

KINETICS OF STRUCTURAL TRANSFORMATIONS AND  
FRACTURE OF HEAT-RESISTANT ALLOYS IN  
LONG-TERM TESTING

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In the past fifty years thermal power engineering in the USSR has changed greatly both quantitatively and qualitatively. We have made remarkable progress, manifest in the substantial increase in working temperatures and pressures, the use of more complex thermal cycles, continuous improvement in the designs of steam boilers and turbines, a sharp increase in their power, and, finally, the development of single units with a capacity of 300, 500, and 800 MW and gas turbines up to 100 MW.

The basis of our progress in thermal power engineering has been the scientific research in the area of heat resistance, the successful resolution of problems in research on heat-resistant materials, and the development of technological processes of manufacturing and treating parts of power equipment and their introduction for commercial use.

The efforts of many scientists in all branches of science in collaboration with engineering and technical personnel in our plants have provided thermal power engineering with heat-resisting steels and alloys for operation up to 585°C and 255 gauge atmospheres in steam plants, using low carbon, low-alloy Cr-Mo-V steels of the pearlitic class, steels with 12% Cr of the martensitic-ferritic class, and austenitic steels. For gas turbines we have developed austenitic steels and alloys based on nickel which have long working lives at temperatures up to 800°C, making it possible to foresee steam and gas turbines with still higher operating parameters.

The provision of reliability and increased working life of parts and units of power plants is the central problem in power engineering. The greatest danger is the failure of parts or units in operation, particularly in single units, the shutdown of which leads to immense losses. For this reason the study of the origin and development of fracture is one of our most important tasks. Unfortunately, our studies are still far from complete.

Kinetics of the Process of Fracture. At our institute (the work of I. L. Mirkin, Yu. A. Rybakova, A. A. Yudin, M. A. Endzelin) we have investigated the kinetics of the formation and development of centers of fracture in creep tests of materials, using copper as a model, and single-phase and multi-phase nickel alloys, including the commercial alloy KhN70VMYuT. The samples were tested in tension, at low stresses, at temperatures near operating temperatures for 4,000-16,000 h. The fractures were mainly perpendicular to the stretching direction in the grain boundaries. We observed the formation of micropores, later passing into intergranular boundaries (Fig. 1). A special etching method made it possible to avoid etching the pores and to determine their number and actual size. Usually one observes first single pores and then lines of pores. In alloys with fine precipitates of hardening phases in the grain boundaries the development and merging of pores is inhibited and the pores slowly flow around the particles of the hardening phase, not fracturing them. Great numbers of measurements have shown that the sizes of the pores are quite similar, ordinarily being 0.5-5  $\mu$ , and always less than 10  $\mu$ ; in 80% of all the pores in one sample the size differs by no more than a factor of three (Fig. 2). When the

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The Institute of Metallurgy of the Academy of Sciences of the USSR and the section on physical chemistry and technology of inorganic materials held a jubilee scientific meeting April 4-6, devoted to the 50th Anniversary of the Great October Socialist Revolution, in which prominent metal scientists of the USSR reviewed the heat resistance of metals and alloys. One of these reports is published here - Editor. The present article is based on data obtained by the authors together with Yu. A. Rybakova, A. A. Yudin, E. F. Dubrovskaya, L. S. Marinenko, Z. N. Petropavlovskaya, S. A. Il'inykh, and M. I. Solonouts. TsNIITMASH. Translated from *Metallovedenie i Termicheskaya Obrabotka Metallov*, No. 8, pp. 8-19, August, 1967.

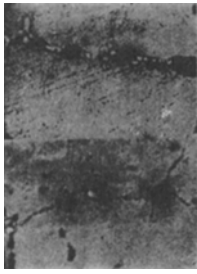


Fig. 1. Micropores in grain boundaries of KhN70VMYuT alloy under a stress of  $6 \text{ kg/mm}^2$  at  $850^\circ\text{C}$ . Time before fracture 4200 h.  $\times 2000$ .

stress is increased three to five times the size of the pores remains almost unchanged, but with increasing time and, particularly, temperature the size increases sharply.

Analysis of the results made it possible to formulate a general rule of the process of fracture. The pores, visible in the microscope, form in the very early stages of creep (less than 0.1 of the time before complete fracture and less than 0.1 of the total elongation before fracture). The size and, particularly, the number increase throughout the entire process of fracture, the hydrostatic density of the sample decreasing continuously (Fig. 3). The size of the micropores increases during testing by the rule  $d_{\text{pore}}^{\text{modal}} = k\tau^{1/8}$ , i.e., the volume of pores is a linear function of time.

This rule remains the same in tests up to 16,000 h (Fig. 4a). Such a time function was also established for coalescence and agglomeration, but of the opposite sign, which shows that the growth of micropores is connected with diffusion processes. The curve of the distribution of pore sizes with time does not shift very far in the direction of larger sizes. Thus, as the testing time increases from 980 to 10,200 h, i.e., 10 times, at  $750^\circ\text{C}$  the modal pore size increases from 1 to  $2.5 \mu$  (Fig. 4b).

The results obtained from varying the stress are particularly interesting. To produce pores at the same temperature and time but at different stresses, conical samples with a small taper were used. Examination of different microsections showed that an increase of the stress has a very slight effect on the pore size but sharply increases the number of pores by the quadratic law  $N_{\text{pore}} = a\sigma^2$  or  $L = 1/\sqrt{n} = 1/k\sigma$ , where L is the distance between pores (Fig. 5).

When the chemical composition is varied, and particularly in the presence of dispersed hardening inclusions, the formation rate of pores decreases sharply and the number of pores is substantially lower. In 1000 h at  $700^\circ\text{C}$  with  $\sigma = 6 \text{ kg/mm}^2$  about 144 pores per  $\text{mm}^2$  are formed in the single-phase alloy; in the alloy with an addition of about 15% hardening phase only 36 pores per  $\text{mm}^2$  are formed in the same time period. In other words, to create the same number of pores requires double the stress under the same conditions in the given case. This explains the mechanism of the hardening influence of the dispersed phase.

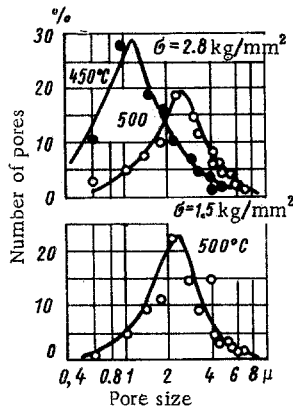


Fig. 2

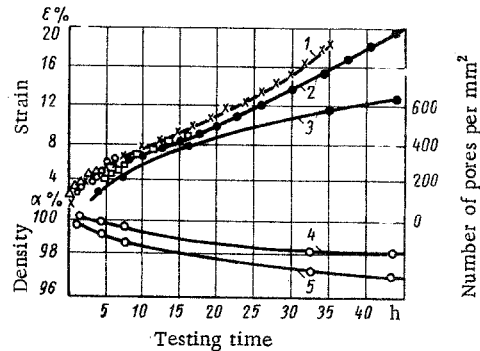


Fig. 3

Fig. 2. Size distribution curves of micropores in copper after testing 4 h at different temperatures and stresses.

Fig. 3. Variation of number of micropores and density of copper samples with testing time at  $450^\circ\text{C}$  under tensile stress of  $3.9 \text{ kg/mm}^2$ . 1, 2) Creep curve; 3) number of micropores per  $\text{mm}^2$ ; 4) calculated density; 5) hydrostatic density.

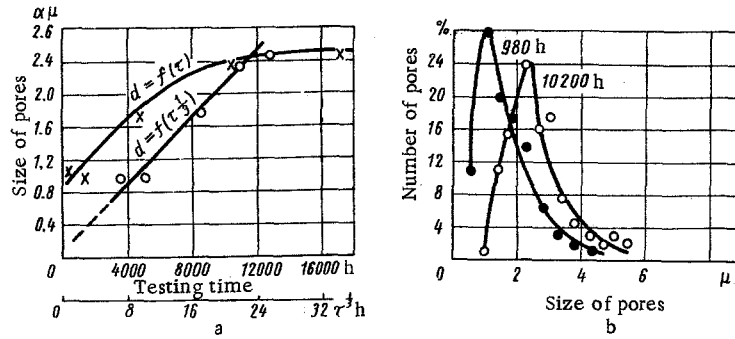


Fig. 4. Variation of micropore size in the KhN70VMYuT alloy with testing time at 750°C under a tensile stress of 15 kg/mm<sup>2</sup> (a) and distribution of pore sizes (b).

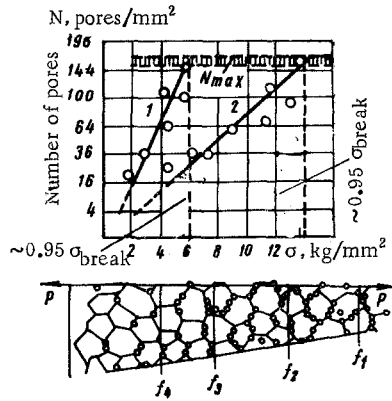


Fig. 5

Fig. 5 Variation of number of micropores with stress at 700°C in 1000 h. 1) Single-phase solid solution (Ni + 14% Cr + 2% Al); 2) two-phase alloy (Ni + 16% Cr + 3.4% Al).

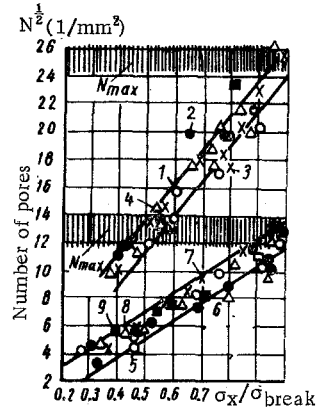


Fig. 6

Fig. 6. Variation of number of pores with  $\sigma_x/\sigma_{break}$ .

It is very significant that the transition to rapid, avalanche fracture of the sample begins at the time when the pore "density" in the grain boundaries\*  $1/L$  (where  $L$  is the distance between pores) reaches a definite and constant value for the entire group of nickel alloys. It is striking that this critical pore density is the same under different testing conditions although the rate at which it is reached varies substantially with changes in the testing conditions and the structure of the alloy. For example (see Fig. 5), in single-phase and two-phase alloys avalanche fracture occurs and the same critical pore density is reached at respective stresses of 6 and 14 kg/mm<sup>2</sup>, other conditions being equal. The constant critical pore density ( $N_{max}$ ) in different alloys for different testing conditions (Table 1) is shown in Fig. 6, where the number of pores is plotted against the ratio of the applied stress to the breaking stress under the same conditions,  $\sigma_x/\sigma_{break}$ . The experimental points lie in a narrow band for nickel alloys and copper at different temperatures and testing times up to 4,000 h.

It is obvious that the fracture pattern is similar in other heat-resisting alloys, although the parameters determining the rate of development of fracture have not been studied in detail. Research has been done on the kinetics of microfracture in pearlitic steels in the beginning stage, which are the leading steels both here and abroad (more than 90% of all boilers and turbines).

\* The proportion of the dimension of all centers of fracture to the total dimension of the grain boundaries (far from the surface of the fracture) is a small fraction and does not exceed 10%.

TABLE 1

Number in Fig. 6	Metal	Test temperature, °C	Time before fracture, h	$\sigma_{\text{break}}$ , kg/mm <sup>2</sup>
1	Copper	450	43	3.2
2		390	4	6.3
3		450	4	5.0
4		500	4	3.0
5	Ni + 2% Al + 14% Cr	700	1000	6.5
6	Ni + 2% Al + 28% Cr	700	550	16.0
7		700	1000	11.5
8	Ni + 16% Cr + 3.4% Al	700	1000	12.5
9	KhN70VMYuT steel	850	4200	6.0

Pearlitic Steels. For prolonged operation at high temperatures over 30 alloy steels of the pearlitic\* class are used, and over 50 stainless steels, high-temperature and heat-resisting austenitic steels, and nickel alloys.

Cr-Mo-V steels are the most widely used of pearlitic steels, and if we analyze their compositions we find that all possible combinations of alloying elements are used, especially with regard to the requirements for technological properties. Therefore, further development of more refined steels is hardly expedient by a systematic investigation of compositions plotted by arithmetic progression of concentrations.

A series of investigations has been made at TsNIITMASH in recent years (I. L. Mirkin, L. P. Trusov, L. S. Marinenko, E. F. Dubrovskaya, and V. A. Uporova) on the following points:

1. Increasing and retaining for a prolonged period the heat resistance and substructure of the basic  $\alpha$  solid solution.
2. Creating more crystallochemically stable and more slowly coalescing precipitation-hardening phases ensuring a smaller mean free path of dislocations, this referring to the carbide phase (and, prospectively, to intermetallic phases). Apart from the composition of the steel, the heat treatment is of great importance in creating such a structure.
3. Strengthening grain boundaries by microalloying.

The requirements of high technological properties at all stages of manufacturing restrict the overall alloying of the steel, reducing the investigation to a search for efficient proportions of the basic alloying elements – chromium, molybdenum, vanadium – effectively distributed between the phases. An increase of the molybdenum concentration increases the heat resistance of the steel but only on condition that most of it occurs in ferrite and is retained there (not passing into the carbide phase) for a prolonged period of time. For this purpose it is necessary that the major portion of the carbon in the steel be combined with other alloying elements into stable carbides (Table 2).

Thus, by doubling the vanadium concentration in 15Kh1M1F steel we increased the amount of molybdenum in the ferrite and sharply increased the amount of vanadium in the carbide phase, the basic type of carbide becoming VC, which coalesces slowly.

In testing for 10,000 h at 585°C the long-term strength increased 150%. After testing 3,000 h at 635°C, which is equivalent to 30,000 h at the working temperature of 585°C, phase analysis showed the following: In the steel with 0.35% V, and V/C = 2.7, about two-thirds of the molybdenum in the steel passes into the carbide phase. At a concentration of 0.24% V, with V/C = 1.6, the ferrite is still more impoverished in molybdenum. It is remarkable that with an increase of the vanadium concentration to 0.58%, with V/C = 4, nine-tenths of the molybdenum is found in the ferrite and this concentration is

\* We shall not deal with the problems of alloying and transformations of austenitic and nickel alloys.

TABLE 2. Carbide Analysis and Long-Term Strength of Pearlitic Steel with Different Amounts of Vanadium

Steel	Composition, %					Carbide analysis				Basic carbide (x-ray analysis)	Long-term strength, kg/mm <sup>2</sup> , at 585°C, 10,000 h
	C	Cr	Mo	V	V/C	% Mo		% V			
						in carbides	in solid solution	in carbides	in solid solution		
A	0.14	1.20	0.92	0.30	2.1	50	50	50	50	M <sub>23</sub> C <sub>6</sub>	12
B	0.16	1.22	1.06	0.57	3.6	40	60	75	25	VC	18

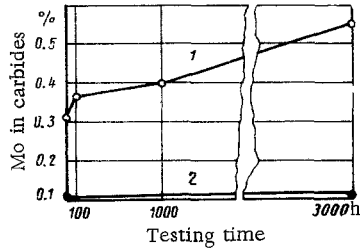


Fig. 7. Changes in chemical composition of carbide phase in long-term tests of 15Kh1M1F steel at 635°C.  
1) 0.35% V, V/C = 2.7;  
2) 0.58% V, V/C = 4.

retained in the ferrite throughout the entire testing time (Fig. 7).\* It is interesting that in passing to a working temperature of 565°C such stability of the phase composition should, in theory, be retained for a period of more than 500,000 h.

In steel of more complex composition (0.19% C, 2.1% Cr, 1.3% Mo, 0.14% Nb, 0.8% Si) with increased concentrations of chromium and molybdenum the rule remains essentially the same, the distribution of molybdenum between the ferrite and carbide phases becoming still more favorable. A relationship between the vanadium concentration and the time before fracture was also found, and it was established that the optimum vanadium concentration is about 0.5% (Fig. 8). In this case the long-term strength at 580°C for 10<sup>5</sup> h is substantially increased (Fig. 9).

Fineness of Carbides in Pearlitic Steels. The fineness of the carbide phase and its influence on the heat resistance of the steel is controlled by the heat treatment. At constant austenitizing temperatures and tempering conditions the main factor becomes the rate of cooling from the austenitizing temperature, which in our experiments was varied by a factor of 1000 (from annealing to quenching). Regardless of the high tempering temperature (750°C, 5 h) the fineness of the carbides varied essentially as a function of the preceding cooling rate. The number of carbide particles is also inversely proportional to the mean free path of dislocations (Fig. 10, curve 3) and varies in a complicated manner. There is a sharp peak at a cooling rate of 500°/h for 20Kh1M2FL steel. It is remarkable that the curve of the time before rupture (curve 1, Fig. 10) in long-term strength tests almost exactly follows the path of the variation of the carbide particle size (curve 3, Fig. 10).

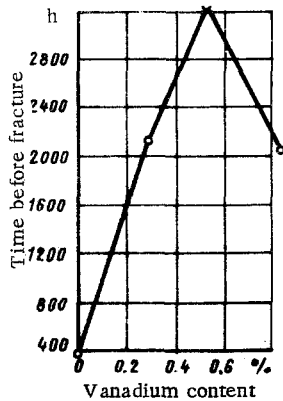


Fig. 8. Time before fracture of steel samples (2.1% Cr, 1.6% Mo, different vanadium concentrations) at 580°C,  $\sigma = 17 \text{ kg/mm}^2$ .

\* Even if nine-tenths of the molybdenum is retained in the ferrite throughout the testing time, this does not necessarily mean that the concentration is constant. The concentration of molybdenum (and any other element) depends not only on that portion of the molybdenum occurring in the carbides but also that in the solid solution and also on the quantitative proportion of the phases (Editor's note).

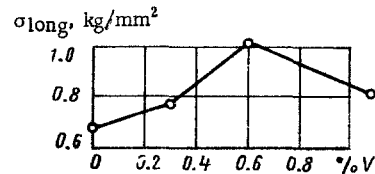


Fig. 9. Influence of vanadium concentration on long-term strength of pearlitic steel at 580°C, 10<sup>5</sup> h.

The parallelism of the curves was noted in the entire series of tests at stresses of 13–20 kg/mm<sup>2</sup>. The greatest durability corresponds to the largest number of carbide particles and changes several times with a change in the particle size. This indicates that the carbide particle size is the principal structural factor.

It should be noted that with sharply differing cooling rates the distribution and condition of the ferrite also varies. However, this does not disturb the parallelism between the particle size and the durability. The decisive importance of the distance between the dispersed particles of the hardening phase could be seen in a number of properties of different classes of heat-resisting alloys. Studies of austenitic steels in which the Laves phase precipitates during aging showed a substantial increase of the yield stress and other properties as a function of the number of precipitated particles (the work of I. L. Mirkin and M. I. Fantaeva). This connection was also established for the relaxation rate and the creep rate of complex ferritic chromium steels alloyed with tungsten and molybdenum, in which a dispersed intermetallic phase of the Laves type (Fe<sub>2</sub>W) is precipitated during aging not from the austenite but from the ferrite. Because of the variation of the distance between the hardening particles resulting from various aging conditions the mean free path of dislocations changes; the rate of creep and relaxation at 580°C varies by a factor of five to ten (the work of Z. N. Petropavlovskaya and S. A. Il'inykh).

Thus, it was demonstrated that the distribution density of hardening particles is one of the chief factors determining the heat-resisting properties in the three basic types of alloys – pearlitic, ferritic, and austenitic. It is striking that after prolonged aging equivalent to an operating time of more than 500,000 h at 565°C the carbide particle size decreases substantially. The size of the carbide particles in pearlitic steel 20Kh1MFL almost doubles, which reduces the durability almost three times (curve 2, Fig. 10). However, the form of the heat resistance curve is similar even after such long aging of the steel and the peak still corresponds to the same rate of initial cooling (in quenching).

These results demonstrate conclusively that the structural differences resulting from heat treatment and their effect on the heat resistance are not attenuated even after many years of operation. This contradicts the widespread opinion that after long periods at high temperatures the differences in the initial structure of the steel are not significant and lead to the same properties.

Since the structural differences strongly influence the heat resistance properties during prolonged operation, we propose a substantially different method of increasing the thermal stability of the structure. It involves the prevention of any martensite or bainite in the steel after quenching or their decomposition products after tempering by the use of appropriate alloying (chromium, molybdenum, vanadium) to create a steel not requiring quenching or rapid cooling but acquiring a structure of alloyed ferrite with dispersed stable carbides during slow cooling (300°/h). Such a steel, of course, will have a lower yield stress and hardness, but on the other hand will have a substantially higher thermal stability in operation and better technological properties than steel with an initial structure of temper bainite.

As a result of our research a steel was created with a lower molybdenum concentration and higher vanadium concentration which has a ferritic structure with dispersed carbides (diameter about 600 Å) and a creep strength at 585°C which is 25% higher than that of the same steel after normalization.

Microalloying. Analysis shows that horophilic elements capable of enriching the grain boundaries are effective in extremely small amounts. It can be calculated that 0.002–0.003% is sufficient for a continuous "wall" of the element 10 atomic layers thick along all the grain boundaries for an average grain size (grade 5). However, in view of the fact that the element will collect within the grains as well as in the grain boundaries, this amount should be higher. On the other hand, to create a boundary layer not of the pure element but of a chemical compound (e.g., FeB) or solid solution the concentration can be lowered. That is why 0.005% B often proves to be extremely effective. Thus, for pearlitic steel of complex composition the addition of 0.005% B produces a manifold increase of the time before fracture while doubling the long-term plasticity (Table 3, Fig. 11).

It is interesting that the addition of boron and niobium together doubles the long-term plasticity in tension. A similar effect of microalloying is quite clear in the case of austenitic alloys. For example, the long-term strength of the complex-alloyed austenitic cast alloy Kh16N25V5TYu2 at 700°C for 10,000 h increases gradually from 14 to 21 kg/mm<sup>2</sup> when up to 0.1% B is added (Fig. 11). Straight tests for 14,000 h confirmed this result. At higher boron concentrations the long-term strength increases hardly at all. Similar results were obtained for the same alloy after forging.

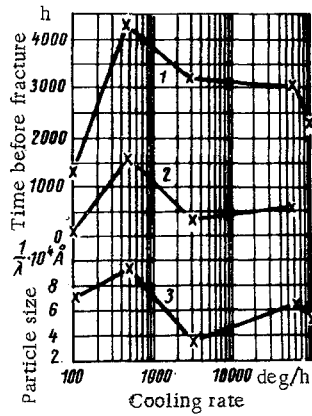


Fig. 10

Fig. 10. Variation of time before fracture of 20Kh1M2FL steel and carbide particle size with cooling rate after austenitizing at 1020°C. Tested at 565°C,  $\sigma = 13 \text{ kg/mm}^2$ . 1) After heat treatment; 2) after aging 300 h at 630°C (equivalent to 700,000 h at 565°C); 3) particle size,  $1/\lambda$  ( $\lambda$  is average distance between carbide particles).

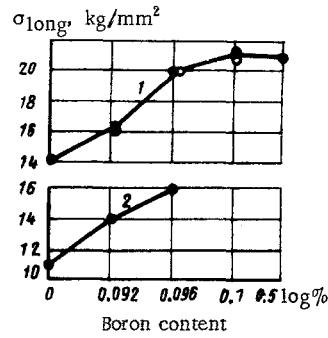


Fig. 11

Fig. 11. Variation of long-term strength ( $10^4 \text{ h}$ ) of austenitic steel Kh16N25V5TYu2R with the boron concentration. 1) Tested at 700°C, cast steel; 2) tested at 725°C, forged steel.

TABLE 3. Influence of Nb, B, and Nb + B on Heat-Resistant Properties of Pearlitic Steel\* at 560°C,  $\sigma = 16 \text{ kg/mm}^2$

Composition, %		Time before fracture, h	$\epsilon_K, \%$	$\psi_K, \%$
Nb	B			
—	—	177	5.9	9.2
0.14	—	1080	8.4	11.4
—	0.005	1703	10.4	23.0
0.14	0.005	1583	20.6	42.7

\* Basic composition: 0.16% C, 0.6% V, 1.9% Cr, 1.5% Ni, 1.4% Mo.

The importance of the boundary layer of a horophilic element and the direct connection with the grain size of heat-resisting alloys is particularly clear in the case where the basic structural components are not completely transformed, i. e., in austenitic steels and nickel alloys.

Influence of Grain Size on Creep Rate of 1Kh18N12T Steel. At a relatively low temperature (550°C) the steady creep rate (Fig. 12) increases noticeably with the grain size. In logarithmic coordinates this dependence is almost linear, but it takes on a different character at a higher testing temperature, and at 650°C the creep rate increases by almost two orders with decreasing grain size. The creep rate increases sharply with a grain size of 0.05 mm (grade 8).

The substantial importance of the grain size is also clear in nickel-base alloys. Grains of gigantic size are often created in the commercial alloy KhN70VMYuT (ÉI765). Macrograins 0.05–10 mm were produced in this alloy by changing the heating conditions after cold working. The short-term and fatigue characteristics were studied and it was found that gigantic grains have a negative effect on the heat resistance and other properties. Similar results were obtained also for other alloys, for example Kh15N65V10M5TYu (ÉI893).\* The best general properties are obtained with grains 0.05 mm in diameter; sizes down to 0.25 mm are permissible.

\* For this and other alloys, drop forging and heat treatment eliminates the differences in grain size and increases the heat resistance. (Editor's note).

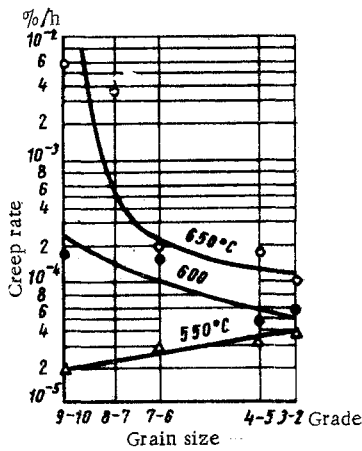


Fig. 12. Variation of creep rate of austenitic steel Kh18N12T with grain size at  $\sigma = 10 \text{ kg/mm}^2$  in range of 1000-3000 h.

Results of 100,000-h Tests. The minimum guaranteed period of operation of stationary power equipment exceeds 10 years. Thus, it is of interest to determine by direct tests the actual changes in the structure and properties of heat-resisting steels and the actual magnitude of the distortion and the time before failure at working temperatures. Such a unique investigation, in spite of the technical complications and high costs, has been completed at TsNIITMASH. The length of the experiments exceeded 95,000 h. The investigation was conducted with a series of 14 samples of two austenitic steels at 580°C.

It was found that, in spite of the comparatively low temperature for austenitic steels (580°C), phase and structural transformations occurred continuously and were not completed for some tens of thousands of hours.

The first of several unexpected results was that the average austenite grain size doubled in 50,000 h at 580°C, although in the usual tests for 1-3 h the recrystallization threshold is determined to be 850°C. There is a good theoretical explanation for this discrepancy, since by the exponential law of recrystallization allowing for the activation energy of the process the increase of the time by a factor of  $10^4$  is equivalent to such a sharp reduction of the recrystallization threshold.

Phase and Structural Changes. Substantial changes also occur in the carbide phase. In spite of the fact that the tests were made with metals previously stabilized, the amount of carbide phase increases greatly in 20,000 h at 580°C and, even though slowly, the phase composition also changes up to 80,000 h (Fig. 13). At the same time, the composition of the solid solution also changes, the lattice constant decreasing continuously. This indicates how slowly equilibrium is established in such alloys at the working temperature.

Along with the changes in the composition there are continuous changes in the structure because of the coalescence of carbides of the  $M_{23}C_6$  type and alloyed tungsten and molybdenum. The average particle size after 95,000 h at 580°C is  $3 \cdot 10^{-5} \text{ cm}$  (Fig. 14). The calculated average linear growth rate of the particles is  $10^{-9} \text{ cm/h}$  – an increment of one atomic layer in the carbide particle requires 100 h at the given temperature. Apparently this characterizes the very high degree of stability observed in the steel and is explained solely by the low diffusion rate of tungsten and molybdenum in austenite at 580°C. Furthermore, the extreme complexity of the lattice of the  $M_{23}C_6$  carbide in comparison with the simple austenite lattice determines the high surface tension at the interface and the high energy of formation of two-dimensional nuclei; this also slows down the growth rate of the particles.

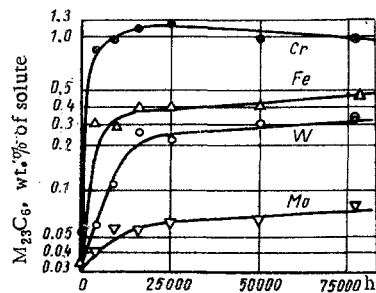


Fig. 13

Fig. 13. Variation of the composition of the carbide phase in 1Kh14N14V2M steel at 585°C with testing time.

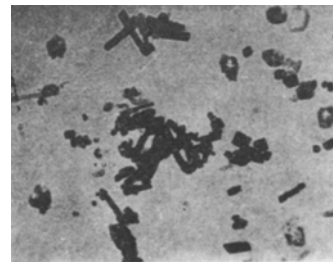


Fig. 14

Fig. 14. Carbide particles in 1Kh14N14V2M steel after 95,000 h at 585°C.  $\times 30,000$ .



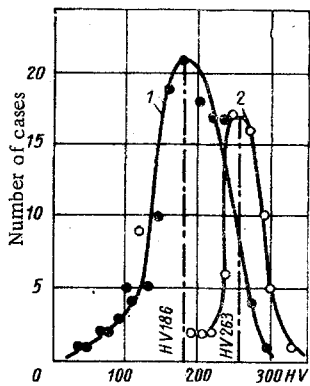


Fig. 15. Distribution of microhardness of 1Kh14N14V2M steel after testing at 580°C under a tensile stress of 14 kg/mm<sup>2</sup>. Time before fracture, 87,780 h. 1) Near fracture surface (layer less than 100 μ thick); 2) far from fracture surface (over 10 mm).

TABLE 4. Microhardness of 1Kh14-N14V2M Steel during Testing at 580°C (About 50 Measurements)

Time, h	H	
	grains	grain boundaries
0	206	210
10 700	254	286
31 000	254	291
50 000	230	196
78 500	194	172

The results obtained confirm the effectiveness of multicomponent alloying of heat-resisting alloys even for comparatively low operating temperatures.

Relationship of the Strength of the Boundary and Body of the Grain. As is well known, the trend of research has been toward strengthening the grain boundaries, since failure in prolonged operation occurs typically in the grain boundaries. An investigation of the microhardness of grains and grain boundaries (Table 4) showed that before the tests these values were almost equal. In the first 10,000 h at 580°C there was a substantial increase of the hardness of the grains and grain boundaries due to the additional precipitation of carbides, which we

mentioned earlier. The microhardness remains high up to 30,000 h, the microhardness of the boundaries being noticeably higher because of the high diffusion rate and intense precipitation of dispersed carbides in the boundaries. From 30,000 to 80,000 h the hardness of the grains and grain boundaries decreases continuously. Because of the high diffusion rate in the grain boundaries, overaging begins earlier in the boundaries, in which the hardness falls below that of the grains in 50,000 h. Thus, from 10,000 to 30,000 h at 580°C the grain boundaries are stronger than the grains.

Creep, Loosening, and Fracture of Metals in 100,000-h Tests. A study of the samples which failed after prolonged operation at 580°C (~ 90,000 h) at 14 kg/mm<sup>2</sup> also gave valuable results (Fig. 15). Statistical treatment of a great number of microhardness measurements in the vicinity of the fracture produced a sharp distribution curve with a modal value of HV 263; in the immediate vicinity of the fracture surface (layer of about 100 μ) the hardness distribution curve is twice as wide and the modal value is HV 186, which indicates that in prolonged operation the metal is loosened irregularly in the process of fracture. The drop of the "microhardness" of numerous particles below HV 100 can be explained only by the presence of pores.

Of great importance is the accurate prediction of the creep and fracture of steels. Because of the lack of a rigorous analytical theory of this process, data from direct long-term experiments are especially valuable and make it possible to check and refine any hypothesis or method of calculation. The primary creep curves up to 90,000-95,000 h in Fig. 16 show the complex forms of the curves for one alloy, which apparently are due to the substantial phase and structural changes.

Multistage curves were obtained rather than the classic three-stage creep curve. On the creep curve for a stress of 14 kg/mm<sup>2</sup> one notes at least six successive stages of accelerated and decelerated creep not including the final avalanche stage leading to fracture. In the strict sense of the word there is no steady-state creep but rather three different approximations of it. At a stress of 7 kg/mm<sup>2</sup> (close to the actual working stress for the main steam pipe of one plant) it can be considered that the first stage lasts 40,000 h; after that the creep rate remains nearly constant.

The preliminary estimate of the creep rate made on the basis of long-term strength tests up to 10,000 h was that the creep rate of the steel would be near  $1 \cdot 10^{-5}$  %/h at a stress of 7 kg/mm<sup>2</sup>, whereas it was actually one whole order lower after 40,000 h. The total distortion after 10<sup>5</sup> h at a stress of 7 kg/mm<sup>2</sup> was 0.4%, which was also somewhat lower than the estimate. Thus, because of the additional

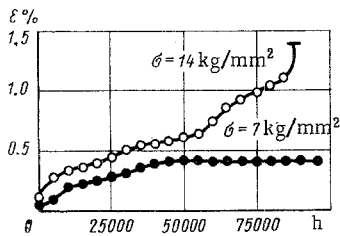


Fig. 16. Primary creep curves of 1Kh14N14V2M steel at 580°C. Testing time over 90,000 h.

phase changes and the retardation of processes which were not taken into account in shorter tests the actual performance of austenitic steel under stress for a period of 10 years proved to be considerably better than predicted.

Strength of Welded Joints. Samples of austenitic steel Kh14N14V2M (EI257), were tested under a stress of 14 kg/mm<sup>2</sup> at 580°C, i. e., double the actual working stress. Of the four samples, two failed after 85,000 and 88,000 h and two have not failed up to the present time (over 95,000 h). Under the same conditions all samples of welded joints failed after 55,000-72,000 h. From the heat-resistance diagram it can be calculated that the breaking stress for the welded joints is about 90% of that for the base metal. Since the tests were conducted at 14 kg/mm<sup>2</sup>, while the actual operating stress amounts to 7-9 kg/mm<sup>2</sup>, the results do not give rise to any apprehensions concerning the useful operating life of welded joints.