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Effects of {VIM+EBCHR} Refining for IN-738 Alloy

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Alloy IN-738 has gained acceptance as a durable turbine blade material for the gas turbine power industry. Typically, IN-738 material is a vacuum-cast, precipitation-hardened nickel-base alloy exhibiting excellent high-temperature creep-stress rupture strengths for extended service in a harsh thermal environment. This alloy also demonstrates an outstanding hot-corrosion resistance superior to that of other high-strength superalloys containing lower chromium content. Normally, IN-738 is vacuum-melted and investment-cast where typical casting conditions are problematic. For example, melts are superheated above the lower-liquidus temperature to about 1454 °C (2650 °F) and casting molds are preheated to temperatures of 981 °C (1800 °F) or higher. For an alloy, which exhibits a melting range of (1231-1315) °C (2250-2400) °F, this abusive melt practice leads to an excessively large dendritic crystals and gross coring of the microstructure with associated defects and deleterious phases. The intent of this article is to show that modern-day refining by {VIM+EBCHR} methods can be used to create a better alloy 738 for turbine blade applications. Cross-comparisons of selected properties for both investment-cast and ingot-cast test samples are here given.

Keywords electron beam cold hearth refining, IN-738, ingot-cast, investment-cast, mechanical properties, optical and electron microstructures, vacuum induction melting

1. History

Clarence G. Bieber of the International Nickel Company (INCO) is given credit for the original research leading to what was initially called "INCONEL 738" alloy (Ref 1). During the 1960s, INCO was oriented more toward research and development work instead of long-term and detailed application studies. As a result, this alloy was removed from the "wrought" INCONEL list, and subsequently, reclassified as IN-738. According to one high-level executive at INCO, this action was justified because INCONEL 738 had too many processing "nuances" and INCO management felt that this alloy could be more suitably developed as a casting material by the Foundry Industry. Hence, the designation of IN-738 originated. For a number of years, this alloy was relatively unknown and sparingly used. But, that was prior to the time that a company called Extex became directly involved. In an Extex comparison study using different alloys on a Rolls Royce 250 turbine engine, IN-738 more than doubled the service life over that for the OEM material, a cobalt-containing alloy (Ref 2). Since then, IN-738 has become an intelligent choice for first stage turbine buckets at repair stations, which rebuild older GE Frame 6 and Frame 7 power units (Ref 3).

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2. Materials

Investment-cast IN-738 property data came from two (2) sources; retired turbine blade samples and the open literature; e.g., from the ALLOY DIGEST data sheet (Ref 4).

Ingot-cast IN-738 properties were determined by MTEC Laboratories in Houston, TX for HMEC. The materials sponsor was BTEC Turbines of Houston, TX and ingots were produced by TIMET of Vallejo, CA using the {VIM+EBCHR} melting process.

3. Methods

Sample preparation for both photon and electron metallography used an ABRAPOL-2 machine (Struers, Westlake, OH) where grind-polish surfaces were processed. Grinding was done using SiC papers, rough polishing used diamond laps and fine, smooth scratches were removed utilizing vibratory methods (SYNTRON, Model LP-01C). Electroetching was accomplished using Petzow's EM-4 etchant (Ref 5) that is diluted with 475-mL of OZARKA Spring Water.

Global (macroscopic) determinations of chemical composition were obtained by the methods of optical emission spectroscopy, wet chemistry and vacuum fusion. Local (microscopic) determinations of chemical composition were achieved using a scanning electron microscope (SEM) in this operating mode: SEM:EDAX:EDS.

Quantitative metallography for the amounts of gamma prime in an austenitic-gamma matrix were executed using an HMEC-proprietary point counting procedure, especially developed for the nickel-base superalloys.

All mechanical tests were done in full accordance with the associated ASTM Specification; e.g., ASTM A 370 SPEC among others. Tests done for this article included the following: hardness, tensile (ambient plus elevated), true stress-true strain, Charpy (unnotched) impact strength, stress rupture and microstructure, including both optical and electron procedures.

4. Discussion and Results

4.1 Composition

Table 1 lists the global composition for the ingot-cast material that was produced by this state-of-the-art melt practice, {VIM+EBCHR}. Vacuum induction melting is used for consolidation of select scrap and electron-beam cold-hearth refining provides a higher-purity level in an ingot-cast IN-738 product. See Fig. 1 for an illustration, which defines the EBCHR process. Table 1 also reveals the global composition for a typical investment-cast blade made from alloy IN-738.

Statistical differences were calculated for the two different types of IN-738 in Table 1, investment-cast and ingot-cast. Significant differences do here exist between these two (2) alloy types. Shaded areas in Table 1 reveal large percent differences for these alloy elements: Co, Fe, C, B, V, Si, Mn, Hf, Cu, O, N, and (O+N). In terms of statistical variations, ingot-cast is much less variant than the investment-cast product.

Table 2 is a statistical study of variation about the observed values expressed as severity (S) and relevance (R) where

$$S = (\text{Standard Deviation}) \div (\text{Arithmetic Mean})\ 100$$

and

$$R\,(\text{Significant}) = S > 5.0$$

or

$$R\,(\text{Insignificant}) = S < 5.0$$

Table 1 Average composition of IN-738 alloy (wt.%)

Investment-cast	Element	Ingot-cast
Balance	Nickel, Ni	Balance
9.14	Cobalt, Cu	8.169
0.06	Iron, Fe	0.170
0.11	Carbon, C	0.122
0.01	Boron, B	0.009
16.27	Chromium, Cr	17.22
0.01	Vanadium, V	0.005
0.04	Zirconium, Zr	0.042
0.09	Silicon, Si	0.078
0.01	Manganese, Mn	0.006
0.02	Hafnium, Hf	0.023
2.09	Molybdenum, Mo	1.967
3.70	Aluminium, Al	3.505
3.59	Titanium, T1	3.427
7.29	(Al+Ti)	6.932
1.20	Columbium, Cb	1.255
1.71	Tantalum, Ta	1.675
2.91	(Cb+Ta)	2.930
3.04	Tungsten, W	3.022
4.75	(Ta+W)	4.697
0.002	Sulfur, S	0.002
0.01	Copper, Cu	0.005
5.0 ppm	Hydrogen, H	5.0 ppm
30.5 ppm	Oxygen, O	4.5 ppm
52.0 ppm	Nitrogen, N	22.0 ppm
82.5 ppm	(O+N)	26.5 ppm

Problematic results were observed only in the investment-cast alloy type for the elements Co and (O+N). It is apparent that excessive variations in the investment-cast composition can be insensitive to the needs and priorities of a dedicated metallurgical designer who demands a better turbine blade for a longer service life.

Arguably, this specification for investment-cast has not changed in more than 30 years. However, the ingot-cast specification was modeled after eight (8) different heats of IN-738 alloy during late 2004 and later revised in 2006 with tighter restrictions on the elements that contribute to the problem of severe coring, (Ta+W).

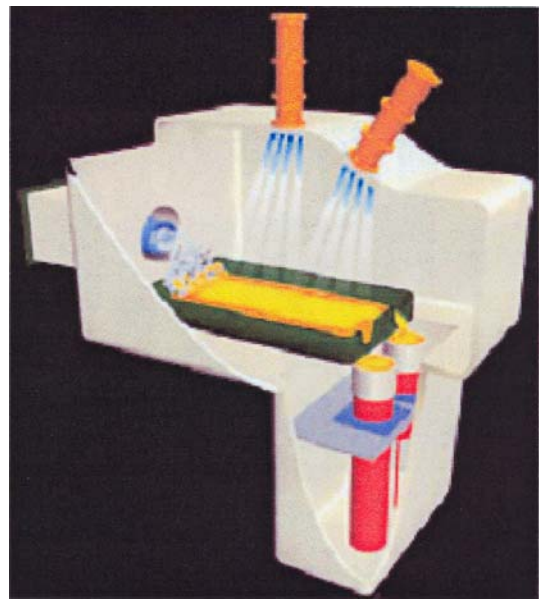

Fig. 1 Artist's drawing of the EBCHR chamber courtesy of the TIMET CORPORATION, Villejo, CA. *Note:* For this study, 203-mm (8-inch) diameter molds were used

Table 2 Statistical variations for both alloy types

Constituent (effect)	Percent severity	Relevance based on boundary limit of 5%
Nickel (GF)	0.34	Insignificant
Chromium (GF)	2.80	Insignificant
Cobalt (GF)	5.60	Slightly significant (*)
Molybdenum (GF)	3.10	Insignificant
Aluminium + Titanium (GPF)	2.50	Insignificant
Columbium + Tantalum (CF)	0.34	Insignificant
Tantalum + Tungsten (CF)	0.55	Insignificant
Oxygen + Nitrogen (IF)	51.4	Very Significant (+)

Notes: (*) Caused by a slightly high cobalt content in the investment-cast samples. (+) Caused by the extremely high interstitial content in the investment-cast samples

Definitions: (GF) = Gamma Formers, (GPF) = Gamma Prime Formers, (CF) = Carbide Formers and (IF) = Interstitial Formers

4.2 Structure

Some technologists might think that a casting is a casting, irregardless of the casting type [ingot vs. investment]. However, this kind of logic would suggest that these two microstructures are similar and, they are not. Figure 2 is a macrograph of investment-cast IN-738 as taken from a failed turbine blade at an original magnification (M_o) of about 6×. Large dendritic crystals exist, some of which are as large as the thickness of the turbine blade. This violates the design requirements of MIL-STD-1870 (Ref 6).

Figure 3 is an optical micrograph at (M_o) = 800× demonstrating a large variation in dendritic crystal sizes, shapes and orientations. Here, the primary problems created by this investment-cast product are extreme coring, degenerative carbides ($M_{23}C_6$) and a deleterious Mu phase ($M_{12}C$) which is known to drastically lower ductility. This embrittlement phase (Mu) is caused by slow-cooling gradients during solidification and subsequent heat treatment. In certain instances, these samples also confirmed the existence of microcracking.

Figure 4 is an electron micrograph at 2000× to represent precisely how the gamma prime precipitates appear in this alloy IN-738 type. These evidentiary observations are germane to these issues: (1) An abnormal distribution of coherent and incoherent gamma prime types, (2) Large differences attributed to atypical size/shape attributes and (3) Degenerative carbide particulates as a result of excessive superheating and extensive soaking times. Ingot-cast IN-738 alloy type also has its own structural problems, but these are not as formidable as those for the investment-cast category, which were just discussed. Generally, ingot-cast microstructures are characterized by three (3) distinct structures along the ingot cross section following all first principles of solidification. At the outermost region, a thin chilled territory exists as a result of the reactions by liquid metal in contact with a much colder mold. In the {VIM+EBCHR} process, this would be the water-cooled copper mold that receives the liquid metal from the cold hearth. Further, and toward the center, is a columnar area caused by directional solidification of the melt. Last, and at the centermost district, there is an equi-axed zone where the remaining liquid is undercooled. See Fig. 5 for a schematic of this freezing process (Ref 7).

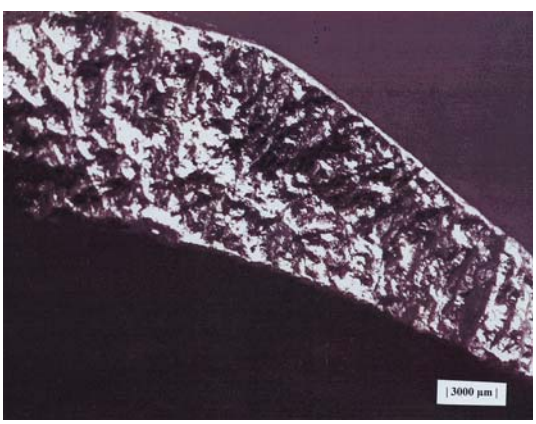

Fig. 2 Fractograph of a failed turbine blade at (M_o) = 800×

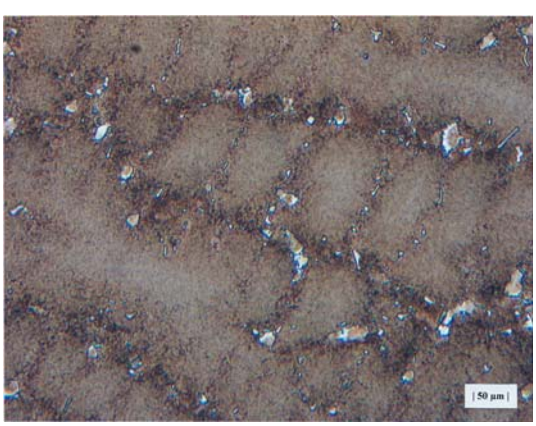

Fig. 3 Optical micrograph of investment-cast IN-738 at (M_o) = 800×

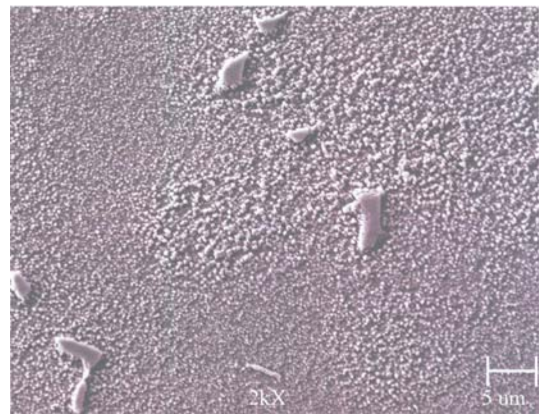

Fig. 4 Electron micrograph of investment-cast IN-738 at (M_o) = 2000×

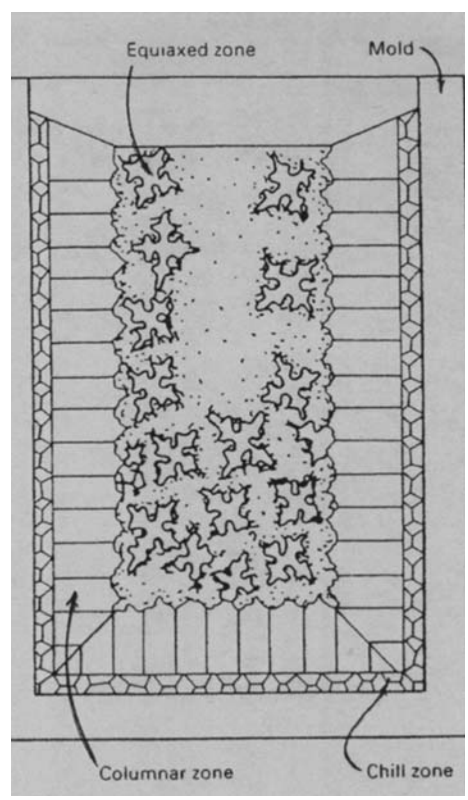

Fig. 5 Engineering sketch of the solidification process for ingot-cast IN-738 after *ASM* (Ref 7)

Figure 6 is an optical photomicrograph at (M_o) = 800×, which describes some of the prime differences between the two (2) alloy types, investment-cast vs. ingot-cast. Here, dendritic crystals do exist but they are smaller and more numerous. Degenerative carbides ($M_{23}C_6$) also exist and, these are about the same size, shape, and distribution as that was exhibited by Fig. 4. This be the case, because of the similar global compositions for carbon, 0.110 vs. 0.122—See Table 1. Admittedly, ingot-cast has more of the (M_6C) Mu phase present but, that is because of this very slow refining process, where impurities float to the top of the hearth. And, this point is technically unimportant, because the Mu phase in ingot-cast can be partially transformed by thermal methods or completely removed by hot deformation. Complete transformation of an investment-cast blade is an impossible task, because one cannot apply forging or extrusion stresses to the net-shape dimensions of a near-cast-to-size turbine bucket.

Figure 7 is an electron micrograph at [M_o] = 2000x. Here, the gamma prime precipitate is of a uniform size and shape. This is an important point because annealing, aging or rejuvenation heat treatments are controlled by the starting conditions. These summary observations are important: (1) A uniform distribution of gamma prime exists, (2) Small differences are attributed to both size/shape issues, and (3) This gamma prime is coherent with the gamma matrix and degenerative carbides are widely spaced.... Even though the lenticular Mu phase ($M_{12}C$) is apparently abundant. On the basis of structure and composition, ingot-cast IN-738 is preferred over investment-cast IN-738.

4.3 Properties

Hardness values for these two alloy types were unremarkable for this experimental study; for example, investment-cast IN-738 = HRC 36.8 ± 0.75 and ingot-cast IN-738 = HRC 37.7 ± 1.0. This same trend was also apparent for calculated values dealing with these elastic constants; Young's modulus (E), shear modulus (G) and Poisson's ratio (v) where

$$E = 224.7\,\text{GPa}\ (32.6 \times 10^6\,\text{Psi})\ [\text{investment} - \text{cast}],$$

$$E = 226.1\,\text{GPa}\ (32.8 \times 10^6\,\text{Psi})\ [\text{ingot} - \text{cast}],$$

$$(G) = 86.8\,\text{GPa}\ (12.6 \times 10^6\,\text{Psi})\ [\text{investment} - \text{cast}],$$

$$(G) = 87.5\,\text{GPa}\ (12.7 \times 10^6\,\text{Psi})\ (\text{ingot} - \text{cast}),$$

$$(v) = 0.2936\ (\text{investment} - \text{cast})\ \text{and}$$

$$(v) = 0.2944\ (\text{ingot} - \text{cast}).$$

Tables 3-5 characterize the routine mechanical properties that was determined for this research activity. The short-term tensile properties of Table 3 indicate that investment-cast is stronger but less ductile than the ingot-cast product. This tensile work is further justified by the principle that, for most alloys, yield strength varies inversely with ductility.

Table 4 confirms this same trend with respect to 100-h stress-rupture test results; viz., investment-cast is stronger than ingot-cast at either of the two-test temperatures that are cited.

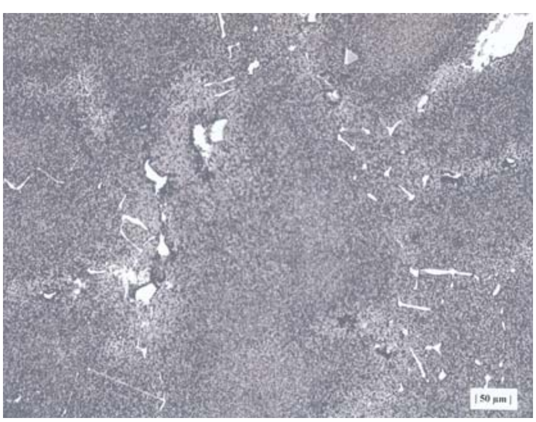

Fig. 6 Optical micrograph of ingot-cast IN-738 at (M_o) = 800×

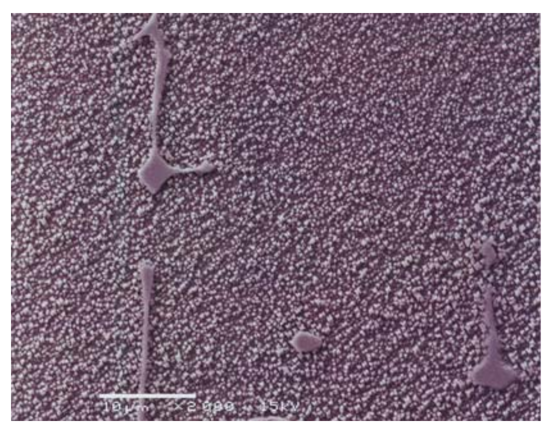

Fig. 7 Electron micrograph of ingot-cast IN-738 at (M_o) = 2000×

Table 3 Short-term tensile properties

Test temperature, °C (°F)	Tensile strength, MPa (Ksi)	0.2% Yield strength, MPa (Ksi)	Percent elongation	Percent reduction of area
For investment-cast IN-738				
Room	1096 (159)	952 (138)	5.5	5.0
538 (1000)	1048 (152)	896 (130)	4.3	4.5
734 (1350)	1000 (145)	841 (122)	3.0	4.0
For ingot-cast IN-738				
Room	945 (137)	765 (111)	12.5	17.8
538 (1000)	889 (129)	696 (101)	11.8	15.9
734 (1350)	883 (128)	662 (96)	16.2	9.8

The most positive effect on IN-738 alloy for the {VIM+EBCHR} refining concerns the parameter of impact toughness—see Table 5. For unnotched CVN comparisons, investment-cast offers 64 J (47 Ft-lbs) while ingot-cast yields a much higher value of 145 J (107 Ft-Lbs). And, this is also attributed to the higher ductility of a more purified product.

Table 5 also deals with the important parameter of flow stress which is more critical than ductility if forging or extrusion is planned. Ingot-cast has a much higher-strength constant (K) than investment-cast. Both alloy types, investment-cast or ingot-cast, have about the same strain hardening coefficient (n) of 0.092 vs. 0.105 and these results compare very favorably with those for quenched and tempered steels. This suggests that ingot-cast IN-738 can be converted to a wrought condition.... See Fig. 8 for proof that an ingot-cast IN-738 alloy type is fully transformed to the wrought condition. Here, M_o = 2000×.

In terms of composition, structure, and properties, ingot-cast IN-738 alloy type is preferred over that for investment-cast IN-738 alloy type.

Fig. 8 Electron micrograph of the wrought condition for ingot-cast IN-738 after extrusion at (M_o) = 2000×

Table 4 100-h stress-rupture properties

Test temperature, °C (°F)	Stress rupture strength, MPa (Ksi)
For investment-cast IN-738	
734 (1350)	662 (96)
815 (1500)	421 (61)
For ingot-cast IN-738	
734 (1350)	579 (84)
815 (1500)	386 (56)

5. Conclusions

Most of the important conclusions have already been discussed in detail but, some additional points do require an appropriate treatment at this time.

(1) The eta (η) phase does not exist in the ingot-cast variety of Alloy 738, because there is no ceramic mold or dye penetrant reactions for the metallurgical designer to be concerned about (Ref 8).

(2) The sigma (σ) phase can exist in both alloy types if the (W+Mo) content is below 3.8 atomic percent, according to Sims (Ref 9). All {VIM+EBCHR} melts of the future will be properly adjusted to avoid this potential problem involving (σ).

(3) The Mu (μ) phase exists in both alloy types but mechanical work followed by a properly tailored heat treatment cycle can annihilate the presence of (μ). This is not a viable solution for investment-cast material.

(4) Originally, cast turbine blades were an economic choice but, in recent years, that situation does not still apply. Forged or extruded products can be cost-effective and the price margin for cast vs. wrought is ever decreasing. For example, in 2004, an OEM Pratt & Whitney turbine blade for the Rolls Royce 250 engine cost $1,100.00 while a replacement blade of the wrought type (forged) made by Extex cost only $400.00.... Thanks to CNC machining (Ref 10).

(5) Extex was able to also show that a wrought blade (forged) can produce a 4% increase in power output with a 2% reduction in specific fuel consumption. In the gas turbine business, power and fuel are the designer's prime concerns.

Finally, it is concluded that the recently deceased Clarence G. Bieber, would certainly enjoy knowing that his beloved INCONEL 738 still survives and that most annoying "nuances" have been dutifully addressed. This article is dedicated to the memory of that legendary metallurgist.

Table 5 Miscellaneous mechanical properties

Room temperature impact properties (unnotched Charpy)		
Test temperature, °F	Impact strength, Joules (Ft-Lbs)	
Room	64 (47) [Investment-cast IN-738]	
Room	145 (107) [Ingot-cast IN-738]	
True stress-true strain results		
Test temperature, °F	Strength constant, (K)	Strain hardening coefficient, (n)
Room	1020 MPa (148-Ksi) [Investment-cast]	0.092 [Investment-cast]
Room	1310 Mpa (190-Ksi) [Ingot-cast]	0.105 [Ingot-cast]

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