness. They are therefore being more and more widely used in modern powerplant construction for fabricating rotors, wheels, and other components intended to operate at temperatures of up to 580-600°C. Large turbine forgings from stainless steels are being produced by the Bochumer Rhein plant (FRG) and other foreign firms [2]. In the Soviet Union, a number of high-hot-strength high-chromium steels have been developed by the Central Steam Turbine Institute, the Central Scientific Research Institute of Technology and Machine Building, the Leningrad Metals Plant, and other organizations. Specifically, 1X12ВНМФ (ЭИ802), 1X12В2МФ (ЭИ756), and 15X11МФБ steels have been developed, investigated with large forgings, and used industrially for fabricating steam- and gas-turbine components.

Large forged wheels and rotors of 1X12ВНМФ steel are characterized by the following mechanical properties: $\sigma_{0,2} = 60-70$ kgf/mm^2, $\sigma_e = 80-87$ kgf/mm^2, $\delta = 14-17\%$, $\psi = 30-40\%$, $a_H = 5-7$ kgf·m/cm^2

Type 15X11MΦБ steel, which was developed by the LMP laboratory, is intended principally for fabrication of cast blanks but is also used for certain steam-turbine fittings (T-joints, pipes, etc.) to operate at temperatures of up to 580°C. Such forged components weigh up to 3-5 t. Tests conducted with forgings having diameters of 300-350 mm after normalization and tempering showed them to have the following mechanical properties (tangential specimens): $\sigma_{0.2} = 55-65$ kgf/mm², $\sigma_b = 75-80$ kgf/mm², $\delta = 18-25\%$, $\psi = 50-60\%$, $a_n = 6-8$ kgf·m/cm².

A general characteristic of high-hot-strength high-chromium steels is their high sensitivity to various deviations in the metallurgical process cycle.

3. STEELS OF THE AUSTENITE CLASS

Chromium-nickel steels of the 18-9 type, with titanium, niobium, and molybdenum added, are most widely used as corrosion-resistant austenite steels. The development of power engineering and the continuous increase in machine working parameters has necessitated the creation of many high-hot-strength austenite steels with varied and complex chemical compositions [3], which are employed for high-temperature operation.

A common high-hot-strength stainless steel with good stability in corrosive media is X18H9T. One can get some idea of the mechanical properties of this steel at normal temperatures from the following data, which were obtained in tests on forgings of the wheel type after austenization and tempering (tangential specimens): $\sigma_{0.2} = 30-34$ kgf/mm², $\sigma_b = 53-60$ kgf/mm², $\delta = 47-50\%$, $\psi = 65-70\%$, $a_n = 16-20$ kgf·m/cm². Forgings of the same type of steel containing no titanium (X18H9) usually have a lower yield strength ($\sigma_{0.2}$ = 24-28 kgf/mm²) with δ and ψ falling in the range 50-60%.

The corrosion resistance of a steel depends largely on its composition and structural state, which is governed by the heat-treatment regime. The lower the carbon content of a steel, the higher is its corrosion resistance. Titanium also has an effective action, the resistance of a steel to intercrystalline corrosion depending on the quantitative ratio of the titanium to the carbon. Technical specifications therefore relate the minimum permissible titanium content to the carbon content. The carbon-titanium ratio shown in Table 1 does not conform to GOST 5632-51, since production experience with critical forgings from X18H9T steel has shown that the minimum titanium content stipulated by this GOST, equal to (C% -0.03)·5, is insufficient to impart effective corrosion resistance to large forgings (e.g., compressor wheels) subjected to heat treatment in the form of austenization and subsequent tempering. Colombier and Hochman [4] recommend that the ratio of Ti to C be raised to 6.

However, X18H9T steel does not have the necessary corrosion resistance in a number of chemically active media. A chromium-molybdenum-nickel steel of the 16-13-3 type with titanium added or X18H12M2T steel, which exhibit a smaller tendency toward intercrystalline corrosion, are sometimes employed in such cases. The mechanical properties of X18H12M2T steel are similar to those of X18H9T steel.

Type Х18Н22В2Т2 steel is dispersion-hardening and is distinguished by good mechanical properties. After austenization and aging, large forgings of this steel (from initial ingot weighing up to 11 t) have the following mechanical properties: $\sigma_{0.2} = 40-55$ kgf/mm², $\sigma_в = 80-90$ kgf/mm², $\delta = 25-35\%$, $\psi = 35-50\%$, $a_н = 10-15$ kgf·m/cm².

Type ЭИ405 steel can undergo slow, prolonged aging and is characterized by the following mechanical properties in forged wheels: $\sigma_{0.2} = 34-38$ kgf/mm², $\sigma_в = 58-60$ kgf/mm², $\delta = 40-50\%$, $\psi = 45-65\%$, $a_н = 10-12$ kgf·m/cm². This steel is highly resistant to corrosion and scale formation at temperatures of up to 750°C, but it has a tendency toward embrittlement as a result of precipitation of a σ-phase, which limits its application in components intended for prolonged service at temperatures of up to 600°C [1].

Like ЭИ405 steel, ЭИ395 steel can undergo long-term aging and is dispersion-hardening. The mechanical properties of this steel at 20°C are distinguished by high strength indices: $\sigma_{0.2}$ — to 50 kgf/mm², $\sigma_в$ — to 88 kgf/mm² with $\delta = 23\%$ and $\psi = 37\%$. Its mechanical properties are also high at elevated temperatures; for example, the yield strength $\sigma_{0.2} = 30-32$ kgf/mm² at 650°. The stability of properties during prolonged holding characteristic of this steel permits it to be used with confidence in components intended to operate at temperatures of up to 650°C [1]. However, it is very difficult to forge. There are great obstacles to production of large forgings from this steel.

The most economical austenite steel is ЭИ572, whose hot strength exceeds that of many other steels with a similar degree of alloying. This steel is also dispersion-hardening, with hardening achieved by austenization and subsequent aging. The following typical mechanical properties have been found in tests on commercial wheels: $\sigma_{0.2} = 35-40$ kgf/mm², $\sigma_в = 65-75$ kgf/mm², δ and $\psi = 20-30\%$, $a_н = 5-8$ kgf·m/cm².

The average yield strength $\sigma_{0.2}$ of ЭИ572 steel is 20-26 kgf/mm² 600°C and 16-20 kgf/mm² at 650°C. The relative elongation δ, the reduction in cross-sectional area ψ, and particularly the impact strength a_k decrease abruptly when the steel is held at 650-700°C, which is due to its structural instability: it has a tendency toward embrittlement resulting from formation of a σ-phase.

Type ЭИ572 steel has high long-term plasticity at temperatures of up to 600-650°C and, in view of its hot strength, is usually employed for components intended for prolonged service at such temperatures. It is unsuitable for use at higher temperatures, becoming brittle.

Type ЭИ726 chromium-nickel-tungsten-niobium steel with boron added, which was developed by the Central Scientific Research Institute of Ferrous Metallurgy, is an austenite steel with relatively high hot strength. It has good plastic properties during long-term tensile testing and high stability of structure and properties during prolonged heating; however, it has a comparatively low yield strength at normal and elevated temperatures.

In large forged wheels fabricated from ingots weighing 4 t, the yield strength $\sigma_{0.2}$ in the tangential direction at 20°C is 24-28 kgf/mm^2 and the ultimate strength σ_b = 50-57 kgf/mm^2.

The chromium-nickel-tungsten-titanium steel ЭИ612, which was developed by the Central Scientific Research, Planning, and Design Steam Turbine Institute, has a hot strength exceeding that of many other austenite steels. This steel has good mechanical properties at 20°C: $\sigma_{0.2}$ = 40—55 kgf/mm^2, σ_b = 75--90 kgf/mm^2, δ = 20--28%, ψ = 25--45%, a_n = 6—10 kgf·m/cm^2. The yield strength $\sigma_{0.2}$ is no less than 40 kgf/mm^2 at 600-700°C.

The principal advantages of this steel are its long-term plasticity, which makes it insensitive to stress concentrators, and the fact that it has no tendency toward intercrystalline corrosion. Type ЭИ612 steel is recommended for wheels intended for prolonged service at temperatures of up to 650°C.

An improved version of ЭИ612 steel is ЭИ612K, whose chemical composition differs in additional alloying with 3-4% cobalt and boron. The mechanical properties of this steel are characterized by high strength indices and satisfactory plasticity: $\sigma_{0.2}$ = 50—60 kgf/mm^2, σ_b = 90--95 kgf/mm^2, δ = 25--27%, ψ = 30—35%, a_n = 7—9 kgf·m/cm^2.

Type ЭИ612K steel has high plasticity during prolonged tensile testing at 700°C and can withstand prolonged service at working temperatures of up to 650-700°C. However, fabrication of large forgings from this steel is a complex process.

Chapter 2

ROLE OF METALLURGICAL FACTORS IN DETERMINING THE PROPERTIES OF FORGING

The metal in special-steel forgings should be compact and physically homogeneous, with a minimum content of gases and nonmetallic inclusions. The higher the purity of the metal, the better are its physical, mechanical, and technological properties.

Defects in the metal, in the form of cracks, flakes, and pores, reduce its properties, since they are strong stress concentrators and serve as foci of component failure. A high gas content in steel reduces its plastic properties and often causes flakes to develop. Nonmetallic inclusions, which are inevitably present in any steel, reduce its structural strength and serves as foci of corrosion and fatigue failure.

Many failures in attempts to fabricate large forgings from high-alloy steels are due to internal defects in the ingot. Proper selection of the basic parameters and casting conditions for the ingot is therefore a factor no less important than smelting of high-quality steel.

Most special steels have low plastic properties at forging temperatures and are difficult-to-deform. A number of smelting and casting defects, such as contamination with interstitial impurities (lead, antimony, and arsenic) and poor deoxidation, reduce the workability of steel and sometimes cause it to completely lose its hot deformability. The presence of flakes, large scales, or slag inclusions at the ingot surface also greatly reduces the deformability of steel when ingots are forged without preliminary cleaning.

Forging of steel under favorable thermomechanical conditions with optimum stress patterns sometimes correct metallurgical defects in the ingot or at least minimizes their detrimental effect on the quality of the finished forging.

For example, it has been established that shrink holes and pores in ingots of many perlite and austenite steels are well filled under definite forging conditions and leave no detrimental traces in the macrostructure of the forged component, while preliminary cleaning of the ingot almost totally eliminates the detrimental effect of surface defects. However, deformation plays a very limited role in correcting metallurgical defects. First of all, a number of steels are difficult to weld up (these include such high-chromium complex-alloyed martensite steels as ЭИ802, 15Х11МФБ, etc.). This process takes place satisfactorily only when the steel is of high purity. Secondly, not all forgings (particularly those that are large or have complex configurations) permit the requisite technological measures to be taken for effective welding-up of internal pores.

As for mechanical removal of certain superficial metallurgical defects, this process is expensive, laborious, and not always technologically feasible.

The examples given show that it is unreliable and often impossible to correct metallurgical defects by any mechanical method, although special solutions can theoretically be very effective.

We must therefore again emphasize that the basis of high forging quality lies in the initial ingot, i.e., the latter should have a compact and homogeneous structure, minimum gas saturation, and minimum contamination with detrimental impurities and nonmetallic inclusions.

4. SMELTING AND CASTING OF STEELS USED FOR CRITICAL FORGINGS

Smelting of steel. Analysis of the characteristics of various steel-smelting processes and many years of experience in the production of large forgings convincingly shows that use of an open-hearth furnace with a basic bottom for smelting steels intended for fabrication of critical forgings is impermissible, since the steel smelted in such a furnace does not provide the high metal quality stipulated by technical requirements. Metallurgical defects, in the form of flakes, a high nonmetallic-inclusion content, and reduced plasticity in transverse and tangential specimens, are usually present in most large critical forgings fabricated from basic-open-hearth alloy steels smelted by current techniques, even when special technological procedures are employed to prevent high gas saturation and metal contamination.

Relatively reliable results with respect to both purity and minimum gas saturation are yielded by acid-open-hearth steel, which has certain advantageous physical characteristics, particularly low flowability and low slag hydrogen permeability [5]. Such steel can be smelted by using an especially pure charge and handling the technological process carefully; the duplex process (reductive silicon method) is used for most critical forgings. This process is used at the Urals Machine Plant, where turbine-component forgings are fabricated from special steels of the perlite and martensite classes.

Steel smelted in a basic arc furnace by oxidation or remelting is used for a broad range of forgings. An electric furnace is the only possible smelting unit for certain types of high-alloy steel, such as those of the austenite class. Certain of the main metal-quality indices, and particularly the gas saturation, of medium-alloy steels produced by this method are markedly lower than those of acid-open-hearth steels.

Comparison of the hydrogen content in melts of medium-alloy acid steel and medium-alloy basic electric steel shows that the average hydrogen content of the former is about 3-4 $cm^3/100$ g of metal and that of the latter is 5-8 $cm^3/100$ g, i.e., the hydrogen content of basic electric steel is about 1.5-2 times that of acid-open-hearth steel.

The physicochemical properties of the slag and the technological characteristics of the process permit metal with a low content of detrimental impurities and slag inclusions to be smelted in electric furnaces. Thus, arc smelting can yield a steel with a sulfur content of less than 0.010-0.015%, which is very difficult to do in the open-hearth process. However, it must be kept in mind that forging quality is affected both by the quantitative content of nonmetallic inclusions and by the type and distribution of the inclusions in the steel. Research and analysis of many commercial forgings fabricated from steel smelted by different methods have established that randomly distributed inclusions of the globular type, which do not create stress concentrators, have the minimum effect on the mechanical properties and hot deformability of steel. Inclusions distributed in the form of films, chains, or eutectic networks are more detrimental, essentially giving forgings low mechanical properties in the transverse and tangential directions. From this standpoint, the character of the inclusions in acid-open-hearth steel is generally more favorable than that of the inclusions in basic electric steel, although that in the latter can theoretically be altered by various technological smelting procedures.

In view of the wide use of electric arc furnaces with basic bottoms for smelting a broad range of alloy and high-alloy special steels, the problem of increasing the quality of basic electric steel and particularly that of reducing its gas saturation are particularly pressing.

The experience of the NPL and certain other plants indicates that it is possible to obtain a substantial increase in the quality of electric steel by taking a number of technological measures. Prime among these are selection of a charge with a reduced content of gas-producing elements, use of specially smelted charge blanks for the most critical melts, roasting of the lime, oxidizing agents, and ferroalloys, selection of the optimum melting and slag regimes, use of complex deoxidizing agents, and effective utilization of diffusion deoxidation. These measures permit fabrication of relatively large forgings from various alloy steels (e.g., forged gas-turbine wheels with hubs 400 mm high from ЭИ415 steel) that satisfy all the rigid quality requirements imposed by technical specifications or critical turbine components.

Rare-earth metals [REM](РЗМ), principally Dutch metal and ferrocerium, have recently come into wide use for better deoxidation and modification of special steels, especially those of the austenite class. It has been shown that REM actively interact with the impurities in the metal, facilitating removal of nonmetallic inclusions and dissolved gases, reduction of grain size, and elimination of crystallization defects; they reduce the amount of α-phase in the austenite chromium-nickel and chromium-nickel-molybdenum steels X18H9T and X18H12M2T. As a result, REM increase metal purity, reduce anisotropy of mechanical properties, and have a favorable effect on deformability at forging temperatures.

Casting of steel. Optimum casting methods and regimes facilitate formation of a compact ingot macrostructure, a decrease in the content of gases and nonmetallic inclusions, and minimum development of surface defects. We must emphasize the importance of the latter factor, which is directly related to solution of one problem in the production of special-steel forgings: forging of hot-delivered ingots without cleaning, local removal of defects, or other types of preparation. Hot delivery of low-plasticity, poorly deformable steels, which include most special steels of the austenite, martensite, and ferrite-martensite classes, is permissible only when the ingot surface has no cracks or other large defects, which inevitably have an unfavorable effect on forging, extending as far as ingot fracture during the first few passes. However, provided that the ingot surface is satisfactory, hot delivery has a number of undoubted advantages, including more economical utilization of production areas in the smelting shop, shorter preforging heating of the ingots, and increased technological efficiency. As research and plant experience has established, a continuous hot cycle is an important condition for successful production of high-quality forgings from many alloy and high-alloy steels.

Of the two principal steel-casting methods, top and bottom pouring, special steels are usually cast only by top pouring, which produces fewer axial defects in the ingot and results in less contamination of the metal with slag inclusions.

During top casting, liquid steel spatters on the walls of the mold and, oxidizing, forms large scabs and flakes of metal on the ingot surface. The oxidized metal, reacting with the molten steel, remains as inclusions within the ingot, thus increasing the content of nonmetallic inclusions of the oxide type. This is particularly true of special steels alloyed with titanium, aluminum, and chromium, since these elements have a strong tendency toward oxidation during pouring.

The method by which the steel is poured (directly from the ladle or through some intermediate device) and the atmosphere in the mold during filling have a strong influence on ingot quality. These factors sometimes have a decisive effect on contamination with nonmetallic inclusions, surface quality, and cracking.

Pouring the metal through an intermediate funnel substantially improves mold-filling conditions. Even during the initial stage of pouring, the metal enters the mold with a small ferrostatic head, in a smooth, accurately centered stream. The force with which the stream strikes the bottom at the beginning of casting is substantially reduced. All this promotes an increase in ingot quality. In all cases where the thermophysical state of the molten steel permits the mold to be filled through intermediate funnels, this process therefore is undoubtedly preferable to direct pouring from the ladle. Pouring of special steels directly from the ladle is permissible only when the molten metal has high viscosity and it is difficult to use intermediate funnels.

Possible contamination of the steel with the products of the mechanical and chemical action of the molten metal on the funnel lining causes some misgivings. Serious practical evaluation of this factor is necessary in fabricating forged turbine components and other critical articles from ingots. However, sufficiently stable refractories are now available and their use for funnel linings provides completely reliable results, as has been confirmed by many years of experience in casting alloy and high-alloy steels at the NPL.

The amount of metal oxidized and hence the contamination with nonmetallic inclusions decrease as the liquid-metal surface exposed to the air during pouring is reduced. Oxide formation during casting is the principal factor responsible for surface defects in ingots. Such defects can therefore be eliminated or at least reduced by creating a reducing or neutral atmosphere in the molds. The neutral gas argon is specifically employed for this purpose, being supplied to the mold before the steel is poured; it drives the oxidizing gases out of the mold cavity and reduces the contamination of the molten metal with nonmetallic inclusions. Ingot surface quality is markedly improved. Argon is used in casting high-alloy steels, including the austenite steels ЭИ572, Х18Н22В2Т2, etc.

Various plants have begun to make wide use of the method proposed by Engineer K.N. Ivanov for casting special steels, especially those alloyed with titanium, chromium, and aluminum. This technique consists in introducing magnesium shavings, chips, or powder into the mold (before casting) in amounts of about 75-80 g/t of metal, in order to absorb all the oxygen and set up a nonoxidizing atmosphere in the mold. The latter is tightly covered with a steel lid before casting. When the first portion of metal is delivered, the magnesium burns and binds the oxygen present in the mold. Ingots cast with magnesium are distinguished by a noticeably improved surface.

Parallel experiments on the use of argon and magnesium in casting 0Х18Н10Т, ЭИ572, Х18Н12М2Т, Х18Н22В2Т2, and other high-alloy steels into ingots weighing up to 16 t have demonstrated the advantages of magnesium over argon.

In order to reduce the number of surface defects in high-alloy steel ingots, some plants cast the metal under a layer of molten slag [6,7]. Study of the macrostructure of ingots cast by this method showed that the slag is not included in the metal and that the contamination and porosity of the ingot is, in any event, no worse than that of ingots cast by the usual method. At the same time, there is a severe decrease in surface quality. Experimental ingots weighing from 6.5 to 32 t have been cast under a layer of molten slag with satisfactory results at the Neishil Sapplay plant in Torrence (USA) [8].

Further study and improvement of casting under molten slag as this technique applies to special steels used for large critical forgings is obviously wise.

Cooling of ingots. The thermal conditions under which cooling takes place have a strong influence on the quality of ingots of alloy and high-alloy steels. In particular, the sensitivity of steel to formation of intercrystalline cracks and to propagation of such cracks during transfer of hot ingots from the smelting to the pressing shop is greatly increased when the metal temperature drops (to below 600-650°C). This is true both of martensite steels and of certain types of austenite steels (Х18Н22В2Т2, ЭИ572, etc.). In individual cases, it is therefore best to deliver the ingots to the press shop in the molds, so that their temperature is maintained for a longer period. The NPL has successfully employed this technique for small ingots of ЭИ802, 15Х11МФБ, Х17Н2, Х18Н22В2Т2, ЭИ572, and certain other steels. Statistical analysis showed that forgings fabricated from cooled ingots exhibit a higher percentage of rejects for foliation and other metal discontinuities detected ultrasonically. There is also a marked decrease in deformability during forging.

If the technological process for production conditions require that the ingot be cooled to normal temperatures, this is done in heat-treatment furnaces from a temperature of 600-700°C, following a special regime that depends on the specific type of steel.

Special methods for producing high-quality steel. The development of gas-turbine building and other branches of modern technology has necessitated smelting of special-purpose steels with high physical homogeneity, a compact macrostructure, no nonmetallic inclusions, and high mechanical properties. The usual technique for steel production in open-hearth and electric furnaces cannot always provide metal that satisfies such high requirements, even when special technological procedures are employed. To some extent, this problem has been solved by use of new steel-production methods: evacuation during casting, electric slag remelting, and vacuum smelting. Wide use of these smelting and casting techniques in special-steel production is planned for the immediate future.

Vacuum casting of molten steel is one way to effectively increase metal quality. This procedure has come into wide use in the USSR and abroad. About 80% of the steel intended for forgings in the USA is subjected to vacuum treatment [9].

Vacuum casting promotes degasification of the metal, specifically reducing the hydrogen concentration, decreasing the content of nonmetallic inclusions, and providing a more uniform inclusion distribution in the ingot.

According to the data of the Urals Machine Plant [10], the hydrogen content of steels of types such as Р2 is reduced by [illegible]-40% during exposure to a vacuum, while the total nonmetallic-inclusion content decreases by a factor of 2-3. Vacuum casting has been found to have a similar influence on degasification and nonmetallic-inclusion content in ЭИ415 steel at the Novo Kramatorsk Machine Building Plant [NKMP](НКЗМ) [11]. Foreign data [9] have also confirmed that the content of hydrogen and oxide inclusions in vacuum-cast steel is far lower than in steel cast by the usual method. For example, it has been found that the hydrogen content of treated steel is reduced by 40-60% and its oxide-inclusion content is reduced by no less than 60-70%. There is a decrease in both the number and size of the nonmetallic inclusions, which has a favorable effect on the physicomechanical properties of the steel.

A radical technique for improving metal quality is molten-slag remelting, which provides high purity from nonmetallic inclusions, high compactness and uniformity of macrostructure, and increased forgability.

Considerable experience has now been amassed in the production of forgings from special steels obtained by molten-slag remelting. An extremely large molten-slag furnace, intended for production of ingots weighing up to 12-14 t, was put into operation at the Novo Kramatorsk Machine Building Plant (in the Donbass) in 1962. Furnaces for production of ingots weighing up to 40-50 t are in the planning stage.

Vacuum smelting is another way to solve the principal problems in the production of high-quality steel. Vacuum arc furnaces with consummable electrodes are the most widely used. It has been established that, even when the furnace vacuum is low, there is a substantial decrease in the amount of dissolved hydrogen, oxygen, and nitrogen in the steel. The content of nonmetallic inclusions, especially those of the oxide type, is also greatly reduced and their distribution over the ingot becomes more uniform. Vacuum-steel ingots are distinguished by reduced liquation and porosity and good deformability. Detrimental impurities (tin, lead, antimony, and bismuth) are removed during vacuum smelting [12]. In combination, the aforementioned factors promote an increase in the mechanical and special properties of steel, including hot strength.

Steel of especially high quality, with high, stable hot strength and long-term plasticity, is produced by two or even three vacuum arc remelts, which improves metal quality to a substantially greater extent than a single remelt.

5. DISTRIBUTION OF NONMETALLIC INCLUSIONS AND HYDROGEN IN INGOTS

Numerous experimental investigations have established that most of the nonmetallic inclusions of the oxide type are located in the lower portion of the ingot (the lower third of its height). The ingot content gradually decreases from the bottom of the ingot to the top. In cross-section, the nonmetallic-inclusion content is higher in the axial portion of the ingot.

The qualitative character of the inclusion distribution along the ingot axis remains roughly the same in ingots of different weights, the only difference being that the maximum oxide-inclusion content lies closer to the bottom of the ingot as its weight increases.

The preferential accumulation of nonmetallic inclusions in that portion of a forging corresponding to the lower third of the ingot has been confirmed by many years of experience in the fabrication of large alloy-steel forgings. Under plant conditions, this pattern is very clearly seen during ultrasonic quality control of critical components: according to statistical data, the overwhelming majority of the forgings rejected at the NPL for impermissible accumulations of nonmetallic inclusions were fabricated from the lower portion of an ingot. This situation also holds for other plants.

In individual cases involving fabrication of very critical components, a special quality-control check is made on the intermediate blanks by ultrasonic defectoscopy, rejecting the portions of the ingot with the highest concentration of nonmetallic inclusions. This procedure is specifically employed in forging large numbers of smooth plates from ingots of P2 steel weighing up to 145 t. The ingot, reduced to a plate over its entire length, is subjected to ultrasonic defectoscopy and then laid out for cutting in accordance with the results obtained. It was found that the overwhelming majority of the defects (nonmetallic inclusions) in all the ingots were located in the bottom portion, in a zone occupying from 1/4 to 1/3 of the ingot height. This example is quite interesting, since the nonmetallic-inclusion distribution was found to hold for large ingots on a broad production scale.

The concentration of oxide inclusions in the lower portion of the ingot apparently results from secondary oxidation of the steel when it is poured into the molds. The liquid metal that spatters when the stream strikes the bottom of the mold is oxidized in the air and, entering the metal as it is poured in, ultimately takes the form of inclusions.

The contamination of ingots with nonmetallic inclusions can be reduced by pouring the steel in a nonoxidizing atmosphere. This is one of the main factors responsible for the sharp decrease in the nonmetallic-inclusion content of ingots cast in a vacuum.

The hydrogen distribution in an ingot has the following character: the hydrogen content along the ingot axis increases from bottom to top, while that over the ingot cross-section increases from the periphery to the center. According to data obtained in investigating a chromium-nickel-molybdenum steel ingot weighing 2 t [13], the hydrogen content along the ingot axis increase from 3.7 cm^3/100 g in the lower portion to 6.4 cm^3/100 g in the center and 7.4 cm^3/100 g in the upper portion; the content over the ingot cross-section increased from 1.7 cm^3/100 g at the periphery to 6.4 cm^3/100 g at the center, the highest hydrogen content occurring in the liquation zones. A number of other investigations have shown the hydrogen distribution to have a similar character.

A tendency for the hydrogen content to increase from the periphery to the center has also been observed in ingots of austenite steel, but the cross-sectional hydrogen distribution is more uniform than in ingots of structural pe[illegible]e steel. This can be illustrated by data for an ingot of Х18Н9Т austenite steel weighing 2.1 t, in which the hydrogen content ranged from 6.6-7.8 cm^3/100 g in the peripheral zone to 9-14 cm^3/100 g in the center [5].

This hydrogen-distribution pattern has also been confirmed by the results of quality-control tests on alloy-steel forgings under plant conditions: the highest hydrogen content generally occurs in forgings produced from the upper portions of large ingots.

6. SELECTION OF OPTIMUM INGOT PARAMETERS

The optimum initial-ingot shape is one of the decisive conditions for high forging quality and, at the same time, governs the economy of the technological process.

The principal requirement imposed on an ingot are minimum development of shrinkage and liquation phenomena and a compact central structure. These requirements are especially important for alloy and high-alloy steels containing nickel, chromium, vanadium, titanium, and other elements, since the nature of the forgings fabricated from special steels not only makes it impossible to remove the axial zone of the ingot in many cases but also necessitates its use for forming the most critical portions of the component (e.g., gas-turbine wheels without central holes). Shrinkage defects in the ingot are difficult to weld up during deformation and even special complex forging techniques often do not yield satisfactory finished products.

Development of metallurgical defects in an ingot is largely governed by its weight and shape, whose principal parameters are the ratio of height to average diameter (*H/D*) and taper, which is determined by the direction of ingot crystallization. The combination of these factors has a decisive influence on the physical structure and chemical homogeneity of the ingot. An increase in relative height reduces extracentral liquation as a result of rapid solidification and limitation of diffusion processes on the one hand and promotes development of axial cores on the other, since horizontal solidification proceeds more rapidly than vertical solidification; the situation is reversed in ingots with a reduced relative height. The less distinct solidification direction in ingots with a high *H/D* ratio causes formation of a less compact axial zone. An increase in ingot taper improves its internal structure: the zone of axial porosity is reduced. By selecting an optimum *H/D* and taper, it is possible to produce ingots that most fully satisfy requirements for physical and chemical homogeneity. However, there is also an economic factor that must be taken into account: an elongated ingot is usually more economical to work than a short ingot. This is due, first of all, to the higher yield of finished products resulting from the relatively lower shrinkage head and, secondly, to the increased forging-equipment productivity in cases where the ratio of the cross-sectional area of the initial ingot to that of the drawn billet is sufficiently high for specific forgings.

Selection of an optimum ingot shape should take into account the steel used, the forging configuration, and the degree of criticality. For example, high compactness of ingot macrostructure is not of as decisive importance in fabricating special-steel forgings of the ring or bushing type as in forging gas-turbine wheels without central holes. In the first case, the axial zone of the ingot is to some extent removed during broaching; in the second case, this zone forms the portion of the wheel subject to the most severe structural loading, i.e., the hub, as determined by etching and ultrasonic and mechanical tests on tangential specimens. Individual forgings, even those of complex shape, fabricated from steel that has been thoroughly welded up during forging can be produced with complete reliability from ingots with a less compact structure, but such ingots are sometimes totally unsuitable when a steel that is difficult to weld up (e.g., ЭИ802 or 15Х11МФБ) is used.

It is thus best to orient oneself toward a general-purpose ingot shape for special-steel forgings, with their diversity of types, classes, and varieties. Study of the characteristics of forgings and of the types of steel used under specific production conditions will thus permit completely reliable classification of steels with respect to the required axial-zone compactness in the initial ingot and selection of the ingots that are best with respect to both forging quality and production economy for each group of forgings.

Table 2 shows the principal dimensions and other parameters of ingots used for fabrication of critical special-steel forgings. All the ingots are octagonal in cross-section.

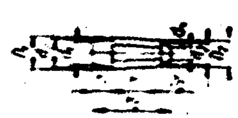
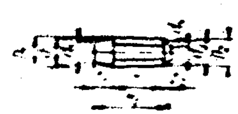
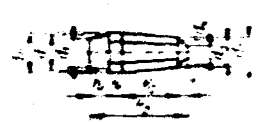
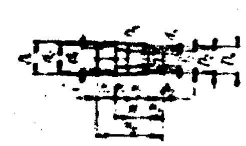
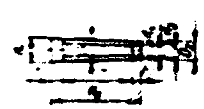
The NPL makes extensive use of high-taper ingots having an H/D ratio of about 1.7. Ingots weighing up to 5.7 t are produced with a stationary adapter, while those of greater weight are produced with a floating adapter. This ingot shape creates the requisite conditions for a given crystallization direction with minimum development of shrinkage phenomena. Prolonged production testing has established that the results obtained in using ingots with high taper or complex forgings from many alloys and high-alloy steels of the perlite and austenite classes are completely satisfactory, as is confirmed by the relatively compact and homogeneous ingot structure. The manner in which axial defects develops in the ingot is such that they can be successfully welded up under normal forging conditions, including forging of rotors from 2Х13 steel, solid wheels without central holes from ЭИ415, Х18Н9Т, and ЭИ572 steels, and even components of austenite steel with a very complex composition (Х18Н22В2Т2). The forgings exhibit no negative phenomena that might be attributed to increased liquation nonuniformity of the ingot.

However, use of ingots with this shape does not always have a favorable effect. We have observed some cases in which residual defects, in the form of foliation and unwelded pores, were detected in the central zones of forgings despite extensive machining of the metal. This was true principally of forgings from high-chromium complex-alloy steels of the martensite class (ЭИ802 and 15Х11МФБ).

Sectioning of an ingot of ЭИ802 steel weighing 3.75 t actually confirmed that the axial zone contained rather well-developed shrinkage pores, principally in the central third of the ingot height. The most serious defects, in the form of concentrated shrinkage pores in a vertical band about 150 mm long and 40-50 mm wide, were located about halfway up the ingot. The shrinkage cavity did not extend beyond the shrinkage head, below which was a layer of compact metal. There were no traces of porosity in the

TABLE 2

Characteristics of Ingots Used for Special-Steel Forgings

1 Эскиз слитка	2 Форма слитка	3 Вес слитка, т	4 Весовые соотношения частей слитка, % 5 Прибыль	6 Корпус	7 Низ	8 Размеры слитка 5 Прибыль d_1	d_2	h_1	6 Корпус D_1	D_2	D_3	D_4	H_1	H_2	H_3	H	7 Низ d_3	d_4	h_3	H_0	$\frac{H}{D_{cp}}$	9 Конусность корпуса слитка (на обе стороны), %
	1 0 Обычная	1,0	18	78,5	3,5	350	396	240	416	346	—	—	—	—	—	825	312	200	120	1186	2,50	5,3
		1,5	18	78,5	3,5	370	460	330	484	414	—	—	—	—	—	935	400	250	125	1360	1,97	8,0
		2,0	16	78,5	3,5	460	506	305	448	448	—	—	—	—	—	1030	440	260	130	1465	1,98	7,6
		3,0	18	78,5	3,5	514	580	379	616	584	—	—	—	—	—	1115	492	260	156	1641	1,83	7,7
	1 1 С повышенной конусностью и стационарной надставкой	3,75	18	78,3	3,2	560	630	370	700	590	—	—	—	—	—	1060	450	250	170	1620	1,64	10,4
		4,1	18	79,1	2,9	570	700	390	720	610	—	—	—	—	—	1095	450	250	170	1655	1,65	10,4
		5,2	18	79,1	2,9	650	760	400	780	660	—	—	—	—	—	1170	520	250	190	1780	1,68	10,8
		5,7	18	79,0	3,0	670	790	410	810	680	—	—	—	—	—	1250	530	250	190	1850	1,69	10,8
	1 2 С повышенной конусностью и плавающей надставкой	6—7,5	18	78,4—79,1	2,9—3,6	550	680	455	850	725	—	—	264	1170	—	—	500	334	210	—	1,48—1,75	10,7
		8—10,5	18	78,4—79,2	2,8—3,6	700	800	495	950	815	—	—	352	1255	—	—	550	334	250	—	1,42—1,7	10,7
		12—15	18	78,4—79,1	2,9—3,6	700	820	646	1076	915	—	—	340	1490	—	—	660	334	280	—	1,48—1,76	10,9
		16—20	18	78,4—79,1	2,9—3,6	810	930	615	1150	1020	—	—	360	1600	—	—	720	334	310	—	1,45—1,71	10,6
	1 3 С переменной конусностью и уменьшенным отношением $\frac{H}{D_{cp}}$	1,3	29,3	68,2	2,5	436	470	325	520	350	510	450	200	370	103	673	330	153	130	1128	1,45	1 5 См. эскиз
		2,1	26,6	70,7	2,7	515	576	367	615	420	600	530	250	450	110	810	400	190	153	1330	1,49	
		5,0	24,7	72,8	2,5	696	762	436	837	637	819	726	333	592	149	1074	607	285	167	1677	1,42	
		7,0	24,5	73,0	2,5	819	875	465	939	715	919	809	355	665	167	1187	690	260	164	1816	1,40	
		11,5	23,5	74,0	2,5	961	1030	530	1100	845	1096	960	431	792	197	1420	820	320	195	2145	1,42	
	1 4 Удлиненная	1,18	15	81,5	3,5	411	—	180	400	325	—	—	—	—	—	1200	250	212	145	1525	3,32	6,29
		1,65	15	81,5	3,5	458	—	210	445	360	—	—	—	—	—	1350	270	212	155	1715	3,35	6,29
		2,17	15	81,6	3,4	500	—	255	484	390	—	—	—	—	—	1500	300	246	160	1915	3,36	6,26
		3,53	13	83,5	3,5	630	—	280	580	504	—	—	—	—	—	1750	370	320	210	2240	3,34	6,29
		7,5	14	83,5	2,5	530	700	400	760	578	—	—	—	—	—	2240	460	334	204	—	3,34	8,00
		8,75	14	83,5	2,5	600	730	450	790	608	—	—	—	—	—	2400	480	334	230	—	3,40	7,50
		11,0	14	82,5	2,5	630	800	460	860	678	—	—	—	—	—	2460	530	334	260	—	3,20	7,30

1) Sketch of ingot; 2) shape of ingot; 3) ingot weight, t; 4) ratio of ingot-section weights, %; 5) shrinkage head; 6) body; 7) bottom; 8) ingot dimensions; 9) taper of ingot body (both sides), %; 10) normal; 11) with increased taper and stationary adapter; 12) with increased taper and floating adapter; 13) with varying taper and reduced H/D_{sr} ratio; 14) elongated; 15) see sketch.

lower portion of the ingot. The transverse axial-defect zone had an extent of 50 mm in only one section, being less distinct and no more than 20-30 mm wide in the other sections.

The poor weldability of ЭИ802 and similar steels during forging means that ingots with the shape under consideration cannot justifiably be used for critical forgings in which the axial zone of the ingot is not removed. Only use of especially pure melts and complex deformation techniques lead to complete welding up of the shrinkage defects in the ingots.

The group of ingots (Table 2) whose shapes are distinguished by triple taper and a low H/D ratio are of interest with respect to complex high-alloy steels used for production of forgings with continuous cross-sections.

The taper increases sharply (from 5.7 to 53% on both sides) as we move from the upper portion to the center and from the center to the lower portion of an ingot with triple taper. The H/D ratio equals 1.5. The bottom is spherical in shape. The shrinkage head is larger than in ordinary ingots, amounting to 28% of the total weight. Ingots of this shape were planned and first used for high-alloy steel forgings in the Urals Machine Plant. They subsequent came into use in certain other plants that produce special-steel forgings. The principal advantage of triple-taper ingots is the smaller extent of the shrinkage pores and hence the greater ingot compactness and homogeneity.

Sectioning and study of an ЭИ405 steel ingot weighing 1.3 t [14] established that there was no highly developed axial porosity and only very fine, discontinuous intercrystalline cracks in a narrow region of the axial zone, i.e., the shrinkage porosity was very limited in extent. A similar pattern was observed when a triple-taper ingot of 15X11MФБ steel weighing 2.1 t was investigated at the NPL. The fact that there were no large shrinkage defects at the surface of a longitudinal section distinguished it from other ingots.

Analysis of production results permits us to look favorably on triple-taper ingots from the standpoint of the quality of finished forgings of complex-alloyed steels. This is particularly true of metallurgically complex forgings of ЭИ802 and 15X11MФБ steels, which were checked by very precise type of defectoscopy.

However, triple-taper ingots are uneconomical: the large weight of the shrinkage head greatly increases the specific metal consumption in comparison with ingots having normal parameters, while the relatively large diameter (when not required by forging conditions) complicates the forging cycle, necessitating additional heating and greater labor consumption in forging fabrication. The metal consumption per ton of finished forgings is increased by an average of 15-17% in comparison with ordinary ingots.

Elongated ingots (Table 2) are most economical, greatly increasing the main technological-economic indices of forging production. Their use reduces metal consumption by 10-15%, raises forging-equipment productivity by 10-20%, and increases heating-furnace capacity. As a result, analysis and technical evaluation of the feasibility of using elongated ingots for a broad range of special-steel forgings is of great practical interest.

Ingots weighing up to 6 t with H/D = 3.3-3.4 and increased taper have a modified shrinkage head, which forms a unit with the mold and extends upward, which yields a more compact ingot structure and reduces the depth to which the shrinkage cavity extends.

Longitudinal sectioning and study of an ingot of chromium-molybdenum perlit steel weighing 2.17 t showed that the shrinkage cavity was dish-shaped and, at its greatest depth, occupied about 2/3 of the shrinkage head. The polished section surface exhibited no pores and only deep etching of the central portion, within a radius of 30 mm from the axis, revealed defects, which took the form of a fine cellular structure and small pores. The defect zone occupied about 30% of the ingot height. The extracentral liquation zone was less highly developed than in ordinary ingots.

Elongated ingots are widely used at the NPL for many forgings with different weight and configurations (including wheels without central holes) fabricated from perlite steels. Even with reduction factors less than those employed in forging ingots with normal parameters, no defects associated with axial pores were generally found in critical forgings produced from these ingots.

We also obtained favorable results in using elongated ingots for austenite-steel forgings. Specifically, we produced complex forged gas-turbine wheels with a continuous hub (ЭИ572 austenite steel) from elongated ingots weighing 2.17 and 3.5 t. No traces of residual ingot defects were detected when an experimental batch of such forgings and then many production batches were checked. In investigating an elongated ingot of Х18Н9Т austenite steel weighing 2.17 t, a fine but quite distinct intercrystalline crack extending over 2/3 of the ingot height and localized along the geometric axis of the ingot, was found in the axial zone. It began below the compact metal in the upper portion of the ingot and ran to 400-450 mm from the flange in the lower portion. The crack length in transverse sections of the ingot did not exceed 2-3 mm; there were almost no axial cells or pores in the radial direction. One characteristic of an elongated austenite-steel ingot is therefore very limited development of shrinkage defects in the transverse direction and a considerable extent in the longitudinal direction.

Comparison of research data with the results of quality-control tests on forgings produced from ingots of this type leads us to conclude that the lack of internal defects in austenite-steel forgings is due not so much to the high compactness of the axial zone of an elongated ingot as to the ease with which defects are welded up during forging. Welding up is facilitated both by the physical nature of the steel and by the complete isolation of intercrystalline cracks in the ingot, which protects their walls from oxidation during heating. As detailed investigations similar to those described in Chapter 8 showed, the minimum reduction in area necessary for effective welding up of shrinkage pores in an elongated ingot is usually small and achievable by ordinary forging processes.

At some plants, use of triple-taper ingots for austenite-steel forgings without central holes is regarded as a necessary condition for successful fabrication of such components. The experience of the NPL in using elongated ingots of high-alloy steels, including austenite steel, has shown that they can be widely employed for complex, very critical forgings.

The use of elongated ingots of high-chromium martensite steels (ЭИ802, 15Х11МФБ, etc.) is a considerably more complex problem.

Attempts to use even small elongated ingots (2.17 t) for forgings of continuous cross-section must be regarded as unsuccessful in practical terms: isolated forgings were of high quality but stable and satisfactory results were not obtained, despite the fact that a number of defective forging processes, involving alternate deformation directions, large reductions in area, and favorable deformation temperatures and rates, were employed. The defects detected in the forgings by ultrasonic and macroscopic examination (in sections of the component) were found to be related to unwelded ingot pores.

One ingot of 15Х11МФБ steel weighing 2.17 t was sectioned. A porous shrinkage structure with discontinuous cracks and cavities elongated along the ingot axis was found along the entire length of the ingot center, except in the area beneath the shrinkage head and in the lower portion of the ingot. The zone of pronounced axial porosity was 50-70 mm wide. Analysis of the microstructure of the metal in the central zone showed chains of carbide and oxides to be present along the grain boundaries, while defects in the form of intercrystalline discontinuities were also found. Complete welding up of axial defects in a steel of this type can apparently take place only when metal of very high purity is used. Since this is difficult to achieve when complex-alloyed, high-chromium steels are smelted in arc furnaces, selection of the optimum ingot shape is especially critical. Despite the increased metal consumption, it is best to use triple-taper ingots for complex forgings of such steels, since they are most reliable with respect to final production results; at worst, high-taper ingots should be employed (Table 2).

Among the additional factors affecting the compactness of ingot structure are the character of the riser and the thickness of the mold walls. Two types of risers are employed in producing special-steel forging ingots: the presence of a floating riser set close to its lower position has an unfavorable effect on ingot quality, since it causes substantial loss of heat into the upper portion of the mold, promoting formation of a "bridge" in front of the shrinkage head and poorer crystallziation conditions in the central zone. In casting ingots for critical applications, it is therefore desirable to avoid having a floating riser in this position and even better to use molds with stationary risers.

Use of thin-walled molds is of considerable interest. Research in this area [15, 16] has shown that use of such molds for ingots of high-alloy steels increases the compactness of the axial zone and improves forging quality.

Experimental work on the use of thin-wall molds for elongated ingots of 15Х11МФБ steel has been conducted at the NPL. Investigation of an ingot weighing 2.17 t established that there was a marked decrease in the volume of the axial shrinkage porosity. Batches of experimental forgings produced from such ingots were of satisfactory quality. However, the experimental results require additional investigation and more thorough verification of the expediency of using thin-walled molds under production conditions.

As ingot weight rises, there is an increase in the extent of the liquation and shrinkage defects. It is therefore necessary to attempt to use ingots with the minimum necessary weight. However, the general rule that it is preferable to select an ingot of minimum weight for a single forging is not always the optimum one. The point lies not only in the greater process economy that sometimes accompanies use of one ingot for several forgings and the better organization of the operations involved in casting the molten steel, but also in the characteristics of the forging process, which affect the efficiency with which the metal can be machined and hence the quality of the forgings. An example is one of the technological processes for forging gas-turbine wheels from ЭИ802 steel, in which it was found necessary to subject the ingot to preliminary drawing (before intermediate upsetting), in order to weld up axial defects. It was found that this was feasible only when the ingot was twice the necessary weight, i.e., could be used for two forgings. In some cases, it is best to select an ingot of greater weight than necessary for the final forging because use of large reductions in area is undesirable. This situation occurs, for example, in forging certain components of the ring type from poorly deformable steels, where substantial upsetting of the ingot or blank before rolling is impermissible for reasons of component quality. In this case, use of an ingot of greater weight (equivalent to 2-3 components) permits a sharp reduction in the upsetting of each blank (as a result of the increased cross-sectional area) and improves steel deformability with no detrimental effect on component quality. Finally, when the central portion of the ingot is to be removed, use of an ingot equivalent to 2-3 forgings and having a favorable configuration is sometimes expedient on the basis of technical-economic considerations.

The general rule that an ingot of minimum weight should be used must thus not be regarded as excluding other solutions in individual cases, especially with respect to ingots weighing up to 4-5 t.

Chapter 3

THERMOMECHANICAL CHARACTERISTICS OF THE FORGING OF HIGH-ALLOY STEELS

7. CHANGE IN METAL STRUCTURE AS A FUNCTION OF DEFORMATION TEMPERATURE AND RATE

High-alloy steels of the austenite group are distinguished by the most important features with respect to the thermomechanical factors involved in forging, since they are known to undergo phase transformations. The requisite metal structure can be achieved in these fields only through the deformation conditions, as there is no heat treatment that can reduce the grain size produced during forging. The quenching and subsequent aging generally employed for austenite steels can, at best, merely maintain the grain size created during forging. The opposite situation is often observed, however, in which heat treatment leads to an increase in grain size and there is no way to modify it by subjecting the metal to any thermal process.

The mechanical properties and hot strength of steel depend on its grain size. The mechanical properties are reduced as grain size increases, while long-term strength is slightly raised if the grain structure is relatively uniform. The principal requirement imposed on the structure of high-hot-strength austenite steels is thus homogeneity of grain structure. In most cases, forgings of high-hot-strength steels should have a fine- or medium-grained structure in order for them to have the required characteristic.

The grain size of a deformed metal depends on the recrystallization process, whose course is determined by the degree of deformation and the deformation temperature. When a metal is forged at critical degrees of deformation, its structure after recrystallization is coarse-grained. The final operations of forging should therefore provide a reduction regime for each press or hammer stroke that ensures production of normal grain size. The degrees of deformation recommended for forging are based on the recrystallization diagrams that have been worked out for specific types of steel.

The critical degree of deformation for most high-hot-strength austenite steels is 5-12%. Critical deformation occurs at all forging temperatures, but the grain size undergoes a particularly large increase at high temperatures. As a result, forging cannot terminate in the high-temperature region when the degree of deformation is small (close to critical) in cases where the forgings must have a fine-grained structure. However, use of very low final forging temperatures also produces an unsatisfactory structure: the grain size varies as a result of the nonuniform deformation conditions produced by incomplete recrystallization.

One condition for production of a uniform grain size throughout a forging is that there should not be a composite deformation mechanism, i.e., recrystallization should not terminate during forging. Studies of the influence of deformation temperature and degree of deformation on the recrystallization of ЭИ572 austenite steel made at the[NPL] (НЗЛ) and the Central Scientific Research Institute of Technology and Machine Building showed that recrystallization does not occur with any reduction in area when the final forging temperature is below 800°C. Working recrystallization begins after forging at 900°C or above. Formation of recrystallized grains at this temperature requires deformation by 80-85%, but working recrystallization is not complete even at this degree of deformation. The deformation at which working recrystallization begins decreases as the forging temperature is raised, amounting to about 50% at 950°C and 40% at 1000°C; recrystallization is complete throughout the entire forging volume at 1100-1150°C, regardless of the degree of deformation. When working recrystallization is not complete (at 1000°C or below), the structure of ЭИ572 steel contains grains of greatly differing size. It was established that selective recrystallization begins at 1050-1100°C but is very weak, having only a slight effect on grain growth even during prolonged holding. In practice, marked grain growth resulting from selective recrystallization occurs only after holding at a temperature of about 1150°C for 2-3 h. For other austenite steels, the specific relationship between the recrystallization conditions on one hand and the deformation temperature and degree of deformation on the other depends on the characteristics of the steel, but the general pattern is the same.

The lower temperature limit for forging of austenite steels is thus dictated both by the plasticity of the metal and by grain-uniformity requirements.

Aggregates of enlarged grains are found in the most diverse areas of austenite-steel forgings subjected to production quality-control checks. In wheels, for example, a coarse-grained structure is most frequently observed in the regions adjoining the faces, but it is also often encountered in the inner layers. The presence of large, nonuniform grains near the contact surfaces is due to nonuniform deformation and cooling of the metal in these areas: the deformation temperature is often below the recrystallization-initiation temperature, so that recrystallization takes place only during subsequent heat treatment and large grains are formed. The coarse-grained structure in the deep regions of a forging, where the metal temperature is relatively high during the forging process, is apparently formed as a result of prolonged holding in the high-temperature region and of selective recrystallization. There is no heterogeneity of grain structure when working recrystallization throughout the entire volume of the component being deformed takes place (and is completed) at a relatively low temperature.

The metallurgical nature of high-alloy steels also acts to produce nonuniform recrystallization [18]. A highly contaminated steel has a broader range of critical deformation and recrystallization usually involves formation of large crystals in local areas of te forging. Production of a uniform metal structure can therefore be facilitated by using high-purity steels with a minimum content of non-metallic inclusions and other impurities.

The influence of deformation rate on working recrystallization is of great practical interest. Research has established [17, 18, 19] that there is a substantial difference in the kinetics of steel recrystallization at different deformation rates, especially in the region close to the lower limit of the forging-temperature range. Softening processes occur with time, so that recrystallization is more complete at low deformation rates. In this case, the critical-deformation range is generally smaller, recrystallization begins at smaller deformations, and the deformation mechanism is to a larger extent purely a hot one. Hence it follows that, with identical temperature conditions, hammer forging presents a greater danger of incomplete recrystallization and a mixed deformation mechanism than pressed forging. In view of this pattern, different temperature-time regimes have been established for pressed and hammer deformation. For example, it is recommended that the same section of a forging not be deformed discontinuously in hammer-forging certain high-alloy steels at relatively low temperatures in order to avoid suppression of recrystallization processes.

The influence of deformation rate on the recrystallization conditions for austenite steels is greatly reduced when the forging temperature is raised; the recrystallization rate is relatively high and the steel is not hardened at all during hot working, even under dynamic-deformation conditions. In practice, hammer-forging of austenite steels is no less successful than press-forging at elevated temperatures. The substantially more uniform temperature throughout the blank cross-section, which is maintained during deformation as a result of the dynamic action of the hammer, also helps make this possible.

The grain size and uniformity in a finished forging are dictated principally by the thermomechanical deformation factors acting during the final operation.

In addition to the deformation temperature and degree of deformation, the final grain size is to some extent affected by the initial state of the metal [20], but this effect is most noticeable at small degrees of deformation. The initial structure ceases to have any effect on the final structure as the degree of deformation is increased: a nonuniform or large-grained structure can be almost completely eliminated by deformation during the final operation and subsequent recrystallization.

Heating a forging from the final deformation temperature to the temperature at which extensive recrystallization takes place causes more uniform structural changes throughout the forging volume, which promotes a decrease in grain nonuniformity [21].

Use of ultrasonic defectoscopy before heat treatment is frequently limited by the structural heterogeneity of forgings, since sound transmission is hampered when areas with a coarse-grained structure are present, the latter scattering the ultrasonic waves and creating false defect pulses. Reliable ultrasonic quality control can be carried out only when there is no intereference, in the form of large unrecrystallized grains.

The ultrasound transmission of coarse-grained areas resulting from incomplete recrystallization during forging can be greatly improved by heating the forgings to a temperature somewhat above the recrystallization-initiation point and then cooling them. The same effect can be achieved by quenching (austenization), provided that the heating temperature and the holding time at this temperature do not cause selective recrystallization.

Poor ultrasound transmission in forgings that have not been heat-treated is characteristic both of austenite steels and of high-hot-strength steels of other classes, such as the high-chromium, hardened martensite steels ЭИ802 and 15Х11МФБ. There have been many cases in practice where complete ultrasonic quality control of forged disks, rotors, and other components was impossible before heat treatment, because of the interference set up by regions of uncrystallized grains. Ultrasonic testing before heat treatment has sometimes led to incorrect conclusions regarding the presence of internal defects in forgings and caused improper appraisal of metal quality. The poor ultrasound transmission of such forgings is due to the temperature conditions under which forging is conducted: as a result of the high recrystallization temperature of high-hot-strength steels, final deformation at reduced temperatures leads to incomplete recrystallization and local areas with a nontransmitting coarse-grained structure developed in the forging.

Recrystallization of the metal is completed after heat treatment and ultrasound is transmitted without interference (provided that the forging contains no areas with an insufficiently deformed cast structure). For technological reasons, however, ultrasonic defectoscopy of forgings should often be carried out before heat treatment. This is true, for example, of seamless forged rotors, wheels heat-treated after cutting, or components of the T-joint and piping type, whose irregular shape does not permit ultrasonic testing after heat treatment. As production experience has shown, introduction of a special thermal operation, i.e., holding at a temperature above the recrystallization temperature, into the overall cycle of primary heat treatment and cooling improves ultrasound transmission in such cases and permits quite reliable ultrasonic quality control of forgings before final heat treatment.

8. FORGING-TEMPERATURE RANGES AND INGOT HEATING REGIMES

The optimum forging-temperature range is determined by analysis of experimental plasticity diagrams, which show the variation in the mechanical or technological properties of a given steel as a function of deformation temperature. Each type of steel, which is distinguished by its own chemical composition and metallurgical and structural characteristics, has a most favorable forging-temperature range, within which the metal can be efficiently deformed and the requisite forging quality obtained. The basic characteristic of a steel with respect to high-temperature plasticity are shown by appropriate plasticity diagrams but, in establishing forging-temperature regimes, particularly the maximum heating temperature, it is necessary to take into account possible deviations from the predetermined temperatures under furnace-operation conditions, e.g., during brief overheating. It is best for technical working specifications to stipulate heating temperatures somewhat (20–30°C) below the maximum established from the plasticity diagram. This is especially important for high-alloy steels containing low-melting elements. For example, overheating ЭИ726 steel by 10–15°C causes the metal to fail during forging, since it contains a low-melting boride eutectic. Moreover, the plastic properties of steel from different melts (even of the same type) may vary in accordance with the metallurgical characteristic of the production process, a factor not taken into account by plasticity diagrams.

The heating temperature for high-alloy steels should, as far as possible, ensure complete solution of the carbide and intermetallic compounds in the solid solution. However, the efficiency of solution depends both on the temperature and on the properties of the phases present in the steel. Some carbides (e.g., niobium carbides) are very stable at forging temperatures. Complete solution of these compounds requires heating at a temperature close to the solidus point, which is obviously impermissible because it weakens the intercrystalline bond. Excessive heating of austenite steels is also undesirable because it produces a coarser grain structure as a result of active collective recrystallization, although this factor is not decisive if the metal is subsequently subjected to effective deformation. The purity of the steel to be deformed is therefore of some significance in selecting the heating temperature. The presence of lead, antimony, tin, and other low-melting elements or of nonmetallic inclusions ... the grain boundaries reduces to capacity of a steel for hot deformation. The detrimental effect of impurities on steel plasticity can reduced by lowering the temperature at which forging is begun.

In some cases, a heating temperature quite satisfactory from the standpoint of metal plasticity must be adjusted in the light of special requirements imposed on the phase composition of a given type of steel. This is particularly true of X18H9T steel, whose α-phase content is stipulated. It has been established [22] that the α-phase content of this steel is increased by high-temperature (up to 1250°C) preforging heating. As a result, high-temperature heating is not recommended if the initial melt has a high α-phase content. The plasticity of a steel with a high ferrite-phase content is greatly reduced during forging, which causes discontinuities and cracks to develop in the blank.

The technological plasticity of an ingot is always lower than that of a deformed blank. Chemical and physical inhomogeneities in the metal are more severe in the cast state: these include zone and dendritic liquation, local aggregates of carbides and nonmetallic inclusions, and areas of noncompact structure. Deformation compacts the metal, breaks up and shifts the grains, and causes the brittle constituents to be more uniformly distributed throughout the deformed blank. All this promotes an overall increase in metal plasticity, so that initial heating of a deformed blank can be carried out at a higher temperature than that of an ingot for certain high-alloy steels. This relationship between technological properties and optimum hot-deformation temperature on one hand and the initial state of the metal on the other is illustrated by the plasticity diagram for ЭИ395 steel shown in Fig. 1. The quantity $(h_0 - h)/h_0$ corresponds to the

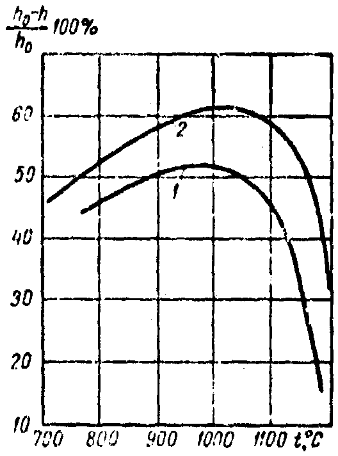


Fig. 1. Plasticity diagram of ЭИ395 steel [17]. 1) Cast steel; 2) deformed steel.

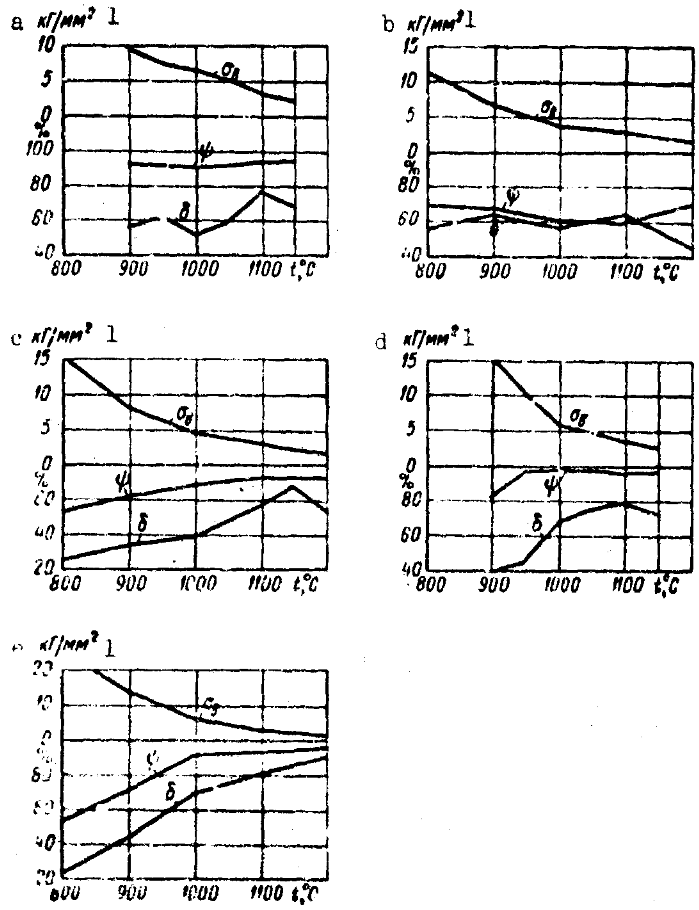


Fig. 2. Graph representing variation in mechanical properties of certain high-alloy steels as a function of forging temperature t. a) 15X11MФБ steel; b) X18H9 steel [22]; c) X18H9 steel [22]; d) X18H12M2T steel; e)X18H22B2T2 steel. 1) kgf/mm^2.

degree of upsetting deformation before the first crack appears.

The large complex of factors affecting selection of the initial heating temperature for a steel thus confirms the need for ferro verification of the optimum temperature established on the basis of plasticity diagrams under actual production conditions, using several typical melts and taking into account the initial state of the metal and the special requirements imposed on finished-forging quality.

There are different methods for evaluating metal plasticity at different temperatures in order to construct appropriate diagrams. The most widely employed of these are upset testing and tensile testing at forging temperatures.

Construction of diagrams from the results obtained in upsetting specimens (Fig. 1) corresponds most closely to free-forging conditions. However, it does not give consideration to all the actual production factors involved in deformation. For example, cracking during upsetting of blanks under shop conditions is often due not to reduced metal plasticity but to the presence of surface defects, which are exposed during deformation or to additional tensile stresses produced in the lateral processes by greater cooling of the ingot butt than under laboratory conditions. This factor must be taken into account in using plasticity diagrams constructed from upsetting results.

Tensile testing and subsequent construction of plasticity diagrams from the relative elongation and reduction in cross-sectional area of specimens at high temperatures does not provide an absolute characterization of metal plasticity applicable to production deformation conditions, since the stress patterns in the specimen and production blank usually are quite different from one another. However, such diagrams, which characterize relative metal plasticity, make it possible to compare the behavior of different types of steel at different temperatures and thus to establish quite well-justified forging-temperature ranges.

Figure 2 shows graphs representing the variation in the mechanical properties of certain high-alloy steels as a function of test temperature .

Type X18H9 steel has high plasticity at temperatures of 900-1250°C. Steel of this type but with titanium added (18 9) has lower plasticity over the forging-temperature range, especially when its α-phase content is high. In the latter case, the steel has satisfactory plasticity at 900-1200°C. This is also true of the same type of steel with a reduced carbon content (0X18H9T) and a relatively high α-phase content. The best deformability is achieved for 1X13 and 1X17H2 steels by heating them to 1250-1270°C.

The final forging temperature for high-alloy steels is established from their plasticity and recrystallization diagrams, proceeding from the basic premise that a reduced forging temperature should not lead to a sharp drop in plasticity, high deformation resistance, or formation of a heterogeneous structure with different grain sizes. Undesirable phase transformations occur in certain austenite steels when they are forged at low temperatures, reducing the service characteristics of the forgings. This is particularly true of ЭИ572 steel, which tends toward rapid σ-phase precipitation over the temperature range 930-650°C. Despite the adequate metal plasticity, the final forging temperature is set above the temperature at which intensive σ-phase formation occurs, i.e., 930°C, in such cases. It should be noted that, as a result, forgings of ЭИ572 steel are subjected to accelerated cooling in water or at least in moving air. Holding of the blanks at temperatures of 650-930°C is undesirable. Table 3 shows the forging-temperature ranges for certain high-alloy steels.

In order to avoid development of substantial thermal stresses or thermal cracks, high-alloy steels should be heated solely to 600-700°C, since they have low thermal conductivity. Further heating to the forging temperature differs little from that for ordinary alloy steels, since the two types of steel have almost the same thermal conductivity over this temperature range. The temperature zone requiring slow heating is neglected in practice for hot upsetting of ingots.

The most important element of the preforging heating regime for high-alloy steels, which affects deformability and metal quality, is holding at the forging temperature. As for ordinary steels, the holding time is governed by the period necessary for the entire ingot volume to be heated to the requisite temperature. The optimum holding time at a given temperature is also governed by the kinetics of the phase transformations.

TABLE 3

Forging-Temperature Ranges for Certain High-Alloy Steels

1 Марка стали	2 Максимальная температура нагрева металла, °C	3 Температура конца ковки (не ниже), °C
1X13	1260	800
2X13	1220	800
1X17H2 (ЭИ268)	1250	800
15X12BMФ (ЭИ802)	1220	880
X18H9	1250	850
X18H9T	1190	850
ЭИ572	1200	960
ЭИ612, ЭИ405	1180	900
ЭИ726	1160	900

1) Type of steel; 2) maximum heating temperature, °C; 3) final forging temperature (not below), °C.

The monophasic state is known to be most favorable for deformation of a metal. Homogeneous structures have a higher plasticity at forging temperatures, as a result of the absence of hardening phases and the more uniform deformation of individual grains. When ingots of alloy structural steels are heated, the individual structural components usually have time to go into solid solution during temperature equilibration over the ingot cross-section and no additional holding is generally required to convert the heterogeneous structure to a homogeneous one.

The conditions obtaining for constituent-solution kinetics in high-alloy steels heated to forging temperatures are more complex. The diffusion processes associated with solution of the excess phases and structural homogenization proceed slowly and require a longer time. High-temperature holding of such steels should therefore be of a duration sufficient to provide both complete heating of the ingot cross-section and maximum solution of the excess structural components.

When ingots are held at the forging temperature for the appropriate time, the hardening phases, in the form of carbides and intermetallic compounds, are partially dissolved and partially coagulated, being dispersed throughout the structure rather than forming a compact chain along the grain boundaries. This arrangement of the low-plasticity phases is more favorable with respect to plasticity and deformability. However, ingots and blanks should not be held at high temperatures for extremely periods, since certain negative factors also act in this case; in particular, the crystal structure of blanks becomes coarser as a result of collective recrystallization. This factor is not of great importance for ingots, since, as a result of the characteristics of their cast structure, grain growth does not occur even when they are heated to very high temperatures, or for deformed blanks, provided that heating is followed by a high degree of forging. However, the structure and properties of forgings deteriorate if deformation after prolonged high-temperature holding is slight. Moreover, prolonged holding reduces the plasticity of steel containing large amounts of oxide and sulfide impurities [17].

With respect to initial heating of ingots and blanks of special high-hot-strength steels, where subsequent deformation is relatively large and the sulfide and oxide content of the metal is small, prolonged holding at forging temperatures not only increases deformability but improves the structure and mechanical properties of the metal in the finished articles. This has been verified at the NPL with a large number of ingots of ЭИ572 steel from different melts, which were used for forged gas-turbine wheels.

Long-term holding at the forging temperature has also been found to have a favorable effect on the deformation of other high-hot-strength alloys [23]. The technical specifications for heating regimes for austenite-steel ingots should therefore provide the minimum holding time at the forging temperature that will give an excess of 2-3 h over the time required to heat the entire ingot cross-section. There is no danger in more prolonged holding (up to 6-8 h) but it is undesirable because of the large amount of scale formed on the metal and for technical-economic reasons: the increased furnace load and fuel consumption.

9. DEFORMATION RESISTANCE OF CERTAIN SPECIAL STEELS

High-alloy high-hot-strength steels have a substantially higher deformation resistance than perlite structural steels; it usually increases with their alloying-element content. Austenite steels are particularly noteworthy in this respect, their deformation resistance being several times that of ordinary alloy steels. This property of austenite steels is related to the characteristics of the deformation mechanism in the forging-temperature region: hardening of the metal during forging begins at relatively high temperatures (900-1000°C), while ordinary steels undergo slight hardening at 800-850°C. From the standpoint both of production of a homogeneous crystal structure and of deformation resistance, which affect the efficiency and productivity of the forging equipment, the final deformation temperature for austenite steels should be considerably higher than that for structural steels, while the forging-temperature range should be narrower. The power of the forging equipment at the lower limit of the permissible temperature range is far higher for deformation of austenite steels.

TABLE 4

Ultimate Strength of Certain Special Steels at Forging Temperatures

1 Марка стали	2 Предел прочности $\sigma_в$ (кГ/мм²) при температуре 800° C	900° C	1000° C	1100° C	1150° C	1200° C	3 Скорость деформации, мм/мин
ЭИ415	—	8,0	4,2	2,3	1,9	—	4
2X13	7,5	—	6,3	3,7	—	—	16
1X17H2	—	7,9	5,7	2,4	1,9	1,3	4
15X11MФБ	—	9,6	6,6	3,4	2,4	—	4
X18H9	12,2	6,9	3,9	3,1	—	1,6	1,1
X18H9T	18,5	9,1	5,5	3,8	2,9	1,8	1,1
X23H18	14,1	9,2	5,5	5,2	—	2,9	1,1
X18H12M2T	—	16,7	6,5	3,5	2,6	—	4
ЭИ572	32,0	22,0	12,7	8,2	—	5,0	16
ЭИ405	—	—	—	8,2	—	5,0	16
ЭИ612	—		9,4	—	—	3,8	16
X18H22B2T2	27,6	14,6	6,1	3,2	2,3	1,5	1,1

1) Type of steel; 2) ultimate strength σ_b (kgf/mm²) at temperature of; 3) deformation rate, mm/min.

In addition to temperature, the deformation resistance is materially affected by the rate and degree of deformation. Experimental investigation of a broad range of high-alloy steels [19, 22, 24] has established that the deformation resistance of metals increases with the deformation rate; this pattern is maintained at all temperatures and degrees of deformation. Press forging is therefore more favorable in this respect than hammer forging, in which the deformation resistance of austenite steels is greater by a factor of 2-4. The higher dynamic-deformation resistance of the metal is due to the fact that softening processes are less complete as a result of the slow recrystallization rate of high-alloy steels subjected to hot plastic deformation.

It must be kept in mind, however, that there is not always a noticeable manifestation of this phenomenon when press and hammer forging of a metal are compared, since a large amount of heat is evolved when a blank is deformed under the dynamic action of a hammer and there is a substantial rise in metal temperature.

When identical forgings are produced with a press and a hammer of equal power, the total work consumed may be less for hammer forging as a result of the temperature factor, despite the higher deformation resistance.

Like an increase in deformation rate, a rise in degree of deformation increases deformation resistance; the effect of this factor becomes stronger as the alloying of the steel increases and is greatest in the lower portion of the forging-temperature range. A steel undergoes hardening during both dynamic and static deformation as the degree of deformation is increased. The general mechanism of this phenomenon is related to the combination of two simultaneous processes, hardening and recrystallization, which govern the deformation mechanism and the extent to which deformation is suppressed by softening processes.

In testing specimens at small deformation rates, deformation resistance is determined from the yield strength $\sigma_{0.2}$. Ultimate strength σ_b is often used in place of $\sigma_{0.2}$ for calculations, introducing the correction factor 0.85-0.90, i.e., $\sigma_{0.2} = 0.85\text{-}0.90\sigma_b$. Table 4 gives the values of σ_b for certain special steels at forging temperatures, taken from the data of the NPL and the Central Scientific Research Institute of Technology and Machine Building, as well as from the literature [22, 25]. A uniform deformation rate is an indispensable condition for comparability of such data at high temperatures.

The deformation resistance of a metal σ_{st} during hydraulic-press forging is assumed to equal $\sigma_{0.2}$, which is determined in low-speed test machines for the steel in question. In hammer forging, the approximate value of σ_{din} is determined by taking into account the rate factor m: $\sigma_{din} = m\sigma_{st}$, where m can be assumed to equal 4 (according to S.I. Gubkin).

Among the additional factors affecting deformation resistance σ are the type of stress pattern during forging and the scale factor. All other conditions being equal, the value of σ increases with the hydrostatic pressure [26]. The scale factor is operative in the fact that the actual value of σ decreases as the deformed blank becomes larger. This phenomenon is due to the different thermal conditions obtaining during deformation of a laboratory specimen (in determining the main deformation-resistance constant $\sigma_{0.2}$ or σ_b) and a production blank, since the surface-to-volume ratios differ. The amount of heat removed from the metal is greater and the average deformation temperature is lower for deformation of a small specimen, which results in an increase in $\sigma_{0.2}$ and σ_b. In fabricating large commercial forgings, the deformation resistance of the metal is therefore determined with a correction factor for the forging scale. This factor ranges from 0.70 to 0.55 for blanks 100-1000 mm in diameter [27].

10. INFLUENCE OF REDUCTION IN AREA ON STRUCTURE AND MECHANICAL PROPERTIES OF AUSTENITE-STEEL FORGINGS

The reduction in area is one of the most important factors affecting the structure and mechanical properties of forgings; it also influences the reliability of ultrasonic quality control of austenite steels (in connection with the characteristics of their recrystallization), since the structural heterogeneity due to inadequate reduction

TABLE 5

Technological Regimes for Forging Experimental Wheels of ЭИ572 Steel

1 Схема ковки	2 № диска	3 Фактические размеры заготовки под осадку, мм		6 Размеры поковки диска, мм		8 Степень укова	
		4 Диаметр	5 Длина	4 Диаметр	7 Высота	9 при протяжке	10 при осадке
	1*		210	—	—	1,1	—
11 Биллетирование по низу корпуса слитка. Рубка заготовок. Осадка на окончательный размер	2		270	550		1,1	1,8
	3	400	300	570		1,2	2,0
	4		430	680	150	1,3	2,8
	5		450	700		1,15	3,0
	6		675	850		1,15	4,5
12 Промежуточная осадка сбиллетированного слитка. Протяжка. Рубка заготовок. Осадка на окончательный размер	7	440	300	620	150	2	2

*The blanks were not upset.

1) Forging process; 2) disk No.; 3) actual size of blank before upsetting, mm; 4) diameter; 5) length; 6) size of forged wheel, mm; 7) height; 8) reduction in area; 9) during drawing; 10) during upsetting; 11) billeting along bottom of ingot container. Trimming of blanks. Upsetting to final size; 12) intermediate upsetting of billeted ingot. Drawing. Trimming of blanks. Upsetting to final size.

NOT REPROD

is the principal obstacle to complete ultrasound transmission.

The influence of reduction in area on the structure, mechanical properties, and ultrasound transmission of forged wheels of ЭИ572 steel was investigated at the NPL.

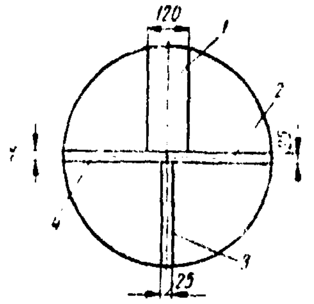


Fig. 3. Pattern for cutting sections from experimental wheels. 1) For tangential specimens; 2) for axial specimens; 3) for radial specimens; 4) for macroscopic examination.

GRAPHIC NOT REPRODUCIBLE

a

b

RODUCIBLE

c

d

Fig. 4. Macrostructure of radial sections cut from wheels. a) No. 1; b) No. 2; c) No. 6; d) No. 7.

The initial material consisted of elongated ingots weighing 2.17 t (see Table 2). The technological regimes for the experiments (Table 5) provided for forging of wheels with reductions in area of from 1.8 to 4.5, with and without intermediate upsetting. In the latter case, the reduction in area during drawing was due solely to billeting of the ingots and ranged from 1.1 to 1.3, depending on the location of the blank. One of the blanks was not upset but underwent all the subsequent machining and test stages together with the forged wheels. All the experimental forgings were fabricated from steel of the same melt (containing 0.30% C, 0.55% Si, 1.2% Mn, 18.8% Cr, 10.3% Ni, 1.36% Mo, 1.29% W, 0.33% Nb, 0.31% Ti, 0.01% S, and 0.023% P), which was smelted in a 10-ton arc furnace with a basic hearth.

The ingots were supplied hot to the press shop, with a surface temperature of 640-720°C, and subjected to preforging heating under a hot-upsetting regime. All the forging operations were carried out in a press exerting a force of 2000 t, drawing was conducted in composite die blocks (with the upper block flat and the lower block cutout), and upsetting was carried out in a broad, flat die block with successive deformation of individual sections of the ingot butt. Forging was carried out over the temperature range 1180-950°C, with a final degree of deformation of no less than 30%. The forced disks were cooled in water.

The forgings were then trimmed and subjected to heat treatment (quenching from 1150°C and aging at 760°C), ultrasonic defectoscopy, and thorough investigation of their macrostructure and mechanical properties.

Ultrasonic testing of the forgings was carried out at a frequency of 2.5 MHz, with the defectoscope calibrated from a standard having an aperture 3 mm in diameter. Complete ultrasound transmission was observed in wheels Nos. 6 and 7. No defects were found in these wheels. The other forgings, which were produced without intermediate upsetting of the ingot to a reduction of 3.0 or less, exhibited no ultrasound transmission in their central zones, whose radii increased as the reduction decreased. Wheel No. 2 was ultrasound-transmissive only within the inner half of its diameter, while wheel No. 1 was nontransmissive throughout its entire volume.

All the wheels were cut in accordance with the pattern shown in Fig. 3 and the resultant sections were used for macroscopic examination and testing of mechanical properties in different directions.

The results of macroscopic examination of the sections showed that wheel No. 1 had an almost undeformed cast structure (Fig. 4a), while wheel No. 2, which was forged with a reduction of 1.8, was distinguished by substantial cast-structure areas in the central zones of inhibited deformation near the contact points (Fig. 4b). The volume occupied by the cast structure in the contact zones decreased as the degree of upsetting increased. Almost the entire metal volume had a deformed structure at a reduction of 4.5 (Fig. 4c) and only isolated weakly deformed grains were found in one of the contact areas. No traces of a cast structure were detected in wheel No. 7, which was produced with intermediate upsetting of the ingot (Fig. 4d).

Table 6 presents the results of mechanical tests performed on the experimental wheels in the tangential, radial, and axial directions. Wheels Nos. 1-6 differed principally in degree of upsetting. The reductions in area during drawing K_n were almost the same in this group of wheels and apparently had no material influence on the mechanical properties of the metal. The degree of upsetting K_o had a very marked effect on plasticity, which rose in the tangential and radial directions as the value of K_o increased. This pattern was especially pronounced when the mechanical properties of the wheels in the vicinity of the horizontal axial plane (vertical midline) of the forgings were compared. The plasticity properties δ and ψ in the tangential direction were more than double those in the initial metal at K_o = 1.8 and reached their absolute maximum (35-40%) at K_o = 3-4.5. There was a simultaneous rise in impact strength and only very slight changes in other strength characteristics.

It is interesting to note that the mechanical properties of the metal in this region of the wheels was almost completely equilibrated at reductions of 2.0 or more, a high degree of homogeneity being achieved in the central and peripheral areas. A similar situation was observed for the mechanical properties of the wheels in the radial direction when specimens were taken from the zone close to the horizontal axial plane. The absolute values of the mechanical properties observed in this zone in the radial and tangential directions were almost equal when the reduction was more than 2.0.

The mechanical properties of the wheels in the axial direction were distinguished by low values of δ and ψ, whose absolute magnitudes in the central region decreased as the degree of upsetting increased. Since the variation in plasticity characteristics in the axial and tangential directions in the central zone of the wheels as a function of K_o was reciprocal in character, the ratios $\delta_{osev}/$ $/\delta_{tang}$ and ψ_{osec}/ψ_{tang} decreased substantially as K_o increased (Fig.5).

The variation in the mechanical properties of the metal as the degree of upsetting increased had a quite different character in the central zone of the wheel faces. Here, in contrast to the middle zone (the region close to the hozizontal axial plane), the plasticity characteristics in the tangential and radial directions were markedly elevated only at K_o = 4.5. At smaller reductions, the indices δ and ψ were equivalent to those for the undeformed blank and were characterized by large scattering of values, a feature characteristic of the metal in the cast state.

The mechanical properties of the peripheral zone were rather high in the blank for wheel No. 1, which was not upset. Plasticity increased with K_o; the indices δ, ψ, and a_k were roughly the same in the face and central portions of the wheels. This phenomenon also occurred at relatively small degrees of upsetting.

TABLE 6

Mechanical Properties of Wheels Forged from ЭИ572 Steel with Different Reductions in Area

1 № диска	2 Коэффициент уковки при вытяжке	3 Степень осадки	4 Направление образца	5 Испытуемый участок по радиусу	6 Место отбора образцов по высоте диска	$\sigma_{0,2}$	$\sigma_в$	δ	ψ	8 $a_н$, кГ·м/см²: 9 1-й образец	10 2-й образец
						7 кГ/мм²		%			
1	2	3	4	5	6	7	8	9	10	11	12
1	1,1	—	1 1 Тангенциальное	1 2 Центр	1 5 1-й торец	39,5	63,7	16,4	22,5	4,8	4,7
					1 6 Середина	36,9	63,7	16,4	15,6	5,4	5,3
					1 7 2-й торец	36,9	63,7	18,2	19,7	4,8	4,5
				1 3 Средняя треть радиуса	1 5 1-й торец	39,5	66,2	15,0	13,6	4,1	4,1
					1 6 Середина	40,1	68,8	18,0	15,4	5,3	4,1
					1 7 2-й торец	36,9	63,7	15,2	17,2	4,2	3,8
				1 4 Периферия	1 5 1-й торец	38,2	66,2	20,0	18,7	5,6	6,1
					1 6 Середина	36,9	68,1	21,8	20,8	5,4	5,6
					1 7 2-й торец	35,6	68,1	23,4	26,0	5,8	6,4
			1 8 Радиальное	1 2 Центр	1 5 1-й торец	—	—	—	—	4,2	—
					1 6 Середина	36,1	63,7	16,0	22,0	—	—
					1 7 2-й торец	38,9	62,6	14,0	16,0	—	—
				1 3 Средняя треть радиуса	1 5 1-й торец	—	—	—	—	7,1	—
					1 6 Середина	35,3	66,5	21,6	22,0	—	—
					1 7 2-й торец	33,6	61,8	16,6	22,0	—	—
1	1,1	—	1 8 Радиальное	1 4 Периферия	1 5 1-й торец	—	—	—	—	7,7	—
					1 6 Середина	34,6	67,8	25,0	30,6	—	—
					1 7 2-й торец	37,5	71,8	25,0	30,6	—	—
			1 9 Осевое	1 2 Центр	1 6 Середина	36,3	53,5	7,8	7,8	4,1	4,4
				1 3 Средняя треть радиуса		36,3	56,7	11,2	7,8	4,8	4,1
				1 4 Периферия		38,5	61,4	14,0	12,2	5,3	5,0
2	1,1	1,8	1 1 Тангенциальное	1 2 Центр	1 5 1-й торец	39,5	67,5	18,8	17,2	3,8	3,9
					1 6 Середина	38,8	68,8	31,2	31,2	6,4	5,9
					1 7 2-й торец	35,6	62,4	18,4	18,7	2,7	3,1
				1 3 Средняя треть радиуса	1 5 1-й торец	39,5	69,4	22,6	23,5	4,7	5,6
					1 6 Середина	39,5	68,5	32,0	34,3	7,2	7,2
					1 7 2-й торец	39,5	64,0	21,2	24,8	4,2	5,3
				1 4 Периферия	1 5 1-й торец	40,7	71,9	28,6	27,8	7,5	5,6
					1 6 Середина	39,5	71,3	36,6	40,7	5,8	6,6
					1 7 2-й торец	40,7	71,9	23,0	27,8	6,6	5,9

TABLE 6 CONTINUED

№ диска	Коэфф. укова при вытяжке	Степень осадки	Направление образца	Испытуемый участок по радиусу	Место отбора образцов по высоте диска	$\sigma_{0,2}$	$\sigma_в$	δ	ψ	$a_н$, кГ·м/см² 1-й образец	$a_н$, кГ·м/см² 2-й образец
						кГ/мм²	кГ/мм²	%	%		
1	2	3	4	5	6	7	8	9	10	11	12
2	1,1	1,8	Радиальное	Центр	1-й торец					4,1	
					Середина	39,5	68,1	28,2	27,8	—	—
					2-й торец	38,8	63,7	23,6	22,5	—	—
				Средняя треть радиуса	1-й торец					6,2	—
					Середина	36,8	65,5	21,3	19,0	—	—
					2-й торец	38,2	72,2	31,0	24,9	—	—
				Периферия	1-й торец		—	—	—	8,6	—
					Середина	36,5	70,7	36,7	30,6	—	—
					2-й торец	40,3	75,3	21,4	43,5	—	—
			Осевое	Центр	Середина	37,5	53,5	7,6	8,6	2,8	2,4
				Средняя треть радиуса	Середина	39,5	54,7	8,2	10,6	2,7	3,0
				Периферия	Середина	42,0	61,1	10,0	10,6	3,8	3,2
3	1,3	2,0	Тангенциальное	Центр	1-й торец	38,2	64,9	13,0	17,2	2,7	3,1
					Середина	39,2	66,2	32,4	37,8	5,6	5,3
					2-й торец	38,2	61,1	14,4	18,0	4,1	3,9
			Тангенциальное	Средняя треть радиуса	1-й торец	40,1	63,7	23,8	22,5	4,1	4,1
					Середина	36,9	66,2	33,4	36,0	5,6	5,4
					2-й торец	41,4	71,9	32,8	36,0	5,6	5,0
				Периферия	1-й торец	40,7	68,8	19,6	26,0	5,1	5,4
					Середина	42,0	70,0	33,8	32,8	5,0	8,1
					2-й торец	42,0	71,9	34,8	40,6	6,7	6,1
3	1,3	2,0	Радиальное	Центр	1-й торец	—	—	—	—	6,9	—
					Середина	38,8	64,9	34,8	34,4	—	—
					2-й торец	36,9	57,3	11,4	15,4	—	—
				Средняя треть радиуса	1-й торец	—	—	—	—	7,5	—
					Середина	36,4	69,7	34,0	36,0	—	—
					2-й торец	36,1	67,7	32,6	33,0	—	
				Периферия	1-й торец		—	—		8,1	—
					Середина	35,0	69,0	36,0	38,6	—	—
					2-й торец	38,2	69,0	22,0	24,9		—
			Осевое	Центр	Середина	36,9	45,8	5,0	6,0	3,1	2,8

— 45 —

TABLE 6 CONTINUED

№ диска	Коэффициент уковки при вытяжке	Степень осадки	Направление образца	Испытуемый участок по радиусу	Место отбора образцов по высоте диска	$\sigma_{0,2}$, кГ/мм²	$\sigma_в$, кГ/мм²	δ, %	ψ, %	$a_н$, кГ·м/см², 1-й образец	$a_н$, кГ·м/см², 2-й образец
1	2	3	4	5	6	7	8	9	10	11	12
3	1,3	2,0	Осевое	Средняя треть радиуса	Середина	38,2	54,7	9,2	11,6	3,4	3,2
				Периферия		38,8	61,7	11,1	17,2	3,9	4,8
4	1,4	2,8	Тангенциальное	Центр	1-й торец	40,1	59,5	13,0	17,8	4,1	3,8
					Середина	37,5	67,5	37,0	57,6	8,2	7,9
					2-й торец	36,3	61,1	10,2	13,6	4,2	3,9
				Средняя треть радиуса	1-й торец	38,0	62,2	16,2	21,5	6,2	6,7
					Середина	36,9	64,3	34,8	37,6	8,7	8,4
					2-й торец	36,3	65,6	31,4	35,9	7,1	6,6
				Периферия	1-й торец	38,7	69,8	36,8	45,2	8,4	8,2
					Середина	39,3	66,3	34,0	32,3	7,5	7,1
					2-й торец	38,2	70,0	27,4	27,8	8,2	8,7
			Радиальное	Центр	1-й торец	—	—	—	—	8,6	—
					Середина	38,2	67,5	33,0	40,6	—	—
					2-й торец	36,2	61,7	19,0	22,5	—	—
				Средняя треть радиуса	1-й торец	—	—	—	—	6,0	—
					Середина	38,2	68,8	35,0	40,6	—	—
					2-й торец	36,9	66,8	31,6	32,8	—	—
				Периферия	1-й торец	—	—	—	—	5,8	—
					Середина	38,2	68,8	34,8	36,0	—	—
					2-й торец	38,2	[illegible]	35,4	43,7	—	—
4	1,4	2,8	Осевое	Центр	Середина	36,8	42,5	3,7	7,2	3,4	2,8
				Средняя треть радиуса		38,7	52,3	6,2	9,8	3,1	3,5
				Периферия		40,1	62,8	12,4	16,6	3,6	4,2
5	1,2	2,0	Тангенциальное	Центр	1-й торец	38,8	62,4	16,0	[illegible]	3,9	3,9
					Середина	38,8	67,5	39,6	40,6	6,6	8,1
					2-й торец	38,2	50,9	9,6	7,8	4,1	5,8
				Средняя треть радиуса	1-й торец	38,8	67,5	36,2	31,2	8,1	7,5
					Середина	39,5	67,5	35,2	32,8	6,9	6,6
					2-й торец	38,8	64,9	19,8	20,8	5,0	4,4

TABLE 6 CONTINUED

1 № диска	2 Коэф. уков[illegible]	3 Степень осадки	4 Направление образца	5 Испытуемый участок по радиусу	6 Место отбора образцов по высоте диска	$\sigma_{0,2}$	$\sigma_в$	δ	ψ	8 $a_н$, кГ·м/см² 9 1-й образец	2-й образец 10
						7 кГ/мм²		%			
1	2	3	4	5	6	7	8	9	10	11	12
5	1,2	3,0	1 1 Тангенциальное	1 4 Периферия	1 5 1-й торец	39,5	71,9	35,6	37,6	8,1	7,5
					1 6 Середина	39,5	66,8	35,0	37,6	6,9	6,6
					1 7 2-й торец	39,5	71,3	33,2	32,8	5,0	4,4
			1 8 Радиальное	1 2 Центр	1 5 1-й торец					3,5	—
					1 6 Середина	36,3	64,7	25,2	24,7	—	—
					1 7 2-й торец	36,9	60,5	13,2	13,6	—	—
				1 3 Средняя треть радиуса	1 5 1-й торец			—		5,0	—
					1 6 Середина	38,2	66,8	34,0	37,6	—	—
					1 7 2-й торец	40,1	69,4	29,0	37,6	—	—
				1 4 Периферия	1 5 1-й торец	—	—	—	—	7,1	—
					1 6 Середина	37,1	72,2	35,0	36,0	—	—
					1 7 2-й торец	40,3	76,0	42,3	43,5	—	—
			1 9 Осевое	1 2 Центр		36,9	44,5	3,4	6,0	3,2	2,6
				1 3 Средняя треть радиуса	1 6 Середина	39,5	53,5	5,6	13,6	3,9	3,8
				1 4 Периферия		43,9	71,3	23,6	22,5	4,1	4,1
6	1,2	4,5	1 1 Тангенциальное	1 2 Центр	1 5 1-й торец	38,8	68,1	26,4	22,5	4,7	5,9
					1 6 Середина	38,8	70,7	35,8	31,2	9,7	8,9
					1 7 2-й торец	36,9	66,8	20,0	17,2	4,1	4,2
				1 3 Средняя треть радиуса	1 5 1-й торец	38,2	70,0	36,0	37,6	8,4	7,5
					1 6 Середина	38,2	70,7	38,0	43,7	9,9	8,1
					1 7 2-й торец	39,5	73,8	38,4	42,2	7,9	7,2
				1 4 Периферия	1 5 1-й торец	40,7	73,2	36,6	46,7	7,5	8,7
					1 6 Середина	36,9	68,8	38,6	46,7	8,9	9,9
					1 7 2-й торец	38,2	73,2	38,0	48,2	8,2	8,7
			1 8 Радиальное	1 2 Центр	1 5 1-й торец	—	—			6,7	5,1
					1 6 Середина	36,3	68,8	38,0	46,7	—	—
					1 7 2-й торец	33,7	60,5	23,4	21,6	—	—
				1 3 Средняя треть радиуса	1 5 1-й торец	—	—	—	—	8,6	8,9
					1 6 Середина	36,9	66,2	36,6	43,7	—	—
					1 7 2-й торец	34,4	67,5	33,0	45,2	—	—

TABLE 6 CONTINUED

1 № диска	2 Коэффициент уковы при вытяжке	3 Степень осадки	4 Направление образца	5 Испытуемый участок по радиусу	6 Место отбора образцов по высоте диска	$\sigma_{п.ц}$	$\sigma_в$	δ	ψ	8 $a_н$, кГ·м/см²	
						7 кГ/мм²		%		9 1-й образец	2-й [10] образец
1	2	3	4	5	6	7	8	9	10	11	12
6	1,2	4,5	1 8 Радиальное	1 4 Периферия	1 5 1-й торец 1 6 Середина 1 7 2-й торец	— 38,0 38,2	— 67,5 64,9	— 32,4 32,0	— 37,6 30,6	7,2 — ·	8,9 ·
			1 9 Осевое	1 2 Центр	1 6 Середина	37,5	40,7	1,2	4,3	3,1	3,8
				1 3 Средняя треть радиуса		36,9	45,8	4,0	6,0	3,8	3,6
				Периферия		40,7	71,3	22,8	20,8	5,0	4,5
7*	2,0	2,0	1 1 Тангенциальное	1 2 Центр	1 5 1-й торец 1 6 Середина 1 7 2-й торец	40,7 36,9 38,8	69,1 70,0 70,0	21,4 36,2 20,8	24,7 37,6 26,0	8,7 8,9 5,6	7,2 8,1 6,6
				1 3 Средняя треть радиуса	1 5 1-й торец 1 6 Середина 1 7 2-й торец	39,5 40,7 40,1	72,6 71,3 72,6	32,0 29,0 28,0	37,6 33,5 26,0	8,9 8,7 7,2	8,1 8,9 8,1
				1 4 Периферия	1 5 1-й торец 1 6 Середина 1 7 2-й торец	42,0 41,4 41,4	74,5 73,8 75,4	33,7 33,8 37,0	40,6 36,0 40,6	7,5 7,4 7,5	7,2 7,5 7,2
7*	2,0	2,0	1 8 Радиальное	1 2 Центр	1 5 1-й торец 1 6 Середина 1 7 2-й торец	— 38,5 38,2	— 68,3 70,0	— 27,0 28,4	— 26,5 22,5	6,7 — —	— — —
				1 3 Средняя треть радиуса	1 5 1-й торец 1 6 Середина 1 7 2-й торец	— 35,7 37,5	— 72,0 72,3	— 13,1 25,5	— 38,6 24,9	7,7 —	— — —
				1 4 Периферия	1 5 1-й торец 1 6 Середина 1 7 2-й торец	— 33,9 39,6	— 71,3 78,8	— 38,0 40,0	— 38,6 43,5	8,7 — —	— — —
			1 9 Осевое	1 2 Центр	1 6 Середина	36,3	50,9	6,6	6,0	4,4	3,8
				1 3 Средняя треть радиуса		36,9	61,7	9,6	11,6	4,5	3,8
				1 4 Периферия		36,9	57,3	11,2	18,7	5,9	5,6

*With intermediate upsetting.

1) Wheel No.; 2) reduction during drawing; 3) degree of upsetting; 4) specimen direction; 5) area of radius tested; 6) sampling site on disk height; 7) kgf/mm²; 8) kgf·m/cm²; 9) 1st specimen; 10) 2nd specimen; 11) tangential; 12) center; 13) middle third of radius; 14) periphery; 15) 1st face; 16) center; 17) 2nd face; 18) radial; 19) axial.

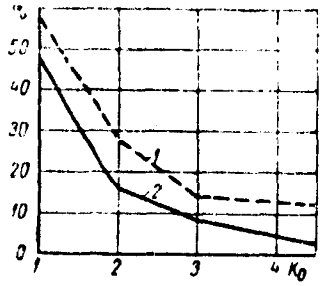


Fig. 5. Variation in ratio of plasticity characteristics in axial and tangential directions in central zone of wheels (vertical midline) as a function of degree of upsetting K_o. 1) ψ_{osev}/ψ_{tang}; 2) $\delta_{osec}/\delta_{tang}$.

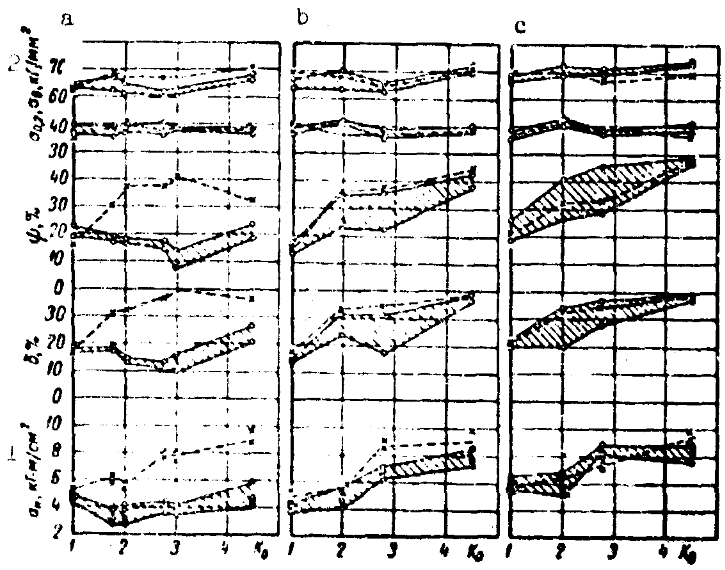


Fig. 6. Variation in mechanical properties of wheels in tangential direction as a function of degree of upsetting K_o. a) Central zone; b) middle third of radius; c) periphery. The shaded areas represent the scattering of the mechanical properties of the disk faces; the dashed curves represent the mechanical properties of the metal in the vicinity of the horizontal axial plane. 1) kg·m/cm²; 2) kgf/mm².

The mechanical properties of the wheel faces in the middle third of their radius occupied a position intermediate between those for the central and peripheral sections. It is noteworthy that the properties of the middle zone (especially δ, ψ, and a_k) were higher in the central region at relatively small degrees of upsetting (2.0-3.0) than in the middle third of the radius or at the periphery.

Another striking phenomenon was the substantial difference in metal plasticity in the two faces of wheels forged with a reduction of 2.0-3.0. For example, the values of δ and ψ in the tangential direction in a wheel forged with K_o = 3.0 were 8.6 and 7.8% for the central section of one face, while δ = 16.0% and ψ = 13.6% for the opposite face. For the central third of the radius of one face, δ = 19.8% and ψ = 20.8%; for the other face, δ = 36.2% and ψ = 31.2%.

Figure 6 presents graphs representing the variation in the mechanical properties of the wheels in the tangential direction as a function of the degree of upsetting.

These patterns in the plasticity indices of the metal in different sections of wheels produced with different degrees of upsetting are directly related to the macrostructure of the forgings and the characteristics of the forging process. Because of the nonuniform deformation during upsetting, the greatest actual reduction in area occurs in the vicinity of the horizontal axial plane in the central zone of the forging, while the smallest reduction occurs near the faces in the central zones, as a result of the inhibitory action of the contact surfaces. Appropriate experiments have established that, when wheels are forged from ЭИ572 steel with a total geometric reduction of two, the actual degree of upsetting varies vertically through the central section, ranging from 1.05 near the faces to 6.5 in the vicinity of the horizontal axial plane.

Even at a very small degree of upsetting, i.e., a small ratio of initial-blank height to wheel height, the middle zone of the forging thus undergoes very efficient mechanical working. The macrostructure of this section therefore contains no traces of the cast structure and the grains are radially oriented. On the other hand, the macrostructure of the metal near the faces is characterized by almost undeformed crystals, which is reflected in the reduced plasticity of these areas. As we move farther from the center of the forging, the difference in the macrostructures of the face and middle sections decreases, so that their mechanical properties become more similar. As the average degree of upsetting increases, there is a rise in the actual reduction in area near the faces and a simultaneous improvement in plasticity.

Assuming the acceptable values of δ and ψ for the central zone of a wheel to be 20-25%, the investigation described above gives us grounds for stating that the minimum necessary degree of upsetting for wheels of the type under consideration is about 4-5. However, taking into account the vertical deformation distribution through the forging and comparing the mechanical properties obtained in testing the middle zones with the actual reductions in these areas, it must be assumed that the highest metal plasticity in the central region is achieved with an actual degree of upsetting of 6.0-8.0.

The difference in the mechanical properties of the metal near the two faces of a wheel is due to the nonuniform deformation of these areas during upsetting. The blanks were deformed with the die block on the lower plate; the face of the blank was deformed in individual sections rather than as a whole, which promoted better working of the metal. The opposite (lower) face of the blank, which was in contact with the plate, was less highly worked, as a result of the greater deceleratory action of the contact surface, and was found to have lower plasticity indices during mechanical testing. Hence it follows that the best plasticity indices are obtained for the faces of a wheel when blanks are upset without a pressure plate, individual sections being clamped against a die block or other forging tool.

A wheel forged to a reduction of 2.0 after preliminary drawing of the ingot (Table 6, wheel No. 7) had relatively high mechanical properties in the tangential and radial directions throughout the entire forging volume. Comparison of this wheel with others forged to different reductions without preliminary upsetting of the ingot shows that its mechanical properties and their uniformity were equivalent to those of wheel No. 6 (produced with a degree of upsetting of 4.5) and markedly exceeded those of wheel No. 3 (produced with a degree of upsetting of 2.0). The principal advantage of wheel No. 7 lay in the higher metal plasticity in the face zone. The mechanical properties of this wheel in the axial direction were roughly the same as those of wheel No. 3.

The positive effect of intermediate upsetting of the ingot on the mechanical properties of the forgings was confirmed by their macrostructure (Fig. 4d). Intensive drawing of the upset block resulted in effective working of the entire volume of metal and no large cast grains were found even near the wheel faces, i.e., in the zones of impeded deformation, as occurred in wheel No. 3. This explains the higher plasticity in the face areas of wheel No. 7.

The technological processes employed in forging wheels should therefore include intermediate upsetting of the ingots in all cases where a relatively high degree of upsetting cannot be achieved for austenite-steel blanks and special measures cannot be taken to greatly reduce the extent of the inhibition zone. The importance of this technological factor increases as the D/H ratio of the wheel decreases.

Selection of the optimum reduction in area for austenite-steel forgings of the cast type is also technologically important. Investigation of forgings of this type produced from X18H9T steel with drawing reductions $K_n = 1.5$-4 and without intermediate upsetting of the ingots (which weighed 3.8 t and had an H/D ratio of 2.08 and a taper of 6.04%) showed the following pattern in the variation of mechanical properties at normal temperatures [28]: strength remained almost unchanged in both the longitudinal and transverse directions as the reduction was increased, while plasticity in longitudinal specimens rose markedly until $K_n = 2.5$ was reached and remained almost unchanged at higher values of K_n. Plasticity in transverse specimens was almost constant below $K_n = 2.5$ and decreased slightly when this factor was further increased. The macrostructure of the metal was coarse-grained and not destroyed by forging at $K_n = 1.5$, while it was medium-grained and had a radial orientation at $K_n = 2.5$ and fine-grained at $K_n = 4$.

The forgings were ultrasound-transmissive (at a frequency of 2.5 MHz) only with a reduction of 4.0.

Forgings produced from Х18Н9Т steel with the same drawing reductions but with intermediate upsetting of the ingot were distinguished by higher (by 10-30%) plasticity indices in the transverse direction and somewhat lower (by 5-10%) indices in the longitudinal direction. Intermediate upsetting reduced the anisotropy of mechanical properties in the forged shafts, especially at small reductions. Experiments conducted on forgings of Х18Н9Т steel also showed intermediate upsetting of the ingots to have a favorable effect on the structure and ultrasound transmission of the metal. The structure was fine-grained and uniform even at $K_n = 1.5$ and no scattering of ultrasound was observed.

Similar work has been conducted with ЭИ405 steel [17]. Stepped shafts were forged from ingots weighing 1.6 t under regimes that provided for different reductions (from 2.0 to 10.0), the ingot being subjected to one or two intermediate upsettings or to none. It was found that the strength indices in the longitudinal direction remained almost unchanged as the reduction was increased, while the plasticity characteristics rose rapidly as the reduction was increased to 4.0 and only very slightly when it was further raised.

Mechanical tests in the tangential direction established that the mechanical properties of forgings produced with intermediate upsetting were more uniform. Metallographic examination of the same forgings revealed no elongated carbide inclusions, which are characteristic of forgings produced without intermediate upsetting.

The research conducted at the NPL on the relationship between the structure and mechanical properties of austenite steel and the reduction to which they are subjected and other investigations in this area thus indicates that deformation in alternate directions during fabrication of shaft and wheel forgings has a positive effect. The influence of this factor increases with ingot weight. In some cases, as when a large wheel has an unfavorable shape (a large H/D ratio), effective working of the metal, good ultrasound transmission, and the requisite properties in a given direction can be obtained only by one or more intermediate upsettings. The reduction after intermediate upsetting can be limited to 3.0-4.0 for shaft forgings (depending on ingot weight) and to 2.0-3.0 for wheel forgings. The reduction must be increased by a factor of 1.5-2 when intermediate upsetting is not employed.

Chapter 4

INFLUENCE OF FORMING CONDITIONS ON SPECIAL-STEEL DEFORMABILITY AND FORGING QUALITY

Most high-alloy steels, especially those of the austenite class, have low reserve plasticity, which is manifested in poor ingot deformability and affects the physical state, structural compactness, and macrostructural homogeneity of the finished forgings. Such phenomena as the welding up of axial defects in the ingot, which is of prime importance for some groups of forgings, are directly related to the plasticity of the metal.

11. ANALYSIS OF BASIC TECHNOLOGICAL OPERATIONS IN FORMING AND SPECIAL FORGING PROCEDURES FOR HIGH-ALLOY STEELS

The plasticity of an ingot depends to a large extent on its surface quality, since superficial metallurgical defects serve as stress concentrators and foci of metal failure during deformation. Moreover, superficial defects, in the form of erosion-crack networks, cavities, impurities, blisters, and so forth are more highly developed in ingots of low-plasticity high-alloy steels. The work that has been done on improving the surface quality of ingots merits serious attention.

Solution of many of the technological problems involved in forging is associated with ingot surface quality. Among the most important of these is the practical feasibility of hot delivery of ingots or heating before the first withdrawal. Actually, when ingots of high-alloy steel with a low reserve plasticity contain large superficial defects, they cannot be fully reduced without preliminary trimming, removal of local flaws in the metal, or other types of adjustment in order to avoid failure during the first few reductions.

Preliminary trimming of the ingot has a very great effect in increasing the technological plasticity of low-deformability steels. Experience in forging austenite steel has established that mechanical removal of surface defects results in a metal plasticity such that the ingots can be deformed as well as ordinary structural steel in many cases.

Figure 7 shows two smooth forged wheels fabricated from billeted ingots of ЭИ572 steel, one of which was completely trimmed and the other trimmed over half its height. Both ingots had relatively poorly developed surface defects. The ingot with half its surface untrimmed was upset quite satisfactorily, without deep propagation of cracks

GRAPHIC NOT REPRODUCIBLE

Fig. 7. Forged wheels of ЭИ572 steel fabricated from billeted ingot, completely trimmed (upper wheel) and trimmed for half of height (lower wheel).

or tears. However, there was still an obvious difference in the surface state of the wheels: trimming greatly reduced the number and depth of the defects, which permitted a decrease in the number of passes required to give the forgings the characteristics provided for ordinary-steel forgings by the GOST. In other cases, where the character of the surface defects in the ingots is less favorable, failure of the metal during forging can be avoided only by including a trimming operation, especially when the technological process involves upsetting.

Trimming of multifaceted ingots requires installation of special lathe equipment, so that the billeted blanks are often trimmed instead of the ingots. Billeting carried out under appropriate temperatures and deformation conditions usually does not lead to noticeable development of surface defects and permits ordinary lathes to be used for intermediate trimming of the blanks.

Whether or not trimming of the ingots or billets is necessary and expedient is determined by the physical nature of the steel to be deformed and technical-economic considerations. Trimming is obligatory in forging low-deformability steels, whose low reserve plasticity makes it impossible to forge them without removing the surface defects from the ingots, even when substantial tolerances are left for machining and the forging deformation regime during the basic forming operation is favorable. In all other cases, trimming is best if it conforms to economic production principles.

The main economic advantage of such a technology lies in the greater reliability of the forging process, i.e., the relatively smaller number of forgings rejected for cracks and other surface defects and the possibility of a substantial reduction in forging tolerances and allowances, bringing them close to the norms established for similar forgings fabricated from ordinary structural steels. The tolerance problem becomes especially important in connection with the introduction of automatic devices of the СШТ-11 type [29] in many presses, since these greatly increase forging precision and hence reduce machining tolerances. However, use of such devices for forgings of high-alloy, low-deformability steels makes sense only when the ingots or blanks are preliminarily trimmed. Economical forging shapes can also be introduced under such conditions, using power-hammer dies and other forging attachments, which results in a substantial reduction in allowances and forging weight.

However, use of trimmed ingots or intermediate blanks is complicated by the need for additional cooling of the ingots in the furnace (or even for special heat treatment), repeated heating of the cold blanks, and prolongation of the technological cycle. Additional heat treatment is required both for forgings of martensite special steels (ЭИ802, 1X17H2, etc.) and for forgings of certain austenite steels, e.g., X18H22B2T2. The experience of the Neva Machine Building Plant [NPL](НЗЛ) has established that cooling of such steels in molds leads to development of shrinkage defects in the ingots and unsatisfactory forging quality. Air-cooling of billeted blanks causes detrimental phenomena, since the metal is only slightly deformed.

The expenditures associated with the loss of metal during trimming and the labor consumed in lathing is usually compensated for by savings in tolerances and in machining of the finished forgings. Different technological processes for forging identical components (with and without trimming of the billeted blanks) were compared at the NPL. It was found that the weight of the initial ingots was almost constant: the amount of metal removed in the form of scrap during trimming was roughly equal to the amount that went for the increase in tolerances required by forging of crude blanks. The total labor consumed in lathing (taking into account the trimming of the billet) on one hand and the decrease in tolerances and allowances in the forgings produced from clean blanks on the other were somewhat less for trimmed blanks.

Use of trimmed blanks is thus economical in many cases from the standpoint of metal consumption and total lathe load. Among other factors, it is necessary to take into account the specificity of the concrete production conditions: forging shape and production technology, the surface state of the cast ingots, the production scale for forgings of low-deformability steels, and the load on the preheating and heat-treatment furnaces. For example, the following pattern of operations has been found suitable for conditions at the NPL: only a limited number of blanks of X18H22B2T2, 0X18H10T, and other low-deformability austenite steels that are to be upset with a high degree of deformation during forging are generally trimmed. In individual cases, when the ingot surface is satisfactory, the blanks are trimmed after drawing of the upset workpiece rather than after billeting, provided that the forging process includes intermediate upsetting of the ingot. This sequence of operations is employed, for example, for forged wheels of ЭИ572 steel formed in power-hammer dies. The purpose of trimming the blank after the last forging operation (upsetting in the power-hammer die) is to create the most favorable conditions for production of forgings with minimum tolerances.

Individual examples of the tolerances employed at the NPL for different types of forming with certain special steels are given in

Billeting of the ingot is of material importance in forging ingots of low-deformability steel. Special emphasis must be laid on this factor, since some forging-pressing operations omit billeting in fabricating many types of forgings, not evaluating the nature and characteristics of the specific type of steel or the metallurgical characteristics of the ingot and not running reliable practical trials of a new process.

The feasibility of forging ingots without billeting depends on the metal plasticity, which is characterized by the type of steel and the surface state of the ingot. Practice has shown that billeting can be omitted for many structural steels, provided that the ingot surface contains no large cast-structure defects. For example, upsetting without billeting has been successfully employed at the NPL for ingots of 40, 34XM, 34XH3M, and certain other steels weighing up to 12-15 t. However, deformation of ingots of certain melts cast under less favorable conditions and having larger superficial metallurgical defects during upsetting without billeting caused formation of deep tears; similar ingots subjected to preliminary billeting were upset without producing visible defects. Reliable results were obtained on a large scale only after production of ingots with a relatively stable surface quality had been mastered.

An attempt to upset unbilleted ingots of high-alloy X18H9T, ЭИ572, and other steels was substantially less successful. Even selective upsetting of ingots cast with magnesium and having a comparatively clean surface yielded unreliable results: a substantial portion of the upset workpieces exhibited a larger than normal number of tears.

This type of behavior on the part of superficial defects during upsetting of unbilleted ingots of high-alloy steels is due to the strong influence exerted by the stress concentrators present in all low-plasticity steels. In addition to defects in the ingot that coincided with the upsetting direction, transverse defects also characteristically served as foci of metal failure, expanding and disrupting the continuity of the metal.

Experimental investigations of ingot billeting under laboratory and production conditions [30] have shown that this process, which creates a sheet of deformed metal at the ingot surface, substantially increases the plasticity of the outer zone, especially at forging temperatures. This is basically responsible for the less expensive development of superficial defects during upsetting of billeted ingots of low-deformability steel.

It follows from the data given above that omission of billeting for ingots of high-alloy low-plasticity steels greatly increases the probability of metal failure during forging and is unwise until ingot quality has been radically improved. Elimination of billeting during forging of medium-alloy steels can be recommended only after ingots of stable quality have been developed and the new process has been subjected to thorough practical testing with production batches of forgings.

Ingot drawing is one of the main operations in forging, being of fundamental importance for austenite-steel forgings of the shaft type and for components in disk form, as a result of the special characteristics of the behavior of such steels during upsetting. Production practice has established that large austenite-steel wheels cannot be reliably fabricated by normal forging techniques without preliminary drawing of the ingot or upset workpiece, during which the cast structure undergoes an extremely rapid transition to the deformed structure and shrinkage defects in the ingot are effectively welded up.

However, drawing can have a positive effect only when deformation conditions are favorable, i.e., when the entire volume of metal is directly subjected to multilateral nonuniform compression and substantial tensile stresses do not develop in the blank. This condition is also the main prerequisite for failure-free deformation of low-plasticity steels.

Among the technological parameters responsible for a favorable stress pattern in the metal during drawing, we should note the shape of the die blocks, the relative feed, and the reduction regime during the press stroke or hammer blow.

A flat die block is simplest and best for drawing ingots of special steels having relatively low reserve plasticity. Drawing is carried out on the square-square or, even better, square-rectangle-square pattern. This provides the greatest depth of deformation through the ingot cross-section and creates favorable conditions for vigorous working of the central zone of the ingot. The effectiveness of this drawing pattern has been confirmed by much research and production experience. Specifically, investigation of the macrostructure of two blanks of austenite chromium-nickel-manganese steel forged from identical ingots with the same reduction ratio ($K = 2.7$) but different drawing regimes showed that remnants of the cast structure and concomitant porosity were present in the center of the blank forged in shaped die blocks on a round-round pattern; the blank forged on flat die blocks by a square-rectangle pattern exhibited a compact deformed structure throughout its entire cross-section [17]. The square-rectangle-square pattern is employed both for forging that are to undergo final forming by drawing and for certain critical components of the wheel type, for which drawing precedes upsetting. In such cases, the shift from a square to a circle is effected with a minimum cross-section ratio. Drawing in flat die blocks is widely employed particularly in forging rotors and large wheels of high-hot-strength steels of the perlite and martensite classes (Р2М, ЭИ802, etc.). A similar regime has been successfully employed at the NPL for forging large shafts from 1Х17Н2 steel.

Flat die blocks are completely unsuitable for drawing on a circle-circle pattern or for shifting from a square to a circle, since transverse tensile stresses develop in the blank, especially at small degrees of reduction, and the shrinkage defects in the metal are not welded up but extend still farther; new internal foci of metal failure not associated with defects in the ingots may develop.

A combined drawing pattern with flat die blocks cannot be used for low-deformability steels (principally those of the austenite class) because of the low technological plasticity of the metal and the entire drawing process, from the initial to the final cross-section, is carried out on a circle-circle pattern. Die blocks with a shape profile are employed in such cases. The most favorable profile is one that provides contact with the blank over the maximum perimeter of its cross-section. This factor is most effective from the standpoint of the stress-deformation regime: the pressure applied to the blank by the die block and directed perpendicular to the contacting surfaces causes development of compressive stresses, provides more or less multilateral compression, and creates the most favorable conditions for deformation of the inner regions of the ingot and a simultaneous increase in the total reserve plasticity of the steel.

The best die-block shapes in this sense are notched or rhombic. However, their design (angle of blank entry, apical radius of curvature, and ratio of die-block radius to blank radius) is of material importance. The optimum regime is one in which the die-block radius equals the blank radius, but, since this impossible in practice, such blocks being suitable for forging blanks of only one size at best, a minimum productively feasible range of blank diameters has been established for notched and rhombic die blocks with a definite radius or entry angle. The technological series of diameters was established from the condition $R_z \geq R_b$ (Fig. 8a), with a restriction imposed on the maximum blank radius. The variant in which $R_z < R_b$ is unfavorable (Fig. 8b). At worse, this variant approximates to drawing of round blanks in flat die blocks.

Composite die blocks (with the upper block flat and the lower notched or rhombic) are often used in practice. Such blocks are convenient in production terms but substantially less effective than notched or rhombic blocks from the standpoint of the stress pattern in the blank. Drawing blanks of high-alloy steels with low reserve plasticity in composite blocks often leads to cracking. Such blocks are also less effective with respect to welding up of internal defects. These characteristics of the stress pattern in low-plasticity steels during drawing in die blocks of different shapes were specifically confirmed by Ye.P. Unksov, who studied the stress distribution during drawing of shafts by the optical-polarization method [31].

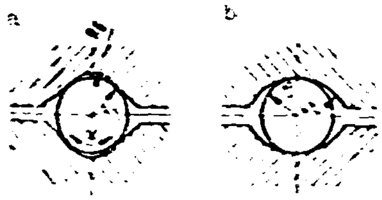


Fig. 8. Diagram of reduction of round blank with notched die block: a) $R_z > R_b$; b) at $R_z < R_b$; R_z is the blank radius and R_b is the die-block radius.

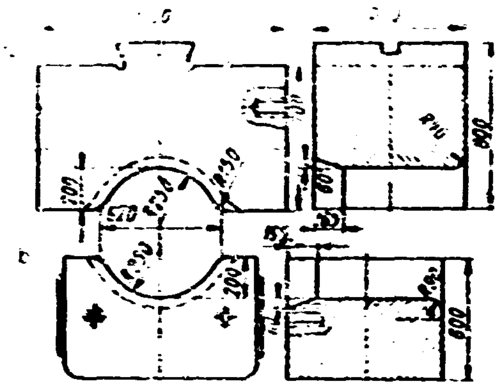


Fig. 9. Notched rhombic die block with chamfer for forging press exerting pressure of 3000 t [33]: a) Upper block; b) insert for lower block.

The angle at which rhombic blocks are cut away is usually 90-110°, but the optimum block-shape parameter requires refinement. Experimental work conducted in this area in recent years has led to the conclusion that the cutaway angle is obviously best established in accordance with the character of the technological operation to be performed (billeting, drawing, or smoothing), since specific stress-deformation conditions are operative in each case.

In drawing low-deformability steels, creation of the optimum deformation pattern during the initial period of forging, when the cast steel has its lowest reserve plasticity, is especially important. It is during the first few times that the die block is pressed against the ingot surface that tears and cracks appear in the metal, then often leading to complete failure. Slight reduction of the ingot markedly increases the plasticity of the metal and permits subsequent drawing of the blank to be carried out under conditions that are less favorable in terms of stress pattern but technologically better. For example, after an ingot is removed from the press after its initial reduction in notched blocks, shifting to composite blocks or even to forging on a circle-square-circle pattern is permissible in many cases. Special attention should therefore be paid to the initial reduction operation and the die block and preparatory procedures should be selected accordingly. The results of work conducted by L.V. Prozorov [32] indicate the significance of the plastic sheath formed on the ingot as a result of the initial deformation for subsequent forging-forming conditions. Fabrication of forgings from ingots with an extremely low reserve plasticity, which made it impossible to force them with die blocks of any shape, was carried out by forcing them through special feeders in which the primary cast structure was deformed by nonuniform multilateral compression. Reduction of the ingots by 8-10% permitted subsequent forging without any serious difficulties.

In forging low-plasticity steels, transverse cracks are preferentially formed where the deformed portion of the ingot passes into the undeformed portion. The smaller the radius of die-block curvature (given the same degree of reduction), the greater are the tensile stresses that arise in this area and the more rapidly do transverse cracks develop. A pattern of this type is consistently observed in forging high-alloy steels. As a result, the die blocks must be fabricated with larger radii of curvature in the areas where the lateral surface passes into the working surface (up to 80-100 mm) and the block must cover the region where the undeformed portion of the blank passes into the deformed portion during each reduction.

The tensile stresses that develop in the transition areas (as a result of elongation) can to some extent be compensated for by compressive stresses produced by reduction of the transition zone with a chamfered die block (Fig. 9). Experimental investigations in this area and work carried out under production conditions [33] has yielded positive results: cracking was substantially reduced during forging of low-plasticity steel ingots. The best results were obtained when the blanks were reduced with die blocks having a taper angle of 30°.

The relative feed l/h (where l is the feed and h is the blank thickness within the geometric focus of deformation during a given press stroke or hammer blow) plays a large role in creating a favorable stress pattern during drawing of ingots.

As numerous theoretical and experimental studies [34 and others] have shown, the value of l/h affects the uniformity of reduction over the blank thickness; the nonuniformity of reduction increases as the relative feed decreases. As a result, longitudinal tensile stresses act in the central zone of the blank when l/h is low, not only failing to promote welding up of axial defects in the ingots but actually creating conditions for enlargement of defects and even for formation of transverse cracks in weakened areas of the blank. The metal is in this state when $l/h < 0.5$. Both tensile and longitudinal compressive stresses are operative in the central portion of the blank cross-section at $l/h > 0.5$ and the possibility of effective working of the metal in the central zone is greatly increased. However, the maximum value of l/h should be limited to 0.8-1.0, since transverse tensile stresses are dominant in the blank at $l/h >$ > 1 and, in forging low-plasticity steels, can lead to formation of longitudinal axial cracks and surface defects. The optimum value of l/h for drawing is therefore $0.8 > l/h > 0.5$. It is important to satisfy this condition both during drawing in flat or composite blocks or when notched blocks are used, i.e., when the deformation pattern is most favorable.

The influence of the l/h ratio has been widely checked under production conditions, particularly in forging rotors and wheels from high-hot-strength steels. In most cases, switching to broad die blocks and large feeds led to an increase in forging quality.

The mechanical deformation pattern during drawing is also affected by the reduction, i.e., the degree of deformation per press stroke or hammer blow. This technological factor, which has no material effect on the stress and deformation distribution during drawing of blanks with a square or rectangular cross-section in flat die blocks, greatly alters the stress-deformation pattern of the metal during drawing of round blanks in flat or notched blocks [35].

An increase in the relative reduction lengthens the contact line between the die block and blank perimeter, which promotes a decrease in the tensile stresses in the center of the blank and may even convert them to compressive stresses, i.e., produce a more favorable stress-deformation pattern.

The depth to which deformation extends increases with the reduction. This factor is responsible for one of the main functions of drawing: increasing the density of the metal in the central zone and welding up axial defects in the ingot, which is possible only if the deformation extends to the center of the blank, i.e., through its entire thickness.

Small reductions cause the flow in the outer layers of the metal to lead that in the inner layers when forging steels, especially those with high alloying-element content. This is manifested in craters formed on the butt surfaces of the blank, their presence indicating ineffective working of the deep region. Coupled with other detrimental factors, cratering causes increased metal consumption, a higher labor consumption in forging (since, in some cases, it is necessary to smooth the butts), and occasional appearance of constrictions and surface cracks in the finished forgings. It should be noted in passing that formation of butt craters during drawing of upset blocks can to some extent be prevented by use of spherical plates during upsetting of the ingots. The convex face of the block compensates for the preferential metal flow in the outer regions and promotes smoothing of the blank butt during the final drawing operations.

The relative reduction during drawing therefore ought to be as large as possible. Its maximum value should be limited principally by a single factor: the reserve technological plasticity of the steel.

Use of large reductions for special steels of the perlite class usually presents no particular difficulties. Some problems are encountered with low-deformability steels. For relatively uniform metal flow in the outer and inner layers of a blank of diameter d during drawing, the reduction during a single press or hammer stroke should be no less than 0.08-$0.1d$ with a die-block width of no less than 0.6-$0.8d$ [32]. In practice, some austenite steels, such as X18H22B2T2 and X23H18, can undergo a relative per-pass reduction of no more than 0.05-$0.08d$ (during drawing in composite blocks). In such cases, it is especially important to create conditions for increased steel plasticity by preliminary trimming of the ingot, use of the optimum die-block shape and elevated forging temperatures, and employment of other measures that permit an increase in permissible reduction during drawing.

Increased reductions have a varying final influence on the quality of forgings for components with different shapes and different degrees of criticality. Thus, this problem is not as fundamental for forgings of the wheel type or other hollow components as for rotors or wheels without holes, in which effective working of the central region of the blank during drawing of the upset block is the principal prerequisite for high component quality.

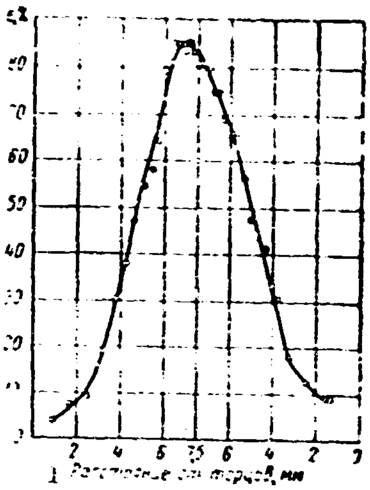


Fig. 10. Graph showing distribution of actual degrees of deformation ε over height of axial zone of specimen of ЭИ572 austenite steel after upsetting to half its height (data from Central Scientific Research Institute of Technology and Machine Building). 1) Distance from butts, mm.

The permissible per-pass degree of deformation established at the NPL for drawing of round articles in rhombic and composite die blocks is 15-20% for 2X13, 1X17H2, ЭИ802, and X18H9 steel, 10-12% for 0X18H10T, X18H12M2T, and ЭИ572 steel, and 5-8% for X23H18 and X18H22B2T2 steel. The relative reduction is reduced by a factor of about two in drawing square blanks in flat die blocks. Similar degrees of deformation are employed at the Urals Machine-Building Plant [36].

Upsetting is widely employed in forging production, serving as an auxiliary operation to ensure the required reduction ratio during subsequent drawing of the upset blank and as the main forming operation in fabricating forgings of the wheel type.

Forging occurs with nonuniform metal flow over the blank thickness, a consequence of the varying deformation resistance that is produced mainly by the influence of the frictional forces at the surfaces where the metal comes into contact with the deforming tool (die block, plate, or punch). The nonuniformity of deformation results from the fact that metal movement in the radial direction in the regions adjoining the contact surfaces lags behind deformation in sections far from the butts, so that the lateral surface of the blank is distorted: it changes from rectilinear to curvilinear and comes to resemble a barrel. The larger the frictional forces, the greater is the inhibition of radial metal flow in the butt regions and the more barrel-like the blank becomes.

The inhibition surfaces in the contact planes are bases of segment-like zones of impeded deformation (inhibition zones), within which minimum metal flow occurs. The metal lying between the zones of impeded deformation is most rapidly deformed, since contact friction does not act on it. The peripheral zone of the blank, along the curve and in adjoining regions, undergoes mostly tangential deformation. It is in this area that the greatest tensile stresses develop, causing cracking at the curved surface during upsetting of low-plasticity steels [37].

Figure 10 is a graph representing the distribution of the actual degrees of deformation over the height of the axial zone of a specimen of ЭИ572 austenite steel after upsetting to half its height (with a reduction ratio of 2.0). The graph was plotted from the change in the thread pitch of a screw inserted into the specimen. The temperature to which the specimen was heated before upsetting was 1200°C. It can be seen from the diagram that the actual degree of deformation was 3-8% near the contact surfaces and about 85% in the vicinity of the horizontal axial plane. This curve, obtained for an experimental specimen, also characterizes the general trend of the phenomenon during upsetting of production blanks.

The presence of zones of impeded deformation during upsetting is an extremely important factor, making the fabrication of wheel forgings from low-deformability steels a complex process. Actually, remnants of the undeformed cast structure are detected in wheel faces even when a billeted austenite-steel ingot is upset 4 or 5 times, this being completely impermissible in cases where these areas of the metal are not removed during subsequent machining. The metal in the poorly deformed areas of the wheel near the faces have a substantially lower plasticity than that in zones far from the faces, particularly that in the vicinity of the horizontal axial plane, where metal flow is most rapid. The influence of zones of impeded deformation on the mechanical properties of different areas of wheels was shown in Chapter 3 for forgings of ЭИ572 austenite steel.

It is important in practice to determine the factors exerting the greatest influence on the size of the zones of impeded deformation and the feasibility of reducing the size of these zones through the action of technological factors. Any conditions that reduce the contact frictional forces simultaneously promote a decrease in the size of the zones of impeded deformation. The following factors have been found to be instrumental in reducing friction during hot upsetting: the surface state of the working tool and the contact surfaces of the blank, the lubrication employed, and the deformation temperature and rate.

A decrease in the surface roughness of the pressure tool markedly reduces the coefficient of friction μ, although there has as yet been no quantitative evaluation of μ as a function of the surface-finish class of the tool. The importance of this factor is confirmed by the anisotropy of friction, i.e., the difference in deformation parallel to and perpendicular to the working direction. I.M. Pavlov [38] established that the coefficient of friction in the working direction is approximately 20% less, even when the tool is subjected to double grinding and a lubricant is used. The anisotropy of friction is considerably greater during upsetting with a tool having a coarse-ground surface and no lubricant, while the elliptical shape of an upset cylindrical blank becomes quite pronounced as a result of the greater inhibition of metal flow perpendicular to the working direction. Beads of metal, notches, and other defects formed on the working surfaces of the die blocks and plate during operation have a strong influence on the value of μ. Use of a working tool with a clean, smooth working surface is a necessary condition for increased uniformity of deformation.

The physicochemical state of the blank surface also has a considerable effect on the value of μ: the presence of oxide films and especially of scale increases μ.

Use of a lubricant greatly reduces the coefficient of friction during upsetting. Lubrication is regarded as an almost obligatory condition for successful upsetting during forging of high-alloy high-hot-strength steels.

The change in coefficient of friction is not strictly related to that in deformation temperature and is sometimes opposite in direction. All the same, the experimental work and observations of Ye.P. Unksov [31] have established the following relationship (for any metal): the coefficient of friction increases with rising deformation temperature (which Unksov attributes to scale formation), reaches a maximum in the vicinity of 600-800°C, and then begins to decrease to its initial level (as a result of the substantial increase in plasticity).

One of the main factors responsible for the impeded deformation in the areas adjoining the contact zone is the tendency of the metal to shrink from contact with the cold tool; preliminary heating of the die blocks and plates (to 350-400°C), which ensures less rapid cooling of the faces, therefore promotes a decrease in contact friction.

The coefficient of friction decreases as the deformation rate increases, which is due principally to a reduction in the contraction of the metal adjoining the contact region. The coefficient of friction is therefore lower during hammer upsetting than during press upsetting.

The development of zones of impeded deformation during upsetting is directly related to the ratio of the initial blank diameter to the blank height: the size of the inhibition zone decreases as D/H increases. Under production conditions, this is manifested in an increase in homogeneity of macrostructure and mechanical properties over the radius of forged wheels as their diameter is increased or their height is reduced.

The technological difficulties associated with formation of inhibition zones in wheels affect principally austenite-steel forgings, as a result of their thermomechanical characteristics. The high degrees of upsetting required to convert the cast structure to a deformed structure necessitate rapid drawing (before the final upsetting operation), even for wheels of relatively small size. Preliminary drawing is also sometimes employed in cases where it is not favorable from the standpoint of coincidence between the fiber direction and the direction of the main working stresses in the component, in order to avoid the still more detrimental effect of a central zone that undergoes only slight deformation during upsetting and of the presence of remnants of the cast structure.

In this sense, the intermediate operations involved in upsetting (ingot drawing in forging austenite-steel wheels) are very critical operations, often determining the ultimate success of forging.

The main condition for positive results in drawing and upset billet is preparation of the workpiece itself during upsetting of the ingot, which is important not so much for its immediate influence in working the metal as for its role in the effectiveness of subsequent drawing. Inadequate or nonuniform heating, a high H/D ratio in the billeted ingot, inadequate press force, forging with unjustifiably small reductions, upsetting with die blocks rather than plates, improper aperture diameter in the upsetting plate, and other factors that can to some extent distort the billet shape are impermissible. The minimum billet diameter through any cross-section should ensure that a given reduction ratio (usually no less than 2) and an appropriately worked structure are obtained during subsequent drawing. For example, upsetting often ingot into a "wine glass" shape, as often occurs in practice, usually leads to an impermissibly small degree of working for the metal in the constricted portion of the billet. Production experience has established that this is one of the causes of the traces of cast or poorly deformed structure and reduced mechanical properties sometimes observed in wheels. An overly large aperture in the upsetting plate sometimes also causes poor working of the metal in the shank (the portion of the ingot under the shrink head). It has been established that the zone of impeded deformation becomes larger [35] and the entire upsetting process takes place under less favorable conditions as the aperture diameter increases.

In hammer-forging small wheels from light ingots, where intermediate upsetting is practically impossible, the wheels must be formed either from a blank reduced with a reduction ratio of no less than 2.0 or from a billeted ingot with a relatively high degree of upsetting (no less than 4-5). The latter process is usually employed only when the D/H ratio is large (no less than 6-8), in which case the metal in the central zone is effectively worked.

The recommendations made above are for forged wheels without axial apertures. The upsetting requirement can be somewhat less stringent for components with central bores or more favorable configurations (rings), depending on the amount of metal removed from the central zone during broaching or boring and the subsequent deformation to which the blank is to be subjected (rolling, etc.). In some cases, the forging procedure can be simplified by omitting the intermediate upsetting of the ingot and reducing the degree of deformation during the final upsetting.

Achievement of as uniform a metal deformation as possible during upsetting of high-alloy steels is important from the standpoint both of effective working of the cast structure and of the effect of recrystallization processes and fiber texture in wheels.

As the nonuniformity of metal flow during upsetting increases, there is a rise in the probability that the steel will recrystallize (in individual areas of the wheel) at critical degrees of deformation, a process accompanied by impermissibly large grain growth. This phenomenon is often observed when commercial forgings are sectioned and examined. At the same time, the fibers formed in the central zone of the wheel during drawing of the ingot or upset billet have an unfavorable arrangement with respect to the direction of the working stresses. The greatest stresses in wheels under operating conditions usually develop in the tangential and radial directions. The preferential fiber arrangement in the axial direction that occurs in the central areas of a wheel near the contact zones when deformation is markedly nonuniform during upsetting therefore reduces the service life and reliability of the component.

One very effective way to increase the uniformity of metal flow during upsetting is use of various technological lubricants. Thick greases or pastes containing machine oil and graphite and mixtures of molten glass with graphite yield satisfactory results. For convenience of industrial use, the lubricant is applied uniformly to previously prepared cardboard or asbestos liners on the lower upsetting plate and the upper face of the blank to be upset. Powdered glass or glass wool are also employed in upsetting austenite steels, simultaneous serving as heat insulation between the metal and tool as a result of their low thermal conductivity [17].

The experience of the NPL in forging ЭИ572 steel blanks with and without lubricants has demonstrated the undoubted advantages of the former procedure. In practice, these are manifested in a substantial increase in metal plasticity when tangential specimens from forged wheels are tested. In many cases, use of lubricants is of fundamental importance in switching to stamping of austenite-steel wheels in power-hammer dies. At the NPL, switching from production of large wheels from ЭИ572 steel by free forging to stamping, which permits an increase in complexity of component shape but entails a substantial reduction in weight, proved possible only when an effective lubricant was employed; in addition to increasing the uniformity of deformation, lubrication promoted a decrease in deformation resistance and better filling of the die.

It should be noted that exertion of lateral pressure on the walls of the forging tool (e.g., a die) yields a less severe stress pattern and an increase in the technological plasticity of the steel. In general, maximum limitation of the free surface of the forging to be deformed and use of closed or semiclosed forging techniques promotes an increase in metal plasticity. Deformation in the presence of lateral pressure improves the macrostructure, microstructure, and mechanical properties of the metal (provided that the required reduction ratio is achieved) [19].

In some cases, special procedures can be employed to favorably alter the stress pattern in forging wheels from low-deformability high-hot-strength steels. For example, use of hot liners of soft cast steel and paired upsetting of the blanks can be very effective.

Upsetting of blanks enclosed in plastic liners on both faces causes forced radial flow of the metal in the areas of the blank near the contact zones. The action mechanism of a plastic layer between the faces of the blank and the die blocks [39] consists in the following: the plastic liner, filling unevennesses in the contact surfaces and being displaced in the radial direction at stresses below the yield strength of the blank, is deformed by triaxial compression. As a result of the grabbing of the contact surfaces, a varying triaxial stress pattern is set up in the faces of the blank, promoting deformation of these areas. Since the faces also cool more slowly because of the hot liners, the regions near the contact zone are substantially less strongly decelerated and their plasticity is increased. The principal condition for forced metal flow near the faces is that the liner material have a lower yield strength than the blank material at the forging temperature. This can be achieved by selecting an appropriate liner material and proper heating conditions. It is best to heat the linears together with the blanks and supply them to the press in the form of a stack, in order to avoid rapid cooling. The liner thickness is usually equivalent to 0.07-0.12 times the blank thickness, depending on the D/H ratio. Smaller liner thicknesses correspond to lower values of D/H [32].

Experience in the practical application of this method for forging wheels from high-hot-strength austenite steel [40] has shown its reliability and technical-economic feasibility: use of liners permitted a large decrease in the total reduction ratio necessary to convert the cast structure to a deformed one and thus made it possible to shorten the forging cycle in comparison with the usual technological process by omitting drawing.

In paired upsetting of wheels, the zone of impeded deformation encompasses the metal in only the one face of each blank in contact with plate or block surface; the opposite faces of the blanks, which are in contact with one another, lie in the zone of maximum deformation, like the horizontal plane of an upset monolithic blank. By turning the wheels during upsetting (after heating) so that different sides are opposed, relatively uniform upsetting can be achieved for both wheels. Moreover, paired upsetting of wheels reduces the press force required.

It is also possible to reduce the size of the zone of impeded deformation and increase the plasticity of the steel by using conical die blocks or arbors during the first stage of upsetting.

Upsetting of blanks between two cones promotes greater homogeneity of the deformation pattern and a more uniform stress distribution. When the angle of taper is equal to or close to the friction angle, i.e., $\tan \alpha = \mu$, upsetting becomes relatively more uniform and the stress pattern becomes approximately linear, as opposed to the volumetric pattern observed during ordinary upsetting.

Special bands are sometimes used in upsetting low-deformability steels. Essentially, this process consists in placing a band of limited height fabricated from a plastic metal around the blank and upsetting the two to the required size. The metal in the band, which resists any increase in diameter, exerts a lateral pressure on the blank at the sites of maximum tensile stresses, thus promoting a more favorable pattern of deformation by nonuniform multilateral compression, which hampers formation of radial cracks. The lateral pressure is set up by the different temperatures to which the blank and band are heated, thus giving them different yield strengths in the heated state (lower for the blank). Use of bands markedly increases the plasticity of steel. The band is generally removed with a press after upsetting, knocking out the blank. Any metal that builds up on the faces of the band are removed mechanically.

12. INFLUENCE OF DEFORMATION PATTERN AND REGIME ON WELDING UP OF SHRINKAGE DEFECTS IN INGOT AND FORMATION OF INTERNAL DEFECTS IN METAL DURING FORGING

The possibility of welding up metallurgical defects in an ingot during production of forgings (especially large ones) from special steels is directly related to the aforementioned stress patterns and the ways in which they develop in the metal during forging.

In most cases, internal defects in the ingots (shrinkage pores in the axial zone and, sometimes, networks of fine axial cracks) develop during crystallization. One characteristic of such defects (discontinuities in the metal) is their contamination with liquates and nonmetallic inclusions, whose presence hampers or even prevents welding up of the metal during forging. The purity of the melted metal is thus often the decisive factor in successful welding up of internal defects in the ingot, this being particularly true of complex-alloyed steels. A second condition for the metal contact required for welding is isolation of the defects in compact metal, so that their surface is not oxidized by atmospheric air or furnace gases. The arrangement of axial defects in an ingot is usually favorable in this respect: they are generally separated from the shrinkage cavity by the "bridge" below the shrink head, i.e., by a layer of compact metal.

Ingots also sometimes contain internal defects produced by thermal factors rather than by shrinkage: these take the form of transverse tears and can be welded up during forging if they do not extend very deeply in the radial direction. Such tears, which are most often encountered in ingots of monophasic (particularly austenite) and high-chromium martensite steels, are produced by uneven cooling of the ingot. A characteristic example is the cooling of ingots during hot delivery to the forging-pressing shop (at temperatures of up to 300-400°C).

Determination of the mechanisms by which internal discontinuities in the metal are closed and then welded up during various forging operations is of great practical interest.

Drawing is known to be the most effective operation in terms of welding up of internal defects. With optimum deformation regimes that produce no internal tensile stresses and favorable forging temperatures, welding up of axial defects in plastic-steel ingots takes place at a relatively small reduction ratio. Experimental work and the practical experience of the NPL indicate that severe shrinkage defects in an ingot of ЭИ415 perlite steel are completely welded up at a reduction ratio of 1.5-1.7. These data are for ingots weighing up to 3-5 t; the experience of the Urals Machine Building Plant

Studying the welding up of internal defects in large forgings (up to 1000 mm in diameter) of 34XH2M steel is of interest with regard to ingots of greater weight [41]. Periscopic examination of the axial channel of a rotor revealed a large number of longitudinal defects of the crack type arrayed in a barrel-shaped region 1000 mm in diameter. In all cases, subsequent drawing of the rotor body (a carbon-steel rod was pressed into the channel before forging) into a multistage forging on a circle-circle pattern with reduction ratios of 2.8 and 3.8 and on a circle-plate-circle pattern led to complete welding up of the defects.

Welding up of defects was also investigated during drawing of a 34XH2M steel ingot weighing 19.5 t. A transverse section was cut from the center of the blank after drawing with a reduction ratio of 1.5; its surface exhibited numerous cracks ranging from 2 to 16 mm in length. Half of the previously produced blank was later drawn into a multistage forging with reduction ratios of 1.5, 2.5, 4.0, and 8.0. Careful ultrasonic monitoring and macroscopic investigation of sections showed no defects in any of the forging stages. Thus, drawing of a large chromium-nickel-molybdenum steel ingot with a total reduction of about 2.0-2.5 will ensure that it contains no unwelded axial defects.

Axial defects in ingots of certain readily welded austenite steels, such as ЭИ572, are welded up at relatively small reduction ratios. For example, it has been established at the NPL that shrinkage defects (fine intercrystalline cracks) in an elongated ЭИ572 steel ingot weighing 2.17 t are completely welded up during effective drawing with a reduction ratio of no less than 1.5. At the same time, it was noted that certain other complex-alloyed austenite steels and alloys require substantially greater reduction ratios for welding up of internal defects. Thus, it was found that defects in an ingot of a nickel-based alloy weighing 700 kg are welded up only when the reduction ratio is about 2.0, while an alloy with an iron-chromium-nickel base requires a reduction of about 3.5 [42]. A reduction ratio of no less than 2.0-2.5 is necessary for complete sealing of axial defects in an ingot of the dispersion-hardening austenite steel Х18Н22В2Т2 weighing up to 5-6 t. The experience of the NPL has demonstrated that high-chromium complex-alloyed steels (ЭИ802, 15Х11МФБ, etc.) are among those whose internal defects are difficult to weld up. Only drawing with a reduction ratio of about 2.5 eliminates traces of unsealed shrinkage pores from blanks forged from ingots weighing about 4 t.

The reduction figures given above are for press-forging of the blanks, which permits use of greater single reductions and application of more prolonged deforming forces. Both factors facilitate successful welding up of defects by creating a more stringent multilateral compression pattern and more complete recrystallization, which is especially important for high-alloy austenite steels. The minimum reduction ratio necessary for welding up of internal defects in ingots of high-alloy steels and alloys is greater for hammer-forging. Specifically, when the aforementioned ingots of nickel and iron-chromium-nickel alloys weighing 700 kg are hammer-drawn rather than press-drawn, the minimum reduction necessary for welding up of defects is almost doubled, reaching 3.5 and 7.0 respectively [42].

The influence of the deformation rate (press- or hammer-forging) is obviously different for different steels: this factor is less important for plastic steels and more important for difficult-to-weld high-alloy steels.

It must be kept in mind that the diversity of factors influencing defect welding during drawing (the extent, direction, and location of the defect in the ingot, the nature of the steel, the extent of the liquational inhomogeneity in the steel, the geometric shape and final dimensions of the blank, the stress pattern during forging, etc.) makes it impossible to establish general norms for the minimum reduction ratios that will ensure complete welding up of internal defects during drawing. Nevertheless, the data given above, which are based principally on the results of industrial research and practical experience, enable us to evaluate the feasibility of welding up shrinkage defects in ingots of different types of steel. For

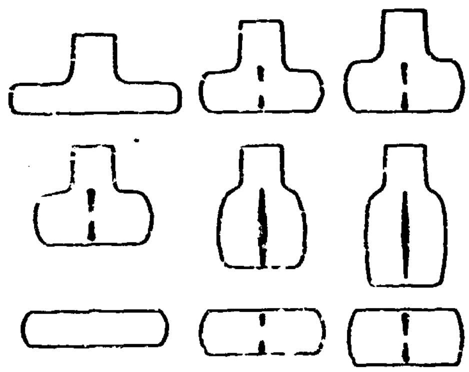


Fig. 11. Diagrams showing sealing of axial defects during upsetting of blanks with different shapes.

example, it can be more or less assumed that, given favorable press-drawing conditions, shrinkage defects in ingots of normal size weighing up to 4-5 t will be completely sealed up with the following reduction ratios:

a) 1.5-1.7 for perlite steels and readily weldable austenite steels (ЭИ572, etc.);

b) 1.6-2.0 for stainless steels (2Х13);

c) 2.2-2.5 for high-chromium complex-alloyed steels (ЭИ802 and 15Х11МФБ) and dispersion-hardening austenite steels (Х18Н22В2Т2).

Welding up of internal defects in the ingot takes place under more complex conditions during upsetting than during drawing, since the shrinkage defects are usually oriented in the upsetting direction, thus necessitating greater force and a larger degree of deformation to seal the discontinuities in the metal, and the contact frictional forces and associated nonuniform deformation distribution in the blank of a substantially greater influence. The latter factor is particularly important for low-deformability austenite steels, in which the actual degree of deformation in the areas near the contact zone during upsetting differs greatly from the average degree of deformation. The nonuniform deformation during upsetting causes the effectiveness with which discontinuities in the metal are sealed to vary over the height of the blank. Defects in the vertical midzone are welded up at a substantially smaller total reduction than defects in the zones of impeded deformation. The D/H ratio, which determines the stress pattern during upsetting and the specific pressure exerted on the metal, plays a large role in the effectiveness with which defects are sealed. The higher the D/H ratio, the more effectively are internal defects welded up at a given degree of upsetting, since the multilateral volumetric compression is greater in this case.

Experiments conducted to determine the actual degrees of upsetting necessary to weld up internal defects, which were conducted by modeling the process under different forming conditions [35, 43], showed that sealing of axial defects during upsetting of tall cylindrical blanks (with an initial ratio $D/H < 1$) requires a local deformation of about 60-70%, while that necessary in upsetting low blanks ($D/H > 1$) is 40-45%.

Welding up of defects in a cylindrical blank begins in the vertical midzone and then extends toward the butt as upsetting continues. In upsetting blanks prepared to simulate billeted ingots (with tailpieces to simulate shrink heads), the defects in the vertical midzone became more extensive until the blank diameter exceeded its height. Gradual closing of the defects began at $D/H > 1$. No expansion of the defect was observed in blanks with an initial ratio $D/H > 1$. It has also been demonstrated experimentally that sealing of local defects in the portion of the ingot below the shrink head requires the greatest degree of upsetting. Figure 11 shows the dynamics of sealing of axial defects during upsetting of blanks with different shapes [62].

In using experimental results to consider the actual feasibility of welding up axial defects in ingots and blanks during upsetting, the following factors must be taken into account.

1. The experimentally determined critical degrees of deformation during upsetting pertain to the plastic-forming stage, during which the defects are only closed; complete welding up of discontinuities in the metal requires additional reduction, whose magnitude depends on the physicochemical characteristics of the steel. The additional reduction is relatively small for readily welded plastic steels, while that required for complete welding up of internal defects in steels of complex composition (e.g., ЭИ802 or Х18Н22В2Т2) can be quite substantial.

2. In upsetting blanks of low-deformability steels, especially those of the austenite class, the actual deformation in the areas adjoining the butt may be far lower than the average degree of deformation. The total deformation necessary to weld up defects over the entire blank height will therefore substantially exceed the critical deformations given above when the metal in both the central portion of the blank and the areas adjoining the faces contains discontinuities (a case very common in practice).

Experimental work (taking into account the corrections that have been made) and long-term industrial experience have shown that the practical feasibility of welding up large shrinkage defects in high-alloy steel ingots during upsetting is very limited. Satisfactory results are usually obtained only in forming forgings of the wheel type with a relatively large final-dimension ratio (*D/H* = 8-10 or more). For example, experience has demonstrated that defects in ingots of high-hot-strength chromium-nickel alloys are completely welded up during upsetting when the degree of deformation is 87% and the ratio *D/H* = 21. Upsetting of such ingots with a deformation of 78% and a ratio *D/H* = 9 does not ensure sealing of internal defects [43]. Discontinuities in forgings produced from complex-alloyed steels with a small *D/H* ratio cannot be welded up by upsetting. In such cases (when ingots having well-developed shrinkage defects are used for forgings without central holes), preliminary drawing of the ingot to weld up axial defects before upsetting is obligatory.

Welding up of defects in an ingot by upsetting alone must be regarded as a real possibility only for readily weldable plastic steels with a ratio *D/H* > 3-4. The minimum necessary reduction ratio is only slightly greater than the critical degree of deformation determined experimentally (for closing defects) and corresponds to an upsetting ratio of about 2.0-3.0.

The influence of intermediate upsetting on the dynamics of the process by which internal defects in an ingot are welded up during subsequent forging operations is very important in the practical sense. The fact that defects in the center of an ingot with a ratio *D/H* ≈ 1 are enlarged during upsetting leads us to conclude that intermediate upsetting makes welding up of defects during drawing more complicated than is the case when the ingot is immediately drawn. This has been confirmed by a number of investigations and by industrial practice. Specifically, experiments involving forging of a 34XH1M ingot weighing 19.5 t under different technological regimes established that a reduction ratio of 2.0-3.0 was sufficient for welding up of internal defects when only drawing was employed, while complete sealing was achieved only at a reduction ratio of about 4.0 when the ingot was subjected to preliminary upsetting [41]. Work conducted at the NPL on ЭИ802 complex-alloyed high-chromium steel showed there to be an even greater difference in the reduction ratios required for welding up defects in drawn and upset billets. More effective welding up of defects during drawing without preliminary upsetting has been repeatedly observed in producing large forgings from other special steels. This is due both to development of defects in the ingot during preliminary upsetting and to the less favorable conditions for drawing of an upset billet in comparison with a drawn ingot.

The principal conditions and deformation regimes required to ensure sealing of internal defects in the ingot have been described above. The same conditions, which are based on the requirement that tensile stresses be absent throughout the entire blank undergoing deformation, are also the main factors that prevent formation of internal tears in low-plasticity steel forgings during drawing.

We should also mention certain characteristics and causes of the formation of defects in forged wheels with hubs fabricated by upsetting and subsequent hammer-chasing of the barrel. Wheels up to 1200-1400 mm in diameter, which are widely employed in electrical machine building, are usually produced in presses, with the barrel formed by narrow separable die blocks after the blank has been upset to a height corresponding to that of the forged wheel at its hub.

Radial cracks are sometimes formed in the central portion of the lower face in forging wheels from austenite and other low-plasticity steels (ЭИ405, ЭИ572, etc.). Formation of annular cracks at the site of the transition from the hub to the barrel is also observed in some cases.

Analysis of this phenomenon [35, 44] has established that the cracks are formed under the action of the radial tensile stresses that develop as a result of the fact that metal flow during chasing of the barrel is principally in the tangential direction. The smaller the ratio of the width of the die block to that of the wheel body, the more rapid is the metal flow in the tangential direction and, consequently, the greater are the tensile stresses produced in the central region of the wheel. This phenomenon is also facilitated by an additional factor, the small reduction ratio, which produces a substantial difference in the degrees of body deformation at the top (facing the narrow die blocks) and at the bottom (facing the plate).

In order to avoid cracking in forging wheels with hubs from low-plasticity steels, it is therefore desirable to form the body by upsetting on a liner ring and chasing from the side opposite the hub rather than by chasing with narrow die blocks from the hub side. This forging pattern prevents development of radial tears in the central portion of the wheel or of annular cracks where the hub meets the body.

When it is feasible in technological and production terms to use the forging pattern described above, i.e., upsetting and chasing of the body on the hub side, the danger of cracking can be substantially reduced by increasing the width of the die block, employing a greater reduction during each press stroke, and reducing the relative difference in the thicknesses of the forged wheel at the hub and body by somewhat increasing the thickness of the latter. In forging thin wheels, it is desirable that the body be chased at relatively high forging temperatures with the blank subjected to uniform heating, in order to avoid development of additional tensile stresses.

Cases are often encountered in industrial practice where discontinuities in the metal of special-steel wheels, particularly annular cracks, are detected after the wheel has cooled or undergone heat treatment rather than during deformation or immediately after forging. This happens when the tensile stresses produced during chasing of the body do not exceed the ultimate strength of the metal and large residual internal stresses are present in the forging, summing with the thermal stresses that develop during cooling or heating of the blank and eventually being manifested in defects in the metal.

Chapter 5

HEAT TREATMENT OF FORGINGS

13. COOLING AND PRIMARY HEAT-TREATMENT REGIMES

Steels that do not undergo phase transformations, particularly those of the austenite class, are cooled in air after forging and are usually not subjected to any type of preliminary heat treatment. Most forgings produced from special steels of the perlite class and high-chromium martensite steels (especially those with large cross-sectional areas) are cooled in the furnace under a regime that can be combined with one of the operations involved in primary heat treatment: relief annealing or annealing accompanied by phase recrystallization. Small forgings, for which accelerated cooling presents no danger of formation of flakes or thermal cracks, are cooled in air and then subjected to softening or recrystallization annealing or sent for machining or final heat treatment without any preliminary thermal operations, depending on the hardness of the forging, the purpose for which it is intended, and the subsequent processing that it is to undergo.

The main purpose of primary heat treatment of forgings fabricated from high-hot-strength perlite steels (ЭИ415, 15Х1М1Ф and Р2) is to prevent flaking. Long-term industrial experience and numerous investigations have established that the principal factor responsible for flaking in steel is hydrogen, which creates large pressures in the cooled steel and reduces its plastic properties as a result of hydrogen embrittlement. Forgings produced from acid open-hearth steel, whose hydrogen content is lower than that in steels smelted in basic open-hearth and arc furnaces by an average factor of 1.5, are consequently less susceptible to flaking. Internal stresses in the steel, particularly those created during phase transformations, are an additional factor that intensifies flaking.

On this basis, the main role of heat treatment of forgings to prevent flaking is to ensure a maximum hydrogen-diffusion rate and a reduction in the average hydrogen content of the steel, as well as a more uniform hydrogen distribution over the forging cross-section as a result of diffusion from areas with high concentrations to those with low concentrations. The latter factor is especially important, since the hydrogen content in the most brittle liquation zones of the forging is greatly reduced. The primary heat-treatment regime should simultaneously reduce residual stresses (structural, thermal, and forging) to a minumum.

Alloy steels containing nickel, particularly in amounts of more than 2% (34XH3M, 18X2H4BA, etc.), have the greatest susceptibility to flaking. Such steels are distinguished by high austenite stability in the perlite region and large forgings fabricated from them therefore require quite prolonged heat treatment and cooling. The basic regime involves long-term isothermal holding at 600-650°C after cooling at 220-300°C. The isothermal holding time has been set at 10-15 h for large forgings of chromium-nickel-molybdenum steels and 20-30 h for forged turbine rotors with a diameter of up to 100 mm [5]. Large forgings are cooled from the isothermal annealing temperature to 100-300°C with the furnace. In most cases, forgings that are to undergo subsequent refinement are not subjected to recrystallization annealing.

High-hot-strength perlite steels (15X1M1Ф, ЭИ415, and P2), which do not contain nickel, are substantially less susceptible to flaking than chromium-nickel-molybdenum steels. The austenite in these metals has a relatively low stability and is converted to perlite, i.e., a structure in which hydrogen diffusion occurs at high speed during relatively short holding at temperatures in the perlite region [45]. These types of steel therefore do not require special prolonged cooling, as do chromium-nickel-molybdenum steels, and the isothermal holding time can be relatively short.

Rapid cooling before isothermal annealing is carried out only in order to obtain accelerated or more complete cooling of the central region of the forging to the temperature at which the austenite transition is most rapid. The isothermal annealing is the most important element of the heat-treatment regime, governing its effectiveness. In establishing holding times for forgings, it is necessary to keep them proportional to the forging cross-section, except for especially large or especially critical forgings (e.g., forged rotors of P2 steel), for which the relative holding time should be somewhat longer. Forgings of comparatively small diameter can be cooled from the isothermal annealing temperature in air. It must be noted, however, that not all plants have adopted the same regime for the initial and high flaking treatment. As P.V. Sklyuyev quite correctly notes [46], only the method used to select proper regimes can be universal, the actual regimes differing substantially in accordance with the method used to smelt the steel and other metallurgical factors. Drawing on experience is therefore the critical factor and the proposed regime should be subjected to thorough experimental verification under the specific conditions that obtain at the plant in question.

Chapter 6 describes the primary heat-treatment regimes for certain typical electrical-machinery forgings produced from ЭИ415, 15X1M1Ф (basic electric steel), and P2M (vacuum-poured acid open-hearth steel) steels.

Forgings fabricated from martensite chromium steels containing 12-14% chromium are usually subjected to modification annealing, which is combined with cooling in the furnace for large components. Such forgings have no tendency toward flaking. The cooling regime therefore provides for no special procedures intended to accelerate hydrogen diffusion or other antiflaking measures. The main purpose of primary heat treatment for simple martensite chromium steels i..

softening, bringing their hardness to a level that permits machining, and prevention of cracking.

There is a transformation to the martensite region when forgings are cooled in air or are not cooled sufficiently slowly with the furnace: the steel becomes quite hard and, as a result of the structural transformations, is in a highly stressed state. If the forgings are not quickly tempered in this case, there is a possibility that defects will develop in the metal.

The tendency toward cracking in forgings of 1X13-3X13 steel during cooling in air depends on a number of factors: the final deformation temperature, the forging cross-section and state, and the metallurgical characteristic of the melt. The decisive factors are the forging dimensions and configuration. The more massive the forging, the larger are the internal stresses that develop in it and, consequently, the more dangerous is cooling of such a forging in air (from the standpoint of possible cracking). The influence of the component configuration is exerted through the fact that cracks always coincide with the direction of least plasticity, so that forgings with marked anisotropy of plasticity have a greater tendency toward cracking.

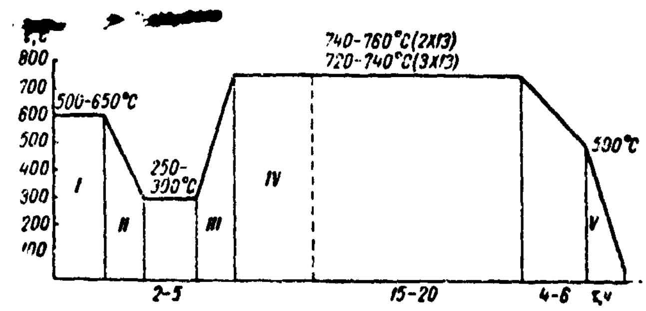


Fig. 12. Typical pattern for primary heat treatment of forgings 500-600 mm in diameter fabricated from 2X13 and 3X13 steels. I) Accumulation of forgings; II) cooling with hood and dampers open; III) Heating to isothermal annealing temperature (heating rate unrestricted); IV) equilibration of metal temperature in charge; V) cooling in air. 1) h.

Forgings of two shapes, shaft and wheel, can be used as an example. Forged shafts, which are usually produced by a single drawing operation, exhibit a pronounced longitudinal fiber direction and, consequently, greatly reduce transverse plasticity. Forged wheels are generally formed by a combination of two operations (drawing and upsetting) and, as a result of the characteristics of the upsetting process, their fibers do not exhibit any marked tendency to run in the same direction. Given the same diameter and similar cooling conditions, cracks should therefore develop principally in forgings of the shaft type (in the longitudinal direction), as has been consistently confirmed by industrial experience.

Forgings with diameters of up to 100-120 mm can be cooled in still air. Accelerated cooling of such relatively small forgings usually does not cause cracking. Forgings cooled in air have high hardness and modificational annealing can be omitted only if they are to be subjected to final heat treatment in untrimmed form. If this is not the case, softening annealing is obligatory.

Different plants have resolved the question of whether high-chromium steels should be annealed with or without recrystallization in different ways. Both types of annealing are employed. However, research has shown that high-temperature annealing of such steels is unwise and even detrimental, since rapid precipitation of the chromium carbides that persist in the metal structure during slow cooling are precipitated at the grain boundaries over the critical temperature range; the carbides do not go into solid solution during subsequent prequenching heating and reduce the impact strength of the steel.

Isothermal annealing at the effective austenite-decomposition temperature, which is about 700-760°C for 2X13 and 3X13 steels, is best. Appropriate holding at this temperature and subsequent cooling in air ensure that the steel has suitable hardness and the best possible preliminary structure.

Large forgings are cooled under a special regime combined with isothermal annealing: the hot forgings are accumulated in a furnace at 500-600°C and, after cooling, are subjected to isothermal holding at 700-760°C. The hardness obtained after heat treatment usually does not exceed 220-240 HB. Figure 12 shows a typical primary heat-treatment regime employed at many plants for forgings of 2X13 and 3X13 steels up to 500-600 mm in diameter.

Forgings of 1X17H2 (ЭИ268) steel and 12 per cent chromium steels are not susceptible to flaking but, as a result of their structural characteristics, require very complex and prolonged heat treatment and cooling in order to obtain satisfactory hardness and to reduce internal stresses to a minimum. Isothermal annealing of these steels yields effective results when the heat-treatment regime includes two cooling periods, the first at 200-250°C and the second at 150-200°C. The decomposition products obtained during the two cooling periods must be tempered at a temperature below Ac_1, at 650-680°C. The hardness obtained for the steel in this case does not exceed 240-250 HB.

Forgings of 1X17H2 steel are not subjected to recrystallization annealing. Post forging cooling is carried out in the furnace in combination with isothermal annealing, taking into account the aforementioned structural characteristics of the steel. Specifically, postforging heat treatment of wheels fabricated from 1X17H2 steel (with a hub height of up to 150 mm) is carried out under the following regime: holding at 650-690°C until the metal temperature is fully equilibrated throughout the entire charge, cooling to 200-230°C at a rate of 30-40 deg/h, holding at 200-230°C for 4 h, heating to 640-680°C, holding at this temperature for 15-20 h, cooling to 160°C at a rate of 30-40 deg/h, holding at this temperature for 5 h, reheating to 640-680°C, holding at this temperature for 15 h, cooling with the furnace to 200°C at a rate of 30 deg/h, and further cooling in air. Accelerated cooling in air (from the forging temperature) of even relatively small forgings often causes cracking of the metal.

In contrast to 1X13 and 3X13 steels, forgings fabricated from high-alloy high-chromium steels (ЭИ802 and 15X11МФБ) have a tendency toward hydrogen embrittlement and formation of flake-type defects in their central regions, which is apparently due to the far higher stability of the supercooled austenite.

On the basis of a study of the kinetics of the isothermal decomposition of austenite, the most effective primary heat-treatment regime for hot forgings of ЭИ802 steel is assumed to include an initial cooling from the accumulation temperature of 600-700°C to 250-400°C, subsequent isothermal annealing at 690-710°C, a second cooling to 180-200°C, isothermal holding, and slow cooling with the furnace to 100-200°C. The presence of two cooling periods and isothermal annealing at 690-710°C ensures gradual and quite complete decomposition of the austenite and tempering of the decomposition products with relatively small structural stresses.

The individual elements of the heat-treatment regime (primarily the isothermal holding time) are selected on the basis of the metallurgical and geometric characteristics of the forging. For example, satisfactory results have been obtained in primary heat treatment of wheels with a hub height of more than 300 mm forged from basic electric steel without vacuum casting under the following regime: an initial cooling from 700 to 370°C with the furnace and then to 250° in air, holding at the latter temperature for 9 h, heating to 710-730°C, isothermal annealing (after heating) for 20 h, a second cooling to 190-200°C, holding at this temperature for 9 h, isothermal annealing at 710°C for 20 h, and slow cooling with the furnace to 110°C.

It is interesting to note that flaking was detected in wheels of this type forged from ingots produced by electric-slag remelting and heat-treated after forging under a regime involving a single cooling at 250-300°C and subsequent annealing at 700-710°C for 15 h. When the hydrogen content and other metallurgical characteristics of the ingots are favorable, there is obviously less reason to expect flaking in the forgings and the primary heat-treatment regime can be less prolonged. All procedures intended to reduce the hydrogen content of complex-alloyed high-chromium steels (use of vacuum-melted ferrochromium, vacuum casting, etc.) ar obviously very important factors in this sense, affecting both forging quality and the duration of the initial heat-treatment process in the forging-pressing shop.

The primary heat-treatment regimes for forgings of complex-alloyed high-chromium steels do not provide for phase recrystallization. Experience has demonstrated that recrystallization at 880-900°C causes an abrupt decrease in the impact strength of the steel. According to the data of the Central Scientific Research and Planning-Design Boiler and Turbine Institute, this phenomenon is caused by precipitation of chromium carbides, which are of low solubility at the quenching temperature, and their accumulation in coagulated form along the grain boundaries.

14. FINAL HEAT TREATMENT

Forgings of High-Hot-Strength Perlite Steels

Selection of heat-treatment regimes for high-hot-strength perlite steels is complicated by addition of an extra requirement above those for ordinary alloy structural steels intended to operate at normal temperatures: the metal must have optimum hot strength under given working conditions. The most favorable combination of mechanical properties for ordinary steels is achieved by quenching and tempering. Modification of high-hot-strength steels does not always yield optimum properties and it is often necessary to seek other, more effective heat-treatment regimes.

Recent investigations [47, 48] have established that steels with a bainite structure have the highest hot strength, so that normalization and high tempering is regarded as a more suitable type of heat treatment than improvement for the very common chromium-molybdenum-vanadium perlite steels. One factor that makes normalization more suitable for large forgings than quenching is the smaller thermal and structural stresses produced during heat treatment, which promotes development of fewer metallurgical defects.

In order to obtain the requisite mechanical properties in large forgings, particularly impact strength, normalization is sometimes supplemented by special rapid cooling of the metal in air. This technique is employed, for example, in heat-treating forged rotors of P2 steel, which are normalized in special chambers with forced air circulation. A more complex processing cycle, involving double normalization and tempering, is sometimes used; it provides greater plasticity and homogeneity of the metal in large forgings and a lower notch sensitivity in long-term strength tests than does single normalization [47].

In view of the other advantages of normalization and tempering (simplification of the heat-treatment process, the possibility of more general use of heat-treatment furnaces, and improvement of working conditions in the heat-treatment shop), this technique is best in all cases where the technical requirements imposed on the forgings permit it. Of the high-hot-strength perlite steels mentioned in Chapter 1, most chromium-molybdenum-vanadium steels of the 15X1M1Φ and P2 types are subjected to single or double normalization. Modification is employed only when the forgings must have elevated mechanical properties, e.g., in heat-treating wheels of P2 steel to obtain $\sigma_{0.2} \geq 75$ kgf/mm^2 or in treating large components of 15X1M1Φ steel. Forgings of ЭИ415 steel are usually subjected to modification. Final heat treatment in the form of a single modification (without subsequent tempering) is completely impermissible, even when there is no danger that large internal stresses will develop, since this procedure greatly reduces the hot strength of the steel.

The prenormalization or prequenching heating temperature for the aforementioned high-hot-strength perlite steels has been set at 100-120°C above Ac_3. Heating to a temperature that substantially exceeds the upper critical point is dictated by the fact that these steels contain vanadium (and tungsten as well for ЭИ415 steel), which forms carbides with a high dissociation temperature, and by the need for maximum solution of these elements in the austenite and enlargement of the grains in order to maintain the hot strength of the metal. Investigations conducted at the Central Scientific Research Institute of Technology and Machine Building have shown that, for example, 15Х1М1Ф cannot be normalized at 930-960° C (as was previously done), since this results in low hot strength despite the very high plasticity it yields [49]. Raising the normalization temperature to 1030-1050°C somewhat reduces metal plasticity but substantially increases hot strength.

Normalization at higher temperatures has also been found to have a favorable effect on research on experimental forgings of P2 steel [48]. The originally stipulated normalization temperature of 920°C was raised to 980°C, which gave the metal a higher hot strength.

The tempering temperature, which should ensure complete decomposition of the nonequilibrium structures and relief of the internal stresses, plays a very important role in determining the hot strength and structural stability of steels. It is of fundamental importance to establish a tempering temperature that is as far above the working temperature of the forging in question as possible. The structural changes in the metal under the action of operating conditions will be minimal in this case. However, the maximum tempering temperature is limited by two factors: the location of the point Ac_1 and the requirements imposed on the strength of the metal. The attempt to obtain as high a tempering temperature as possible therefore inevitably involves treating the component to the minimum permissible strength and, consequently, with a high degree of process accuracy.

The characteristics of each specific component must be taken into account in selecting the optimum tempering temperature. When a component has abrupt changes of shape, for example, a high tempering temperature is desirable to reduce the notch sensitivity of the steel by increasing its plasticity to as great an extent as possible.

The influence of tempering time on the strength of steel at normal and elevated temperatures and on its hot strength is similar to the influence of tempering temperature. Both elements of the tempering regime should therefore be established experimentally in strict correlation with one another.

The Central Scientific Research Institute of Technology and Machine Building [TsNIITmash](ЦНИИТмаш) investigated the influence of tempering temperatures in the range 700-760°C and holding times of from 2 to 12 h on the hot strength of pipe blanks fabricated from 15Х1М1Ф steel [49]. The optimum combination of hot strength and plasticity was obtained after normalization from 1030-1050°C and tempering at 700-720°C for 5 h. However, this regime is recommended only for 15Х1М1Ф steel containing no more than 0.13% carbon. When the carbon content of the melt is greater than 0.13%, the tempering temperature should be raised to 740-760°C.

According to the data of the Leningrad Metal Plant [LMP](ЛМЗ), the optimum heat-treatment regime for large forged flanges of

15Х1М1Ф steel is normalization from 1000°C and tempering at 720°C for 8-10 h or quenching from 1000°C in oil and tempering at 740-750 for 10 h. Both regimes yield satisfactory results with respect to mechanical properties at normal temperatures and hot strength. Large forgings subjected to normalization and temperature and having $\sigma_{0.2}$ = 30-32 kgf/mm^2 exhibit the following approximate characteristics: a yield strength at 565°C of 6 kgf/mm^2, a long-term strength of 7-8 kgf/mm^2, and a very high long-term tensile plasticity. Quenching and tempering raise $\sigma_{0.2}$ to 37-50 kgf/mm^2 and yield a higher long-term strength. Modification is usually employed for large forged flanges, T-joints, pipes, and other components with large cross-sections.

Type P2 steel is used principally for large forged steam-turbine blades. The heat treatment of the forgings consists of double normalization from 970-990°C and 935-945°C and tempering at 680-710°C for 20-30 h (after heating), followed by slow cooling with the furnace. The first normalization produces more complete carbide solution and structural homogenization. The second normalization promotes a decrease in grain size and in the notch sensitivity of the steel. As was pointed out above, double normalization increases metal plasticity and its homogeneity over the forging cross-section.

The cooling rate during the second normalization has a strong effect on the ultimate results obtained in heat-treating large forgings of P2 steel. The initial experiments involving cooling of forged rotors in still air showed this regime to be unsuitable, since it reduced the impact strength of the metal: about 80% of the rotors had an impact strength below the norm stipulated by technical specifications ($\geq$4.5 kgf·m/cm^2). Introduction of forced cooling of forgings by fan-supplied air from a special apparatus at the Urals Machine Building Plant provided the requisite impact strength and increased the homogeneity of mechanical properties throughout the entire forging volume. In this case, the strength and plasticity of the rotors usually were characterized by the following values: $\sigma_{0.2}$ = = 50-65 kgf/mm^2, δ = 17-21%, and ψ = 55-65% [50]. Heat treatment of P2 steel to a yield strength of more than 75 kgf/mm^2 (a process employed for certain types of centrifugal compressor wheels) involves quenching from 940-960°C in oil and tempering at 650-670°C.

The principal type of heat treatment employed for large forgings of ЭИ415 steel is quenching from 1000-1030°C in oil and subsequent tempering at 660-700°C. Narrower ranges are dictated for the tempering temperature and time by the strength requirements imposed on the forgings by technical specifications. Specifically, turbine wheels with a hub height of 300 mm treated to $\sigma_{0.2} \geq$ 62 kgf/mm^2 are tempered at 680-700°C. The actual yield strength reaches moderate values (65-70 kgf/mm^2), while the indices δ, ψ, and a_n are 15-18%, 50-60%, and 6-10 kgf·m/cm^2 respectively.

In order to obtain more complete solution of the vanadium carbide, it is recommended that the heat-treatment cycle be supplemented by preliminary normalization from 1050-1100°C, which promotes an increase in the plasticity and impact strength of the steel. Large one-piece rotors are subjected to such treatment. Forgings of relatively small cross-section can also be treated by normalization and tempering, since the steel has good hardenability.

The post tempering cooling regime for forgings of high-hot-strength perlite steels depends on the size, configuration, and purpose of the components. Small forgings are cooled in air. Large forgings, especially those of complex shape, are cooled slowly with the furnace to 200-300°C, in order to avoid the internal stresses that develop in such forgings when they are rapidly cooled from the tempering temperature in air.

The final heat-treatment cycle is sometimes supplemented by tempering at a temperature 30-50°C below that of the post quenching temperature, in order to relieve residual stresses. This operation is employed for massive forgings in those relatively rare instances where they are heat-treated in untrimmed form (without preliminary machining) or when the component contains elements with very thin cross-sections and complex changes in shape. In such cases, even the relatively small residual stresses produced in the forging by removal of a large amount of metal during machining (especially when the distribution of the material removed is nonuniform) can cause an unfavorable stress distribution and impermissible deformation of the component. When necessary, tempering for stress relief is carried out as an intermediate stage of machining (leaving the minimum tolerance necessary to bring the component to its final dimensions).

Forgings of Chromium Stainless and High-hot-Strength Martensite Steels

Heat treatment of forgings fabricated from chromium martensite steels usually consist in quenching and tempering. Small forgings are quenched in air, while large forgings are generally quenched in oil, in order to obtain a more complete martensite transformation. Double quenching and subsequent tempering is sometimes employed to raise the impact strength of the steel. For more effective working of the metal in the forging, it is usually trimmed before the final heat treatment, leaving only the minimum allowances necessary to compensate for possible deformation of the component. Heat treatment of untrimmed forgings often causes cracking in areas where superficial defects are concentrated, especially during quenching in oil.

The quenching temperature is selected so as to yield the most complete carbide solution, so that it substantially exceeds Ac_3, ranging from 1000 to 1100°C for most high-chromium steels. The minimum temperature is for 1X13 and 3X13 steels, while the maximum is for steels alloyed with strong carbide-forming elements, such as vanadium and niobium. Quenching from lower temperatures reduces the effectiveness of heat treatment, especially for complex-alloyed steels. For example, the TsNIITmash has noted that raising the quenching temperature for wheels of X12B2MΦ (ЭИ756) steel from 1040 to 1060-1070°C increases all their mechanical properties, including their strength. The favorable effect of high-temperature quenching on the mechanical properties of modified high-chromium steels (both short-term and long-term) has also been confirmed by other experimental data [47]. Extremely high quenching temperatures are not recommended, however, since there is a danger of intensive grain growth and an increase in the amount of structurally free ferrite, phenomena that reduce the plasticity and impact strength of the steel. Selection of the optimum quenching temperature for forgings fabricated from a given type of steel requires careful analysis of the location of the δ-ferrite region in the phase diagram (taking into account possible deviations in the contents of individual elements within the limits prescribed by the chemical composition of the type of steel in question) and experimental determination of the critical temperature at which rapid grain growth begins.

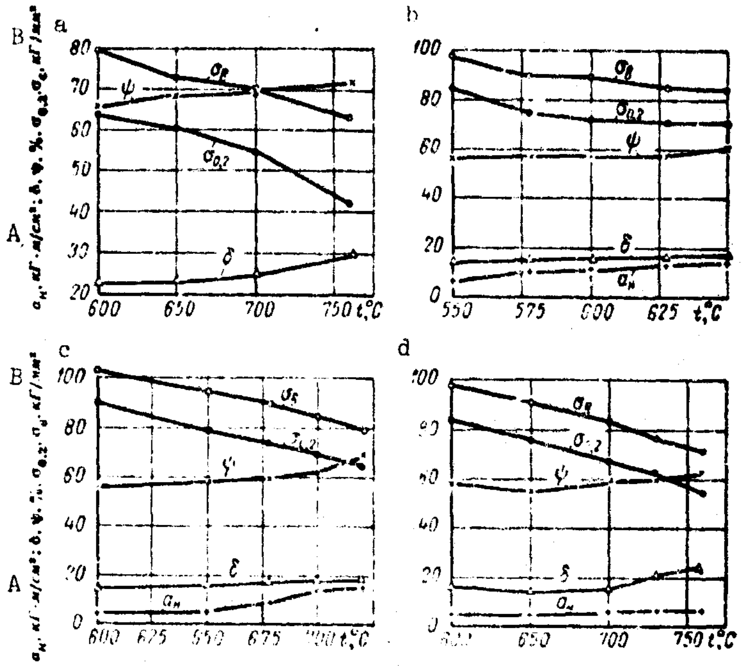


Fig. 13. Mechanical properties of different steels at 20°C as a function of tempering temperature *t*. a) 1X13 steel; b) 1X17H2 steel ([NKMP](HKM3) data); c) ЭИ802 steel [1]; d) X12B2MФ steel (TsNIITmash data). A) kgf•m/cm²; B) kgf/mm².

Laboratory experiments have established that the phase transformations that take place in steels of the type under consideration during prequenching heating go to completion during the first hour [51], so that a longer heating time is not required. However, in view of the possibility of deviations from a precise temperature regime under industrial conditions, the holding time at the quenching temperature should be 2-3 h after heating is complete for large forgings and 6 h for forgings 800-1000 mm in diameter.

During tempering of a steel, the martensite decomposes to form intermediate structures and carbides are simultaneously precipitated from solid solution. The joint action of these two factors affects the mechanical properties of the steel at normal temperatures and its hot strength.

The first criterion to be considered in selecting the tempering regime is the combination of strength and plasticity that the steel must have. The proper combination is achieved by choosing an appropriate tempering temperature and time. Figure 13 shows the variation in the mechanical properties of certain high-chromium steels as a function of tempering temperature. In bringing critical forgings into production, the optimum tempering temperature should be verified experimentally in combination with a definite holding time, using actual components.

The tempering time for forgings of moderate size (up to 300-400 mm in diameter) usually ranges from 2 to 6-8 h and, like the tempering temperature, directly affects the mechanical properties of the component. For example, an increase in tempering time from 2 to 10 h (at 725°C) for 18X11MФБ steel reduces its yield strength from 68-78 to 60-67 kgf/mm² and increases its relative elongation from 12-17 to 16-21% and its impact strength from 10-13 to 13-15 kgf·m/cm² [47]. In setting up and putting into practice a heat-treatment regime, it is therefore very important to correctly determine the instant at which the tempering temperature is equilibrated throughout the entire forging cross-section, so that the actual holding time corresponds to the time provided by the regime.

The general pattern in the variation in the mechanical properties of high-chromium martensite steels as a function of tempering regime is the same at elevated temperatures (up to 500-600°C) as at normal temperatures. The relationship between tempering temperature and long-term strength is essentially analogous to the relationship between tempering temperature and strength during short-term tests [47]. However, research has established [52] that the optimum combination of hot-strength properties in high-chromium steels is yielded by a combination of quenching (which is best double) and high tempering, which yields moderate strength characteristics. Tempering to very high strengths is impermissible from the standpoint of high-temperature deformability. In particular, the technical specifications for 15X11MФБ steel, which the Leningrad Metals Plant employs for steam-turbine housing components, stipulate a maximum yield strength (68 kgf/mm²).

The tempering temperature should naturally be higher than the working temperature of the component in question, in order to avoid softening of the metal. The tempering-temperature range 600-750°C is therefore of practical interest for high-chromium high-hot-strength steels. In addition to development of definite mechanical properties, effective relief of internal stresses takes place at this temperature, which is of no small importance for large forgings. As for steels used principally as corrosion-resistant materials, there is usually a tendency in this case to select as low a tempering temperature as possible, thus achieving maximum corrosion resistance [4]. A low tempering temperature is sometimes selected in order to obtain high steel strength, even though it results in reduced plasticity and impact strength. The large internal stresses that develop in large forgings during quenching and the relatively high tempering temperatures required to relieve them are not always taken into account in this case. As a result, cracks and local discontinuities can develop in large forgings of complex shape under the action of internal stresses. One such case occurred in tempering forged wheels of 2X13 steel about 500 mm high to a hardness HB $\geq$ 280 (the tempering temperature was 530°C). Deep tears appeared in the metal as a result of the substantial internal stresses and their redistribution during finish machining.

Post tempering cooling of chromium martensite steels takes into account the characteristics of the forging: the size, shape, and purpose of the components and the characteristics of the type of steel in question. Types 1X13, 2X13, and 3X13 steel are subject to temper brittleness. A substantial decrease in impact strength occurs at tempering temperatures of about 500-550°C. When these steels are tempered at higher temperatures, accelerated cooling of the forgings (at least in air) is desirable. Depending on the individual characteristics of the melt, slow cooling of the forgings with the furnace causes impact strength to decrease by 5-20% in comparison with the levels obtained after accelerated cooling.

However, the absolute value of a_n is sufficiently high in most cases, even after slow cooling with the furnace, so that large forgings of complex shape should be cooled slowly in order to avoid large residual stresses and the resultant possibility of component deformation during finish machining. As an example, we can cite the case of forged shafts 80-150 mm in diameter and about 2 m long fabricated from 2X13 steel and treated to $\sigma_{0.2} \geq 45$ kgf/mm². In many cases, shafts cooled in air after tempering exhibit impermissible deformation after final machining. Slow cooling with the furnace reduces the impact strength from 8-12 to 7-10 kgf·m/cm² but prevents distortion of the shafts. A similar pattern has been shown to exist for compressor wheels fabricated from 2X13 steel. The presence of a thin body, which tends to warp during machining as a result of redistribution of the residual stresses, makes it necessary to cool the forgings slowly with the furnace, despite the fact that this somewhat reduces their impact strength. Small forgings or more massive forgings of rigid design are cooled in air after tempering.

Complex-alloyed steels based on 11-14% chromium and 1X17H2 steel exhibit no marked tendency toward temper brittleness. They are therefore cooled in air or slowly with the furnace after tempering, depending on the size of the component.

Table 7 shows typical heat-treatment regimes for certain large forgings produced from high-chromium martensite steels and used in electrical machine building.

The prequenching heating regime for large forgings provides for stacking in the furnace at a temperature not exceeding 300-400°C and a slow rise in temperature to 700-800°C, since these steels have low thermal conductivity. As a result of the larger thermal stresses produced, accelerated heating can cause cracking in the central zone of the forgings. Holding at 700-800°C is desirable to equilibrate the temperature over the forging cross-section. Heating from 700-800°C to the quenching temperature is carried out rapidly, since the thermal conductivity of the steel is markedly increased and this factor no longer has any effect. It is desirable that the heating time in the high-temperature region be as short as possible, in order to avoid extreme grain growth.

The time for which the blank is held in oil during quenching should ensure that its central region is cooled to the temperature necessary for complete decomposition of the austenite, e.g., 150-170°C for ЭИ802 steel.

Pretempering heating of large forgings is conducted at slow speeds, not exceeding 60-70 deg/h for components with a cross-section of 500-700 mm and 30-40 deg/h for components with a cross-section of more than 800 mm. The furnace temperature during charging generally does not exceed 400-500°C. Such heating regimes are dictated by the attempt to reduce internal stresses to a minimum, since large stresses can cause cracking and warping of the components. For the same reason, post tempering cooling with the furnace is carried out at rates not exceeding 30-40 deg/h (depending on the blank cross-section). We must again emphasize the pattern found for the influence of cooling rate (from the tempering temperature) on the residual stresses in forgings [53], which decrease with the rate at which the forgings are cooled through the temperature region in which the elastic-plastic state exists, i.e., from about 750 to 400°C. The influence of the cooling rate at lower temperatures is slight in most cases and is of no serious practical significance. It is therefore generally inexpedient to cool forgings with the furnace to very low temperatures, e.g., 100-150°C, as is done at some plants. Even very large forgings can be cooled in air from 250-300°C.

TABLE 7

Heat-Treatment Regimes for Certain Forgings Produced from High-Chromium Martensite Steels

1 Марка стали	2 Характеристика поковки	3 $\sigma_{0,2}$ по техническим условиям, кГ/мм² (не менее)	4 Закалка: 5 Температура, °C	6 Длительность выдержки, ч	7 Среда	8 Отпуск: 5 Температура, °C	6 Продолжительность, ч	9 Охлаждение после отпуска
2X13	10 Вал с диаметром бочки 150 мм длиной 2500 мм	45	1020—1040	1	19 Воздух	760—780	4	21 В печи до 300° С
	11 Диск с размером сечения по ступице 120 мм	55			20 Масло	630—650		
	12 Ротор с диаметром бочки 750 мм	45	1040—1060	5		760—780	8	
3X13	13 Диск высотой 200 мм	62	1010—1030	1		625—635	6	22 На воздухе
1X17H2 (ЭИ268)	14 Диск с высотой ступицы 100 мм	65	960—980			650—670	4	21 В печи до 300° С
X12BHMФ (ЭИ802)	15 Диск с высотой ступицы 340 мм		1040—1050	3		650—680	8	
	16 Ротор с диаметром бочки 700 мм	60	1050—1070	5		690—710	20	23 В печи до 250° С
15X11MФБ	17 Патрубки, тройники и другие детали с поперечным сечением до 400 мм	55 (18 не выше 68)	1100—1130 1060—1080	1.5	19 Воздух	740—760	8	21 В печи до 300° С

1) Type of steel; 2) forging characteristics; 3) $\sigma_{0,2}$ from technical specifications ,gf/mm² (no less than); 4) quenching; 5) temperature,

°C; 6) holding time, h; 7) medium; 8) tempering; 9) post tempering cooling; 10) shaft with body diameter of 150 mm and length of 2500 mm; 11) wheel with cross-section of 120 mm at hub; 12) rotor with bevel diameter of 750 mm; 13) wheel 200 mm high; 14) wheel with hub height of 100 mm; 15) wheel with hub height of 340 mm; 16) rotor with bevel diameter of 700 mm; 17) pipes, T-joints, and other components with cross-sections of up to 400 mm; 18) but no greater than; 19) aid; 20) oil; 21) with furnace to 300°C; 22) in air; 23) with furnace to 250°C.

Forgings of Austenite Steels

The heat-treatment regime for forgings of austenite steels usually consists in two principal operations: austenitization (quenching)and stabilization tempering.

The purpose of the first operation is to dissolve the hardening phases (carbides, intermetallic compounds. etc.) in the γ-solid solution and saturate it with alloying elements. The optimum hardening-phase solution temperature depends on the individual characteristics of the alloy, but a general pattern observed in the process that occurs when an austenite steel is heated is that solution and coagulation of submicroscopic particles begins as the temperature is raised, while solution of the large coagulated particles occurs at a very high temperature (1000-1200°C). There is a simultaneous increase in the size of the austenite grains during high-temperature heating, so that the austenitization temperature should be selected in such fashion as to ensure the requisite hardening-phase solution and yet not cause extremely severe austenite-grain growth. These phenomena are progressive with time and both the austenitization temperature and the holding time therefore play a material role; the holding time at a given temperature should ensure completion of all the reactions associated with solution of the alloying elements in the γ-solid solution and hold the austenite-grain enlargement within suitable limits.

After appropriate holding at the austenitization temperature, the forgings are cooled in a quenching medium at a rate adequate to fix the saturated solid solution produced during heating. The quenching medium for most austenite steels is water. An austenitized steel has a somewhat lower hardness than that yielded by hot mechanical working.

The optimum quenching temperature for many austenite steels is 1100-1200°C. The importance of effective prequenching heating of the forgings must be emphasized. This operation is intended mainly to produce monophasic high-alloy austenite. The higher the temperature to which the steel is heated, the more rapid is solution of the hardening phases and, consequently, the more completely are they driven into solid solution. Since rapid austenite-grain growth begins only after the carbides have dissolved, it is possible to obtain sufficiently complete solution of the phases with only a moderate increase in grain size by selecting the appropriate heating-temperature regime and adhering strictly to it in practice.

The quenching temperature has a strong effect on the mechanical properties of the steel at normal and elevated temperatures, as well as on its hot strength and corrosion resistance. It has been established [47] that quenching from temperatures that ensure complete solution of carbides and other precipitated phases not only substantially increases the plasticity and impact strength of steel at 20°C but also raises its high-temperature strength. However, an overly high quenching temperature is impermissible because of the rapid grain growth and the tendency of some steels in which the γ- and α-forming elements lie along the boundaries of the austenite region toward precipitation of a ferrite.

The holding time at the optimum temperature established for each specific type of steel should ensure that all the various phases are driven into solid solution and yet not be overly long, in order to avoid impermissibly severe grain growth. Proper selection of the heating regimes for forgings of different shapes and configurations, especially those with large cross-sections, is thus very important. The heating regime should be such that the forging temperature at the instant when grain growth begins is virtually the same throughout its entire cross-section, which makes it possible to employ the minimum holding time at high temperatures required for solution of the hardening phases and still obtain relatively small and uniform grain enlargement.

An imperfect understanding of the nature of the internal phenomena that occur in an austenite steel during prequenching heating and the role of the minimum permissible temperature sometimes gives rise to attempts to somewhat reduce the quenching temperature by prolonging the holding time. It must be kept in mind that prolonged holding at a comparatively low temperature leads only to solution of the small hardening-phase particles and coagulation of the large ones [54], having no effect in promoting solution of the large coagulated particles; it is therefore impossible to compensate for a reduction in quenching temperature by prolonging the heating time.

Cooling of the forgings after holding at the quenching temperature should be rapid, in order to fix the supersaturated solid solution, and is usually carried out in water, which ensures that no carbides are precipitated over the range 900-450°C.

The principal elements of quenching regimes for austenite steels are the heating temperature, holding time, and cooling medium, which are established experimentally for each specific type of steel.

There are different reasons for including tempering in the overall heat-treatment cycle for austenite-steel forgings intended for use as high-hot-strength and corrosion-resistant materials. In the first case, tempering (aging) is dictated mainly by the relatively high working temperatures and the need to ensure that the metal has a stable structure and properties under working conditions. During heating and prolonged operation of components at 550-600°C or above, the alloyed γ-solid solution produced by quenching tends to go into a more nearly equilibrium state: carbides and other hardening phases are precipitated from the solid solution, which is sometimes accompanied by various volumetric changes. Moreover, the stress and temperature distributions over the cross-section of a component are nonuniform under machine operating conditions, so that the structural changes that take place are also nonuniform, which reduces component reliability by creating substantial heterogeneity of properties.

Tempering a quenched high-hot-strength steel at a temperature that exceeds the component working temperature by 100-200°C leads to precipitation of the maximum amount of finely dispersed hardening-phase inclusions from the austenite and stabilization of component size and metal structure and properties at lower temperatures. The mechanical properties and hot strength of a quenched steel are increased by aging. The fact that some types of complex-alloyed dispersion-hardening steels, such as ЭИ612, acquire high hot strength is specifically based on this phenomenon.

The principal factors affecting the results of tempering of austenite steels are the tempering temperature and time. These elements of the heating regime are selected on the basis of experimental data, in such fashion as to ensure the requisite dispersion of secondary-phase inclusions and produce the proper quantitative composition, these factors being directly related to the hardening effect and hot strength of austenite steels.

The importance of tempering time must be specially emphasized. Prolongation of tempering is equivalent to an increase in tempering temperature, i.e., the two elements of the regime are interrelated, so that the tempering time should be strictly stipulated when the optimum tempering temperature has been established.

After tempering, relatively large forgings are cooled with the furnace to 200-400°C and then in air. Slow cooling of forgings with the furnace is undesirable in principle, since it can lead to additional solid-solution decay. However, this operation is unavoidable for certain forgings, such as turbine wheels, since tempering also reduces the internal stresses in the component to a minimum.

Large austenite-steel forgings whose decisive characteristic is corrosion resistance rather than hot strength are also tempered after austenitization, but these operations have only one purpose: to relieve internal stresses. The operating temperature for such components is usually low and stabilization tempering is therefore not required. On the other hand, it is very desirable to restrict the heat treatment of the forgings to quenching (austenitization) alone in order to obtain high resistance to intercrystalline corrosion, since steels in this state have the greatest chemical stability. However, large components require tempering to relieve internal stresses, which are impermissibly large after austenitization. Forgings with simple shapes and relatively small cross-sections are subjected only to quenching.

As was pointed out above, the heat-treatment regimes for austenite steels that best conform to component operating conditions are selected experimentally. Table 8 shows quenching and tempering regimes for certain common austenite steels employed as high-hot-strength or corrosion-resistant materials in large forged wheels and other components for stationary gas turbines.

TABLE 8

Heat-Treatment Regimes for Forgings of Certain High-hot-Strength Austenite Steels

1 Марка стали	2 Температура закалки, °C	3 Среда охлаждения при закалке	4 Режим отпуска (старения)	
			5 Температура, °C	6 Длительность, ч
Х18Н9Т Х18Н12М2Т	1080—1100	7 Вода	800—830 750—780	8—15
Х18Н22В2Т2	1050—1080	8 Воздух	720—750	15—20
ЭИ405	1100—1130	9 Вода или воздух	750—770	10
ЭИ572	1150—1180	7 Вода	750—770	15
ЭИ395	1170—1190	7 Вода	700—720 800—820	30 10
ЭИ726	1130—1170	8 Воздух	750—770	25
ЭИ612 ЭИ612К	1080—1100	7 Вода	850—870 700—720	10 35—50

1) Type of steel; 2) quenching temperature, °C; 3) quenching medium; 4) tempering (aging) regime; 5) temperature, °C; 6) time, h; 7) water; 8) air; 9) water or air.

The holding time at the quenching temperature depends on the size of the forging and is usually 2-4 h(after the component cross-section has been completely heated). Tempering is not obligatory for small and medium forgings of X18H9T and X18H12M2T steels but it is necessary for large forgings, in order to relieve residual stresses. In these cases, the tempering time is set in accordance with the size and shape of the component.

The tempering temperature for forgings of X18H9T steel is selected to yield the highest resistance to intercrystalline corrosion. Reducing the tempering temperature to 750-770°C increases the susceptibility of the steel to such corrosion and is therefore impermissible.

The high-hot-strength steels ЭИ612 and ЭИ612К are subjected to double aging at 850-870 and 700-720°C, which is most effective from the standpoint of precipitation of dispersed hardening-phase particles. The duration of the second aging depends on the titanium content of the melt in question: it is reduced as the titanium content increases [2]. Double tempering is generally also employed for ЭИ395 steel.

The preaustenitization or pretempering heating rate and the post-tempering cooling regime depend on the specific type of steel and on the individual characteristics of the forging (size and configuration). Despite the absence of phase transitions, heating to 900-1000°C is carried out at a relatively low rate, because of the low thermal conductivity of austenite steels and the large temperature gradient over the forging cross-section. Further heating at higher temperatures is rapid, while the holding time at the austenitization temperature is minimal. Such heating protects the forged metal from extremely severe grain growth.

The working regimes for heat treatment of gas-turbine wheels with a hub height of about 300 mm fabricated from the high-hot-strength austenite steels ЭИ572 and ЭИ612 (heat treatment after trimming) are given below as examples.

<u>Wheels of ЭИ572 steel</u>

Quenching

1. Loading into furnace heated to temperature not above 600°C.
2. Holding at charging temperature for 2-3 h.
3. Heating to 1000-1050°C at rate of no more than 60 deg/h.
4. Holding at 1000-1050°C for 2 h.
5. Heating at maximum rate permitted by furnace to 1150-1180°C.
6. Holding at 1150-1180°C for 1.5 h.
7. Cooling in water to 200°C.

Tempering

1. Loading into furnace heated to temperature not above 500°C.
2. Holding at charging temperature for 2 h.
3. Heating to 750-770°C at rate of no more than 80 deg/h.
4. Holding at 750-770°C for 15 h.
5. Cooling with furnace to 400°C at rate of no more than 40 deg/h.
6. Further cooling in air.

<u>Wheels of ЭИ612 steel</u>

Quenching

1. Loading into furnace at temperature not above 600°C.
2. Holding at charging temperature for 3-4 h.
3. Heating to 980-1000°C at rate of no more than 50 deg/h.
4. Holding at 980-1000°C for 2 h.
5. Heating at maximum rate permitted by furnace to 1090°C (±10°C).
6. Holding at 1080-1100°C for 1.5 h.
7. Cooling in water to 200°C.

Tempering

1. Loading into furnace at temperature not above 500°C.
2. Holding at charging temperature for 2 h.
3. Heating to 850°C at rate of no more than 80 deg/h.
4. Holding at 850°C (after temperature equilibration) for 10 h.
5. Cooling with furnace to 700°C.
6. Holding at 700°C for 35-50 h (depending on titanium content of melt).

7. Cooling with furnace to 300°C at rate of no more than 30 deg/h.
8. Further cooling in air.

Chapter 6

INVESTIGATION AND EXPERIENCE IN PRODUCTION OF FORGINGS FROM PERLITE SPECIAL STEELS

15. GAS-TURBINE WHEELS OF ЭИ415 STEEL

Among the special features of gas-turbine wheels (Fig. 14) are the lack of a central aperture and a large ratio of hub height to wheel diameter. This wheel shape is regarded as comlex and technologically unsuitable in metallurgical terms, since effective measures for mechanical working of the central zones of the forging, which are formed from the axial regions of the ingot, are very limited.

The wheels are subjected to etching of their faces and ultrasonic quality control. Macrodefects and internal flaws in the metal are impermissible. The metal throughout the entire volume of the wheel should have high mechanical properties in the tangential direction, satisfying the following norms: $\sigma_{0,2} \geqslant 62$ кГ/мм²; $\sigma_в \geqslant 76$ кГ/мм², $\delta \geqslant 11\%$; $\psi \geqslant 35\%$; $a_н \geqslant 4,0$ кГ·м/см².

When the wheel contains no central aperture, these requirements necessitate complete welding up of shrinkage defects in the ingot and creation of deformation conditions that promote radial metal flow in the areas of the forging adjoining the faces.

Solution of these problems for ЭИ415 steel presented no difficulties: use of an ingot weighing 1.5 t and then of one weighing 2.5 t for two forgings and intermediate upsetting of the billet ensured welding up of axial defects in the ingot and provided the requisite working of the central regions of the wheel without any special forging techniques.

Figure 15 shows a forged wheel with its trimmed dimensions, while Fig. 16 shows the technological forging pattern for an ingot weighing 2.5 t (60% yield). All the forging operations were carried out in a press exerting a force of 2000 t over the temperature range 1180-800°C, with four passes. A flat upper die block and a cutaway lower block were used for drawing, while movable die blocks were used for chasing the wheel body. The final forging operations, i.e., upsetting of the blank from 780 to 350 mm and chasing of both faces on a platen and liner ring, were conducted in a single pass.

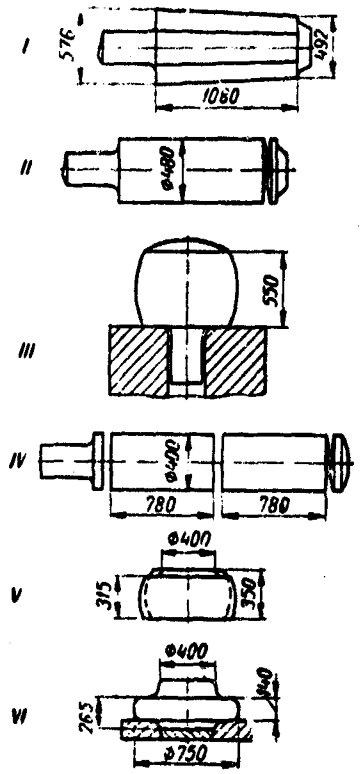


Fig. 16. Technological forging pattern for producing wheel from ingot weighing 2.5 t. I-VI) Forging sequence.

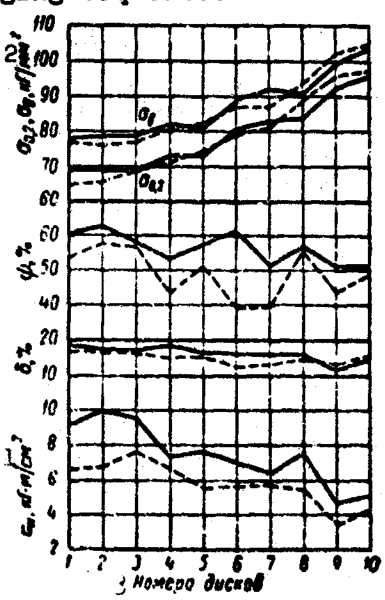


Fig. 17. Mechanical properties of experimental wheels heat-treated to different strengths. The solid line represents data at the periphery of the hub (standard ring for mechanical tests), while the dashed line represents data for the center of the hub. 1) kgf·m/cm²; 2) kgf/mm²; 3) wheel number.

TABLE 9

Mechanical Properties of Wheels Forged from Ingots from Different Melts (ЭИ415 Steel)

Химический состав плавки, %										№ диска	Механические свойства при 20° С						$\sigma_{0,2}$ при 550° С, кГ/мм²
C	Si	Mn	Cr	Ni	Mo	W	V	S	P		$\sigma_{0,2}$, кГ/мм²	$\sigma_в$, кГ/мм²	δ, %	ψ, %	$a_к$, кГ·м/см²	Твердость HB	
0,19	0,20	0,41	2,95	0,30	0,47	0,41	0,74	0,013	0,018	1	67,5 70,0	77,0 79,6	20,2 14,2	63,9 62,7	14,0 13,8	235	41,7
										2	68,5 72.6	81,0 80,3	16,0 16,0	63,9 63,9	11,5 7,0	248	53,7
0,20	0,40	0,43	2,85	0,16	0,48	0,38	0,71	0,024	0,023	1	71,9 71,3	80,9 79,6	15,0 18,6	51,0 57,6	9,3 9,4	255	47,8
										2	71,9 73,8	79,6 81,5	15,6 18,0	56,4 58,8	9,5 9,9	248	46,0
0,22	0,39	0,44	3,00	0,17	0,48	0,50	0,77	0,020	0,021	1	73,2 73,2	80,9 81,5	17,0 14,6	53,8 46,7	5,8 6,1	241	58,8
										2	71,9 75,4	81,5 87,9	17,0 16,6	56,4 52,4	6,5 8,4	248	51,7
0,24	0,36	0,50	2,84	0,05	0,52	0,40	0,85	0,016	0,012	1	63,0 64,3	74,5 75,1	19,0 15,8	60,1 52,4	12,6 11,5	229	48,3
										2	66,2 67,5	76,0 78,3	16,8 19,8	63,9 63,9	13,5 12,5	241	51,9
0,22	0,32	0,47	2,70	0,22	0,44	0,30	0,75	0,021	0,017	1	68,8 70,7	78,9 79,6	16,2 16,0	58,8 55,2	12,5 10,0	241	48,4
										2	63,7 62,4	73,8 72,6	17,4 16,8	65,1 58,8	14,8 14,0	229	45,9
0,20	0,16	0,45	2,83	0,20	0,50	0,32	0,73	0,022	0,017	1	72,6 75,1	82,1 82,8	18,6 15,6	60,1 55,2	7,4 6,0	255	54,3
										2	73,2 73,8	81,5 82,1	11,8 16,0	55,2 57,6	7,5 8,1	255	47,7
0,18	0,15	0,42	2,90	0,15	0,47	0,30	0,66	0,016	0,016	1	71,9 70,0	78,9 78,9	16,0 15,4	51,4 63,9	9,3 10,3	241	49,5
										2	75,1 73,8	82,8 81,5	12,2 14,6	61,4 30,1	11,4 10,6	241	53,6
0,19	0,40	0,45	2,88	0,19	0,51	0,36	0,69	0,020	0,018	1	76,4 78,4	82,8 84,7	15,4 14,4	61,4 61,4	9,4 9,5	255	59,0
										2	77,7 76,4	85,9 83,4	14,4 14,6	19,6 51,0	5,1 6,2	255	56,7
0,18	0,23	0,48	2,80	0,11	0,48	0,32	0,60	0,016	0,013	1	75,8 72,0	83,4 80,3	18,8 16,4	62,7 63,9	19,0 16,5	255	57,8
										2	77,7 82,8	85,3 89,8	16,2 13,0	60,1 42,2	10,6 9,5	269	60,2

1) Chemical composition of melt, %; 2) Wheel No.; 3) mechanical properties at 20°C; 4) kgf/mm^2; 5) $kgf \cdot m/cm^2$; 6) hardness HB; 7) at 550°C, kgf/mm^2.

It was very indicative that the values of δ, ψ, and a_n were not below those stipulated by the technical specifications even in the central portion of the wheel when $\sigma_{0.2}$ was less than 90 kgf/mm^2 (Fig. 17).

After heat treatment and mechanical testing, the wheels are subjected to ultrasonic quality control with an УЗД-7н apparatus operating at a frequency of 2.5 MHz. The entire wheel volume was monitored from one face side of the hub and body. The size of the defects was determined by comparing the amplitude of the signal reflected from the defect with that of the signal from a control aperture in a standard specimen.

Statistical analysis of the results obtained in ultrasonic quality control of 70 wheels yielded the following data: eight wheels exhibited from one to five small scattered defects of the nonmetallic-inclusion type up to 3 mm in diameter, principally in the horizontal axial plane. One wheel contained two defective zones with areas of 50 and 8 cm^2 and a number of isolated defects in the vicinity of the hub, at a depth of 100-200 mm. No internal defects in the metal were detected in any of the other forgings. Of the 60 wheels, only the one with defective zones was considered unsuitable and rejected. This wheel was subsequently sectioned and examined: the defective zones were found to be aggregates of nonmetallic inclusions.

After final treatment, the wheels were etched with 15 per cent ammonium sulfate and 10 per cent nitric acid and the hub surface and the face surfaces of the rim and body were subjected to macroscopic examination. All 59 of the wheels checked were acceptable: no defects that disrupted the continuity or homogeneity of the metal were detected in the etched surfaces.

Impact-test specimens cut from 2-3 wheels fabricated from the steel of each melt were used to check the microstructure of the metal, which was found to consist of sorbite with a weak martensite orientation in all cases. The metal had adequate structural homogeneity.

The fact that the etching and ultrasonic monitoring revealed neither external defects nor internal discontinuities in the metal in any of the wheels confirms the reliability of the technological process, particularly the weight and shape of the ingots employed and the procedures used in forging the wheels. Since most of the forgings were fabricated from elongated ingots, our experience confirms that use of such ingots for very complex wheels of ЭИ415 steel intended for critical applications yields fully satisfactory results.

The variation in strength and plasticity in the interior of the forgings was checked by sectioning and mechanical testing of different portions of one wheel. We also sectioned a second wheel forged from the same steel melt by a modified version of the technological process (without intermediate upsetting of the ingot). This second wheel was investigated in order to study the influence of the reduction ratio on the welding up of shrinkage defects in the ingot and the mechanical properties of the metal in different areas of the forging.

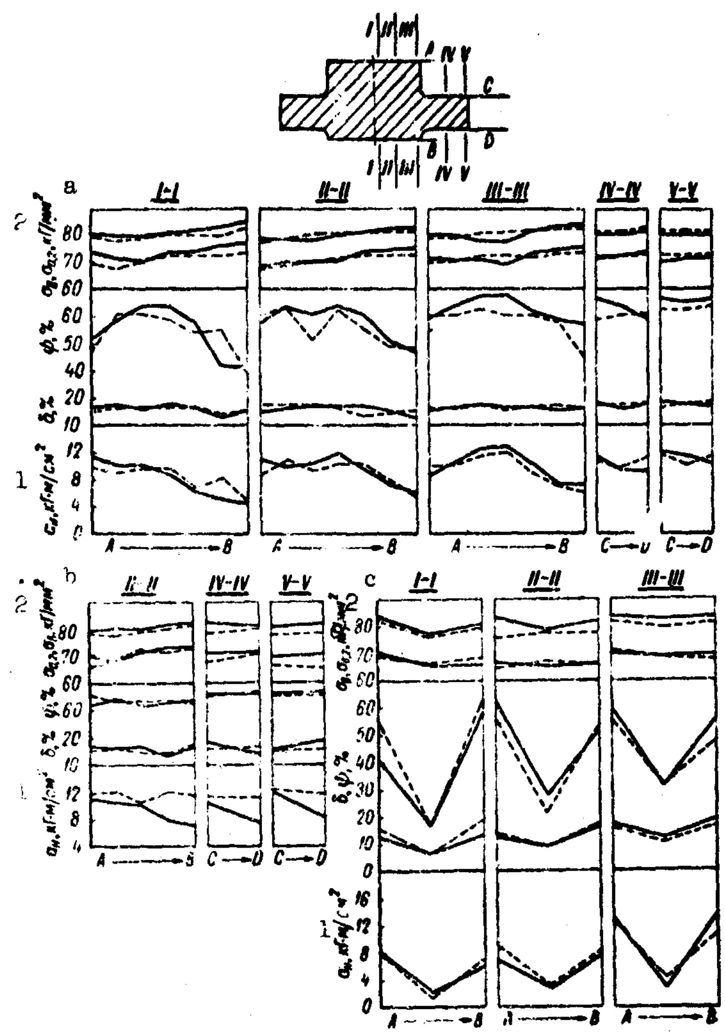


Fig. 18. Graphs representing variation in mechanical properties of wheels forged with (solid lines) and without (dashed lines) intermediate upsetting. a) In tangential direction; b) in radial direction; c) in axial direction. 1) kgf·m/cm²; 2) kgf/mm².

In order to make a reliable comparison, both wheels were forged from elongated ingots of steel from the same melt (containing 0.20% C, 0.28% Si, 0.47% Mn, 3.00% Cr, 0.23% Ni, 0.52% Mo, 0.65% V, 0.33% W, 0.019% S, and 0.013% P). The steel was smelted and cast in accordance with plant specifications. The ingots were delivered hot to the press shop and loaded into the furnace for preforging heating with a surface temperature of 580-650°C.

The heating regimes for the ingots and intermediate blanks, the deformation regimes, and the forging-temperature regimes corresponded to the technological norms established for commercial wheels. The forging proceduré for the second wheel differed from the normal pattern in omitting intermediate upsetting and subsequent drawing of the upset ingot. As a result, the technological cycle was reduced to two passes.

The initial heat treatment, cooling, trimming, and final heat treatment of the experimental forgings were carried out in accordance with the instructions in effect in the shops and therefore ensured identical conditions for the operations governing the mechanical properties of the experimental and production wheels.

Macroscopic examination and ultrasonic defectoscopy revealed no defects in either experimental forging.

Radial sections were cut from the wheels and their surface was carefully examined after grinding and etching. None of them exhibited any macroscopic defects of a metallurgical character.

The results of the mechanical tests performed on the sections are presented in Fig. 18, in the form of graphs representing the variation in mechanical properties over the height and radius of the wheels in the tangential, radial, and axial directions.

The strength characteristics $\sigma_{0.2}$ and σ_b in the tangential direction (Fig. 18a) remained almost constant over the entire wheel volume, exhibiting only a slight tendency to increase from one face of the wheel to the other. The difference in $\sigma_{0.2}$ and σ_b was due to a slight nonuniformity of tempering and did not exceed 5-6 kgf/mm², with the absolute values $\sigma_{0.2}$ = 70-75 kgf/mm² and σ_b = 78-84 kgf/mm². Comparison of the strength of the face and internal zones of the wheels indicated that the metal had high temperability. There was no material difference in the strengths of the wheels forged under different technological regimes (with and without intermediate upsetting of the ingot).

The relative elongation δ had uniform values in all portions of both wheels and did not fall below 13%. The relative necking ψ had its highest value in the vicinity of the horizontal axial plane, where it reached 60-68%, i.e., a level even higher than in the peripheral portions of the wheel body. This was due to the higher reduction actually obtained in this zone. The values of ψ were somewhat reduced in the face portions of the wheels, especially at the center of the hub, but did not fall below 40%, which must be regarded as quite satisfactory when conjoined with $\sigma_{0.2}$ = 73-76 kgf/mm². Over most of the wheel volume, the relative reduction in area was 50-60% in the vicinity of the hub and 60-65% in the body.

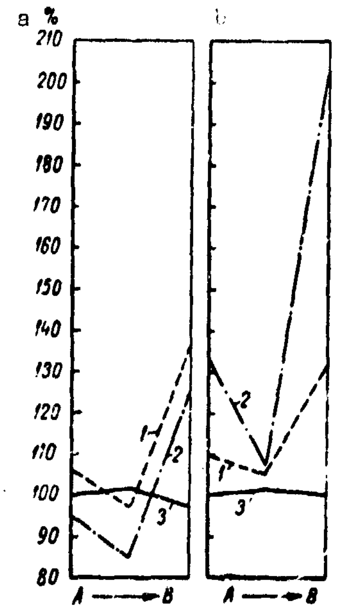


Fig. 19. Ratio of mechanical properties of experimental wheels in radial and tangential directions. a) Wheel with intermediate upsetting; b) wheel without intermediate upsetting.

$$1-\frac{\psi_{рад}}{\psi_{танг}};\quad 2-\frac{a_{н\,рад}}{a_{н\,танг}};\quad 3-\frac{\sigma_{0.2\,рад}}{\sigma_{0.2\,танг}}$$

The forging variant had some effect on the relative necking, causing a slight decrease in this index in the wheel forged without intermediate upsetting. For example, ψ = 42-51% in the face zones of the central portion of the hub in the wheel forged with intermediate upsetting and ψ = 40-47% in the corresponding zones of that forged without intermediate upsetting; the figures for the interior regions of the huc were 60-68 and 50-60% respectively.

The impact strength a_n of the wheels forged in both variants were roughly the same and were sufficiently high. The lowest value occurred in one face of the central portion of the hub, where a_n = = 4.5 kgf·m/cm²; the values for the other portions of the wheels were higher, reaching 8-12 kgf·m/cm².

The overall pattern of the variation in mechanical properties in the radial direction (Fig. 18b) was similar in character to that in the tangential properties, the only difference being that the relative necking had more consistent values over the wheel height. The properties of the two wheels can be regarded as equivalent: the somewhat lower yield strength and ultimate strength obtained for the wheel forged without intermediate upsetting were compensated for by its increased plasticity and impact strength.

Figure 19 shows the ratio of the principal mechanical properties

$\sigma_{0.2}$, ψ, and a_n in the radial and tangential directions. The ratio $\frac{\sigma_{0.2\,\text{рад}}}{\sigma_{0.2\,\text{танг}}}$ remained almost constant over the height of both wheels, at about 100%; the ratios $\frac{\psi_{\text{рад}}}{\psi_{\text{танг}}}$ and $\frac{a_{\text{н рад}}}{a_{\text{н танг}}}$ were minimal in the vertical midsection of the forgings but amounted to 85-97% in the wheel with intermediate upsetting (Fig. 19a) and 105-107% in the wheel without intermediate upsetting (Fig. 19b).

The graph representing the variation in the mechanical properties of the forgings in the axial direction was very indicative. There was a distinct tendency toward an abrupt decrease in plasticity from the faces toward the horizontal axial plane, especially at the center of the hub. For example, the decrease in ψ over this region was characterized by the following data: ψ = 42-63% in the vicinity of the faces and ψ = 16% in the vertical midsection. The impact strength a_n equaled 6-8 and 1-2 kgf·m/cm² respectively. The change in strength characteristics was relatively small.

The decrease in metal plasticity in the vicinity of the horizontal axial plane can be attributed to the physical characteristics of the deformation undergone by this zone during upsetting, since it is the boundary between two opposing metal flows, where deformation of nonmetallic inclusions takes place most rapidly. The thin nonmetallic films formed here have no material influence on the mechanical properties of the forging in the tangential direction, but they are the principal factor responsible for the sharp decrease in metal plasticity in the axial direction.

The actual degree of upsetting decreased from a maximum in the vicinity of the horizontal axial plane toward the faces of the wheels. At the same time, the fibers deviated more and more from the radial direction. There was accordingly a decrease in the total surface of the metallurgical defects in the fracture planes for the axial specimen, which promoted an increase in metal plasticity in the axial direction.

It is natural to presume that the maximum decrease in plasticity in the axial direction in the vertical midsection of a wheel should occur in those areas most highly saturated with liquates and other defects of a metallurgical character, i.e., in the central zone. Except for isolated deviations, this pattern was actually confirmed experimentally. In Fig. 18c, for example, the values of the most sensitive plasticity characteristic, ψ, increase continuously from the center of the hub to its periphery.

The forging variant had little effect on the mechanical properties of the wheels in the axial direction and the overall pattern of the variation in the principal plasticity characteristics in different zones of the forgings was virtually the same for both variants.

The ratio of the mechanical properties of the wheels in the axial and tangential directions (Fig. 20) exhibited markedly reduced values in the vicinity of the horizontal axial plane at the center

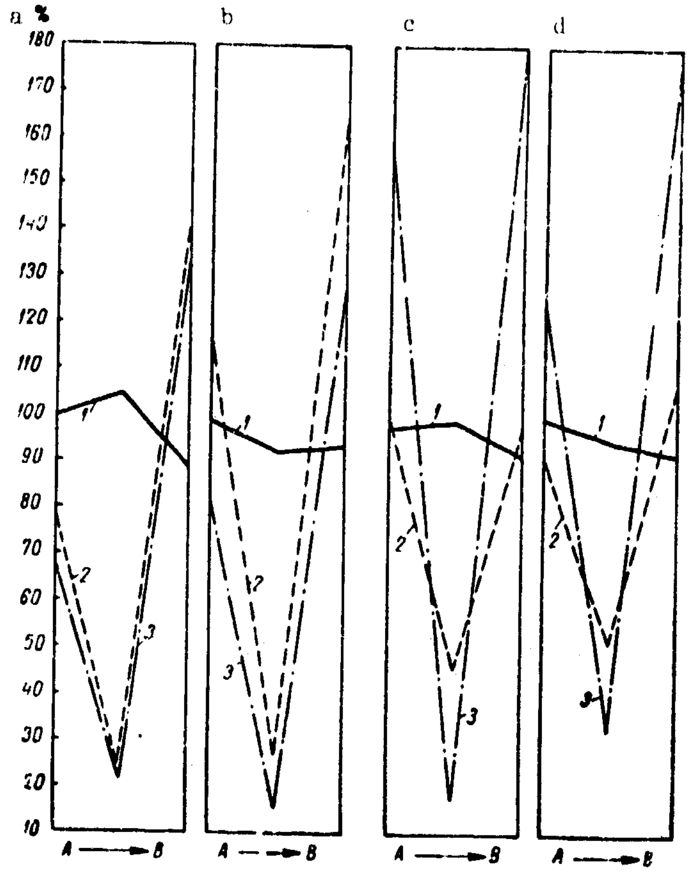


Fig. 20. Ratio of mechanical properties of experimental wheels in axial and tangential directions at center (a and b) and periphery (c and d) of hub. a, c) Wheels with intermediate upsetting; b, d) wheels without intermediate upsetting.

$$1 - \frac{\sigma_{0.2_{осев}}}{\sigma_{0.2_{танг}}};\quad 2 - \frac{\psi_{осев}}{\psi_{танг}};\quad 3 - \frac{a_{н_{осев}}}{a_{н_{танг}}}.$$

of the hub, where $\frac{\psi_{осев}}{\psi_{танг}}$ was about 25% and $\frac{a_{н_{осев}}}{a_{н_{танг}}} \approx 15 \div 22\%$. The ratio $\frac{\psi_{осев}}{\psi_{танг}}$ increased to 45-50% in the peripheral region of the hub, while there was a slight rise in $\frac{a_{н_{осев}}}{a_{н_{танг}}}$.

The ratios $\frac{\psi_{осев}}{\psi_{танг}}$ and $\frac{a_{н_{осев}}}{a_{н_{танг}}}$ reached and, in some cases, even exceeded 100% in the face regions of the wheels. These relationships for plasticity and viscosity in different zones can be attributed to opposite variations in ψ and a_n in the axial and tangential directions over the wheel height: maximum values of ψ and a_n were obtained for the vicinity of the horizontal axial plane when tangential specimens were tested, while minimum values were obtained when axial specimens were tested.

The strength ratio $\frac{\sigma_{0.2_{ocev}}}{\sigma_{0.2_{mang}}}$ was usually 90-100%.

Our experimental results confirmed the high quality of both forgings. In none of the zones did the mechanical properties of the metal in the tangential and radial directions fall below the norms stipulated by technical specifications and the properties in the axial direction were fully satisfactory.

Our investigations established that it is possible in principle to simplify the technical procedure for forging wheels of this type by omitting intermediate upsetting of the ingot and subsequent drawing of the upset billet.

The NPL, working in conjunction with the Central Scientific Research Institute of Technology and Machine Building [TsNIITmash] (ЦНИИТ), has developed and introduced a method for sectional stamping of wheels for the ГТ700-5 low-pressure gas turbine from ЭИ415 steel in a press exerting a force of 3000 t.

The characteristics of this new process [55] made it technologically feasible to reduce the weight of the forgings from 750 to 450 kg, i.e., by 40%, and permitted a substantial reduction in the labor consumed in machining of the wheels.

Figure 21 shows a forged wheel fabricated by sectional stamping. The decrease in forging weight is achieved by an increase in forging precision, a decrease in tolerances and allowances, and a better distribution of sample rings for mechanical testing.

The wheels were stamped in a two-sectional die. The dimensions of the sections were chosen in such fashion that the pressure (14-15 kgf/mm²) was distributed more or less evenly in stamping with the central and outer punches.

The original forging procedure with the sectional die consisted of the following basic operations:

1) heating an ingot weighing 2.0 or 2.17 t to the forging temperature;

2) forging of a journal and billeting of the ingot (the elongated ingots weighing 2.17 t were not billeted);

3) heating of the billet;

4) intermediate upsetting;

5) drawing of the upset billet to a diameter of 330 mm;

6) cutting of three blanks 680 mm long;

7) heating of the blanks;

8) upsetting of the blanks to a height of 260 mm on a flat platen;

9) stamping of wheels in three strokes of press traverse.

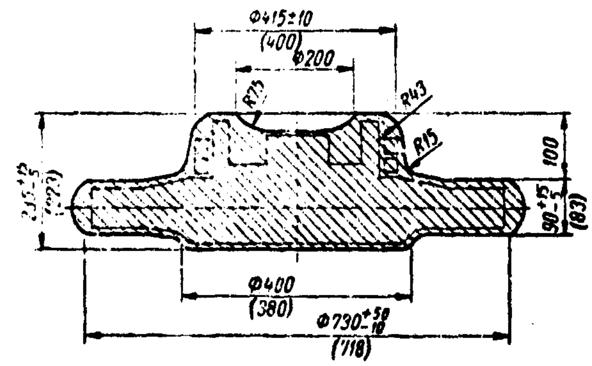


Fig. 21. Forged wheel (weighing 460 kg) fabricated by sectional stamping.

TABLE 10

Mechanical Properties of Wheels Produced from ЭИ415 Steel by Sectional Stamping

№ диска (1)	$\sigma_{0.2}$	$\sigma_в$	δ	ψ	$a_н$, кГ·м/см² (3)	Твердость, HB (4)	$\sigma_{0.2}$ при 550° C, кГ/мм² (5)
	кГ/мм² (2)		%				
1	65.6 65.6	79.5 79.8	16.6 17.6	62.3 57.8	11.0 9.6	241	50.2
2	73.1 74.1	84.1 86.0	16.0 17.3	57.8 57.8	8.0 9.0	255	54.8
3	75.0 77.0	84.0 85.3	14.6 17.0	45.2 53.8	5.5 5.6	263	53.7
4	82.8 84.0	90.4 92.3	14.8 16.0	57.5 57.6	6.6 7.3	285	57.3
5	93.3 91.9	101.0 101.0	11.0 15.0	48.6 55.6	5.3 4.8	321	69.8

1) Wheel No.; 2) kgf/mm²; 3) kgf·m/cm²; 4) hardness, HB; 5) at 550°C, kgf/mm²

All the forging operations, with the exception of the final pass, were carried out in a press exerting a force of 2000 t; upsetting of the blanks and sectional stamping were carried out in a 3000 t press. As experimentation showed, this sequence of operations was the most favorable from the standpoint of effectiveness of metal working in two mutually perpendicular directions throughout the entire wheel.

Two plates were mounted on the movable press bed for the final pass; the blank was placed on one and the sectional-stamping die was fastened to the other. Upsetting was carried out with the traverse platen, to which the sectional punch was then attached with

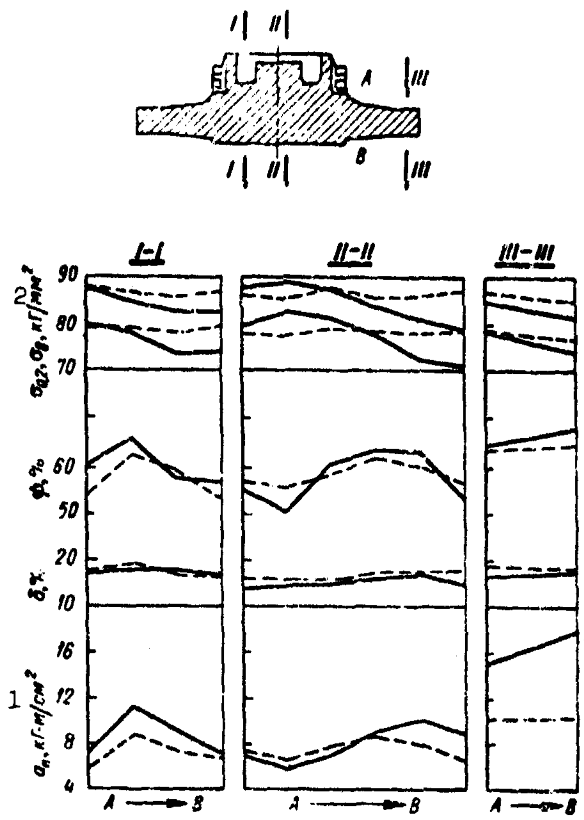


Fig. 22. Graphs representing change in mechanical properties of wheels produced by sectional stamping with (solid lines) and without (dash lines) intermediate upsetting of ingot. 1) kgf·m/cm²; 2) kgf/mm².

TABLE 11

Mechanical Properties of Wheels Forged by Sectional Stamping

1 Вариант технологического процесса	$\sigma_{0,2}$	$\sigma_в$	δ	ψ	3 $a_н$, кГ·м/см²		6 Твердость НВ	7 $\sigma_{0,2}$ при 550° С, кГ/мм²
	2 кГ/мм²		%		4 1-й образец	5 2-й образец		
8 С промежуточной осадкой слитка	79,6 78,3	86,6 84,7	15,0 15,0	60,1 60,4	7,3 10,3	7,8 7,5	269	54,7
9 Без промежуточной осадки слитков	78,9 79,6	87,2 87,9	15,2 16,2	57,6 57,6	6,1 7,6	8,8 7,4	255	58,4

1) Variant of technological process; 2) kgf/mm²; 3) kgf·m/cm²; 4) 1st specimen; 5) 2nd specimen; 6) hardness HB; 7) at 550°C, kgf/mm²; 8) with intermediate upsetting of ingot; 9) without intermediate upsetting of ingot.

hooks.

A 3000 t press proved adequate for stamping wheels by the procedure described above, provided that the final deformation temperature was no less than 1000-1020°C. The total duration of the operations involved in the final pass, taking into account the time consumed in transferring the blank from the furnace to the press (3-4 min), the upsetting and stamping procedures, shifting the bed, and all auxiliary operations (2-3 min) is about 5-7 min. Under these conditions, the initial metal-deformation temperature should be no less than 1130-1150°C and the blank should be heated to about 1180°C in the furnace. When the initial heating is lower and the final temperature of the blank to be stamped drops to 900-950°C, the forgings are understamped and require additional heating to bring their dimensions to those stipulated.

The cooling and heat-treatment regimes for the stamped forgings, the mechanical tests, and the quality-control procedures correspond to those usually employed for wheels of the type in question. It has been established that the mechanical properties and the ultrasonically determined microstructure and macrostructure of the metal in such stamped wheels are not inferior to those in wheels produced by free forging. Table 10 presents the results of mechanical tests on a series of wheels stamped from steel of a single melt (containing 0.28% C, 0.26% Si, 0.45% Mn, 2.85% Cr, 0.35% Ni, 0.46% Mo, 0.46% W, 0.77% V, 0.019% S, and 0.020% P) and heat-treated to different hardnesses. All the blanks were fabricated from elongated ingots weighing 2.17 t. Macroscopic examination and ultrasonic defectoscopy showed all the walls to be acceptable. The microstructure of the metal in the wheels consisted of sorbite with areas of ferrite and very fine carbides located along the grain boundaries and within the grains, i.e., was that normal for ЭИ415 steel.

We studied the feasibility of simplifying the technological forging (stamping) process for these wheels by omitting intermediate upsetting of the ingot. The probability of obtaining satisfactory metal quality in wheels forged by the simplified technology was quite real, since the ingot not subjected to intermediate upsetting were pressed with a reduction ratio of no less than 1.6.

Two wheels were fabricated from the melt whose chemical composition was given above, using sectional stamping with and without intermediate upsetting of the ingots. Both wheels were heat-treated and checked by the established procedures, yielding satisfactory results for all types of tests (Table 11).

In order to make a more thorough examination of the metal, both wheels were cut into rings, mechanical testing of tangential specimens then giving a complete representation of the mechanical properties throughout the entire wheel volume. The test results are shown in Fig. 22, in the form of graphs representing the variation in mechanical properties over the height of the wheels in different cross-sections of the hub and body (Fig. 22). It can be concluded from an analysis of these curves that the wheels have high mechanical properties even in the regions near the axis, where, for example, the minimum relative necking exceeded 50% and the minimum impact strength was more than 6 kgf•m/cm², with a yield strength of more than 73 kgf/mm². The forging variant employed (with or without intermediate upsetting of the ingot) had no material effect on the quality of the metal in the wheels.

16. FORGED COLLARS, T-PIPES, PIPES, AND OTHER COMPONENTS PRODUCED FROM 15X1M1Ф STEEL

The forgings shown in Fig. 23 are employed in fabricating components for steam turbines (300,000 kW) operating at temperatures of up to 565°C and pressures of up to 240 atm (abs). The very critical applications of these forgings requires use of reliable technological processes in their fabrication, ensuring high metal quality in different regions of the components. Blank quality is checked by mechanical testing, metallographic analysis, and ultrasonic defectoscopy.

The presence of central holes in flanges to some extent permit removal of the lowest-quality axial zone of the ingot by broaching and simultaneous examination of the internal surface for defects in the metal. T-pipes, pipes, and similar components with no central holes are metallurgically more complex, the critical portions of the forgings being formed from the inner layers of the ingots. The technological aspect of forging production associated with smelting of high-quality steel, use of the best ingots for the purpose, and effective mechanical working of the metal during forging are decisive for production of final components that satisfy the requirements imposed on metal purity, density, and homogeneity. As for methods for checking the quality of the metal, including that deep within the forgings, the design characteristics of the components make them relatively simple and very reliable: the presence of inclined apertures reaching the central portion of the forging and the feasibility of using special hollow drills to remove cores permits thorough metallographic analysis and mechanical testing of different portions of the forging.

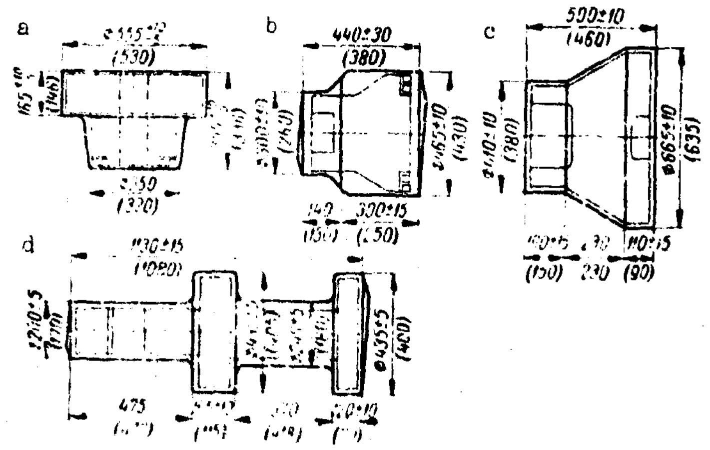


Fig. 23. Forgings of 15X1M1Ф steel. a) Collar (weighing 485 kg); b) T-pipe (weighing 600 kg); c) lower portion of valve housing (weighing 900 kg); d) lid of steam chest (weighing 660 kg).

The steel for the forgings is smelted in a 3-ton or 10-ton electric furnace; the smelting conditions conform to those for the high-quality class. The ingot weight does not exceed 2.7 t. The ingots are delivered hot to the press shop.

Elongated ingots weighing 2.7 t are used for forged collars (four forgings each), with a yield of 72% (Fig. 23a). The forging process consists of drawing the ingot to a diameter of 325 mm, cutting off blanks 740 mm long, and upsetting the collar in a liner ring. The blank is heated to 1050-1080°C before upsetting. All the forging operations are carried out in a 7-ton hammer.

It can be seen from analysis of the deformation pattern that the smallest reduction ratio occurs in a section of the forging 350 mm in diameter, whose mechanical working is limited to drawing with a reduction ratio of 2.0-2.5 and slight upsetting in the cavity of the liner ring (with a reduction ratio not exceeding 1.2-1.3). The results of exhaustive quality control, including mechanical testing of tangential specimens after heat treatment, indicates that the metal in this area is satisfactory in quality. Specifically, with $\sigma_{0,2} > 32{,}5\ \kappa\Gamma/\text{мм}^2$ and $\sigma_в > 55\ \kappa\Gamma/\text{мм}^2$, the plasticity and impact strength of collars from a large production batch had the following values: $\delta = 21{,}5-27\%$, ψ $66-78\%$, $a_н$ $14-30\ \kappa\Gamma\cdot м/см^2$. Macroscopic examination and ultrasonic defectoscopy revealed no defects that could be attributed to an inadequate reduction ratio.

The more massive portion of the forging (555 mm in diameter), produced by upsetting with a deformation of about 70%, is distinguished by good working of the metal in two mutually perpendicular directions. Additional mechanical tests performed on tangential specimens cut from this portion of the collar yielded results close to those obtained for the section 350 mm in diameter. Sectioning showed the mechanical properties of the collar to be highly uniform in the tangential and radial directions.

The forged T-pipes (Fig. 23b) are fabricated from ingots weighing 2 t in a press exerting a force of 2000 t. After billeting and preliminary upsetting, the ingot is drawn into a blank 465 mm in diameter, its terminal portions being pressed to a diameter of 300 mm. The shrink head and bottom of the ingot are removed and the blank is cut into two equal parts, each of which corresponds in size and shape to the forging shown in Fig.23b. All these operations are carried out in three passes.

A power-hammer die must be used for final forming of the lower valve housing (Fig. 23c). Two blanks 950 mm long are cut from an ingot weighing 3 t after preliminary upsetting and drawing to a diameter of 400 mm; the blanks are then upset to the required size in the die. Since the lower portion of a blank 410 mm in diameter undergoes almost no deformation in the die, the temperature to which it is heated before the last half should not exceed 1050°C. The die is satisfactorily filled despite this relatively low heating, which is facilitated by the very favorable profile of the die cavity. The forging is fabricated in four passes in a 2000-t press.

The forged steam-chest lid shown in Fig. 23d has a shape that is difficult to produce, since the fact that a rod 240 mm in diameter bears two flanges 435 mm in diameter separated by a distance of 370 mm creates substantial difficulties in direct fabrication from the ingot. Production of such forgings in a 7-ton hammer from a blank forged in a 2000-t press has been found to be best under the conditions that obtain at the NPL. Blanks 435 mm in diameter and 670 mm long (3 pieces) are produced from an ingot weighing 3 t, with obligatory preliminary upsetting and drawing to the requisite size. This provides the necessary reduction ratio at the flanges, which are almost undeformed in the hammer. The end of the forging, which is 200 mm in diameter, and the area between the flanges are hammer-drawn with the blank heated to 1050°C. Press-forging with a simplified lid configuration (with no annular recess between the flanges) leads both to a large increase in forging weight and in the labor consumed in machining and to rejection of a substantial number of components on the basis of macroscopic examination of the etched surface of the inter-flange area. The defects detected consist principally of aggregates of small liquation and nonmetallic inclusions. This phenomenon results from the large amount of metal removed during trimming and finish machining and the nearness of the surface to the axial zone of the ingot. Actually, simple calculation shows that about 85% of the metal is removed, corresponding to the highest-quality outer zone of the ingot.

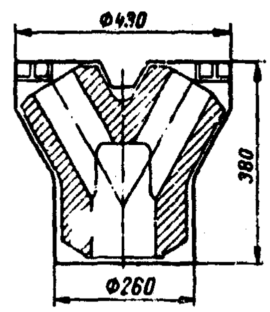


Fig. 24. Trimming of T-pipe shown in Fig. 23b (the fine lines represent the outline of the finished component).

Experience has established that defects are absent after forging without an annular recess only if the steel melt is especially pure.

Forgings produced from 15X1M1Ф steel are cooled with the furnace to 400°C, subjected to isothermal annealing at 650°C for 30 h, and then slowly cooled with the furnace to 300°C.

Despite the relatively large cross-section and unfavorable shape of the forgings under consideration (absence of central holes), formation of flakes in such forgings has not been observed at the NPL. It must be emphasized that 15X1M1Ф steel presents no special danger in this respect, even when there are certain deviations from the predetermined technological regime, as when supercooling before isothermal annealing is omitted or the final temperature to which the forgings are cooled with the furnace is raised. These characteristics of forging condition are even more indicative, since the NPL fabricates the components from steel smelted in a basic electric furnace without vacuum casting.

The forgings are heat-treated after rough machining (trimming with an allowance of 10-12 mm over the finished dimensions and the minimum necessary tolerances). Figure 24 shows an example of trimming of the T-pipe depicted in Fig. 23b. The principal allowance in the blank before heat treatment is the metal in the undrilled oblique openings, which is subsequently used for additional mechanical testing and metallographic analysis.

Forged components with complex configurations (valve-housing elements, T-pipes, pipes) having oblique openings and irregularly shaped cavities are subjected to ultrasonic monitoring from their flat and cylindrical surfaces before heat treatment, in order to show up metallurgical defects. This quality-control procedure, which presents no substantial difficulties, makes it possible to examine the entire forging volume but is still only preliminary, since spurious impulses are often produced as a result of the structural heterogeneity of the un-heat-treated metal. All accessible areas of the component should therefore also undergo ultrasonic monitoring after heat treatment.

As was pointed out in Chapter 5, the type of heat treatment employed for blanks of 15Х1М1Ф steel is determined by their dimensions and the requirements imposed on the mechanical properties of the metal. Both modification and normalization followed by tempering are quite suitable from the standpoint of hot strength. Forgings of the type under consideration should satisfy the following mechanical-property norms: $\sigma_{0,2} \geqslant 30$ кГ/мм²; $\sigma_в \geqslant 52$ кГ/мм²; $\delta \geqslant 20\%$; $\psi \geqslant 50\%$; $a_к \geqslant 5$ кГ·м/см². These properties are provided by normalization and temperature, so that the heat-treatment regime for most components generally consists of normalization from 1000-1020°C and tempering at 730°C. Extremely critical and massive housing-component forgings of the T-pipe, pipe, and other types undergo double normalization and tempering or normalization from 1020-1040°C, quenching from 990-1010°C, and tempering at 750°C for 8-12 h. Post tempering cooling is carried out in air.

Table 12 presents typical results obtained in mechanical tests performed on heat-treated forgings of 15Х1М1Ф steel. The mechanical properties given in this table are for standard sample rings or bars, as provided by the forging plans (Fig. 23). It can be seen from Table 12 that the fluctuations in $\sigma_{0.2}$ produced by varying the tempering temperature cause no substantial decrease in the plasticity indices δ and ψ, whose absolute values remain rather high. For example, steam-chest lids forged from steel of the same melt and heat-treated to yield strengths of 44.5 and 58.0 kgf/mm² have values of 24 and 75% for δ and ψ in the first case and 19 and 70% in the second case; the decrease in impact strength is larger (from 19.6-23.5 to 5.8-7.6 kgf·m/cm²).

TABLE 12

Mechanical Properties of Forgings Produced from 15X1M1Ф Steel

1 Наименование поковки	2 Химический состав, %	3 Направление образца	$\sigma_{0,2}$	$\sigma_в$	δ	ψ	5 $a_н$, кГ·м/см²	
			4 кГ/мм²		%		6 1-й образец	7 2-й образец
8 Фланец (рис. 23, а)	0,15 C; 0,26 Si; 0,49 Mn; 1,23 Cr; 0,95 Mo; 0,25 V; 0,025 S; 0,025 P	12 Тангенциальное	35,5	59,0	26,0	66,0	15,1	14,0
			39,0	55,0	21,5	75,0	30,0	24,7
			36,5	55,0	27,0	78,0	29,0	30,3
			33,5	58,0	26,0	70,0	18,0	15,4
9 Тройник (рис. 23, б)	0,19 C; 0,21 Si; 0,50 Mn; 1,35 Cr; 0,95 Mo; 0,28 V; 0,015 S; 0,016 P		44,0	62,0	26,0	69,0	17,7	19,0
			48,5	62,0	24,0	71,0	21,2	16,2
			51,0	67,0	20,5	72,0	16,6	17,0
			48,0	61,0	23,0	73,0	13,6	12,8
10 Нижняя часть коробки клапана (рис. 23, в)	0,13 C; 0,31 Si, 0,54 Mn; 1,25 Cr; 1,04 Mo; 0,28 V; 0,013 S; 0,023 P		57,0	71,0	19,5	68,0	14,0	16,8
			52,0	64,0	22,5	75,0	18,4	20,2
			41,0	62,0	27,0	75,0	17,3	19,8
			48,0	64,0	23,0	73,0	17,2	18,0
11 Крышка паровой коробки (рис. 23, г)	0,15 C; 0,14 Si; 0,54 Mn; 1,43 Cr; 1,00 Mo; 0,30 V; 0,011 S; 0,018 P	13 Продольное	44,5	62,0	24,0	75,0	19,6	23,5
			51,0	69,0	19,0	72,0	7,0	9,8
			56,0	72,0	20,0	72,0	7,6	5,6
			58,0	74,0	19,0	70,0	5,8	7,6

1) Type of forging; 2) chemical composition, %; 3) specimen direction; 4) kgf/mm²; 5) kgf·m/cm²; 6) 1st specimen; 7) 2nd specimen; 8) collar (Fig. 23a); 9) T-pipe (Fig. 23b); 10) lower portion of valve housing (Fig. 23c); 11) steam-chest lid (Fig. 23d); 12) tangential; 13) longitudinal.

Despite the combination of satisfactory plasticity and elevated strength, heat treatment of forgings to a given yield strength with a minimum reserve is considered expedient. This is because all the structural reserves of the steel are concentrated in its plasticity and the operational reliability of the component is thus increased.

The heat-treated forgings are etched and subjected to ultrasonic monitoring. These quality-control operations reveal macroscopic defects at the component surface and internal flaws in the metal.

In the experience of the NPL, forgings produced from 15X1M1Ф steel generally have had no defects. Their occurrence in individual forgings, in the form of small inclusions or films at the etched surface or impermissible internal flaws, has been random and has always been associated with a higher degree of contamination in the specific steel melt.

TABLE 13

Mechanical Properties in Different Zones of Forged T-Pipes (Fig. 23 b) Produced from 15Х1М1Ф Steel

1 № поковки	2 Место отбора образцов	3 Направление образца в поковке	$\sigma_{0,2}$	$\sigma_в$	δ	ψ	5 $a_н$, кГ м/см²	
			4 кГ/мм²		%		6 1-й образец	7 2-й образец
	Кольцо, предусмотренное чертежом поковки 8	Тангенциальное 10	39,5	59,0	27,0	70,0	15,4	16,4
1	Концевая часть высверленного стержня вблизи осевой зоны 9 поковки	Под углом 45 к горизонтальной плоскости 11	32,0 33,0	56,0 56,0	23,5 27,0	72,0 75,0	18,0 16,8	18,0 18,8
	Кольцо, предусмотренное чертежом поковки 8	Тангенциальное 10	54,0	72,0	20,0	70,0	15,6	10,2
2	Концевая часть высверленного стержня вблизи осевой зоны поковки 9	Под углом 45° к горизонтальной плоскости 11	50,0 46,0	69,0 67,0	16,0 20,0	70,0 73,0	19,5 13,7	14,5 17,0

1) Forging No.; 2) sampling site; 3) direction of specimen in forging; 4) kgf/mm²; 5) kgf·m/cm²; 6) 1st specimen; 7) 2nd specimen; 8) ring provided by forging plan; 9) terminal portion of drilled core near axial zone of forging; 10) tangential; 11) at angle of 45° to horizontal plane.

The quality of the metal in T-pipes, pipes, and other components with apertures is also checked with drilled cores 60-80 mm in diameter, which characterize the state of the interior regions of the forging. The core surface is ground, etched, and checked for macroscopic defects. Specimens for mechanical testing are then cut from different zones of the core in different directions.

The results of additional testing of the metal on a production scale confirmed the high quality of the forgings. Macroscopic defects, in the form of small liquation precipitates, were detected at the surface of the core only in rare instances. The mechanical properties of the metal within the forging exhibited stable plasticity characteristics. For example, moving from the peripheral zone to the interior in the forged T-pipes caused a slight decrease in strength, while the values of δ, ψ, and a_n remained high and almost constant (Table 13).

Our experience in forging production indicates that 15Х1М1Ф steel has good technological properties and that it can be used for critical components while satisfying a broad range of mechanical-property requirements.

17. FORGED ROTORS OF P2 AND P2M STEELS

One-piece forged steam-turbine rotors are among the most critical forgings for power-generation equipment, being subject to stringent requirements for chemical and physical homogeneity, mechanical properties, residual stresses, and other forging-quality factors that govern their operational reliability. These requirements dictate the production technology for forged rotors, which essentially consists in smelting and casting steel that is as pure as possible, conducting forgings under conditions that promote effective welding up of shrinkage defects in the ingot and mechanical working of the metal in the requisite direction, and heat treatment under a regime that ensures absence of flakes, high mechanical properties, and minimum residual stresses. The technological problems involved in satisfying these requirements are complicated by the fact that most rotors are fabricated from large ingots, while the cross-section of the blank often makes it necessary to employ special deformation and heat-treatment procedures.

The metal is checked for external defects and internal flaws by macroscopic examination and ultrasonic defectoscopy; the mechanical properties and their homogeneity over the forging volume are determined by individual mechanical tests on metal from different portions of the rotor. Residual-stress measurements and thermal tests are made to see that there are no substantial internal stresses and that the forgings and ingot are relatively coaxial.

Figure 25 shows typical forged rotors of P2 and P2M steels. Forgings of this type are produced at the Urals Machine Building Plant, which has amassed a great deal of experience in this area. Rotors are also being forged at the Izhorsk Plant.

Let us consider the principal conditions for forging fabrication and the norms for rotor-metal quality stipulated by technical specifications.

The forged rotors are fabricated from acid open-hearth steel and use of steel melted in electric furnaces with basic soles is permitted only when appropriate measures are taken. The steel is deoxidized without using aluminum.

The axis of the forging should more or less coincide with that of the ingot. The heat treatment consists of double normalization (the second normalization being from a temperature of no less than 940°C, after trimming of the blank) and tempering at 670°C, followed by slow cooling. The axial channel of the rotor is generally drilled before heat treatment, but the latter can theoretically be carried out before the channel is drilled, especially when the rotor is of relatively small size.

Longitudinal specimens from the ends of the rotor and tangential specimens from one or both faces of the barrel (depending on its length) are subjected to mechanical testing. Table 14 shows the norms for the mechanical properties of rotors fabricated from P2M steel. The variation in the hardness of specimens taken from different areas of the blank should be no more than 40 HB.

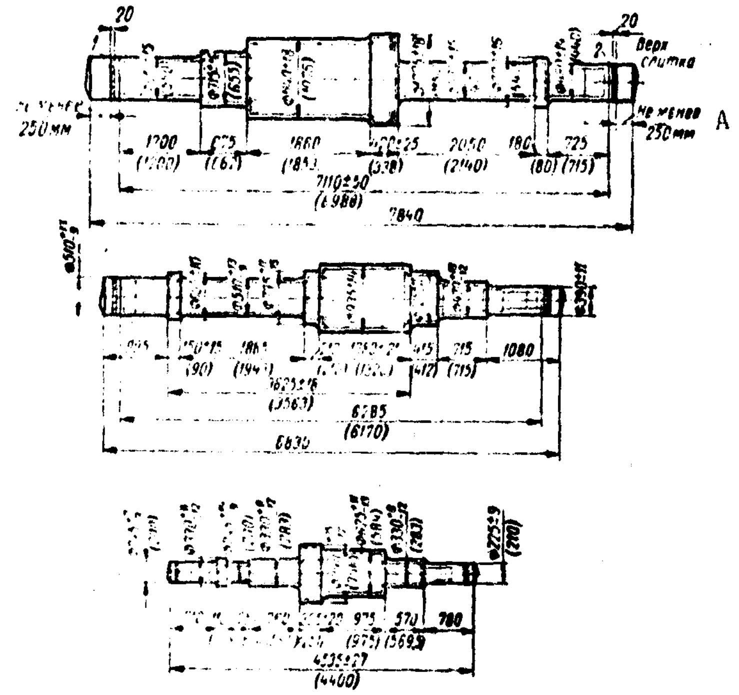


Fig. 25. Typical forged rotors of P2 and P2M steels. 1, 2) Sections for macroscopic examination. A) No less than; B) top of ingot.

TABLE 14

Norms for Mechanical Properties of Rotors Fabricated from P2M Steel

1 Направление образца	2 $\sigma_{0,2}$, кГ/мм²	3 $\sigma_в$, кГ/мм²	δ, %	ψ, %	4 $a_н$, кГ·м/см²	5 Угол при холодном загибе вокруг оправки диаметром 40 мм, град
		6 Не менее				
7 Продольное	≥50	63	16	40	4,5	180
8 Тангенциальное	50–68		13	35	4,0	150

1) Specimen direction; 2) kgf/mm²; 3) σ_b, kgf/mm²; 4) a_n, kgf·m/cm²; 5) angle for cold bending around form 40 mm in diameter, deg; 6) no less than; 7) longitudinal; 8) tangential.

The principal requirement imposed on the macrostructure of the blank is that there be no cracks or flakes. Large nonmetallic inclusions or aggregates of small nonmetallic inclusions in the final component are impermissible. The rotor metal is checked for defects by examination of the fracture surfaces of the mechanical-test specimens, etching of the neck surfaces, barrel faces, and a section cut from the topside of the ingot, use of Baumann impressions, periscopic examination, and ultrasonic defectoscopy.

The Baumann impressions made from the neck surfaces and barrel faces should rate three points on the four-point Urals Heavy Machine Building Plant [UZTM](УЗТМ) scale. The fracture surfaces of the mechanical-test specimens are considered satisfactory when they have a fibrous or fine-grained structure and exhibit no discontinuities, pores, or nonmetallic inclusions.

Periscopic examination with a magnification of up to 1.5 should reveal no cracks, flakes, fissures, shrinkage pores, large nonmetallic inclusions (more than 3 mm), or aggregates of small nonmetallic inclusions (more than 10 inclusions less than 1.5 mm in diameter over an area of 60 cm^2) in the surface of the axial channel. Scattered nonmetallic inclusions ranging from 1.5 to 3 mm in diameter and occurring in groups of less than 10 (over the same area) or small nonmetallic inclusions ranging from 1 to 1.5 mm in diameter and not numbering more than 25 over the entire rotor-channel surface are permissible. The total number of inclusions of all sizes should not exceed 75. When the inclusions in the axial-channel surface are larger in size or greater in number, the channel is subjected to local trimming or boring to a diameter dictated by the rotor design.

The forged rotor is checked for cracks, flakes, shrinkage pores, and other gross defects by ultrasonic defectoscopy over all its cylindrical surfaces. Only isolated nonmetallic inclusions are permitted, their number and size not exceeding the following norms: the total number of defects 2-4 mm in size should be less than 30, including no more than 10 located no less than 50 mm apart in the rotor barrel and the remainder separated by distances of no less than 50, 30, and 15 mm at both ends of the rotor, depending on their specific location. Defects less than 2 mm in diameter are not counted.

The residual stresses, which are determined by cutting rings from one face of the barrel, should not exceed 5 kgf/mm^2.

The rotor blanks are subjected to thermal testing in order to determine the deformation that occurs under temperature conditions similar to those that obtain during turbine operation. The bent blank is retempered and subjected to a second thermal test.

The stringent norms described above for permissible defects in the rotor blanks require that special attention be paid to the metallurgical aspect of forging production. The experience of the Urals Machine Building Plant has shown that even smelting steel by the duplex process in an acid open-hearth furnace with a screened charge, roasted ore, and ferro alloys and employment of other measures of a similar nature cannot ensure reliable metal quality. A large increase in rotor quality and almost complete elimination of rejects or metallurgical defects has been achieved by use of vacuum casting, which is now regarded as a necessary prerequisite for obtaining high-quality rotors that satisfy the norms stipulated by technical specifications. The only drawback of vacuum casting of steel for rotor forging is the increase in carbon liquation, but this factor has no marked effect on the strength of specimens taken fromthe upper and lower ends of the blank when the forging is suitable laid out with respect to the ingot length.

Ingots for forged rotors are usually 12-sided, have a ratio $H/D = 1.5$, and exhibit a bilateral taper of less than 11.4%. It has been established that this ingot shape yields minimum development of shrinkage phenomena in the ingot and a more compact structure in the central zone of the forged rotor. The ingots are supplied hot to the press shop, with a surface temperature of 600-700°C. Special attention must be paid to uniform preforging heating of the ingot.

As an example, we will describe the technology used to forge rotors with a barrel diameter of 1140 mm from ingots weighing 51 t (Fig. 25).

The forging pattern provides for fabricating the smaller end of the rotor from the lowerportion of the ingot, in connection with the carbon distribution over the ingot height. The usual difference between the carbon content of the upper and lower portions of the ingot is markedly increased by vacuum casting and reaches 0.04-0.05% in some cases. Proper utilization of the ingots, taking into account the cross-sectional dimensions of the terminal portions of the rotor, therefore promotes higher mechanical characteristics and greater uniformity of forging strength, since the lower carbon content in the small end is compensated for by the better thermal working of the metal. The shutting pin is fabricated with 50 mm of metal from the ingot body gripped. The edges and facets are pressed to a diameter of 1400 mm, using the full width of the die block and reductions of up to 120 mm per pass. After intermediate upsetting of the ingot, it is drawn to a diameter of 1450 mm through a 1400-mm square. Composite die blocks are used for the round-drawing operations: the upper block is flat and the lower cut-away. A spherical upsetting platen is employed for the intermediate upsettin . A total of six passes are made. The forging-temperature range is 1220-800°C, except for the two final passes, for which the presence of deformed zones in the blanks makes the maximum heating temperature 1050°C or less. All the forging operations are carried out in presses exerting a force of 10,000 or 12,000 t.

We must emphasize the large influence of the deformation pattern, die-block profile, and forging temperature on rotor quality. Experimentation has established that effective welding up of defects in the ingot occurs when the stress pattern is favorable, even when highly developed shrinkage pores and discontinuities of the crack type are present in the metal. Experimental work on welding up of defects in large forged rotors has been very indicative in this respect [41]: repeated drawing of defective rotors under favorable temperature and deformation conditions led to complete sealing of defects with a reduction ratio of 2.0-3.0 (4.0 when intermediate upsetting was employed).

The required reduction ratio and the effectiveness with which defects are welded up depends largely on the deformation pattern.

Drawing of the upset billet to a circle through a square, which increases the deformation in the central zone of the forging and reduces the size of the liquate inclusions [5], and use of relative feeds l/D of no less than 0.5 and as large a reduction ratio as possible must be regarded as obligatory. It is also desirable to employ cut-away die blocks rather than composite blocks in drawing the blank into a round shape, since the entire volume of metal to be deformed is subjected to multilateral nonuniform compression in this case.

The aforementioned elements of the mechanical forging regime create the most favorable conditions for deformation of the internal regions of the ingots, which promotes welding up of pores and other axial defects, since the center of the blank is subjected to compressive rather than tensile stresses. The larger the metallurgical defects in the ingots, the more important is use of an optimum deformation regime, whose influence has been verified by comparing different procedures for forging rotors from alloy structural and special steels. This factor was specifically found to be significant in evaluating the quality of rotors forged from ingots weighing 71.5 t [35].

Among other features of rotor-forging technology, we should note that substantial displacement of the neck axis with respect to the barrel axis (exceeding 15-20 mm) is impermissible. This is dictated by the high-temperature operating conditions and by the thermal test procedure, whose results depend on the asymmetry of the metallurgical structure of the forging. Minimum neck-axis displacement promotes uniform heating of the ingot and intermediate blank and makes it easier to obtain identical per-pass reductions during drawing.

The initial heat treatment and cooling of the forgings is carried out under a special regime (Fig. 26). Normalization at 970-990°C is incorporated into the treatment procedure as part of the general cycle of double normalization and tempering and is not an obligatory component of the initial heat treatment. In practice, it can be carried ou after trimming and before the second (final) normalization. In terms of production, however, it is best combined with the initial heat treatment. The other elements of the procedure (supercooling to 280-320°C, isothermal annealing at 720-730°C, and slow cooling with the furnace to 300°C) constitute a complex of operations that ensure that the forgings will not flake and that the residual stresses will be minimal.

The temperature at which the forgings are loaded into the heat-treatment furnace is specified by the technological instructions and should not be less than 400°C. The working temperature of the furnace is held at 500-650°C. The manner in which the rotors are arranged on the furnace bottom provides for uniform multilateral heating of the metal: the forgings are placed on footing plates no less than 250 mm high at a distance of no less than 500 mm from the front and back walls of the furnace and 300 mm from the lateral walls.

After the initial heat treatment, hardness measurements, and marking out, two transverse sections corresponding to the top and bottom of the ingot are cut from the forgings for macroscopic examination, which shows the presence of flakes, nonmetallic inclusions, liquates, pores, and other metallurgical defects. Samples are then taken from the sections for chemical analysis, determining the car-

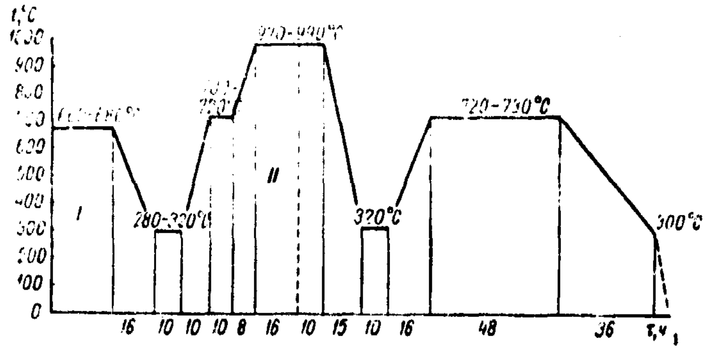


Fig. 26. Graph representing initial heat treatment and cooling of forged rotors of P2 (P2M) steel. I Accumulation of forgings; II) equilibration of charge temperature. 1) h.

bon and molybdenum contents of both ends of the rotor. When the macroscopic examination yields satisfactory results, the forgings go for machining (trimming). The allowance over the final dimensions left during trimming is generally 30 mm in the barrel diameter and 40 mm in the neck diameter.

Trimming is carried out in two stages. The blank is first machined with an allowance of 7 mm per side over the trimmed dimensions indicated in the blank plan. Superficial defects show up during this stage of machining, and, when they are present, the blank is again laid out so that they can be removed during subsequent trimming by slightly shifting the axis.

Before the fins (disks) are cut, the cylindrical surfaces of the blank are ground for ultrasonic defectoscopy. The principal reasons for ultrasonic examination of the rotor before the final heat treatment are the impossibility of thorough checking of a barrel in which fins have been cut and the need for preventive examination of metal quality in order to avoid further processing of blanks that are already unsuitable. Checking the rotor during the intermediate stage of heat treatment does not cause any noticeable distortion of the results through spurious impulses, since the first normalization substantially reduces the structural heterogeneity present in metal heat-treated without recrystallization. When doubtful results are obtained, the ultrasonic defectoscopy can be repeated from all accessible surfaces after the final heat treatment.

Analysis of the results obtained in production monitoring of a large number of rotors showed that switching to vacuum casting almost eliminated rejects for defects detected ultrasonically: no nonmetallic inclusions or other metallurgical defects were detected in most of the forgings and only a few contained isolated inclusions or small groups of inclusions, whose characteristics were within the limits permitted by the technical specifications.

In order to illustrate the number and size of the defects sometimes detected in forged rotors, we will discuss a specific instance from the Izhorsk Plant. A forged rotor weighing 31.5 t and having a barrel diameter of 1140 mm (Fig. 25) was found to contain the following defects of the nonmetallic-inclusion type: eight aggregates of defects 2 mm in size (by comparison with a standard), two defects 4 mm in diameter and an aggregate of defects less than 2 mm in diameter in the barrel region and 5 defects 2-3 mm in diameter in the neck, in the area corresponding to the lower side of the ingot. The location of the defects and the distance between them satisfied the norms set by the technical specifications. The results of ultrasonic defectoscopy showed the forging to be acceptable. The characteristics of the defects detected ultrasonically in other rotors will obviously differ, but, as experience has shown, forged rotors produced by the process described above will either contain no defects more than 2 mm in diameter or contain a number of such defects comparable to that in the case described above.

Blanks shown to be acceptable by ultrasonic defectoscopy are machined before the final heat treatment: annular grooves (slots) are cut in the rotor barrel, the allowances are removed from the necks, the ends are cut in accordance with the trimming plan, and holes are drilled and broached in the shaft for the suspension. An axial channel is also drilled in large rotor blanks before heat treatment, thus providing more effective cooling of the rotor during normalization and to some extent reducing the danger that quenching cracks will be formed.

The final heat treatment of the rotors consists of normalization and tempering. According to the data of A.I. Chizhik [48], this type of heat treatment has certain advantages over modification for forged rotors. Quenching and tempering provides a higher impact strength in the central zone of the forging and a greater high-temperature deformation capacity, but these favorable characteristics are manifested only when the blank is heat-treated to high strength. At the values of $\sigma_{0.2}$ and σ_b commonly employed for rotors of the class under consideration (Table 14), modification has almost none of these advantages. At the same time, there is a material danger that microcracks will be formed and small metallurgical defects will develop as a result of the high thermal and structural stresses produced during quenching. Ultrasonic defectoscopy has specifically shown that there is a noticeable increase in the number and size of the defects in quench-tempered rotors.

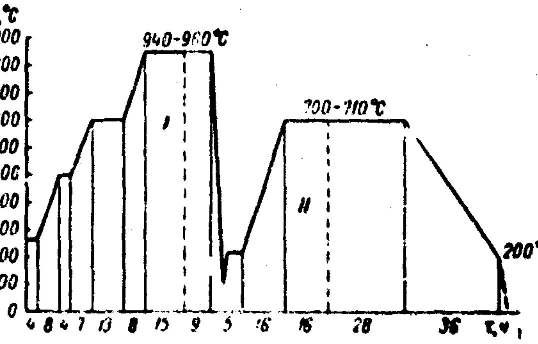


Fig. 27. Final heat treatment of rotor blanks produced from P2M steel. I, II) Equilibration of metal temperature in charge. 1) h.

Normalization is thus the most favorable heat-treatment operation from the standpoint of possible development of metallurgical defects. The required impact strength is achieved by rapid cooling of the blanks, using forced air circulation.

Figure 27 is a graph representing the final heat treatment of rotor blanks (provided that the first normalization is carried out during the initial heat-treatment cycle).

The blanks are heated in a vertical heat-treatment furnace. It must be noted that proper positioning and correct support of the rotors in the universal suspension devices is of great importance. An absolutely vertical position in the furnace is important both with respect to optimum heating conditions and to prevent bending of the rotors, which should be minimal. All other conditions being equal, the noncoaxiality of the heat-treated rotor and initial ingot decreases with the deformation of the blanks, which affects the ultimate results obtained in thermal tests. Deformation that requires the rotor to be straightened, i.e., is outside the actual machining tolerances, is especially undesirable. Such straightening means that additional labor is consumed and the technological cycle is more complex, but it can also lead to bending at elevated working temperatures. It has been found that straightened heat-treated rotors exhibit impermissible bending during thermal tests in many cases, even though they have been tempered for stress relief.

Rotor warping during heat treatment is also affected by a number of other factors, which are associated principally with nonuniform heating and cooling of the blanks during normalization. In order to obtain more uniform heating, the blank is periodically turned by 180°, which increases the uniformity with which the thermal factor acts on the rotor metal.

As has already been pointed out, the rotors are normalized with forced air circulation. A special chamber in the form of a cylindrical shaft is used for this purpose, with fans supplying air through nozzles at a pressure of about 150-200 mm H_2O. The nozzles are arranged in such fashion that the air stream is tangential to the suspended rotor.

Uniform cooling in the chamber is as important for homogeneity of mechanical properties and minimum rotor deformation as is uniform heating before normalization. The rotor suspended in the cooling chamber (in an absolutely vertical position) is therefore periodically rotated by 180° and the airflow over the chamber height is adjusted in accordance with the rotor cross-section in each given area of the chamber.

After cooling to 100-150°C, the rotor goes to a vertical furnace for tempering, during which it is also periodically rotated by 180° (every 10 h after tempering begins). The relatively low final temperature to which the blank is cooled with the furnace is intended to provide maximum relief of internal stresses. In practice, the residual stresses are relatively small when the blank is cooled with the furnace to a higher temperature (250-300°C) and they have no effect on the stability of rotor properties; however, they have a marked influence on deformation when the equilibrium of the residual-stress system breaks down, i.e., when the stresses are redistributed, as during finish machining.

By way of illustration, let us consider the actual final heat-treatment regime (normalization and tempering) for a rotor blank with a barrel diameter of 1140 mm (Fig. 25). After being placed in the universal suspension device, checked for vertical positioning in several planes, and loaded into a vertical heat-treatment furnace at 250°C, the rotor is subjected to prenormalization heating under the following regime: holding at the charging temperature for 4 h, heating to 520°C at a rate of 30 deg/h, rotation by 180°, holding at 520°C for 4 h, heating to 700°C at a rate of 35 deg/h, holding at 700-720°C until the temperature has been fully equilibrated over the rotor cross-section (11 h 35 min), rotation by 180°, heating to 950°C at a rate of 45 deg/h, equilibration of the temperature over the rotor cross-section (14 h) with rotation by 180° after 10 h, and holding at 950°C for 9 h 40 min.

The rotor is normalized in a chamber with forced cooling, rotating it by 180° 1, 2, and 5 h after cooling begins. The final temperature to which the rotor is cooled is 100-140°C. The total cooling time is 10 h 30 min. Contact-thermocouple measurements showed the following temperature distribution at five points over the rotor height at the end of the cooling period (from top to bottom): 140, 130, 140, 100 and 125°.

The rotor is loaded into a vertical furnace for tempering immediately after cooling, with a furnace temperature of 230-240°C. The tempering regime includes holding at 230-250°C for 5 h, heating to 705-710°C at a rate of 25-30 deg/h equilibration of the rotor temperature at 705-710°C for 12 h, holding at 705-710°C for 28h with rotation of the rotor by 180° after 10 h, and cooling to 200°C at a rate of 10-20 deg/h. After cooling with the furnace, the vertically positioned rotor is subjected to additional cooling in air to a temperature of about 150°C.

The actual total duration of the final heat treatment is 176 h 30 min, including 65 h 40 min for normalization, 10 h 30 min for cooling in the special chamber, and 100 h 20 min for tempering.

The heat-treated blanks are subjected to hardness and warping measurements and mechanical tests.

The hardness is measured at two points on both ends of the rotor. The absolute values are not stipulated by the technical specifications and the measurements serve only to check the homogeneity of the metal: the difference in hardness at different points should not exceed 40 HB. The bending of the rotor due to warping is checked with a benchplate laid on roller supports. The outside diameter of the barrel is taken as the measurement base. The maximum bending is determined and its location fixed by turning the rotor on the roller supports. A marker gauge is employed to determine the actual surplus over the established based measurement through all diameters of the blank. The maximum bend in the rotor should not exceed 75% of the finish tolerance. The blank is straightened if greater deformation has taken place.

Long experience at the Urals Machine Building Plant has shown that the curvature of most rotor blanks lies within the maximum permissible deformation. Thus, the results of statistical analyses made over 5 years established that 4.5% of the blanks produced were straightened after final heat treatment. In addition to factors associated with nonuniform heating and cooling of the blanks during heat treatment, increased rotor deformation was also produced by lack of an axial channel in the blank and heat treatment of rotors of different weights in the same charge.

When necessary, the rotors are straightened in a forging press while hot. The prestraightening heating regime includes loading into the furnace at a temperature of no more than 300°C, heating to 650-660°C at a rate of no more than 60 deg/h, and holding at 650-660°C for no less than 20 h. The rotor is placed in special cutaway die blocks and light pressure is applied to straighten it at the point indicated by the marker gauge. After straightening has been completed and checking of the rotor curvature yields satisfactory results, stress-relief tempering is carried out in a vertical furnace, keeping the heat-treatment cycle continuous. The rotor is loaded into the charging furnace at a temperature of 500-600° C, heated to 670-680°C at a rate of no more than 40-50 deg/h, held at 670-680°C for 15 h after the temperature has been equilibrated, cooled to 200°C at a rate of 20-30 deg/h, and then cooled in air while suspended.

Experience has shown, however, that post straightening tempering does not always provide stable satisfactory results in subsequent thermal tests. More reliable results are obtained when the final heat-treatment cycle (normalization and tempering) is repeated after straightening.

Rotor blanks in which the curvature after heat treatment does not exceed permissible levels are subjected to mechanical tests. Table 15 shows the actual results obtained in tests on four rotors forged at the Izhorsk Plant, which are typical of large forged rotors of P2M steel, as well as the actual carbon content in areas of the rotor corresponding to the top and bottom of the ingot.

The blanks have the requisite mechanical properties, with high reserve plasticity and impact strength. The mechanical properties were quite uniform and there was no material scattering of values within each rotor.

It can be seen from a comparison of the actual carbon content in different areas of the rotors with the corresponding mechanical properties of the metal that the decrease in carbon content over the ingot height had no marked influence on strength: the variation in $\sigma_{0.2}$ did not exceed 4-5 kgf/mm^2 in longitudinal and tangential specimens.

In producing forged rotors from P2 and P2M steels, the yield strength sometimes fails to reach the maximum permissible value in tangential specimens. This is undesirable, but a substantial excess over the established value of $\sigma_{0.2}$ is also impermissible, since it reduces long-term tensile plasticity and increases the sensitivity of the metal to stress concentration. In such cases, the rotor blanks are subjected to additional tempering. It is for precisely this rea-

TABLE 15

Results of Mechanical Tests on Four Rotors of P2M Steel (Fig. 25a)

1 № поковки	2 Химический состав (плавочная проба), %	3 Фактическое содержание углерода в конце ротора (%), соответствующем: 4 верхней части слитка, 5 1-й анализ	6 2-й анализ	7 нижней части слитка, 1-й анализ	2-й анализ	8 Механические свойства: 9 продольных образцов от обоих концов ротора, $\sigma_{0,2}$, 10 кГ/мм²	$\sigma_в$, кГ/мм²	δ, %	ψ, %	$a_н$, 11 кГ·м/см²	12 тангенциальных образцов от бочки, $\sigma_{0,2}$, 10 кГ/мм²	$\sigma_в$, кГ/мм²	δ, %	ψ, %	$a_н$, 11 кГ·м/см²
1	0,27 C; 0,34 Si; 0,49 Mn; 1,62 Cr; 1,03 Mo; 0,22 V; 0,012 S; 0,011 P	0,26	0,27	0,25	0,25	58,1	71,9	23,6	62,3	15,0	57,1	72,5	20,0	66,0	12,2
						56,7	72,0	22,0	62,0	17,2	56,7	72,0	21,0	61,4	12,5
						53,8	69,0	23,0	68,4	16,7	56,7	71,5	20,0	63,6	10,5
						53,0	69,1	23,6	68,1	14,5	56,1	71,3	20,0	61,5	10,5
2		0,27	0,27	0,24	0,24	57,4	74,0	21,6	67,1	15,5	57,3	72,6	22,0	61,8	11,5
						56,8	72,5	21,2	68,3	12,7	56,0	71,0	23,8	63,6	11,0
						55,5	71,5	21,6	70,5	18,5	56,1	71,3	23,0	65,1	13,2
						51,6	72,0	21,2	70,0	18,7	56,6	71,6	21,2	61,1	11,7
3	0,25 C; 0,33 Si; 0,51 Mn; 1,50 Cr; 0,97 Mo; 0,24 V; 0,012 S; 0,010 P	0,25	0,26	0,23	0,23	57,8	72,8	20,4	66,7	15,7	60,5	71,4	20,6	63,7	9,7
						57,7	73,4	20,4	66,7	13,0	57,2	71,0	20,1	65,5	11,0
						52,6	68,3	22,4	69,9	17,2	55,0	71,5	21,4	69,5	9,7
						52,6	68,6	22,4	70,5	15,7	57,0	71,0	21,4	69,7	11,0
4	0,27 C; 0,39 Si; 0,62 Mn; 1,73 Cr; 1,03 Mo; 0,27 V; 0,012 S; 0,013 P	0,27	0,27	0,24	0,24	56,0	71,6	22,0	67,8	16,5	58,5	70,0	22,0	63,3	10,0
						55,7	71,0	21,4	67,8	17,0	59,4	70,1	22,0	63,8	12,0
						53,2	71,5	21,4	67,2	15,7	57,7	72,0	21,0	64,2	12,7
						53,4	71,0	22,0	67,8	15,2	57,0	72,1	22,0	64,0	13,7

1) Forging No.; 2) chemical composition (molten sample), %; 3) actual carbon content (%) at end of rotor corresponding to; 4) upper portion of ingot; 5) 1st analysis; 6) 2nd analysis; 7) lower portion of ingot; 8) mechanical properties; 9) longitudinal specimens from both ends of rotor; 10) kgf/mm²; 11) kgf·m/cm²; 12) tangential specimens from barrel.

son that the technical specifications stipulate the maximum yield strength.

Before accelerated postnormalization cooling was introduced, about 80% of the rotor blanks exhibited reduced impact strength [50]. After switching to normalization with forced cooling, the number of ingots with low impact strength was greatly reduced and their occurrence was fortuitous. There was a marked increase in absolute impact strength, which indicates that the structural transformations that take place during normalization, which are directly related to the rate at which the rotors are cooled, have the dominant influence on impact strength.

The high values of the plasticity indices δ and ψ were stable for all the rotors. Statistical analysis of 96 rotors of P2 steel established that the relative elongation for longitudinal and tangential specimens was generally 16-20%, while the transverse necking was 55-65% [50].

Sectioning and testing of a rotor forged from an ingot weighing 47.5 t at the Urals Machine Building Plant showed that the mechanical properties of the central zone, except for individual deviations (principally in impact strength), were sufficiently high: $\sigma_{0.2} = 49-51$ кГ/мм², $\sigma_в = 72-75$ кГ/мм², $\delta = 17-20\%$, $\psi = 4$[illegible]-52%; $a_к = 2.2-7.2$ кГ·м/см² [48].

The reliability of mechanical tests on rotors is specifically governed by the homogeneity of the mechanical properties of different areas of the blank. Hardness measurements on all sides and proper selection of specimens from the rings cut from the rotor barrel are therefore important elements of the test procedure: the specimens should be from diametrically opposed sites, which to some extent ensures that any impermissible discrepancy in properties will show up.

In addition to cutting rings from the rotor barrel for mechanical testing, the residual stresses are also determined (by measuring the ring diameter before and after cutting). Additional tempering is carried out when the residual stresses exceed the established norm for the rotor in question. However, such cases are very rare in practice.

The results obtained in checking rotor macrostructure by making sulfur impressions and etching the surfaces (with ammonium persulfate and 10% nitric acid) are generally satisfactory and it is not difficult to satisfy the requirements imposed by the technical specifications. Periscopic examination of the axial channel is a more complex problem and the results obtained sometimes necessitate rebroaching and careful analysis of the defects observed.

Despite the great improvement in steel quality resulting from vacuum casting, periscopic examination still often shows discontinuities and visible nonmetallic inclusions whose aggregate characteristics exceed the norms established by the technical specifications when the surface of the axial channel is examined (within the limits of the originally stipulated diameter). Trimming to remove localized defects and local or overall broaching of the channel are permissible in such cases. Technical specifications limit the increase in the nominal channel diameter to 10%, but broaching to a larger diameter can be carried out for some blanks if the design characteristics of the rotor permit and the consumer agrees. In practice, this technique is usually employed to remove impermissible axial defects and to bring the aggregates of residual nonmetallic inclusions within the stipulated norms. As an illustration, let us consider the results obtained in checking the axial channels of two rotors forged at the Izhorsk Plant.

First rotor. The original axial-channel diameter provided by the rotor plans was 90 mm. Periscopic examination of the channel surface revealed large aggregates of point and elongated liquates, up to 4 mm in size. After additional broaching of the channel to 100 mm in the portion of the rotor corresponding to the upper part of the ingot (over a length of 3-3.5 m), a chain of inclusions 5-6 mm long was detected. Further broaching to 106 mm did not reduce the defects. Individual inclusions extending over 0.5-1.0 mm remained in the channel surface only after broaching to a diameter of 117 mm. Periscopic examination showed the rotor to be acceptable.

Second rotor. Examination of the axial channel, which was 90 mm in diameter, revealed 8 small cracks 12-40 mm long and groups of small fissures. The channel was successively broached to diameters of 97, 105 and 112 mm. No defects were detected in the channel surface at a diameter of 112 mm.

These examples emphasize the importance of keeping the deviation of the forging axis from the initial-ingot axis to a minimum in order to obtain satisfactory periscopic results, since the defects detected are principally of shrinkage and liquation origin and are directly related to the location of the channel in the central zone of the ingot.

The final quality-control operation for rotor blanks is thermal testing. This procedure shows the deformation (bending) undergone by the rotor under conditions similar to its operating-temperature regime. In essence, the thermal test consists in slowly heating the rotor to about 400-500°C (depending on its characteristics) as it rotates in a lathe, holding it at this temperature, and then slowly cooling it. When bending that exceeds the permissible level occurs (as determined by gauges during all stages of the test), the rotor is again tempered and tested. The final operations are necessary and effective only if the bending initially detected is due to partial relief or redistribution of residual stresses, as indicated by persistence of the curvature after the rotor has cooled. If the curvature observed at elevated temperatures disappears when the rotor is cooled, this is a sign of an asymmetric macrostructure that causes nonuniform distribution of thermal stresses over the rotor circumference. No heat treatment can correct such a defect. Unsatisfactory results in the final thermal tests, which are very rare, are due to disregard of the basic technological rules for obtaining maximum coaxiality between the rotor and initial ingot.

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ABSTRACT (UNCL, 0) SUMMARY OF REPORT. This book is intended for engineering personnel of metallurgical plants, scientific research institutes and design technological institutes. The book deals with problems connected with the production of forgings made of special steel used in power machinery manufacturing and in some other branches of the industry. Processes of forging the hard to deform steels are described and results of experimental research and experience gained in making forgings are analyzed. The quality of forgings is discussed in connection with conditions of their fabrication and practical recommendations on selecting rational technological processes are made. The author thanks V. N. Tokarev, P. M. Libman, E. V. Balayeva and N. I. Belan for their assistance.

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