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DETAILS OF FATIGUE PROPERTY OF ALLOYS AS AFFECTED BY TEMPERATURE

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Abstract

Experimenral investigations of fatigue life dispersion for stainless steels,nickel alloys and blades manufactured from these materials determined that at normal and moderate temperatures the said parameters greatly depend on the level of stresses and the slope of fatigue curve,especially in the region of great life. The condition of the independent accumulation of fatigue and creep damages at the elevated temperature allows to have separate assessment of strength of parts by the vibrational and static stresses. The investigations of the fatigue strength of nickel alloys in the surface layers of which cold-working and residual stresses different in magnitude and sign were obtained showed their influence on the fatigue strength. The fatigue tests indicated that the decrease in strength comes when the crack achieves a definite depth.

The increase of reliability and engine life demand more complete information on fatigue strength of turbine and compressor blades and materials for manufacturing these units for investigating fatigue curve forms and endurance limits dispersion,statistical estimation of guaranteed endurance limits with the given degree of confidence. On the one hand,the variates are: load,temperature,service life,and on the other hand,on the basis of statistical ideas about the nature of fatigue failure the part fatigue performance can be given in the form of the scatter of the variates: endurance limit or life with constant stresses or with stresses in repeated succession.

The scatter of the results of the fatigue tests conducted on compressor and turbine blades is connected with some deviations of the main metallurgical and technology factors observed in the process of blade manufacturing from casting through the final operations of the mechanical treatment which affect the structure and properties of the part metal.

In this connection there were determined fatigue curves plotted using the mean values of the log life for turbine and compressor blades and also for the plain specimens manufactured from nickel alloys and stainless steels. Furthermore, based on the results of fatigue tests at one stress level the values of the roof-square deviation (r.m.s.d.) were determined,which were received at various temperatures. By using these parameters it is possible to determine the guaranteed endurance limits with the given confidence for the known form of the deviation function.*

All fatigue specimen tests at the normal and elevated temperatures were carried out at the cycles of symmetrical bending with rotation at the frequency of 200c.p.s; the diameter of the specimen plain part was 7-8 mm.

The turbine and compressor blades were tested by the electrodynamical vibrator at the oscillations with the symmetrical cycle of bending at the base frequency.

The specimens were manufactured from rods by turning and grinding,the blades were stamped from rods with consequent

*-The work was performed jointly with T.P.Zakharova.

milling and grinding. In some cases electrochemical treatment was used instead of milling. The conditions of heat treatment for the blades and specimens were the same.

The condition of metal at the surface of edges and convex side ("back") of blades is a significant factor effecting the fatigue strength. In these places during the tests and exploitation (in the presence of vibrations) a fatigue crack appears. The surface layers of the blade metal undergo plastic deformation in the process of manufacturing during stamping and mechanical treatment.

For the turbine blades manufactured from deformable nickel alloys the endurance limit relative to the metal endurance limit, determined on plain specimens, decreases by 25-30%.

Fatigue curves for the specimens and the turbine blades manufactured from the nickel alloys EI437B used for the moderately heated turbine stages are shown in fig.I. The curves are plotted using the mean values of life after the statistical treatment of fatigue test results.

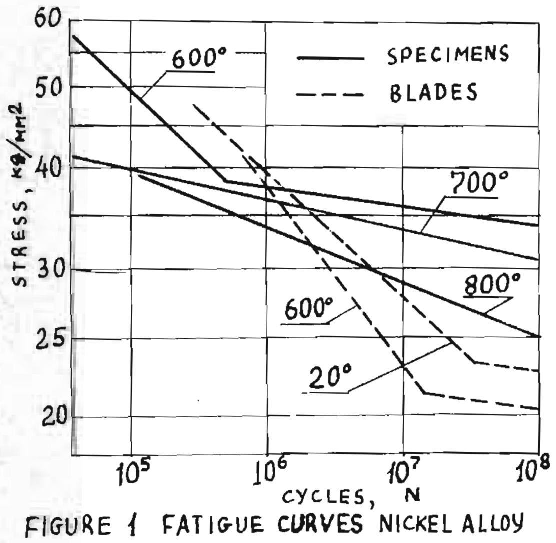

FIGURE 1 FATIGUE CURVES NICKEL ALLOY

The form of the fatigue curve presented as a broken line at the normal and moderate temperatures over the N range of $10^4 \div 10^8$ with the temperature rise becomes one line having a greater slope. The break point is displaced into the region of small cycle numbers. Approximately it may be assumed that at the temperatures equal or exceeding the temperature of the alloy ageing the fatigue curve is described by the straigth line without breaking in the region of low life.

The left of the blade fatigue curves at the temperatures up to 600°C has the same slope as for the specimens while the endurance limit doesn't exceed 70% of the material endurance limit.

Fatigue curves for the specimens and blades manufactured from the more heat-resistant alloy EI867 and cast alloy GS6K are shown in fig.2.

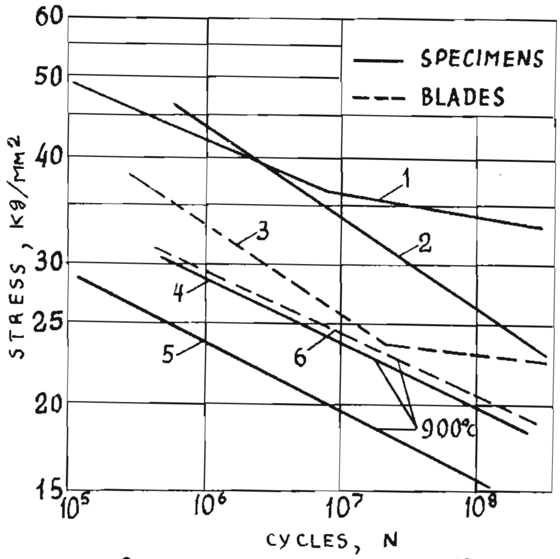

FIGURE 2 FATIGUE OF NICKEL ALLOYS

At temperature 750°C the endurance limit of the blade manufactured from the alloy EI867 also makes up not more than 70% of the plain specimen endurance limit. The fatigue strength of the cast blades manufactured from the alloy GS6K is closely connected with the conditions of metal crystalligation in the casting. The difference of the endurance limits of the blades of various sizes and profiles can become twice as large and relative to the specimen endurance limits their values can prove to be either higher or lower.

The relationship between endurance limit and temperature for the specimens from the steel EI96I (stainless steel of martensite class) and comparizon of the fatigue curves of the specimens with that of the blades are presented in fig.3 and 4. The curves are plotted using the mean values of the life logarithms. The fatigue curves at the failure probability of P=0,I are designated with dotted lines.

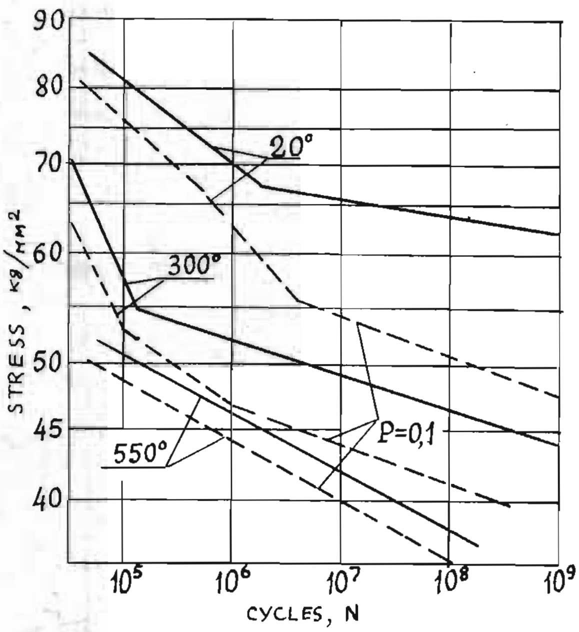

FIGURE 3 SPECIMEN FATIGUE CURVES

The nature of the fatigue curves for the specimens and the compressor blades is the same. At the normal and elevated temperatures the curves have two branches with different slope angles.

The right branch is more flat.Undoubtedly,during the tests the specimens present more homogeneous groups as to the material structure and treatment. Castings for all groups of specimens and the heat treatment were the same and only in special cases the effect of various castings was investigated.

The blades were fabricated from castings of indefinite origion. All groups of blades were of the same design,all the blades were manufactured at one plant but at different times. Depending on the manufacturing conditions these groups had some difference due to quality improving operations (check for surface burn,quality of the surface layer and oth.) The change in the mean values of life and endurance limits indicate to the technology unstability (curves 2 and 3 in fig.4).

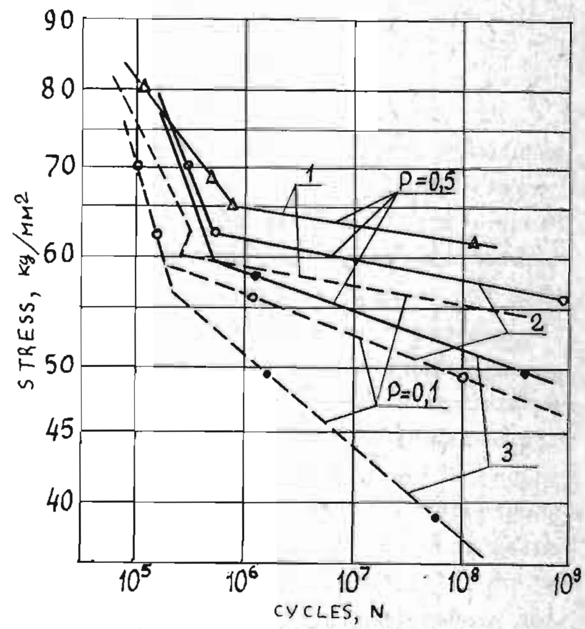

FIGURE 4 FATIGUE CURVES OF STAINLES STEEL

As for the nickel alloys with the temperature rise the straightening of the fatigue curve takes place at the temperature of 550°C,which is approximately equal to the temperature of tempering.

In the left part of the fatigue curves the effect of such technology defects as elevated level of the residual tensile stresses,surface burns,deformation of the exit edge during the correcting treatment, etc,practically is not displayed. The gain in the fatigue strength as result of the advanced technology becomes most evident if the endurance limits correspoding to low probabilities and a large number of cycles are compared. The endurance limits of the blades of the first group (curve 2,fig.4) over the cycle range of $2{,}10^6 \div 2{,}10^9$ practically are close to those of the material determined for the plain specimens,but with the failure probability P=0,I the blade endurance limits are already by 7-8 kg/mm^2 lower due to the greater scatter. It follows that the decrease in the endurance limit of blades is not connected with the effect of the shape,and that the evaluation of the production quality from the values of low lives of the left branch of the curve is unlikely to show any differences.

Thus,more certain determination of the endurance limit can be obtained if the theoretical function of the life and endurance limit dispersion is properly chosen. Numerous tests of specimens and blades with the various stress levels showed that the scatter of the log life is close to normal. Therefore the knowledge of the scatter parameters allows to define the low boundary of the endurance limits. The values of mean log life and root-mean-square deviation were approximately estimated from the scatter curves. An example of plotting the scatter curve for compressor blades is given in fig.5. At high values of life the mean values of lg N were obtained by the linear extrapolation of the scatter curve for broken specimens. Relation of the root-mean-square log life to the stress level from the fatique curves for stainless steel is presented in fig.6.

As the fatigue curves for moderate temperatures have a break it was of interest to estimate the standard deviation for the left branch and especially for the right branch. The stresses were related to the break point and as it can be seen from the graphs the dispersions also have a break at $\sigma = \sigma_0$. In the left part of the curves (this corresponds to the left part of the fatigue curves) the scatter for normal and moderately elevated temperatures is considerably less than that beyond

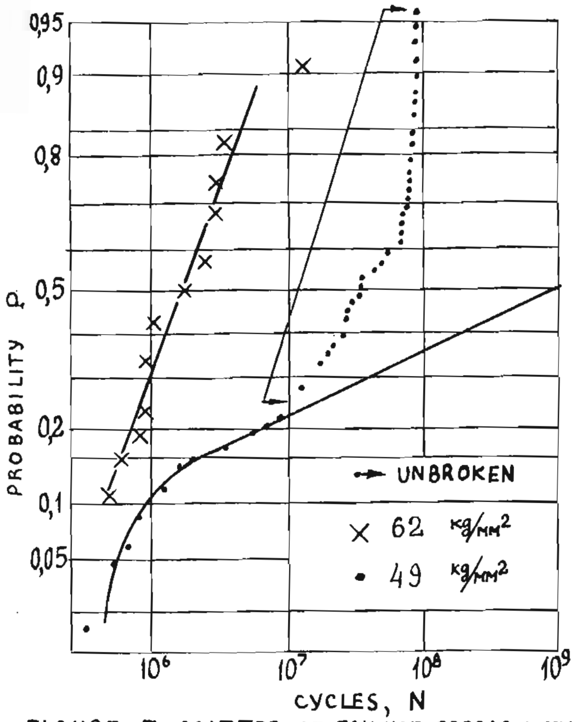

FIGURE 5 SCATTER OF FAILURE PROBABILITY FOR BLADES

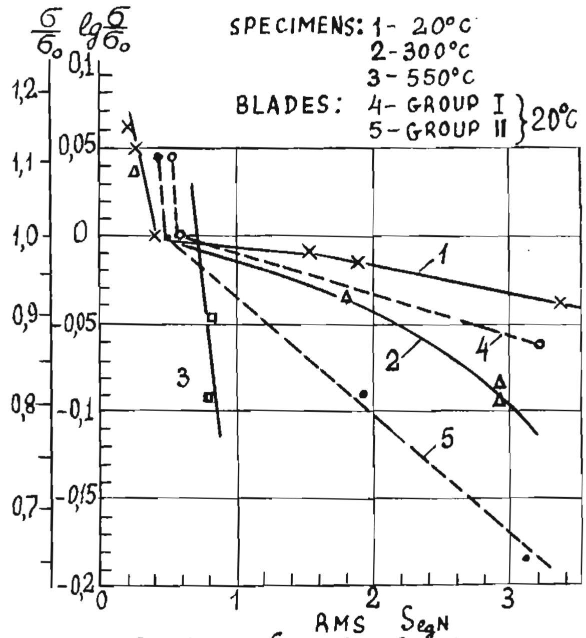

FIGURE 6 RMS S_{lgN} RESPONSE

the break (by 5-I0 times). For compressor blades the scatter value of log N slightly differs from $S_{lg\ N}$ for specimens at the stresses above the break point (S_{lgN} = =0,2 - 0,5). In the right part of the fatigue curves the scatter for the blades manufactured with the advanced technology approaches to the scatter for specimens while the second group (which had defects during the manufacturing process) at a reduced endurance limit has S_{lgN} half as large at one relative level. Such decrease in scattering can be explained by the action of the damaging factor (surface burns, stress concentration and oth.). In this connection test results for a great number of blades surface condition of which was different due to the presence of surface burns,notches as a result of the mechanical treatment,electrical polishing,and accumulation of work hours were analysed. For each series statistical characteristics were determined: $\overline{\lg N}$, S_{lgN} and ν_{lgN} - coefficient of variation and the relationship of ν_{lgN} to the power "m" of the fatigue curve was plotted. This relatioship shown in fig.7 indicates that in the left part of the fatigue curve where the power "m" is low the coefficient of variation is also negligible (up to 0,I). With the increase of the power "m" the coefficient of variation and,consequently, the scatter of log life increase. However, the coefficient of variation at the high values of "m" is stabilized and doesn't exceed the value of 0,3. With the given relationship it is possible to approximately determine the value of S_{lgN} for blades by conducting only standard fatigue tests.

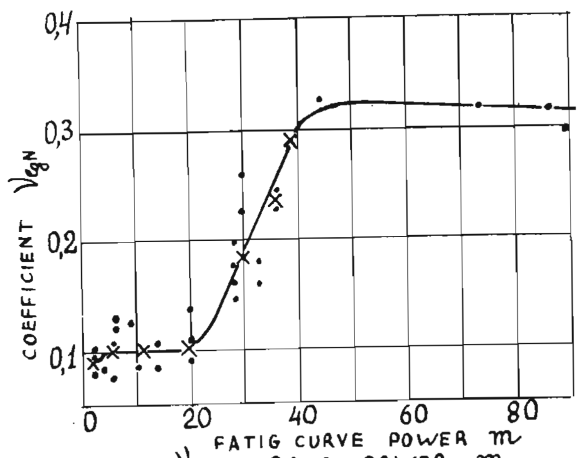

FIGURE 7 ν_{lgN} VERSUS POWER m

With temperature rise the scatter diminishes and at the temperature of tempering it becomes slightly dependable on the stress level and its value approaches that of S_{lgN},which is typical for the left branch of the fatigue curve (see fig.6).

For high-temperature nickel alloys at temperatures of ageing the scatter of log life at the different stress levels is constant and of low value,which simplifies the obtaining of the statistical characteristics. At moderate temperatures the scatter of the fatigue properties follows the obtained regularities for stainless steels.

Comparing at scattering for blades and specimens shows that for the deformable alloys the scatter is approximately the same and over $10^5 \div 10^7$cycles S_{lgN} is 0,2-0,4.

For the mild alloy GS6K S_{lgN}=0,54. These data are given in table I.

Using the above data and the usual statistical procedures it is possible to determine the endurance limits and life of blades with the given safety factor.

Table I

Alloy	t°C	N	S_{lgN} Blade	S_{lgN} Specimen
EI437B	20	$(1\div10)10^6$	0,2-0,4	-
	600	$(1\div10)^5$ 1.10^7 1.10^8	0,4	0,I÷0,4 0,6÷0,65 I,9
	700	$(1\div10)10^6$	0,2÷0,4	0,4
EI867	800	$(1\div20)10^6$	0,25÷0,4	0,3
	900	1.10^8		0,22
GS6K cast	900	$(1\div20)10^6$	0,54	0,I7

The fatique and creep damage of a material

In connection with operation of the turbine blade material under creep conditions and under action of alternating stresses the question of accumulation of fatigue and creep damage and of their mutual influence was of great interest. This phenomenon was investigated on the nickel alloy EI6I7 at the temperature 800°C from the performance test data under axial loading with a mean stress[1].

The specimens were alternatingly loaded with static and vibrational tensile stresses of various durations which stipulated damages from every type of stresses. In fig.8 a typical programme is presented. The time of exposure at the maximum static stress level in a block varied from 5 min to 4 hours, cycle number of vibrational loading in a block was measured in various programmes ($I,5.IO^4 \div 6.IO^5$ cycles).

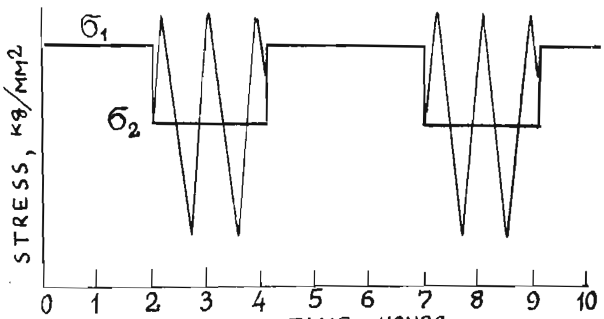

FIGURE 8 LOADING PROGRAMME SCHEME

As a performance of the creep damage programme relative amount of demage in one block was used.

$$\bar{d} = \sum_{j=1}^{m} \frac{\tau_j}{T_j} \quad ; \qquad (I)$$

and for material fatigue performance the following value was used:

$$\tilde{d} = \sum_{j=1}^{\ell} \frac{n_j}{N_j} \quad , \qquad (2)$$

which is proportional to the fatigue damage in one block.

The mean stress at the asymmetrical cycle was considered as a stage of static loading, while N_j for the vibrational loading was determined as a number of cycles to failure with the mean stress.

Where

- m - number of stages of the static loading in a block;
- τ_j - duration of one stage of the static loading in a block;
- T_j - life at the continuous testing for creep rupture strength at the stressing of the corresponding stage;
- n_j - cycle number of stress changing at one stage of the mean stress in one block;
- N_j - number of cycles to failure at the continuous fatigue testing of the corresponding stage.

The life T_j was equal to 50-I00 hours for the maximum stage and 2000 hours for the minimum stage of the static loading. The life N_j was chosen in the cycle range of $2.IO^6 \div I,5.IO^7$. The values T_j were defined as the mean results of testing of 5÷I5 specimens.

Similarly to the procedure of presenting test data with overloading and overheating the material sensivity to the regular change of the load type can be determined by using the characteristics of creep and fatigue damages.

$$\bar{a} = \sum_{i=1}^{K} \sum_{j=1}^{m} \frac{\tau_{ji}}{T_j} \qquad (3)$$

$$\tilde{a} = \sum_{i=1}^{K} \sum_{j=1}^{\ell} \frac{n_{ji}}{N_j} \quad , \qquad (4)$$

where

- m - number of stages of the static loading;
- l - number of stages with the vibration loading;
- k - number of loading blocks.

In simultaneous processes of the fatigue and the creep damages one will observe faster accumulation of damages of a definite type. This part of damages can be expressed through the ratio in a block $\bar{d}/\tilde{d}$.

Usually the number of block was in the range of 20-30, and the ratio $\bar{d}/\tilde{d}$ varied over the range of 0,I÷20.

The test results on the nickel alloy are presented on the graph of the creep damage relation (see fig.9) accumulated by the moment of failure in the specimen ($\bar{a}$), and the fatigue damage for the same test period ($\tilde{a}$). Figures at the points denote ratio $\bar{a}/\tilde{a}$. From the graph it can be noted that the effect of the fatigue damage on the creep rupture strength is slight. The point on the horizontal axis $\tilde{a}$=0, $\bar{a}$=0,7 was obtained at multiple overloadings in the creep tests without including the condition of the vibration loading ($\bar{a}/\tilde{a}=\infty$). To this value of $\bar{a}$ other conditions with $\bar{a}/\tilde{a} > I$ are approaching. The fracture analysis shows that the type of failure is dependent on the ratio $\bar{a}/\tilde{a}$. At $\bar{a}/\tilde{a} > IO$ fracture have the appearance typical to the creep. The condition $I < \bar{a}/\tilde{a} < IO$ corresponds to the compound type of fracture, while in region of $\bar{a}/\tilde{a} < I$ a fatigue fracture is observed due to a faster accumulation of the fatigue damage.

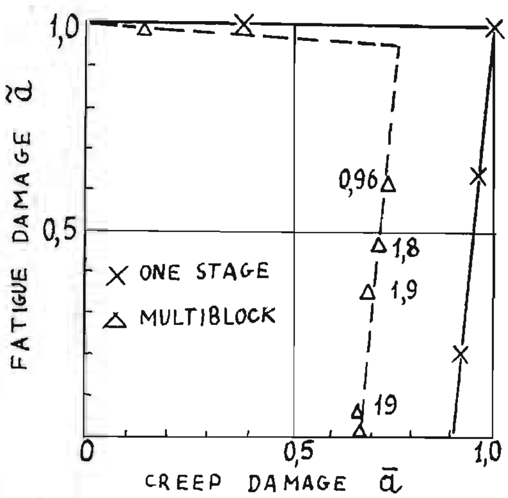

FIGURE 9 RELATION BETWEEN FATIGUE AND CREEP DAMAGES

It must be noted that there is reduced resistance to overloads in the region of fatigue failure at $\bar{a}/\tilde{a} < I$ after cold-working with tension. The intense cold-working results in weakening grain bondaries and formation of fracture of the compound type with the core at grain boundaries at the surface.

Thus it can be concluded that either creep or fatigue damages accumulated to the moment of failure under the conditions of consequent action of static and vibration conditions are practically constant in the region of fracture of the corresponding type. This gives the reason for separate estimation of the part strength by vibrational and static loads under the elevated temperature operating conditions.

Heat-resistant alloy fatigue due to the surface layer condition *

Under service conditions of engines the continuity of the part surface layer can be disturbed due to corrosion damages, notches, intercrystalline cracks at blade edges as the result of overheating, either due to the repeated thermal stresses.

It is very difficult to consider all the questions concerning the initiation of the crack and its propagation in the materials for different purposes. Therefore many investigations are of common and of applied nature as well. The check of the parts working at elevated temperatures indicates that on some of them there are surface cracks with the depth up to some tenths of mm. That is why the question concerning the safe life of such parts, the propagation of such cracks and their effect on the fatigue strength of the part rises.

It follows from the investigations of Cristensen, Forsyth, Thompson and oth.[2,3,4] that the submicrocracks along the face of sliding are being already discovered at the life of the part of 3-IO%. From the surface the crack propagates at an angle to the area of the principal normal stresses. At the second stage the mean crack is formed which propagates along the area of the principal normal stress. A particular case of crack propagation under action of oscillating loads is

*-The work was performed jointly with T.P.Zakharove and G.P.Meschaninova.

observed in the regions of high stress concentration:outturnings,slots and oth.at the stresses below the endurance limit. Here small cracks originate at an early stage of loading by the mode of the second stage, then they propagate into the region of reduced stresses and stop in their growth. When plotting the usual fatigue curve by means of rupture stresses it is necessary to increase the acting stress to the level sufficient for the crack growth throughout the section.

It was shown[5] that the stress for the crack propagation σ_p is constant and independent of α_σ . In the same investigation a calculation for the depth of the original crack capable to grow at the limit of endurance for some alloys is presented. For example,for a chrome-nickel steel this critical depth was equal to 4-5 microns. Such small crack can be detected only at magnifying.

There is a lot of suggestions on the determination of the fatigue crack growth rate by means of the number of cycles. However for turbine blades and other stressed parts of the engine the operation with the developed crack is inadmissible since the crack growth is an arbitrary process and the main attention must be given to the condition of the material before the development of the fatigue crack critical depth,corresponding to the beginning of its propagation in the material. An original surface crack can initiate,as it was stated above,not due to the fatigue. For high-temperature nickel alloys surface cracks mostly appear due to the creep and high temperatures.

Cold-working and structure transformations in the surface layer of the parts from nickel alloys during mechanical treatment may be accompanied by the appearance of residual stresses of both signs reaching at the surface some tens of kilogrammes per square millimeter.

Such value of unfavourable tensile stresses can result in the significant decrease of the fatigue strength of the part. The same factors promote surface cracking due to creep. Below some results obtained from the investigation of the fatigue strength of nickel alloys at elevated temperatures in the connection with appearing residual stresses in the surface layer and cracking under the action of static loads are presented.

Fig.IO shows the results of measurement of residual stresses in flat specimens from the nickel alloy EI6I7 fatigue tested at the symmetrical cycle of bending and at temperature 750°C. The residual stresses were induced by grinding,polishing,electropolishing and lapping-in. It is convenient to compare test results in respect to the specimens electropolished to the depth of 20-30 microns which don't contain any residual stresses in the surface layer and the cold-working as well.

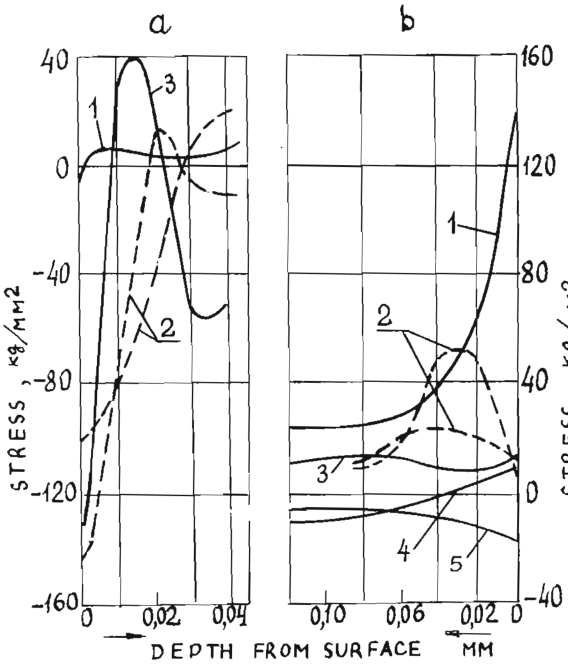

FIGURE 10 RESIDUAL STRESSES

In fig.II fatigue curves are presented. The obtained results allow to make the following conclusions:compression residual stresses of the significant magnitude (up to I20 kg/mm^2),lying to the depth of 20-40 microns for surface cleanness (IO-I2class) due to lapping-in with the cast iron tool to the depth of 0,I5-0,20 mm (curve 3 in fig.IOa) promote obtaining the endurance limit approximately of the same value as for the electropolished specimens without residual stresses and cold-working (curve 2 in fig.II).

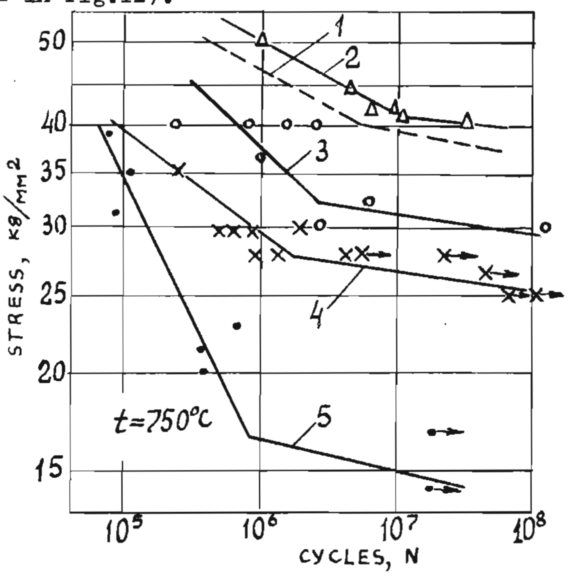

FIGURE 11 FATIGUE CURVES OF SPECIMENS

The explanation to this phenomenon may be sought in the fact that the positive effect of the residual compression stresses to a considerable extent is reduced by the cold-working and undersurface tensile stresses up to 30-40 kg/mm^2 at a small depth.

At the surface deterioration to the 8th class (curve 2 in fig.IOa) and consequently at the increasing of cold-working in depth the residual compression stresses at the surface approximately up to IOO kg/mm^2 contribute to the lowering of the endurance limit by I5-20% (see fig.II curve 3) as a result of the more significant effect of the surface quality,cold-working and undersurface tensile stresses up to 20-40 kg/mm^2.

Residual tensile stresses of 20-40 kg/mm^2 lying in the surface to the depth 20-40 microns (see curve 2,fig.IOb) at the surface quality corresponding to the 8-9 class (with deeper marks)promote decreasing of the endurance limit by I5-30% (curve 4 in fig.II). Such a decrease is the result of the combination of the negative effect of cold-working and the tensile stresses at the surface.

The increase of the residual tensile stresses to 80-IOO kg/mm lying to the depth of 50-200 microns and more (which was obtained by polishing on a ring with hard base) with a high gradient near the surface at the high cleanness of treatment (9-IO class) contributes to still greater reduction of the endurance limit,i.e.by 30-60%.

During the testing of this party of specimens a lot of small cracks (cracking) appeared from which one or two developed into large fatigue cracks.

Fatigue test results on specimens are also confirmed by the testing of parts in particular fir-tree roots of blades. The estimation of the residual stresses in the contact sides of the roots indicates that at some conditions of cutting (milling in one step without forced removal of the cutting instrument as it dulls,drawing of poor quality) in the surface layer of the material to the depth of 0,Imm residual tensile stresses appear (see curves 3,4,5 in fig.IOb).

The value of the said residual stresses IO-I5 kg/mm^2 after annealing is considered to be moderate but taking into account their intense scatter it is suggestive to expect some greater values. Fatigue test results given in table 2 indicate that the decrease of the endurance limit can make up ~40%. Besides the residual tensile stresses at the surface of grooves in the drawn blade roots damages in the form of crumbling to the depth of 0,03-0,04 mm with sharp edges were observed. From these surface damages fatigue cracks initiated.

In micrographs in the place of defect the traces of plastic deformation in the surface regions in form of plastic undulations with the depth 0,05-0,06 mm due to treatment could be seen. During the milling of the blade root the regions of the plastic deformation in the form of shear formations also propagate to the depth of 0,05÷0,08 mm and become apparent more definitely than during the drawing, however metal drawing-on and the surface damages were not present at this condition.

It should be noted that the more correct conditions of treatment of the nickel alloy roots at drawing formed defectless surface layers with residual compression stresses. In this case the fatigue strength of the drawn roots was not below than that of the milled roots. The fatigue tests of the roots were conducted at the alternating bending with the static tension (mean stress σ_m=I6 kg/mm^2) and at temperature t=700^{o}C.

In these experiments which differ from those conducted on specimens by the presence of the mean stress and by the stress concentrations it can be noted that the surface damages in form of grain shear to the depth 0,05 mm in combination with cold-working resulted in significant decrease of the endurance limit.

Table 2

Number	Mode of root manufacturing	Endurance limit (amplitude of bend) kg/mm^2	Residual stresses near surface, kg/mm^2
I	Milling in one step, annealing at 950^{o}C, 2 hours in argon	9	+I0
2	Milling in two steps+annealing	9	-I4
3	Drawing+annealing	5,5	+I5

In order to determine the effect of small surface cracks formed as a result of creep on the fatigue strength of the nickel alloys EI437B, EI6I7 and oth. fatigue tests on specimens with the symmetrical rotating beam were carried out.

Preliminarily specimens were held at the static axial loading and temperature during (0,6-0,9)T_k, where T_k - the mean life in hours at the stress according to the stress-rupture strength curve didn't exceed I00 hours. After the hold on the surface of the specimens cracks with the depth of 0,05-0,I mm were obtained. The subsequent fatigue tests conducted at high temperature 700-800^{o}C stated that the decrease in the endurance limit didn't exceed I0-20%.

The notch and the same hold of specimens with the stress concentration at the static axial loading and at high temperature resulted in increasing of the endurance limit (see fig.I2); the higher the elastic coefficient of the stress concentration the greater the endurance limit. Such increase in the endurance limit can be explained by the hot working in the region of stress concentration and the initiating residual compression stresses. At the surface of the unbroken specimens cracking to a small depth of 0,I5-0,I7 mm was obtained. Crack formation proceeded mainly under the action of alternate stresses. However it will be possible to make more complete judgement concerning the effect of static overloads in the regions of stress concentration on the fatigue strength only after conducting tests with overloads in the region of compression when residual tensile stresses originate.

The preliminary hold of the specimens having the cold-worked layer of various depth from 0,05 to 0,90 mm at static axial loading showed that the surface of flat specimens after the exposure contained a lot of cracks with the depth to 0,4 mm propagated along grain boundaries. The subsequent fatigue test at the symmetrical cycle indicates to the decrease in the endurance limit up to 30%.

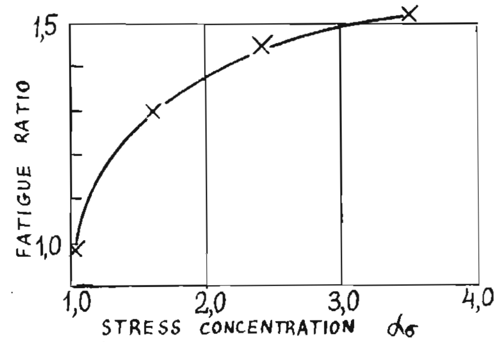

FIGURE 12 FATIGUE STRENGTH RESPONSE

These investigations show that very small cracks with the depth to 0,I mm slightly reduce the endurance limit of nickel alloys and under the condition of absence of factors promoting the propagation of such cracks (overheating,high alternate stresses,intense cold-working) they may be permissible with their consequent removal during repairing.

The surface cold-working promoted the origination of deeper cracks to 0,4 mm in the process of creeping which developed at lower stresses comparing to the initial endurance limit. Evidently the operaton of a loaded part with such cracks is not admissible. Metal condition influences the crack origination and the rate of its propagation as well. Cold-working in combination with the surface damages to the depth 0,05 mm and residual tensile stresses may lead to significant (to 40%) decrease in the endurance limit at the asymmetrical cycle.

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