

HIGH HEAT-RESISTANT CORROSION-RESISTANT SUPERALLOY

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Abstract

Increased requirements for contemporary marine and industrial gas turbines signify economic justifiability of aerospace technologies and materials application in design and manufacture thereof, in particular application of single-crystal heat-resistant alloys, including those alloyed with rhenium, as a material for turbine blades.

The particular feature of the requirements for heat-resistant materials for gas-turbine equipment of marine and industrial application is the necessity to ensure not only high heat resistance, as in aviation alloys, but also sufficiently high corrosion resistance at operating temperatures, which is associated with the use of less pure, compared to aviation kerosene, diesel fuel and associated gas. Besides, presence of sea salt particles in the air is an important factor for marine GTUs.

This paper presents a new advanced heat-resistant nickel-based alloy simultaneously possessing sufficiently high resistance to hot-salt corrosion and heat resistance level comparable to aviation alloys.

Key words: *development of nickel-based heat-resistant alloy, marine alloys, high temperature salt corrosion, creep rupture strength*

INTRODUCTION

Gas-turbine engines initially developed for aviation application, in addition to their main intent, currently obtain an ever-growing use as powerplants for power stations, gas-pumping units, and ship propulsion systems. The reason of this tendency is compactness, high capacity and low vibration of gas turbines compared to internal combustion engines and steam turbines.

The world leading manufacturers of power generation units are actively working to increase efficiency and fuel economy of gas-turbine units (GTU). As of today GTUs of over 450 MW capacity comparable to characteristics of a large hydroelectric power station have already been implemented. The operating gas temperature of up-to-date units at turbine inlet exceeds the level of 1500°C, which by 150-200°C exceeds characteristics of the units produced in the 1980s-1990s of the previous century.

Creation of GTUs of such class is a complex engineering task, which requires application of new materials with increased heat-resistant properties.

To solve the problem of insufficient high-temperature performance through the use of aviation materials traditionally possessing higher heat resistant properties is not possible first of all due to specific operating conditions for GTUs of marine and energy applications. The specific problem is the use of low quality fuel, primarily cheap diesel fuel or associated gas substantially contaminated with sulfur, vanadium and other elements, which affect corrosion resistance and durability of heat-resistant materials. Besides, the GTUs parts operated at sea and in the nearshore zone, in addition to negative factors related to fuel type and grade, are actively corroded by the marine salt medium.

The increased requirements for contemporary marine and industrial gas turbines signify economic justifiability of aerospace technologies and materials application in design and manufacture thereof, in particular application of single-crystal heat-resistant alloys, including those alloyed with rhenium, as a material for turbine blades.

Creation of a new generation of marine gas-turbine engines and competitive gas-turbine power plants through the use of aviation materials have been unsuccessful first of all because of an active corrosion effect of the marine salt medium and low-quality fuel contaminated with sulfur, vanadium and other elements having affecting the heat-resistant materials durability. Therefore, creation of heavy-loaded

GTUs meeting up-to-date requirements is only possible subject to development and industrial application of new special high-heat-resistant domestic alloys, which at the same time possess increased salt corrosion resistance, particularly for hot gas duct parts of marine turbines.

1. ALLOY DESIGN

Currently, significant parts of gas-turbine engines for marine application are manufactured from alloys with face-centered cubic nickel-based matrix, which possess high heat resistance due to a large amount (up to 70 weight %) of strengthening γ' -phase generated during dispersion disintegration of γ -solid solution at alloy cooling. A characteristic feature of the structure generation that ensures long-term rupture strength at (1000 – 1100) °C and above is a close type of γ and γ' matrices, coincidence dimensions of their matrices, and availability of coherent bonding at the interphase boundaries. The specified factors define high thermodynamic and structural stability of these materials, which further ensure their long-time performance at increased temperatures.

At the same time parts of aviation gas-turbine engines (GTE) and marine gas-turbine units (GTU) manufactured from these alloys operate in conditions that considerably differ from each other. Thus, in aviation gas-turbine engines fuels of high purification degree are used; they practically do not contain harmful contaminations. The aviation GTEs mainly operate at high altitudes, where the atmosphere is actually not contaminated. Therefore, the main cause of the blade material performance degradation in these conditions is high-temperature oxidation proceeding at a low rate.

As for blades of the marine GTUs, they operate in essentially other conditions. In these circumstances high-temperature air flowing around the airfoil surface is saturated with marine salt vapors, which contain large amounts of sulfur, sodium, chlorine and other active elements compounds that cause appearance and efficient growth of sulfide corrosion, which by far exceeds high-temperature oxidation on the surface of aviation gas-turbine engine blades. Therefore, alloys designated for marine GTUs differ greatly as far as the alloying level and alloy composition from the aviation GTEs alloys first of all by high chromium concentration, which strongly suppress sulfide corrosion. It should be noted that further heat resistance increase of nickel-based heat-resistant alloys can be ensured by their alloying with elements that possess low diffusibility and high melting temperature, first of all with W, Mo, Re and other elements (Figure 1). However, with high chromium content these elements form lamellar topologically close-packed phases (TCP-phases) that drop performance of the alloys (Figure 1 – 3). Therefore, to ensure structural stability, alloys with increased content of refractory elements should have reduced content of chromium (Figure 2).

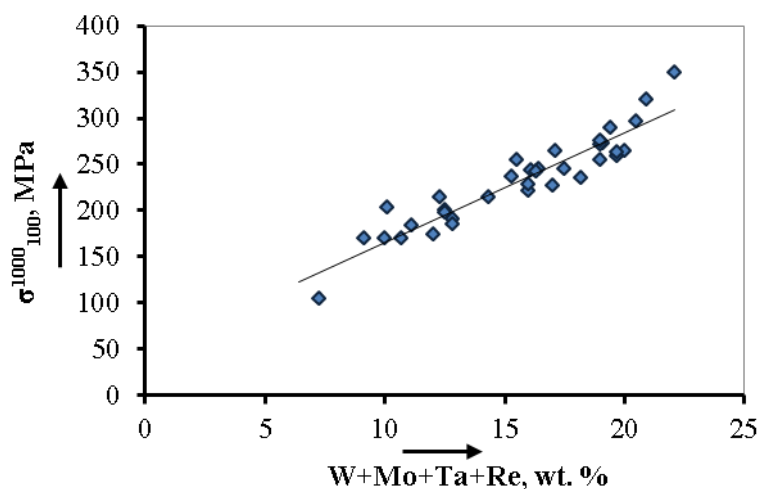


Figure 1. The effect of refractory elements on creep-rupture properties of Ni-based alloys.

That is exactly why heat-resistant properties of alloys for aviation gas-turbine engines significantly exceed the alloy properties for marine GTUs; however, their resistance to sulfide corrosion is one-two times lower.

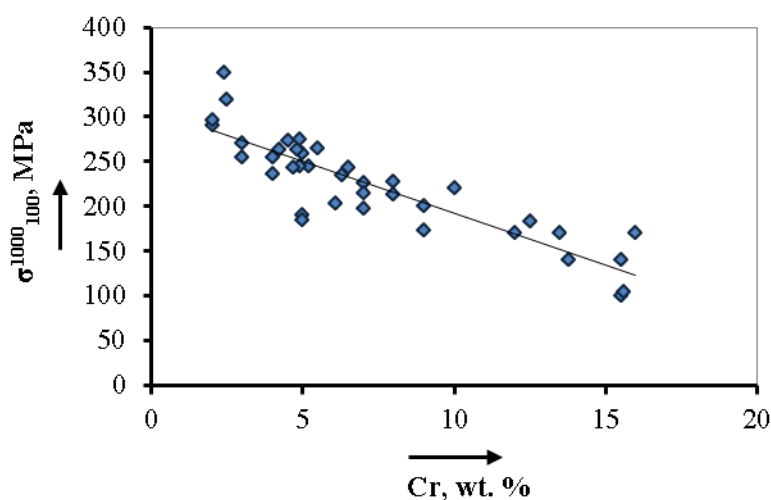


Figure 2. Creep-rupture properties versus chromium content.

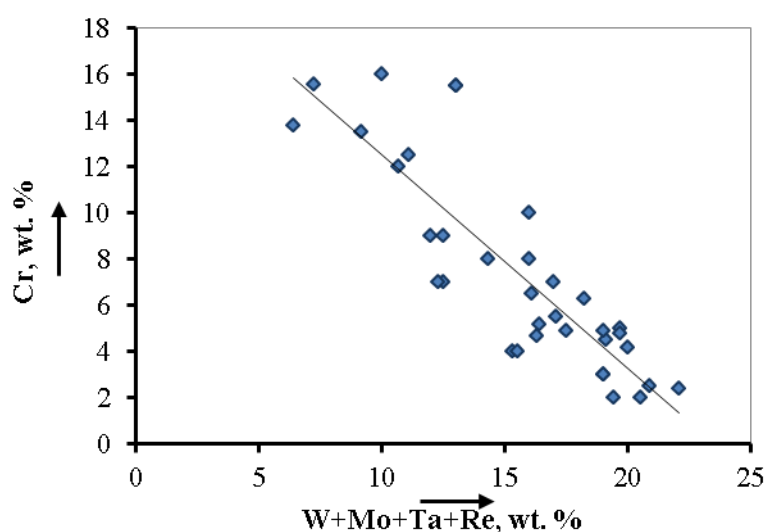


Figure 3. Interrelation of chromium concentration and refractory elements content in Ni-based heat-resistant alloys.

Therefore, creation of the alloys possessing increased resistance to sulfide corrosion and heat resistance level compliant with the alloys for aviation GTEs is a very complex multiparametric task, to solve which it is required to take into account the whole complex of thermodynamic, structural, physical-chemical and strength factors, as only on this basis it is possible to obtain the optimum composition of new alloys. To solve this task, a specialized software package “KMOZhS” [1, 2] developed by “NPO” Saturn” was used, which enables precise determination in automatic mode of the optimum alloy chemical composition versions based on the required characteristics.

While developing a new alloy, to increase the amount of the refractory alloying elements with no impact on the structural and phase stability, a decision was taken to introduce rhenium as an alloying element in the alloy composition. For economic reasons its limiting concentration was initially restricted to

4-4,5 wt.%. The analysis of the chromium effect on the Ni-based alloys resistance to gas corrosion in hydrogen sulfide atmosphere at various temperatures (Figure 4) [3] showed that sufficient minimum content of this element in the alloy should be not less than 12 wt.%.

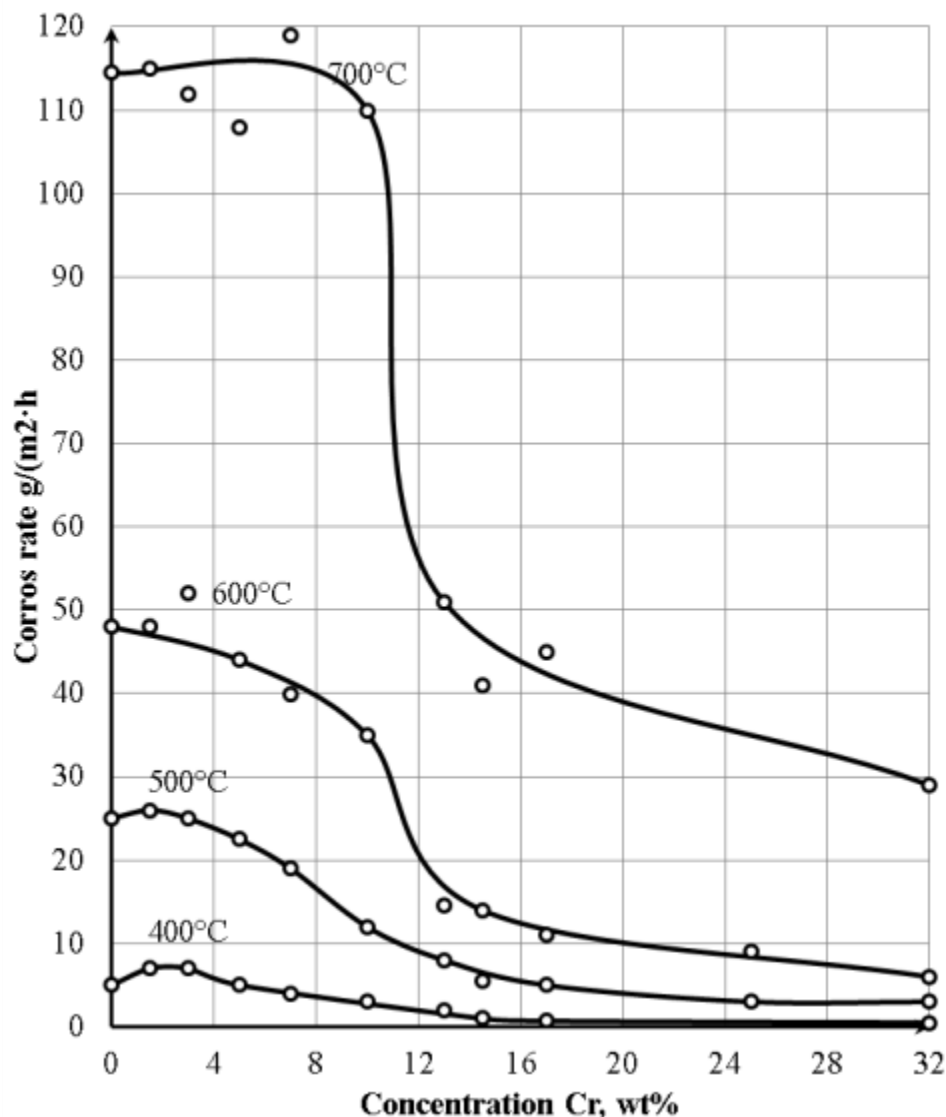


Figure 4. Effect of chromium concentration on gas corrosion resistance in hydrogen sulfide aerosphere.

Taking into account the rhenium propensity to form topologically close-packed (TCP) brittle phases with chrome (Figure 5), to assess the alloy propensity to the precipitation of TCP phases the New PHACOMP method [4] was used, which is based on calculation of the outer-shell electrons concentration in γ -matrix of the alloy. It is assumed that the critical parameter $M(\bar{d})_{\gamma}$ calculated for γ -solid solution of the alloy should be less than 0,93. As for superalloys this parameter should be significantly below, because the TCP-phases in single-crystal alloys generate mainly in the dendritic branches, which is clearly seen in Figure 6. The reason is the dendritic segregation characteristic of single crystal alloys, as a result of which dendritic branches become alloyed with tungsten, cobalt, rhenium and ruthenium, and the interdendritic space – with aluminum, tantalum and molybdenum. To search for solution, as regards correction of the $M(\bar{d})_{\gamma}$ critical value and taking into account real changes in compositions of the dendritic branches and interdendritic areas, the analysis of the literature

data concerning the nature of dendritic segregation of alloying elements of various single crystal nickel-based heat-resistant alloys after casting was carried out.



Figure 5. Single-crystal test specimen fracture zone along the TCP-phases accumulation.

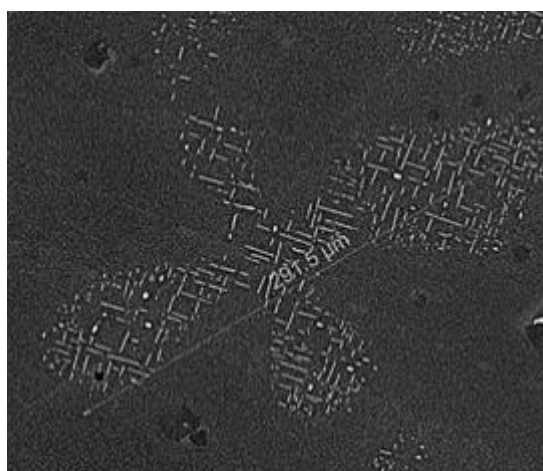


Figure 6. TCP-phases accumulation in dendritic branches of single-crystal Ni-based heat-resistant alloy.

Based on the initial data concerning the chemical composition and segregation factors of the industrial single crystal nickel-based alloys the $M(\bar{d})_{\gamma}$ values of compositions in the dendritic branches and interdendritic areas after heat treatment have been calculated.

Thus, regressional dependences of $M(\bar{d})_{\gamma_{likvDB}}$ parameters for the alloy composition in dendritic branches and $M(\bar{d})_{\gamma_{likvIA}}$ for interdendritic areas against the average alloy composition were plotted. It is obvious that $M(\bar{d})_{\gamma_{crit}}$ should be calculated not against the average alloy composition, but against the composition of the most dangerous area (in interdendritic areas or in dendritic branches) from the formulas (1÷5).

$$M(\bar{d})_{\gamma_{crit}} = 0.93 - \Delta M(\bar{d})_{\gamma_{likv}}, \quad (1)$$

where $M(\bar{d})_{\gamma_{crit}}$ – critical value above which TCP-phases are formed, $M(\bar{d})_{\gamma_{crit}}$ – value above which dendritic segregation.

$$M(\bar{d})_{\gamma_{likvDB}} = -0.00297 + 0.002033C_{Cr} - 0.0018C_{Al} + \\ + 0.00035C_{Ta} + 0.00782C_{Re} \quad (2)$$

$$M(\bar{d})_{\gamma_{likvIA}} = -0.06755 - 0.00205C_W + 0.01063C_{Al} - \\ - 0.00059C_{Ta} - 0.001758C_{Re} \quad (3)$$

where C_i – concentration of the alloying element in the alloy (wt. %).

Then the critical values of the $Md(\bar{d})_{\gamma}$ parameter for compositions in the dendritic branches and in interdendritic areas are respectively equal to:

$$M(\bar{d})_{\gamma_{crit_likvDB}} = 0.933 - 0.002C_W + 0.0018C_{Al} - \\ - 0.00035C_{Ta} - 0.0078C_{Re} \quad (4)$$

$$M(\bar{d})_{\gamma_{crit_likvIA}} = 0.997 + 0.002C_W - 0.0011C_{Al} + \\ + 0.0006C_{Ta} + 0.0016C_{Re} \quad (5)$$

Based on the above theoretical outlays with the use of the computational complex “KMOZhS” (Figure 7) the optimum chemical composition of the new advanced corrosion-resistant high-temperature alloy (**NewAlloy**) was determined, which meets all the requirements and possesses the best property package. The chemical composition of the new advanced corrosion-resistant alloy compared to the counterparts is given in Table 1.

Table 1. Chemical compositions of domestic and foreign ‘marine’ alloys [5 -12]

Alloy	Cr	Co	W	Mo	Al	Ti	Ta	Re	Nb	Hf	σ_{100}^{1000}
ChS-70	14.0	10.8	5.2	1.9	3.0	4.5	0.0	0.0	0.2	0.0	
ChS-88U	15.9	10,8	5,3	2.0	3.0	4.6	0.0	0.0	0.2	0.3	105
CMSX-11B	12.5	7.0	5.6	0.5	3.6	4.4	5.0	0.0	0.01	0.004	184
ZhSKS-1	14.0	9.0	4.0	2.0	4.0	4.0	0.5	0.0	1.0	0.0	140
ZhSKS-2	13.0	9.5	4.0	2.0	4.0	4.75	2.0	1.1	0.0	0.0	170
STAL-15	15,5	5,0	4,0	1,0	4,55	0,0	8,0	0,0	0,0	0,0	100
SCA425	16,0	5,0	4,0	1,0	4,0	2,0	5,0	0,0	0,0	0,0	170
IN792LC	12,5	9,0	4,1	1,85	3,4	3,5	4,1	0,0	0,0	1,0	
“NewAlloy”	12,2	15,0	3,0	1,0	2,9	3,5	3,5	4,0	0,0	0,1	195

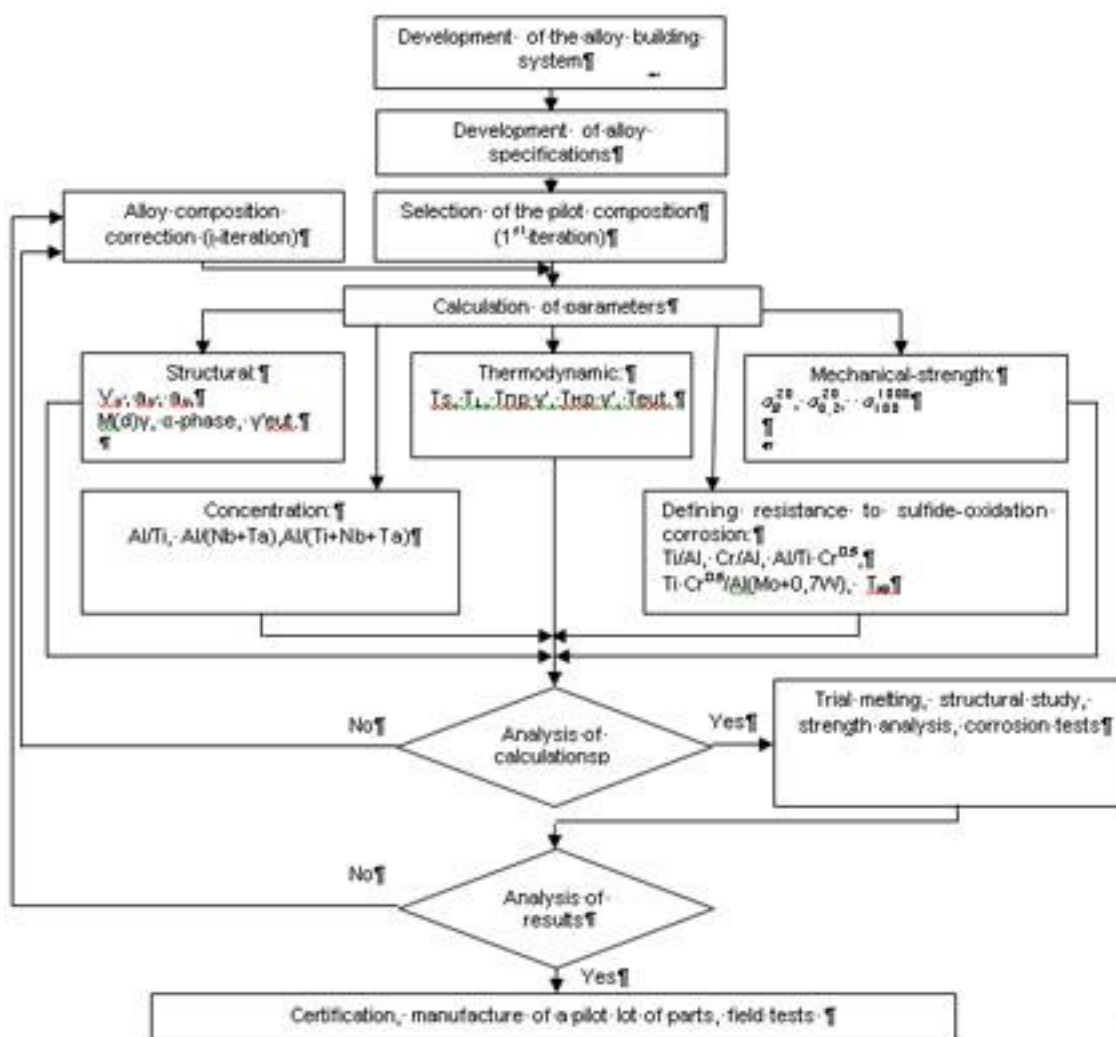


Figure 7. Computer-aided scheme of heat-resistant alloys optimization

Upon calculations a pilot lot of ingots from the advanced alloy was melted; single crystal test specimens to investigate strength and high-temperature corrosion resistance were manufactured from these ingots.

2. RESULT OF ANALYSIS

The obtained single-crystal work pieces had increased eutectic content, which was effectively removed after diffusion annealing at 1220°C for 16 hours.

Figures 8 and 9 show microstructure of a single-crystal work piece of an advanced as-cast alloy without heat treatment and after solution heat treatment, respectively.

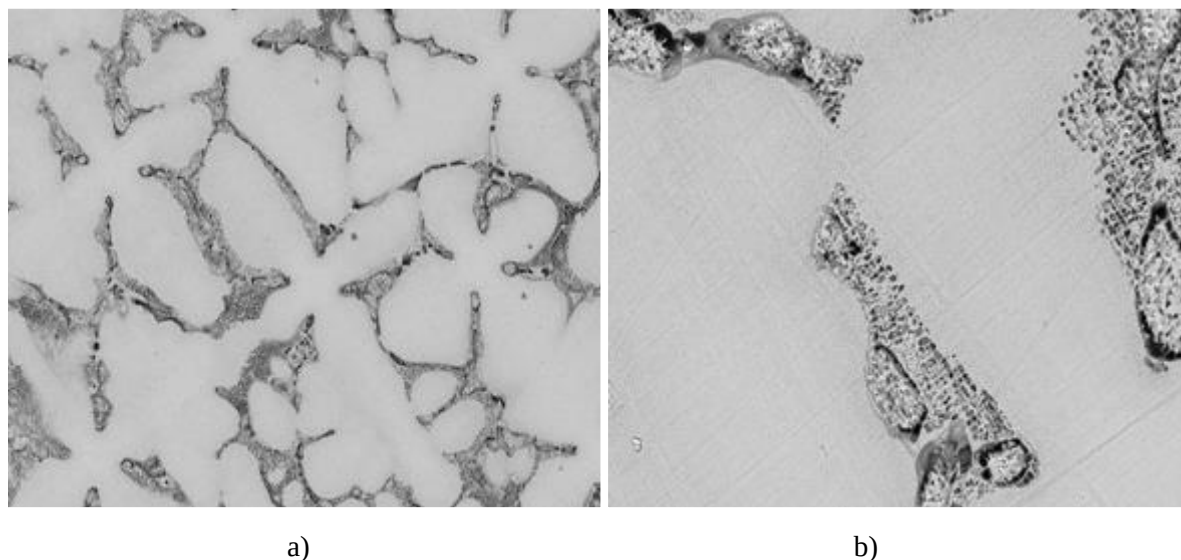


Figure 8. Microstructure of a single-crystal work piece from “New Alloy” as-cast without heat treatment: a) $\times 300$, b) $\times 1000$

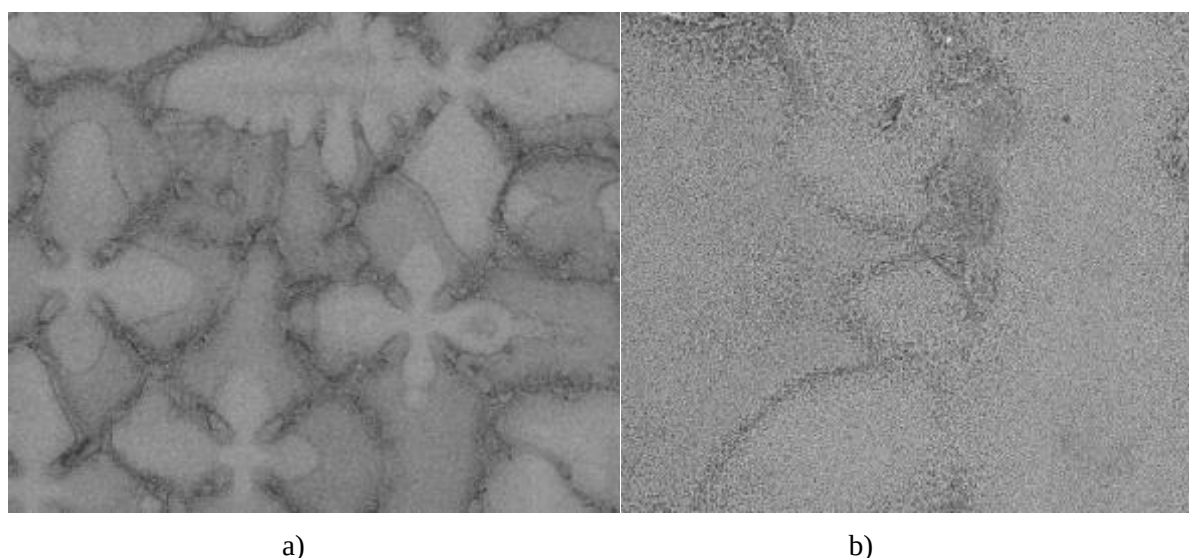


Figure 9. Microstructure of a single-crystal work piece from “New Alloy” after diffusion annealing at 1180°C for 20 hours, at 1220°C for 20 hours and air hardening to a temperature of 700°C at a rate of $80\text{-}100^{\circ}\text{C}/\text{min.}$: a) $\times 300$, b) $\times 1000$

To form an optimal structure and complete precipitation of the strengthening γ' -phase, age-hardening was performed in two stages. As far as the alloy was designed for operation at 1000°C the first age-hardening stage was also performed at 1000°C for 4 hours; after that the structure was fixed by air hardening to 700°C at a rate of $80\text{-}100^{\circ}\text{C}/\text{min.}$ As a result the main amount of the strengthening γ' -phase is precipitated, which comprises about 45% (Figure 10). Air hardening enables fixation of the strengthening phase residue in γ -solid solution, which hereafter is precipitated for 48 hours at 850°C , and this is close to the temperature at the beginning of the strengthening γ' -phase dissolution. Herewith the precipitated γ' -phase growth and generation of new particles occur.

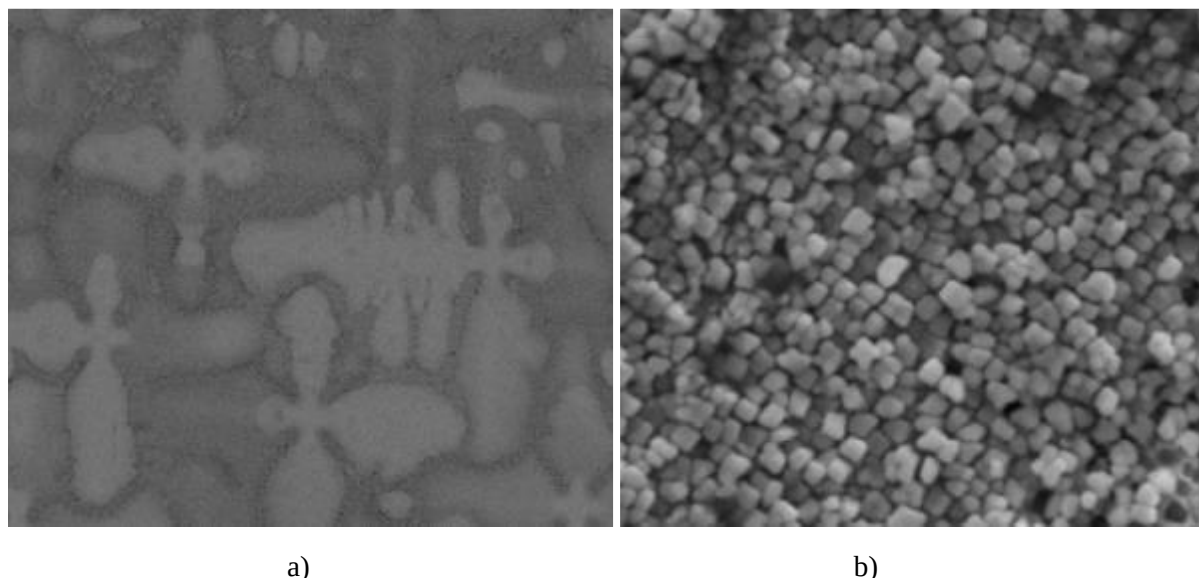


Figure 10. Microstructure of the single-crystal work piece from “New Alloy” after complete ageing cycle at 1000°C for 4 hours and at 850°C for 48 hours:
a) $\times 300$, b) $\times 20000$.

For making experimental alloy the UIPV-0,05 vacuum induction melting unit was used. Micro sections were manufactured and polished on the metallographic laboratory equipment TegraPol-21. The microstructure examinations were performed on the scanning electron microscope FEI Inspect S50 with resolution of 10 nm. Tests for short-time strength were performed on the servohydraulic unit LFV-100 with axial extensometer (High Temperature Extensometer Series HTV – 25mm base) with the use of the control, recording and processing software package DION_Stat. The creep-rupture tests were performed on the AIMA 5-1 testing machine and the creep deformation parameters from the moment of loading to the moment of the work piece rupture were recorded.

To study the alloy salt corrosion resistance, comparison crucible tests were performed on cylindrical test specimens at 900°C in molten salt: 10%NaCl + 90%Na₂SO₄. The industrial alloy ChS88U were used as standards for assessment of corrosion resistance of the experimental alloy.

Tests for resistance to high-temperature salt corrosion were performed to the ‘soft’ test technique. Layout of the unit for high-temperature salt corrosion tests is shown in Figure 11.

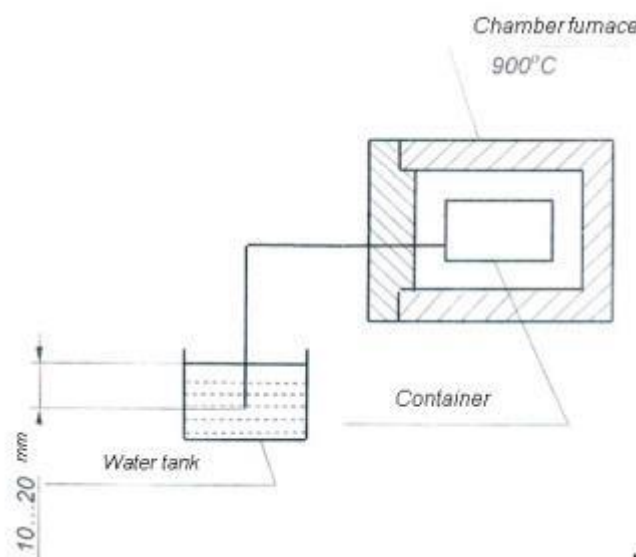


Figure 11. Scheme of the unit for high-temperature salt tests.

The tests were performed on cylindrical tests specimens 20 mm high and 10 mm in diameter.

1. The test specimens were prepared for tests as follows:
 - 1.1. The test specimens were weighed on the analytical balance accurate to 0,0001 g.
 - 1.2. The salt mixture was applied from the water solution; to prepare the mixture (40±1) g of mixture (10% NaCl + 90% Na₂SO₄) and 100 ml of distilled water were taken. The mixture was applied to the test specimens by immersion in the prepared water solution.
 - 1.3. Before immersion the solution was heated up to 80^{±10} °C; the test specimens were heated up to 90^{±10}°C.
 - 1.4. Upon the salt mixture application to the test specimens they were dried in the drying chamber at 80±10°C for 20±5 minutes. For carrying out the tests there should be 5-6 mg/cm² of the salt mixture on the test specimen surface.
 - 1.5. To obtain the specified salt mixture amount, the immersion and drying operations were repeated.
 - 1.6. The amount of the applied salt mixture applications was determined by weighing on the analytical balance accurate to 0,0001 g.
2. The tests were performed in the following sequence:
 - 2.1. The test specimens with the applied salt mixture were weighed on the analytical balance accurate to 0,0001 g and then placed in a special container. The container was installed in the chamber furnace. The test temperature was 900^{±10}°C. (Heating rate 20°C/min., cooling rate 10°/min.)
 - 2.2. The tests were run periodically.
 - 2.3. Every 50 hours of tests the test specimens were removed from the container and weighed on the analytical balance accurate to 0,0001 g. Then the test specimens were cleaned from corrosion products poorly adhered to the surface with a hair brush. The test specimens cleared from the corrosion products were weighed again. With a low oxidation rate the duration should be increased twice. At a sharp rate increase it is recommended to reduce the test time. The total test period should not exceed 300 hours.
 - 2.4. Upon completion of specimens weighing the salt mixture on the specimen surfaces was recovered to the initial state.
3. Corrosion resistance evaluation
 - 3.1. Corrosion resistance of the alloy (coating) was evaluated visually, according to the specimens weight changes at definite test intervals and by metallography to the following criteria:
 - appearance of the specimens compared to the initial state;
 - determination of the average alloy layer (coating) thickness that penetrated the oxide;
 - determination of the maximum corrosion product penetration depth into metal or depth of corrosion damages;
 - determination of the coating corrosion rate according to the kinetic curves.

The thoroughly studied ChS88U alloy was used as a reference standard to compare the level of resistance to high-temperature salt corrosion. To avoid the measurement errors, test specimens of all the alloys were exposed to a complete test cycle simultaneously.

Test specimens from the ChS88U alloy and the advanced heat-resistant alloy were exposed to tests for 250 hours. Microstructure examinations of the test specimens after the corrosion resistance tests of the alloys were performed every 50, 150 and 250 hours.

During the tests it was found that on the surface of these alloys a reasonably dense protective spinel film is generated. This film consists of the aluminum, titanium and chromium oxides; beneath the film a zone without alloying is formed (diffusion layer).

Note that the zone without alloying is increased to the surface, with exception of aluminum and

titanium. As for these elements, sharp decrease of their concentrations at the transition zone from the base metal to the diffusion zone and gradual increase of concentrations closer to the protective skin can be observed. As for the ChS88U alloy such pattern remained unchanged till the end of tests.

A few words should be said about the protective film thickness. After 50 hours of tests its thickness on the surface of the test specimens made from the ChS88U alloy comprised about 30 μm and increased as the experiment went on. The film inhomogeneity also increased due to the oxides buckling and flaking growth.

Starting from 150 hours of tests, areas with corrosion penetration as strands at the grain boundaries were observed on the test specimens from the ChS88U alloy. The strand penetration depth exceeds 1 mm (Figure 12).

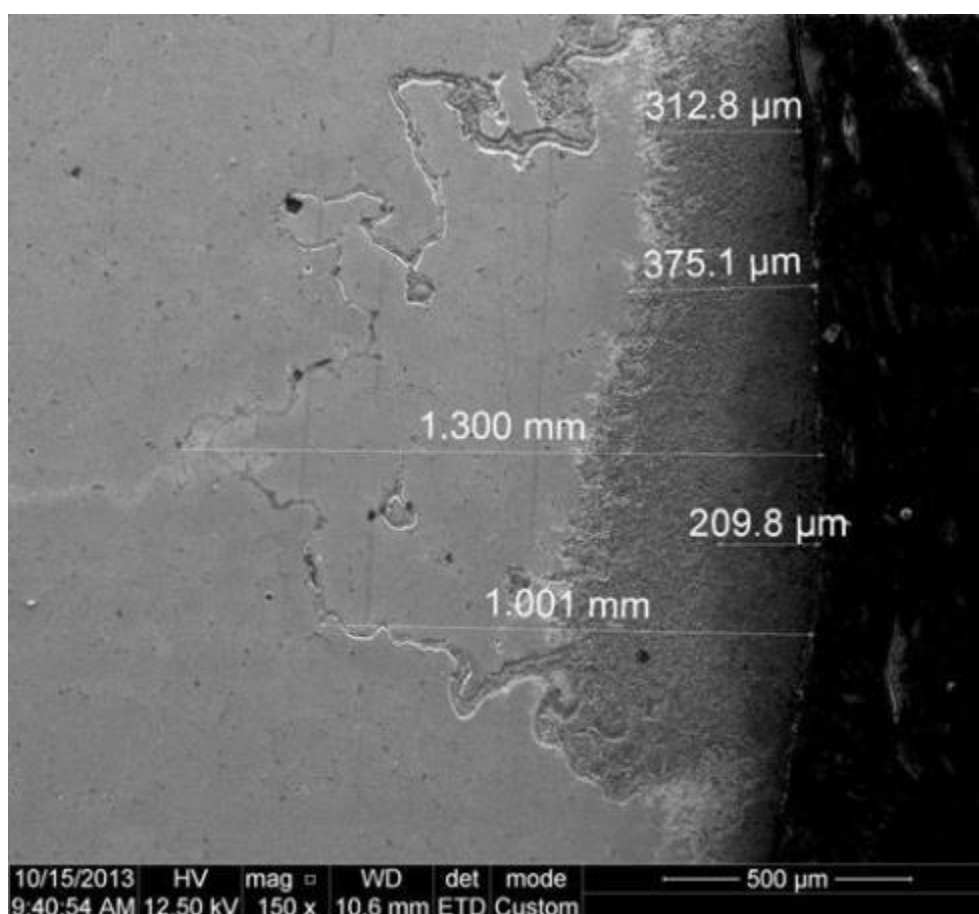


Figure 12. Microstructure of the ChS88U alloy after tests during 150 hours.

By the end of tests accumulation of the TCP-phases based on Cr-W were found in the diffusion zone near the base metal (Figure 13).

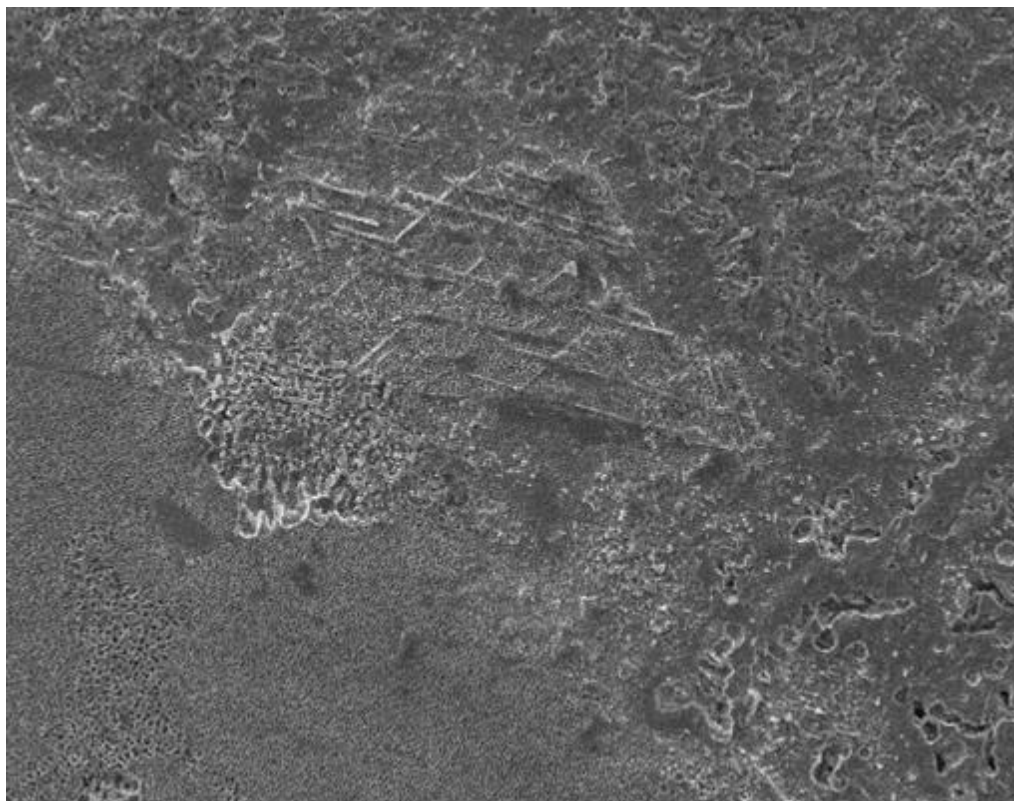


Figure 13. The TCP-phases accumulation in the ChS88U alloy diffusion layer.

The advanced heat-resistant alloy has a similar fracture pattern. Initially a skin of chromium oxide as thick as 40-50 μm is formed on the surface; beneath the skin a layer without alloy building as thick as 100 μm is located, in which concentrations of all the elements, except for titanium and aluminum, decrease in the direction to the surface. Examinations of micro sections from the test specimens after 150 hours of tests showed changes in the chemical composition of the oxide skin. About 1/5 of it is titanium oxide, the rest is chromium oxide. It should be noted that the layer without alloy building starts actively losing aluminum and tantalum located beneath the protective oxide skin. Thickness of the layer rich in tantalum comprises about 30 μm ; besides, 3-4% of tantalum is present in the oxide skin. The main part of aluminum is accumulated between this layer and the chromium and titanium oxide skin as a thin (3-4 μm thick) and dense aluminum oxide interlayer.

As the test went on, the protective skin and the diffusion zone growth was observed; upon completion of the experiment their thickness varied within 100 - 150 μm and 800 - 1000 μm , respectively (Figure 14).

The examinations show the rhenium concentration decrease in the diffusion zone, but its amount in the oxide skin is somewhat higher than its concentration in the alloy, which specifies stability of its oxides. Moreover, simultaneous decrease of rhenium and chromium concentrations makes it possible to avoid formation of topologically close-packed phases in the diffusion zone of the alloy.

Pay special attention to absence of corrosion penetration in single crystal rough parts. The reason is that there are no loosened grain boundaries. Besides, the protective skin has a uniform thickness and a dense structure.

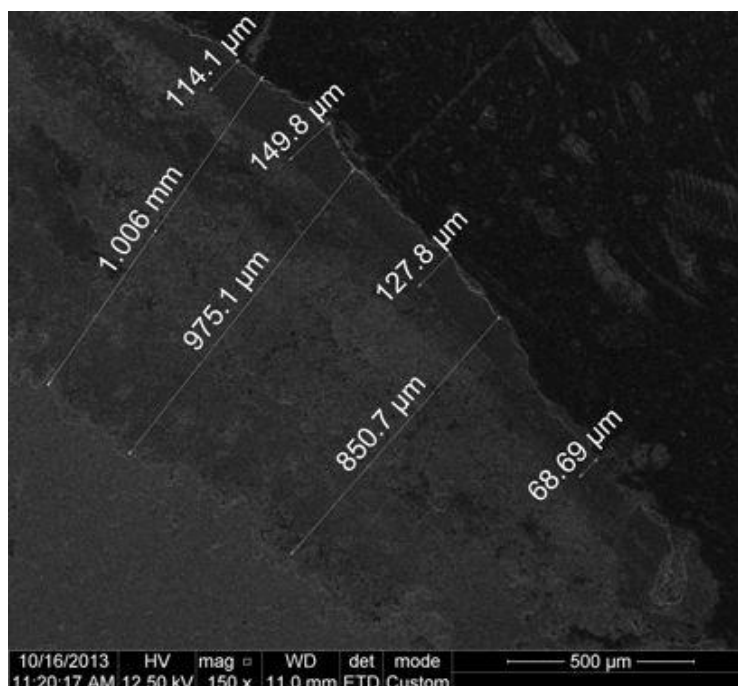


Figure 14. Microstructure of the advanced heat-resistant alloy after 250 hours of tests

Based on the analysis of dependence of the specific weight gain (Figure 15) and the alloy microstructures variation during the test it may be concluded that:

The proposed alloy has precedence over the ChS88U alloy as far as high-temperature salt corrosion is concerned.

Introduction of refractory elements such as tantalum and rhenium in the experimental composition has a positive effect on the alloys resistance to high-temperature salt corrosion thanks to slowdown of the diffusion processes.

Single crystal alloys possess higher corrosion resistance properties thanks to absence of the grain boundaries and as a result they are not susceptible to corrosion penetration.

The obtained results show high performance of the new alloy. A protective dense skin is formed on the metal surface, and as a result the test specimen weight is somewhat increased (Figure 12). However, in alloys of the ChS type on the contrary the weight is reduced, which is related to the metal loss on the test specimen surface due to formation of a loose and less effective protective skin.

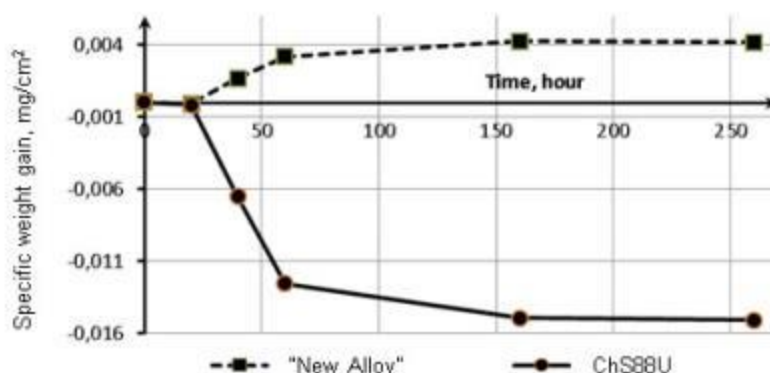


Figure 15. Weight variation of the test specimens made from the new advanced heat-resistant alloy and that of the production alloy ChS88U

Tests for structural and phase stability were also conducted, i.e. high-temperature long-term ageing at 900°C for 5 500 hours.

In the ChS88U alloy after ageing for 1500 h an abundance of TCP-phases precipitated along the grain boundaries has been identified (Figure 16).

The advanced heat-resistant alloy turned out to be more stable. For 5500 h, in its structure (Figure 17) no adverse phase separation that might lead to a sharp alloy softening has been identified.

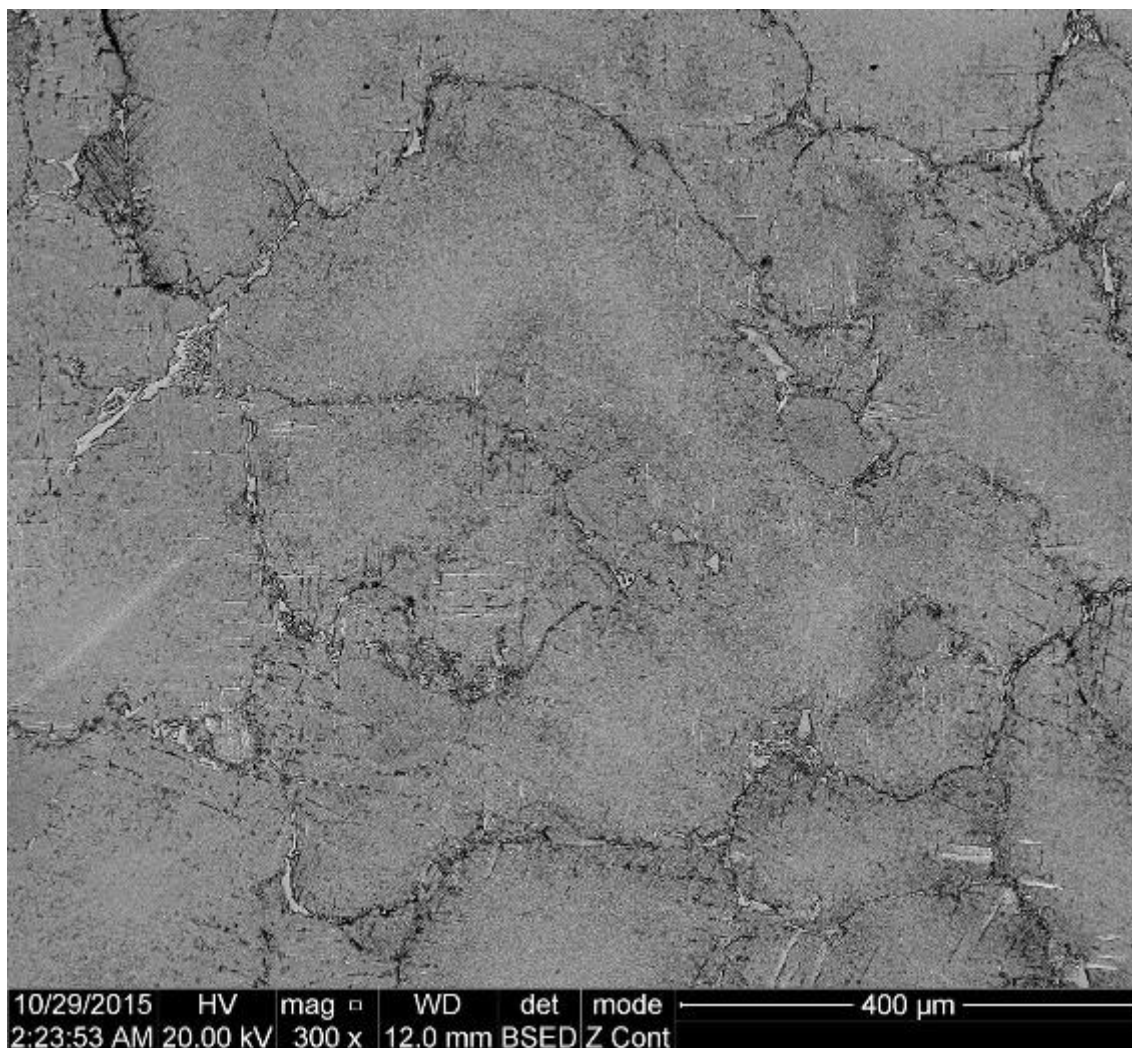


Figure 16. Microstructure of the ChS88 alloy after high-temperature ageing at 900°C, 1 500 hours.

The creep-rupture tests of the advanced alloy were carried out at 900°C and 1000°C. The test results of the new alloy and comparison of its heat resistance with the industrial domestic and foreign counterparts and with the aviation application alloys have been summarized using the Larson-Miller parameter (LMP); they are shown in Figure 18, which enables complex evaluation of long-run performance of the material taking into account the temperature and the test time, based on which the alloy heat-resistant level was determined at temperatures of 900, 950 and 1000°C for time durations of 100, 500 and 1000 hours. The LMP analysis results have been summarized and are given in Table 2.

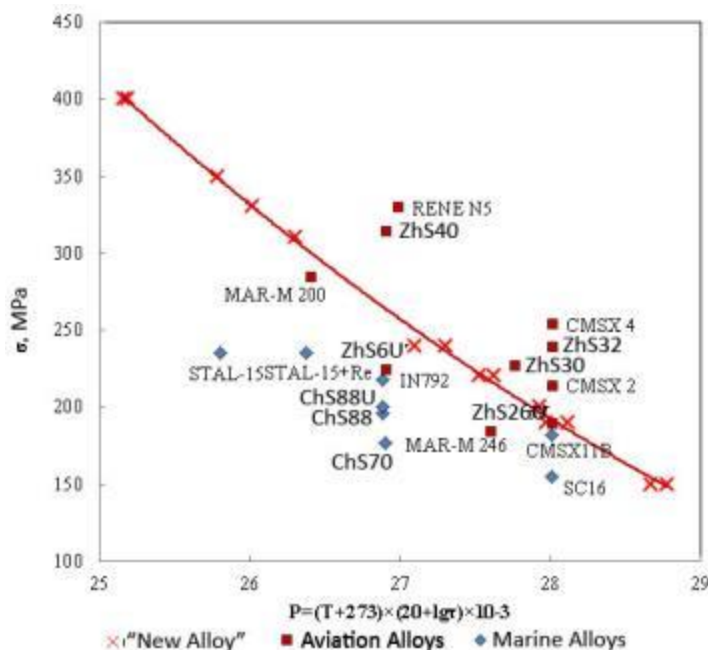


Figure 17. Comparison of the heat-resistant level between the advanced alloy and the counterparts

Figure 17 shows a clear boundary dividing the aviation and corrosion-resistant alloys designated for GTUs of marine and ground application; besides, the advanced heat-resistant alloy has benefits over the existing corrosion-resistant alloys and is positioned within the rupture strength values zone of aviation alloys.

Table 2. Rupture strength values σ^T_τ (MPa) of the advanced heat-resistant alloy for the standard test temperatures.

T, °C	τ , hour		
	100	500	1000
900	348	285	259
950	265	208	188
1000	195	144	122
1050	138	90	N/A

In addition to the creep rupture strength tests, the short-time strength evaluation tests have been performed. The comparative test results are given in Table 3.

It has been established that the advanced heat-resistant alloy possesses a significantly high ductility, which is characteristic of a high operational reliability index of the developed alloy.

Table 3. Comparison of the short-time strength properties between the advanced heat-resistant alloy and the counterparts at 20°C

Alloy	E, GPa	σ_B , MPa	$\sigma_{0,2}$, MPa	Ψ , %	δ , %
ZhSKS-2	13,7	1240	1100	15	11
ChS-88U	21	940	784	4	3
ChS-70	21	880	735	5	3
IN738	N/A	1100	955	5,5	5
IN792-A	N/A	1029	881	10,7	5,8
“New Alloy”	15,2	1245	1000	14,7	7,2

SUMMARY

The analysis of the alloying elements effect has been performed, based on which a complex approach to heat-resistant alloys design for operation under salt and sulfide corrosion is realized. The mathematical model, the algorithm is formulated and the calculation method of nickel-based heat-resistant alloys for marine GTUs chemical composition is proposed; the said method meets the requirements for strength, chemical, thermodynamical, structural and phase characteristics.

The chemical composition of as-cast single crystal heat-resistant nickel-based alloy is proposed. This alloy is characterized by increased resistance to high-temperature salt corrosion and possesses strength characteristics suited for aviation gas-turbine engines applications up to temperatures of 1050°C.

Based on the developed design method a new advanced heat-resistant alloy has been designed, which is superior to the existing domestic and foreign counterparts (including CMSX-11B: $\sigma_{100}^{1000} = 184 \text{ MPa}$) as regards corrosion resistance and heat-resistance, and as of today it has the best characteristics of the 'heat-resistance – high-temperature salt corrosion' complex, which confirms adequacy of the proposed method.

The alloy possesses heat-resistant properties consistent with the aviation materials, which enables using it in design of not only gas-turbine units for ground and marine application, but also in gas-turbine engines for amphibious aircraft and airfoil boats.

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