

Structural characteristics of nickel super alloy INCONEL 713LC after heat treatment

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Materials

ABSTRACT

Purpose: Aircraft industry often uses nickel super-alloys for blades of jet engine turbines. This paper summarises analysis of influence of heat treatment on structural characteristics of nickel super-alloy INCONEL 713LC.

Design/methodology/approach: Experimental investigation castings made of commercially produced nickel super-alloy INCONEL 713LC. Material was after heat treatment subjected to detailed structural analysis with use of methods of light microscopy (LM), or scanning electron microscopy (SEM) on microscopes OLYMPUS IX71 and JEOL JSM 50A.

Findings: Mode of optimum heat treatment was proposed. On the basis of obtained results it is possible to recommend a heat treatment, which would consist of heating and dwell at the temperature exceeding 1240°C (min. 1260°C), so that precipitates at the grain boundaries dissolve completely, with subsequent slow cooling down to the temperature of approx. 940-950°C, so that there occurs intensive intra-granular precipitation of inter-metallic phase γ' .

Research limitations/implications: The experiment was limited by occurrence a void in cast alloys.

Practical implications: Aircraft industry often uses nickel super-alloys for blades of jet engine turbines.

Originality/value: On the basis of obtained results it is possible to recommend the most suitable heat treatment, which produce intensive intra-granular precipitation of inter-metallic phase γ' . It was received a new know-how in this field.

Keywords: Metallic alloys; Nickel super-alloys; Heat treatment; Structural phase analysis

1. Introduction

Aircraft industry often uses nickel super-alloys for blades of jet engine turbines. It is so because this material must satisfy numerous extreme requirements, such as e. g. heat resistance at high temperatures, resistance to fatigue damage, resistance to aggressive effect of combustion products, etc. [1 - 4]. Long-term service life and reliability of material is directly related to its microstructure, i. e. to its stability at long-term exploitation. Used materials are usually alloyed in a complex manner and their structure is very complicated [5 - 10]. One of materials used for these applications is cast, complex alloyed nickel super-alloy INCONEL 713LC on the basis of Cr-Al-Mo-Ti-Nb-Zr. Several strengthening mechanisms take effect in this alloy, the main mechanism is precipitation strengthening by coherent precipitates of intermetallic phase Ni_3Al ,

or $Ni_3(Al, Ti)$ [11 - 14]. This paper is focused on investigation of structure of castings made of nickel super-alloy NICKEL 713LC after various variants of heat treatment.

2. Used material and experimental technique

We have used for experimental investigation castings made of commercially produced nickel super-alloy INCONEL 713LC. Chemical composition of the used melt is given in the Table 1.

Material was investigated in initial as cast state, and also after three modes of laboratory heat treatment (HT):

- 1240°C / 2 hours / water,
- 1240°C / 2 hours / air,
- 1240°C / 2 hours / furnace to 940°C / air.

Table 1.

Chemical composition of nickel super-alloy INCONEL 713LC (mass %)

C	Cr	Al	Mo	Ti	Nb	Zr	Ni	rest
0.05	9.3	6.5	4.7	0.8	2.1	0.1		

Material was after heat treatment subjected to detailed structural analysis with use of methods of light microscopy (LM), or scanning electron microscopy (SEM) on microscopes OLYMPUS IX71 and JEOL JSM 50A. Microstructure of material was developed by chemical etching in solution of cupric chloride, hydrochloric acid and distilled water. Local analysis of chemical composition was made by electron micro-analyser JCXA 733 with use of energy dispersive analyser EDAX.

3. Results of tests and discussion

3.1. Initial as cast state

Material's structure in initial state was somewhat heterogenous and it is documented in fig. 1a, b. Dendritic structure of the alloy is visible very well in fig. 1a. In the areas of dendritic segregations we observed three different types of particles, which were afterward identified by method of local analysis of chemical composition.

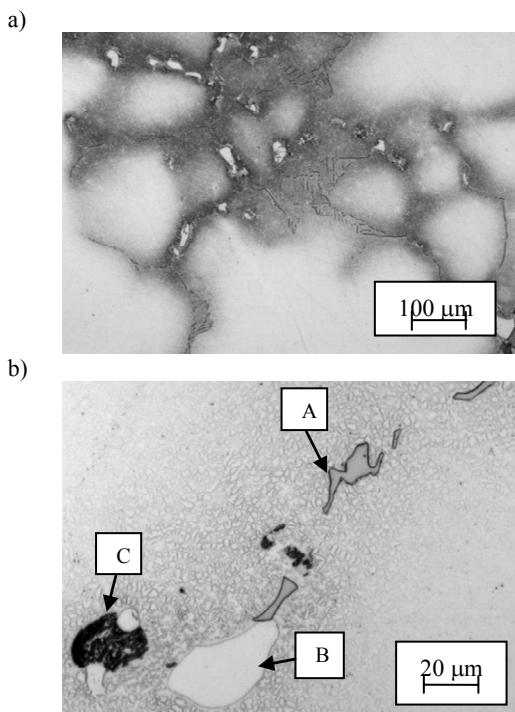


Fig. 1. Microstructure of the alloy in initial as cast state (LM)

Example of particles is shown also in fig. 2 in reflection of secondary electrons (SEM). Individual types of particles are marked A, B, C – identically in figures 1b and 2.

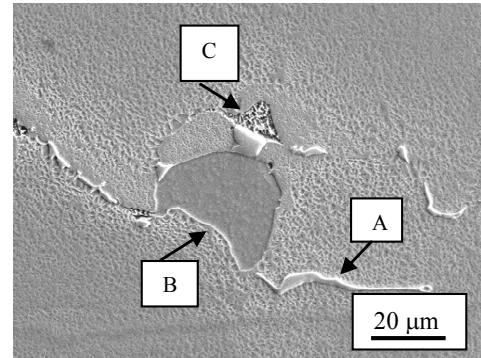


Fig. 2. Microstructure of the alloy in initial state with various types of particles (SEM)

Particles of the type A were identified by method of local micro-analysis as carbides, or niobium and titanium carbo-nitrides of the type (Nb, Ti) (C, N). Niobium prevailed in these particles, relation of niobium and titanium was approx. 8:1. These particles formed locally distinct eutetics, they were locally segregated in discrete rows in inter-dendritic spaces.

We have established in case of the particles marked B, that they are probably inter-metallic particles of the type Ni₃Al, which were formed already during solidification of the casting. Particles were present in samples only locally.

Representation of formations marked as C, with use of method LM as well as SEM, indicates, that this is not a single phase, but products of eutectic, or eutectoid reaction. Dimensions of individual particles in these formations, however, did not make it possible to make their reliable identification. Average chemical composition obtained at local micro-analysis of formations marked as C is shown in Table 2. We would like to draw attention to comparatively high contents of Zr among these values, due to the fact that its average content in investigated alloy is only 0.1 weight %. It can be deduced from the morphology of particles C that they appeared only after formation of the particles of the type A and B, i. e. niobium and titanium carbo-nitrides, or of greater particles of inter-metallic phase of the type Ni₃Al. Particles of the type C were also, similarly as particles of the type B, present only locally, i. e. that their distribution was uneven.

Table 2.
Local chemical composition of formations marked as C (mass %)

Cr	Al	Mo	Ti	Nb	Zr	Ni	rest
18.1	2.9	15.6	0.7	4.1	6.5		

Apart from coarser particles discussed above, which were present in the areas of dendritic segregations, we have observed in the matrix also numerous fine precipitates. In this case they were unequivocally strengthening precipitates of the phase γ'. We have not observed in initial state strings of fine particles of carbides on boundaries of casting grains. These precipitates are dangerous, since they can lead to premature development of defects and reduce significantly reliability and shorten thus

service life of turbine blades [11 - 14]. It is, however necessary to take into account that boundaries of casting grains coincided largely with areas of dendritic segregations and they therefore appeared as dark areas after etching. Identification of fine carbide precipitates would be rather problematic in these areas.

3.2. State after HT 1240°C / 2 hours / water

Microstructure of material after this HT is documented by LM method in fig. 3a, b.

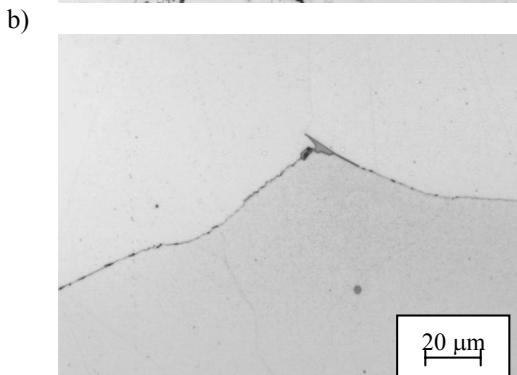
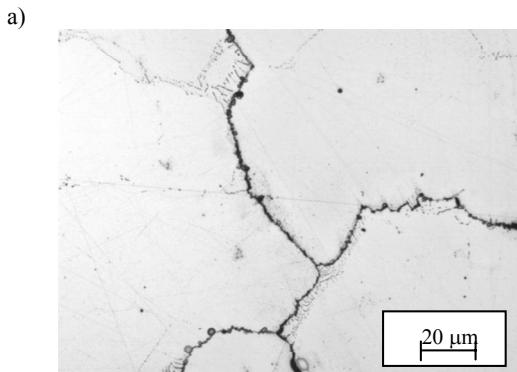


Fig. 3. Microstructures of the alloy after HT 1240°C / 2 hours / water (LM)

It is obvious from photos that after heat treatment dendritic structure of the alloy is almost not etched anymore (fig. 3a). It means therefore, that there occurred at least partial homogenisation of its chemical composition. We have, however, observed in material comparatively frequent inter-crystalline cracks (fig. 3a). We have found after detailed analysis that very fine precipitates covered at part of grain boundaries, at the boundary discernible by LM method (fig. 3b). These are probably very finely dispersive carbides, or niobium and titanium carbonitrides of the type (Nb, Ti) (C, N). Since it is highly improbable that these particles were formed during the heat treatment, we can assume that they were present in the alloy already in its initial state, but we were unable to identify them. The works [15, 16] state, that particles of this type dissolve at temperatures of 1200 – 1260°C. In the given case, nevertheless, dwell of 2 hours at the temperature of 1240°C was probably insufficient for full

dissolving of these precipitates. Weakening of grain boundaries by present particles then resulted in formation of cracks at rapid cooling (water quenching) due to high local stresses in material. We have not observed in matrix after this HT segregation of fine particles of inter-metallic phase γ' .

3.3. State after HT 1240°C / 2 hours / air

Microstructure of the alloy was after this HT identical with the previous state. The only difference was, that after cooling on air no cracks were observed on grain boundaries. This supports a hypothesis that cracks in the previous case originated due to stress at rapid cooling in water as a consequence of local weakening of grain boundaries by fine precipitates of the type (Nb, Ti) (C, N).

3.4. State after HT 1240°C / 2 hours / furnace to 940°C / air

In this case structure of the alloy was already different. There occurred a real etching of structure as a consequence of precipitation of the strengthening inter-metallic phase γ' . No fine precipitates were observed at the grain boundaries. Their presence cannot, however, be completely excluded. In this case as well it is possible that character of structure prevents their visibility. Microstructure of the alloy after HT is shown in fig. 4a, b.

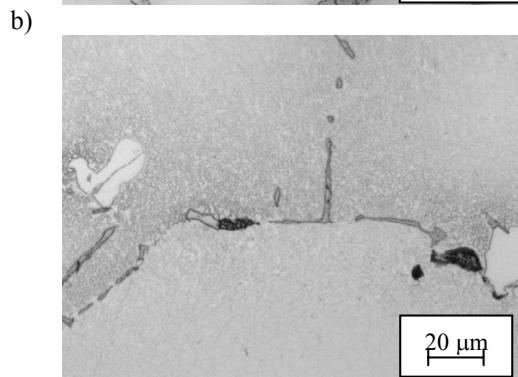
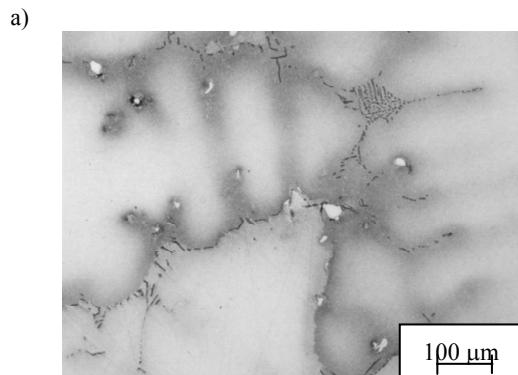


Fig. 4. Microstructure of the alloy after HT 1240°C / 2 hours / furnace to 940°C / air

4. Conclusions

Nickel super-alloy has in as cast state heterogenous structure with distinct dendritic segregations. In the zones of segregations there were found particles of carbides or niobium and titanium carbo-nitrides of the type (Nb, Ti) (C, N), moreover large particles of inter-metallic phase of the type Ni₃Al, and finally multiphase formations with higher contents of Mo, Nb, but particularly of Zr, the reason of which we were unable to identify precisely. Completed laboratory heat treatment changed distinctly material structure. Laboratory annealing at the temperature of 1240°C showed that fine strings of carbide particles were present in material at the grain boundaries, probably already in initial state. They afterwards caused during cooling by water development of inter-crystalline cracks. High-temperature annealing as such did not lead to precipitation of the strengthening inter-metallic phase γ'. Heat treatment 1240°C/2 hours/furnace to 940°C/air already triggered intensive precipitation of the phase γ'. We have not observed precipitates at the grain boundaries, it is, however, to exclude their presence. On the basis of obtained results it is possible to recommend a heat treatment, which would consist of heating and dwell at the temperature exceeding 1240°C (min. 1260°C), so that precipitates at the grain boundaries dissolve completely, with subsequent slow cooling down to the temperature of approx. 940-950°C, so that there occurs intensive intra-granular precipitation of inter-metallic phase γ'.

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