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## (54) A method of forging a nickel base superalloy

(57) A method of forging a nickel base superalloy comprising providing a nickel base superalloy preform (40) with a first predetermined shape, the nickel base superalloy preform having been produced by powder metallurgy. The nickel base superalloy preform (40) is forged to produce a nickel base superalloy forged component (50) with a second predetermined shape. The first predetermined shape and the second predetermined shape are arranged such that the effective strain at the end of the forging is less than 1. The nickel base superalloy forged component (50) is given a supersolvus heat treatment to produce a large grain size in the nickel base superalloy forged component (50).

ponent (50) with a second predetermined shape. The first predetermined shape and the second predetermined shape are arranged such that the effective strain at the end of the forging is less than 1. The nickel base superalloy forged component (50) is given a supersolvus heat treatment to produce a large grain size in the nickel base superalloy forged component (50).

Fig.7

## Pre-Forge

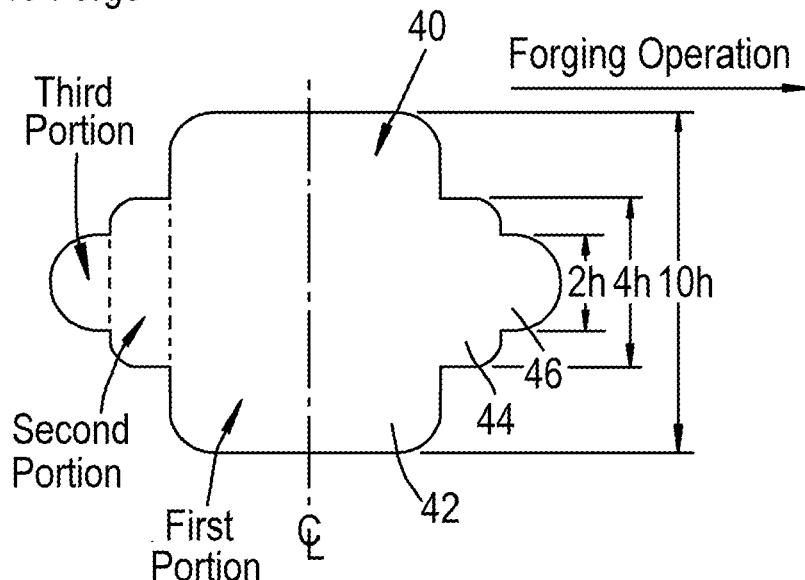
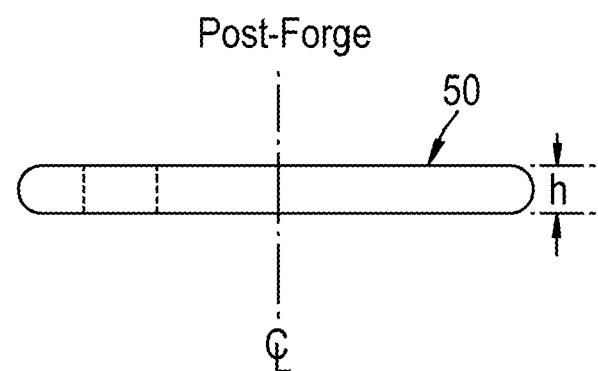


Fig.8



**Description**

**[0001]** The present invention relates to a method of forging a nickel base superalloy and in particular to a method of forging a nickel base superalloy turbine disc or a nickel base superalloy compressor disc.

**[0002]** Currently high strength, powder processed nickel base superalloys such as RR1000, Rene 104, Alloy 10, LSHR, IN100, Rene 88DT and Rene 95 etc may be processed to produce one of three different microstructures or combinations thereof. These different microstructures are a) fine grain size microstructure ASTM 10-12 ALA 7, b) midi gain size microstructure ASTM 7.5-9 ALA 4 and c) coarse grain size microstructure ASTM 5-7 ALA 2.

**[0003]** In addition a single component may be processed to produce a range of grain sizes, with different grain sizes at different locations, e.g. a dual microstructure turbine disc.

**[0004]** The grain size achieved in these nickel base superalloys is influenced by three factors. The first factor is alloy chemistry, in particular the levels of boron and carbon. The levels of boron and carbon dictate the levels of borides or carbides present in the nickel base superalloy and these borides or carbides pin grain boundaries and restrict grain growth. The second factor is processing parameters, principally forging temperature, strain rate and final strain. The third factor is heat treatment. Fine grain nickel base superalloys are heat treated below the gamma-prime solvus temperature, a subsolvus heat treatment, the gamma-prime phase acts to pin the grain boundaries. Coarse grain nickel superalloys are heat treated above the gamma-prime solvus temperature, a supersolvus heat treatment. Midi grain nickel base superalloys are forged at a lower than usual temperature and then heat treated above the gamma-prime solvus temperature, supersolvus heat treatment. In any heat treatment that involves going above the gamma-prime solvus temperature it is necessary to have an understanding of both the chemistry of the nickel base superalloy and prior mechanical working history to avoid critical grain growth.

**[0005]** There is a need within the industry for a nickel base superalloy which has a large grain size in excess of coarse grain without changing the chemistry of the nickel base superalloy.

**[0006]** Accordingly the present invention seeks to provide a novel method of forging a nickel base superalloy which has been produced by powder metallurgy, which reduces, preferably overcomes, the above mentioned problem.

**[0007]** Accordingly the present invention provides a method of forging a nickel base superalloy comprising the steps of:-

(a) providing a nickel base superalloy preform with a first predetermined shape, the nickel base superalloy preform having been produced by powder met-

allurgy,

(b) forging the nickel base superalloy preform to produce a nickel base superalloy forged component with a second predetermined shape, wherein the first predetermined shape and the second predetermined shape are arranged such that the effective strain at the end of the forging is less than 1,

(c) supersolvus heat treating at least a region of the nickel base superalloy forged component to produce a large grain size in the at least a region of the nickel base superalloy forged component, wherein the large grain size in the nickel base superalloy is 80 to 140 micrometers.

15 **[0008]** Preferably in step (b) the effective strain at the end of forging is less than or equal to 0.75.

**[0009]** More preferably in step (b) the effective strain at the end of forging is less than or equal to 0.5.

20 **[0010]** Preferably step (a) comprises providing a stepped cylindrical preform, the cylindrical preform having a first substantially cylindrical portion and a second substantially annular portion arranged coaxially around the first portion, the first cylindrical portion having a first thickness, the second annular portion having a second thickness and the second thickness is less than the first thickness and step (b) comprises forging the stepped cylindrical preform to a substantially disc shaped forged component and after step (c) the second portion having coarser grains than the first portion.

25 **[0011]** Preferably step (a) comprises providing a substantially cylindrical preform, the cylindrical preform having a third substantially annular portion arranged coaxially around the second annular portion, the third annular portion having a third thickness and the third thickness is less than the second thickness and after step (c) the third portion having coarser grains than the second portion.

30 **[0012]** Preferably in the first cylindrical portion the effective strain is about 0.9, in the second annular portion the effective strain is about 0.75 and in the third annular portion the effective strain is about 0.5.

35 **[0013]** Step (c) may comprise a subsolvus heat treatment in a first region of the forged component and a supersolvus heat treatment in a second region of the forged component. The first region may comprise the first portion and the second region may comprise the second portion or the second region comprises the second portion and the third portion.

40 **[0014]** Step (c) comprise supersolvus heat treating all of the nickel base superalloy forged component to produce a large grain size in all of the nickel base superalloy forged component.

**[0015]** The forged component may comprise a forged component for a turbine disc or a compressor disc.

45 **[0016]** The present invention will be more fully described by way of example with reference to the accompanying drawings in which:-

Figure 1 shows a gas turbine engine having a nickel base superalloy turbine disc manufactured according to the present invention.

Figure 2 shows a nickel base superalloy turbine disc manufactured according to the present invention.

Figure 3 shows a micrograph of a nickel base superalloy forged according to the prior art.

Figure 4 shows a micrograph of a nickel base superalloy forged according to the present invention.

Figure 5 is a graph of grain size after heat treatment against effective strain.

Figure 6 is a graph of grain size after heat treatment against time above gamma prime solvus

Figure 7 is a schematic of a nickel base superalloy preform for forging.

Figure 8 is a schematic of a forged component produced from the nickel base superalloy preform.

Figure 9 is a graph showing the grain size against position in the forged component of figure 8 following a supersolvus heat treatment.

Figure 10 is a graph showing the grain size against position in the forged component of figure 8 following a dual microstructure heat treatment.

**[0017]** A turbofan gas turbine engine 10, as shown in figure 1, comprises in axial flow series an intake 12, a fan section 14, a compressor section 16, a combustion section 18, a turbine section 20 and an exhaust 22. The turbine section 20 comprises a turbine disc 24, which carries a plurality of circumferentially spaced turbine blades 26. The gas turbine engine 10 is quite conventional and its construction and operation will not be described further.

**[0018]** The gas turbine engine turbine disc 24, as shown more clearly in figure 2, comprises a hub, or cob, 28, a web 30 and a rim 32. The hub 28 is at the radially inner end of the turbine disc 24, the rim 32 is at the radially outer end of the turbine disc 24 and the web 30 extends radially between and interconnects the hub 28 and the rim 32. The rim 32, in this example, has a plurality of circumferentially spaced slots 34 to receive the roots of turbine blades 26, shown in figure 1, and circumferentially spaced posts 36 are provided on the rim 32 of the turbine disc 24 to define the sides of the slots 34. The slots 34 may be firtree shape, or dovetail shape. The turbine disc 24 comprises a high strength nickel base superalloy, for example RR1000.

**[0019]** There is a need within the industry for a nickel base superalloy which has a large grain size in excess of a coarse grain size (ASTM 5-7). The range of grain sizes for these large grain size microstructures in nickel base superalloy is ASTM 5-2 ALA 00. A large uniform grain size such as this is desirable because increasing the grain size increases the resistance to both creep deformation and fatigue crack growth. However, strength and fatigue properties are reduced.

**[0020]** Figures 3 and 4 show micrographs of high strength nickel base superalloy RR1000 processed by a

powder metallurgy route. RR1000 consists of 18.5wt% cobalt, 15wt% chromium, 5wt% molybdenum, 2wt% tantalum, 3.6wt% titanium, 3wt% aluminium, 0.5wt% hafnium, 0.015wt% boron, 0.06wt% zirconium, 0.027wt% carbon and the balance nickel plus incidental impurities. It is to be noted that RR1000 consists of less than 0.03wt% carbon and it is preferred that the nickel base superalloy consists of less than 0.03wt% carbon.

**[0021]** Figure 4 shows RR1000 with a large grain size of ASTM 5-2 ALA 00 produced according to the present invention whereas Figure 3 shows RR1000 with a coarse grain size of ASTM 5-7 ALA 2 produced according to the prior art.

**[0022]** In the prior art a nickel base superalloy turbine disc is produced by isothermal forging a substantially cylindrical preform to a substantially disc shaped forged component using a large amount of plastic deformation. The effective strain at the end of the forging process is greater than 1. Subsequently, the nickel base superalloy forged component is given a supersolvus heat treatment to produce a coarse grain size in the nickel base superalloy forged component limited to the ASTM 5-7 range of conventional coarse grain nickel base superalloys.

**[0023]** In the present invention the nickel base superalloy preform is provided with a first predetermined shape, the nickel base superalloy preform having been produced by powder metallurgy, the nickel base superalloy preform is forged to produce a nickel base superalloy forged component with a second predetermined shape, wherein the first predetermined shape and the second predetermined shape are arranged such that the effective strain at the end of the forging process is less than 1. Subsequently, the nickel base superalloy forged component is given a supersolvus heat treatment to produce a large grain size in the nickel base superalloy forged component in the ASTM 5-2 ALA 00, 80 to 140 micrometers, which is greater in size than the conventional coarse grain in nickel base superalloys.

**[0024]** Figure 5 is a graph of grain size after heat treatment against effective strain at the end of the forging process, or forging step. It is seen in figure 5 that at effective strains above 1 the grain size remains substantially constant. At effective strains below 1 the grain size increases gradually from an effective strain of 1 down to an effective strain of 0.75. The grain size increases more rapidly from an effective strain of 0.75 to an effective strain of 0.5 and then increases to a maximum grain size below an effective strain of 0.5. We have found that for an effective strain less than 1 the grain size is greater than that for an effective strain greater than 1 for identical supersolvus heat treatments. We have found that the grain size is relatively insensitive to instantaneous strain rate in RR1000 compared to the prior art.

**[0025]** Figure 6 is a graph of grain size against time above the gamma prime solvus for three series of tests, M, P and D>1 where the effective strain was greater than 1 and a single series of tests D<1 where the effective strain was less than 1. This graph shows the effect of

time above the gamma prime solvs on the grain size. It is clear from the data from these tests that the grain size reaches a steady state after a short period of time. Irrespective of effective strain, grain growth does not occur immediately, but rather takes some time to occur. Grain growth occurs rapidly in the nickel base superalloy where effective strain is less than 1 before stabilising at the large grain size. Grain growth occurs more slowly in the nickel base superalloy where effective strain is greater than 1 before stabilising at the coarse grain size.

**[0026]** The definition of effective strain is given by the equation  $\dot{\epsilon} = \sqrt{2/3}[(\epsilon_1 - \epsilon_2)^2 + (\epsilon_2 - \epsilon_3)^2 + (\epsilon_3 - \epsilon_1)^2]^{1/2}$ , where  $\dot{\epsilon}$  is the effective strain at the end of forging,  $\epsilon_1$ ,  $\epsilon_2$  and  $\epsilon_3$  are the principal strains in each of the three principal directions. Thus for an isotropic material, knowledge of the uniaxial tensile test behaviour together with the yield function enables the stress-strain behaviour to be predicted for any stress system. The effective strain equals the square root of two divided by three multiplied by the square root of the squares of the difference in the principal strains.

**[0027]** Figure 7 shows an example of nickel base superalloy preform 40, pre-forge, which is used in a method according to the present invention. In particular the nickel base superalloy preform 40 is a stepped cylindrical preform and the stepped cylindrical preform 40 has a first substantially cylindrical portion 42, a second substantially annular portion arranged 44 coaxially around the first cylindrical portion 42 and a third substantially annular portion 46 arranged coaxially around the second annular portion 44. The first, second and third portions 42, 44 and 46 are integral. In this example the first cylindrical portion 42 has a first thickness 10h, the second annular portion 44 has a second thickness 4h and the third annular portion has a third thickness 2h. The second thickness 4h is less than the first thickness 10h and the third thickness 2h is less than the second thickness 4h. In this example the second annular portion 44 is twice as thick as the third annular portion 46 and the first cylindrical portion 42 is five times thicker than the third annular portion 46.

**[0028]** Figure 8 shows a nickel base superalloy disc shaped forged component 50 which has a thickness of h, which is produced by forging, isothermally forging, the nickel base superalloy preform 40. The nickel base superalloy disc shaped forged component 50 is then given a supersolvus heat treatment, such that the second annular portion 44 has coarser grains than the first portion 42 and the third portion 46 has coarser grains than the second portion 44. The forging temperature is in the range of 1050°C to 1150°C, preferably in the range of 1090°C to 1120°C, and the strain rate is in the range of 0.001 to 0.1 per second, preferably 0.1 per second. The disc shaped forging is produced in a close die shaped to form the nickel base superalloy preform 40 to the desired shape of the nickel base superalloy disc shaped forged component 50.

**[0029]** Considering the strain in the portions 42, 44 and

46 of the stepped cylindrical preform 40 in figure 7, in a simplistic sense by assuming there is no radial growth during forging. The simplistic effective strain in the first portion 42 is  $\epsilon_{eff} = \epsilon/l_0$ , where  $\epsilon_{eff}$  is the effective strain at the end of forging,  $\epsilon$  is the change in dimension and  $l_0$  is the original dimension  $9h/10h = 0.9$ . The effective strain in the second portion 44 is  $\epsilon_{eff} = \epsilon/l_0$ ,  $3h/4h = 0.75$ . The effective strain in the third portion 46 is  $\epsilon_{eff} = \epsilon/l_0$ ,  $h/2h = 0.5$ . Thus, in the first cylindrical portion 42 the effective strain is about 0.9, in the second annular portion 44 the effective strain is about 0.75 and in the third annular portion 46 the effective strain is about 0.5.

**[0030]** It is preferred to use an effective strain  $E_{eff}$  of less than or equal to 0.5 in order to produce a uniform large grain size in the nickel base superalloy. An effective  $\epsilon_{eff}$  of less than, or equal, to 0.75 but greater than 0.5 would produce a slightly finer grain size than the large grain size in the nickel base superalloy.

**[0031]** Figure 9 is a graph showing the grain size against position in the forged component of figure 8 following a conventional supersolvus heat treatment. It is seen that in the third annular portion 46 the grain size is larger than the second annular portion 44 and the grain size in the second annular portion 44 is larger than the grain size in the first cylindrical portion 42. Thus, the nickel base superalloy preform of figure 8 is forged to produce a nickel base superalloy disc shaped forged component 50 and the nickel base superalloy disc shaped forged component 50 is given a supersolvus heat treatment and ageing heat treatment and then machined to produce a nickel base superalloy turbine disc or compressor disc. Thus, the finished nickel base superalloy turbine disc will have an increasing grain size in the turbine disc with a coarse grain size in the rim 32, third annular portion 46, a finer grain size in the web 30, second annular portion 44, and a finest grain size in the hub 28, first cylindrical portion 42. In the case of RR1000 the nickel base superalloy preform is isothermally forged at a temperature of 1050°C to 1150°C, preferably in the range of 1090°C to 1120°C, at a strain rate in the range of 0.001 to 0.1 per second, preferably 0.1 per second. The RR1000 nickel base superalloy disc shaped forging is given a conventional subsolvus solution heat treatment at 20°C to 30°C below the gamma prime solvs temperature for 1 to 4 hours, then air cooled, followed by a supersolvus heat treatment at 20°C to 50°C above the gamma prime solvs temperature for up to 1 hour, then air cooled, followed by an ageing heat treatment at 760°C (1400°F) for 16 hours and air cooled.

**[0032]** Figure 10 is a graph showing the grain size against position in the forged component of figure 8 following a dual microstructure heat treatment. In a dual microstructure heat treatment one region is given a supersolvus heat treatment and one region is given a subsolvus heat treatment. It is seen that in the third annular portion 46 the grain size is larger than the second annular portion 44 and the grain size in the second annular portion 44 is larger than the grain size in the first cylindrical por-

tion 42. Thus, the nickel base superalloy preform of figure 8 is forged to produce a nickel base superalloy disc shaped forged component 50 and the nickel base superalloy disc shaped forged component 50 is given a supersolvus heat treatment and ageing heat treatment and then machined to produce a nickel base superalloy turbine disc or compressor disc. Thus, the finished nickel base superalloy turbine disc will have an increasing grain size in the turbine disc with a large grain size in the rim 32, a coarse grain size in the web 30 and a fine grain size in the hub 28. In this particular example the first cylindrical portion 42 is given a subsolvus heat treatment to retain fine grains in the hub 28, the second annular portion 44 and the third annular portion 46 are given a supersolvus heat treatment to produce coarse grains in the web 30 and large grains in the rim 32 respectively. In the case of RR1000 the first cylindrical portion is given a conventional subsolvus solution heat treatment at 20°C to 30°C below the gamma prime solvus temperature for 1 to 4 hours, then air cooled, followed by subsolvus solution heat treatment at 20°C to 30°C below the gamma prime solvus temperature for times up to 1 hour, then air cooled, followed by an ageing heat treatment at 760°C (1400°F) for 16 hours and then air cooled. The second and third annular portions were given a conventional subsolvus solution heat treatment at 20°C to 30°C below the gamma prime solvus temperature for 1 to 4 hours, then air cooled, followed by a supersolvus heat treatment at 20°C to 50°C above the gamma prime solvus temperature for times up to 1 hour, then air cooled, followed by an ageing heat treatment at 760°C (1400°F) for 16 hours and air cooled.

**[0033]** In an alternative example a stepped cylindrical preform may be provided, the cylindrical preform has a first substantially cylindrical portion and a second substantially annular portion arranged coaxially around the first portion, the first cylindrical portion has a first thickness, the second annular portion has a second thickness and the second thickness is less than the first thickness. The stepped cylindrical preform is forged to a generally disc shaped forged component and then the nickel base superalloy forged component is given a supersolvus heat treatment and ageing heat treatment and then machined to produce a nickel base superalloy turbine disc or compressor disc. The second portion has coarser grains than the first portion.

**[0034]** By controlling the shape of the nickel superalloy preform and the shape of the nickel alloy forged component the effective strain is kept to a level below 1, such that on subsequent supersolvus heat treatment a large grain size is produced in the nickel base superalloy. It is the final effective strain and not the maximum instantaneous strain rate of the forging process, or forging step, which dictates the grain size in the nickel base superalloy. The supersolvus solution heat treatment may be applied directly after the forging step or a subsolvus heat treatment may be applied after the forging step and a supersolvus heat treatment applied after the subsolvus heat

treatment. The forging step may comprise isothermal forging and isothermal application of strain during the isothermal forging step allows the imparted strain levels to be accurately controlled. The present invention may be applied to specific locations of a component. The present invention produces a very creep resistant, fatigue crack growth resistant nickel base superalloy with a high gamma prime volume fraction.

**[0035]** The advantage of the present invention is that it enables critical rotating components to be produced with enhanced high temperature properties, in particular creep resistance and fatigue crack growth resistance. This provides an increase in the operating life of the component or enables the component to operate at higher temperatures and may decrease the weight of the component.

**[0036]** The present invention is applicable to all high strength powder processed nickel base superalloys used for gas turbine engine turbine discs, compressor discs, high-pressure compressor cones and turbine cover plates.

**[0037]** Other suitable nickel base superalloys are Rene 95, Rene 88DT, Alloy 10, LSHR, Rene 104 and IN100. Rene 95 consists of 8.12wt% Co, 12.94wt% Cr, 3.45wt% Mo, 3.43wt% W, 3.42wt% Al, 2.44wt% Ti, 3.37wt% Nb, 0.05wt% Zr, 0.07wt% C, 0.012wt% B and the balance Ni and incidental impurities. Rene 88DT consists of 13.1wt% Co, 15.8wt% Cr, 4wt% Mo, 3.9wt% W, 2wt% Al, 3.7wt% Ti, 0.7wt% Nb, 0.045wt% Zr, 0.05wt% C, 0.016wt% B and the balance Ni and incidental impurities. Alloy 10 consists of 17.93wt% Co, 10.46wt% Cr, 2.52wt% Mo, 4.74wt% W, 3.53wt% Al, 3.79wt% Ti, 1.61wt% Ta, 0.97wt% Nb, 0.07wt% Zr, 0.027wt% C, 0.028wt% B and the balance Ni and incidental impurities. LSHR consists of 20.8wt% Co, 12.7wt% Cr, 2.74wt% Mo, 4.37wt% W, 3.48wt% Al, 3.47wt% Ti, 1.65wt% Ta, 0.049wt% Zr, 0.024wt% C, 0.028wt% B and the balance Ni and incidental impurities. Rene 104 consists of 20.6wt% Co, 13.0wt% Cr, 3.80wt% Mo, 2.1wt% W, 3.4wt% Al, 3.7wt% Ti, 2.4wt% Ta, 0.05wt% Zr, 0.04wt% C, 0.03wt% B and the balance Ni and incidental impurities. IN100 consists of 18.5wt% Co, 12.5wt% Cr, 3.2wt% Mo, 5.0wt% Al, 4.4wt% Ti, 0.06wt% Zr, 0.07wt% C, 0.02wt% B and the balance Ni and incidental impurities.

**[0038]** In general the subsolvus heat treatment for these nickel base superalloys is at a temperature of 20°C to 40°C below the gamma prime solvus temperature for times of 1 to 6 hours and the supersolvus heat treatment for these nickel base superalloys is at a temperature of 20°C to 50°C above the gamma prime solvus temperature for times of up to 4 hours. Alternatively the subsolvus heat treatment is 1110°C to 1150°C for times of 1 to 6 hours and the supersolvus heat treatment is 1160°C to 1210°C for times of up to 4 hours.

## Claims

1. A method of forging a nickel base superalloy comprising the steps of:
  - (a) providing a nickel base superalloy preform (40) with a first predetermined shape, the nickel base superalloy preform (40) having been produced by powder metallurgy,
  - (b) forging the nickel base superalloy preform (40) to produce a nickel base superalloy forged component (50) with a second predetermined shape,
  - (c) supersolvus heat treating at least a region of the nickel base superalloy forged component (50) to produce a large grain size in the at least a region of the nickel base superalloy forged component (50), **characterised in that** in step (b) the first predetermined shape and the second predetermined shape are arranged such that the effective strain at the end of forging is less than 1 and wherein step (c) the large grain size in the nickel base superalloy is 80 to 140 micrometers.
2. A method as claimed in claim 1 wherein in step (b) the effective strain at the end of forging is less than or equal to 0.75.
3. A method as claimed in claim 1 or claim 2 wherein in step (b) the effective strain at the end of forging is less than or equal to 0.5.
4. A method as claimed in any of claims 1 to 3 wherein step (a) comprises providing a stepped cylindrical preform (40), the cylindrical preform (40) having a first substantially cylindrical portion (42) and a second substantially annular portion (44) arranged coaxially around the first portion (42), the first cylindrical portion (42) having a first thickness, the second annular portion (44) having a second thickness and the second thickness is less than the first thickness and step (b) comprises forging the stepped cylindrical preform (40) to a substantially disc shaped forged component (50) and after step (c) the second portion having coarser grains than the first portion.
5. A method as claimed in claim 4 wherein step (a) comprises providing a substantially cylindrical preform (40), the cylindrical preform (40) having a third substantially annular portion (46) arranged coaxially around the second annular portion (44), the third annular portion (46) having a third thickness and the third thickness is less than the second thickness and after step (c) the third portion having coarser grains than the second portion.
6. A method as claimed in claim 5 wherein in the first

- 5 cylindrical portion (42) the effective strain is about 0.9, in the second annular portion (44) the effective strain is about 0.75 and in the third annular portion (46) the effective strain is about 0.5.
- 10 7. A method as claimed in claim 4, claim 5 or claim 6 wherein step (c) comprises a subsolvus heat treatment in a first region of the forged component and a supersolvus heat treatment in a second region of the forged component.
- 15 8. A method as claimed in claim 7 wherein the first region comprises the first portion and the second region comprises the second portion or the second region comprises the second portion and the third portion.
- 20 9. A method as claimed in any of claims 1 to 6 wherein step (c) comprises supersolvus heat treating all of the nickel base superalloy forged component (50) to produce a large grain size in all of the nickel base superalloy forged component (50).
- 25 10. A method as claimed in any of claims 1 to 9 wherein the forged component comprises a forged component for a turbine disc (24) or a compressor disc.
- 30 11. A method as claimed in any of claims 1 to 10 wherein the nickel base superalloy consists of 18.5wt% cobalt, 15wt% chromium, 5wt% molybdenum, 2wt% tantalum, 3.6wt% titanium, 3wt% aluminium, 0.5wt% hafnium, 0.015wt% boron, 0.06wt% zirconium, 0.027wt% carbon and the balance nickel plus incidental impurities.
- 35 12. A method as claimed in any of claims 1 to 11 wherein step (b) comprises forging at a temperature in the range of 1050°C to 1150°C and at a strain rate in the range of 0.001 to 0.1 per second.
- 40 13. A method as claimed in any of claims 1 to 12 wherein step (c) comprises supersolvus heat treating at 20°C to 50°C above the gamma prime solvus temperature, or a temperature of 1110°C to 1150°C, for up to 4 hours.
- 45
- 50
- 55

Fig.1

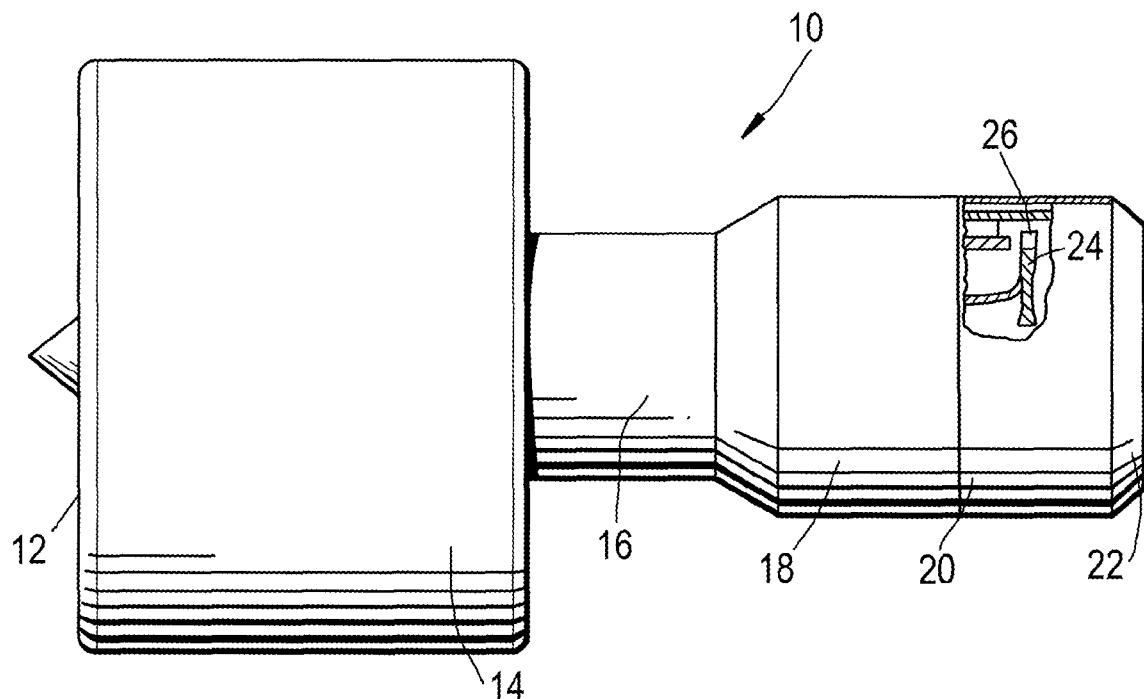


Fig.2

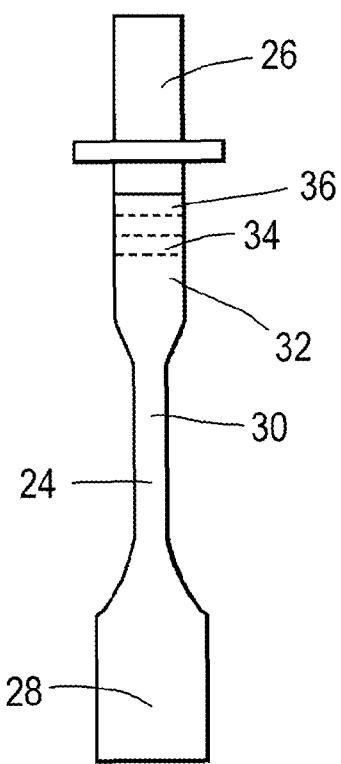


Fig.3

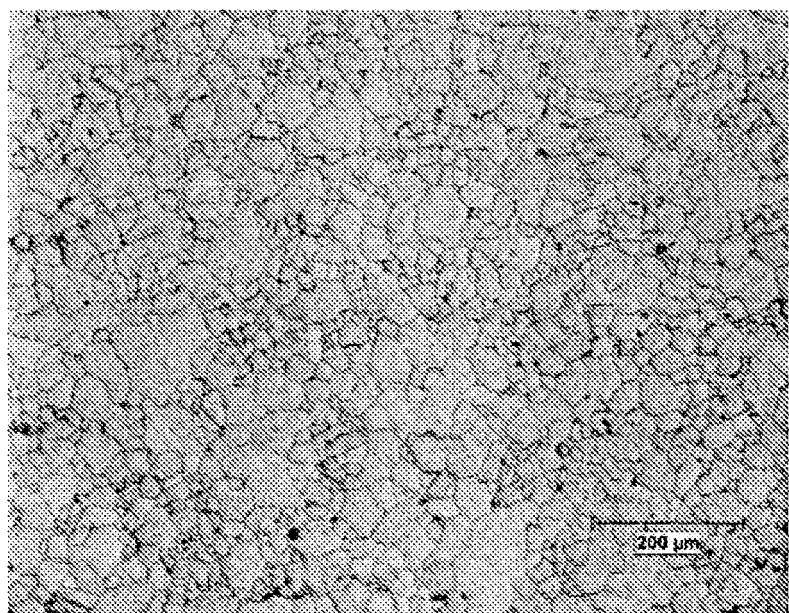


Fig.4

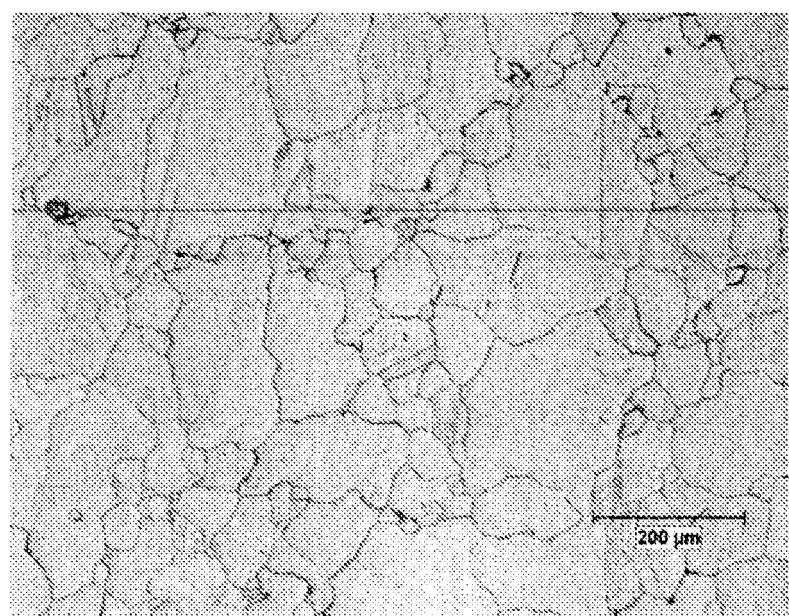


Fig.5

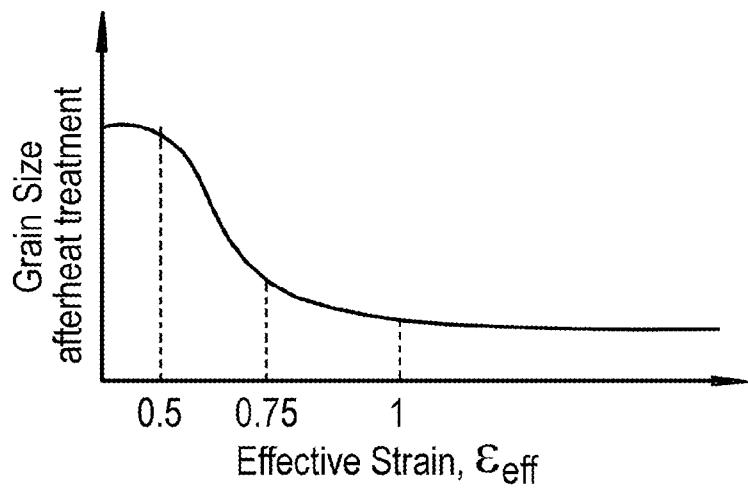


Fig.6

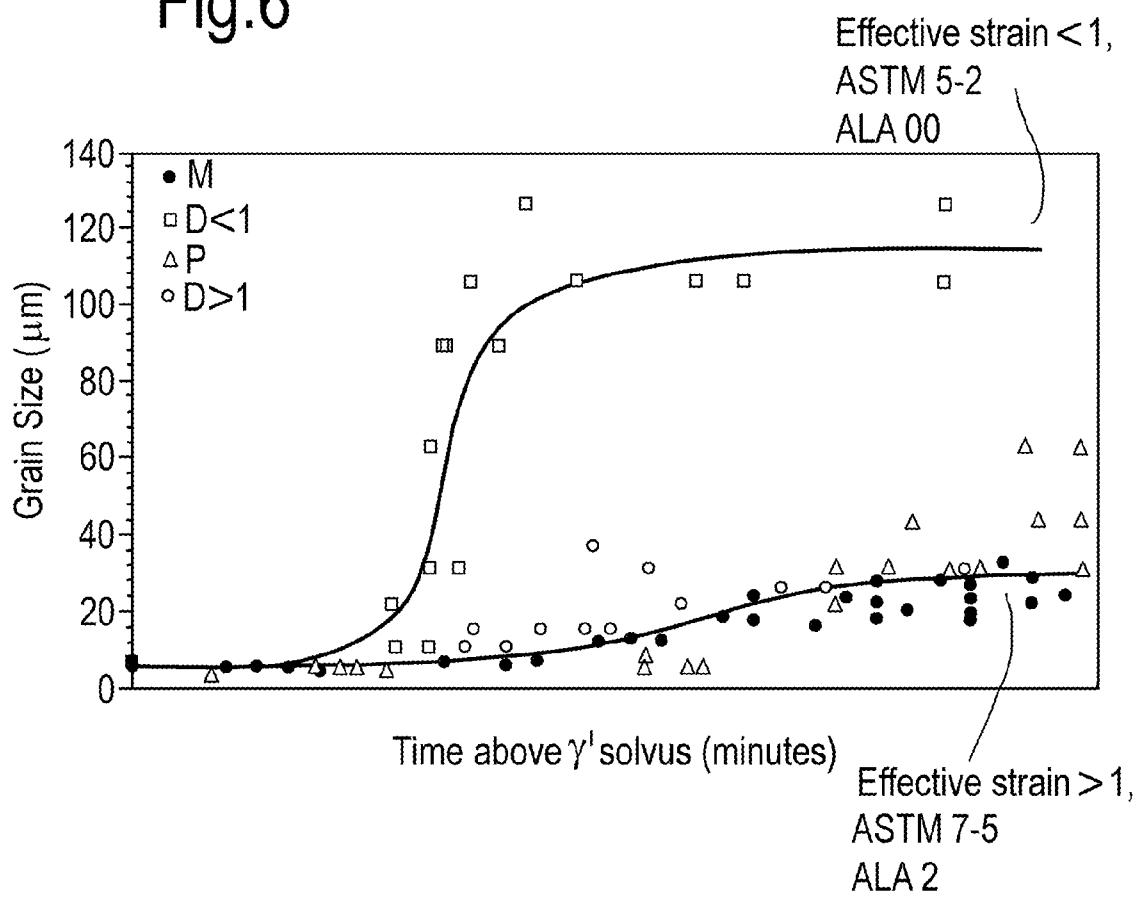


Fig.7  
Fig.8

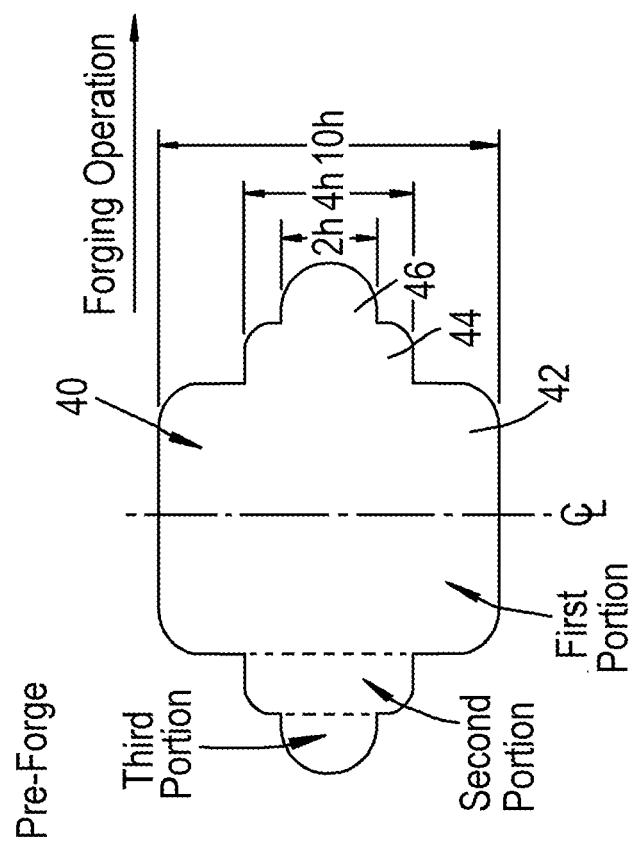


Fig.9

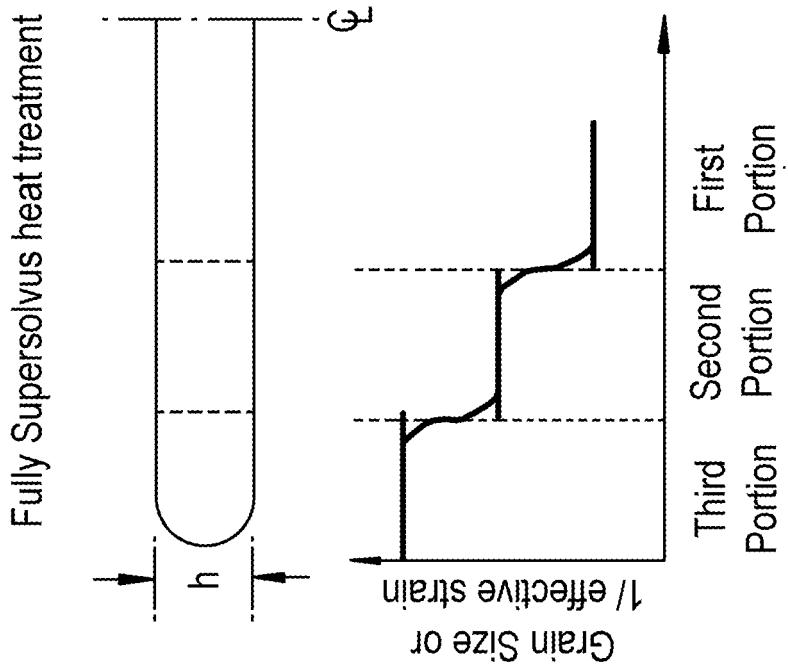
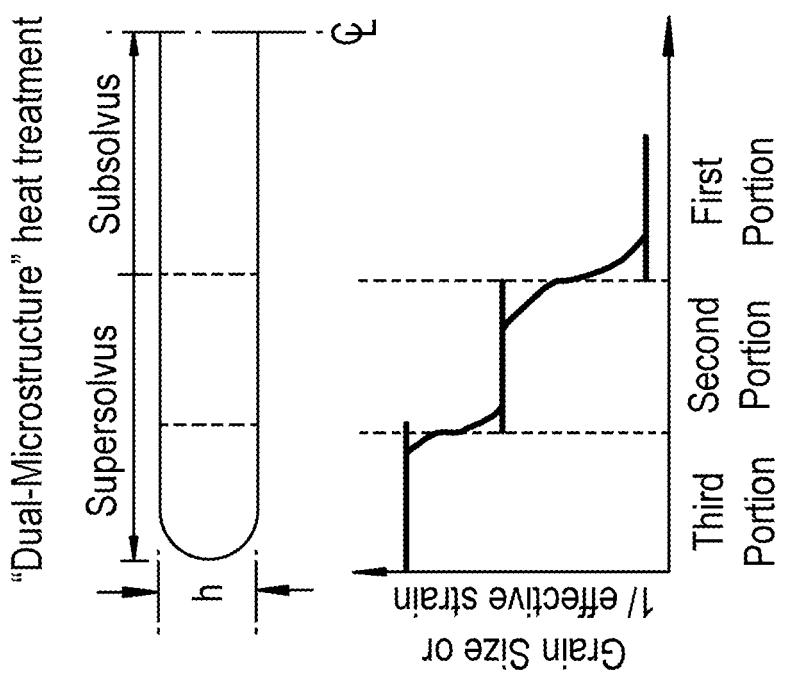


Fig.10





## EUROPEAN SEARCH REPORT

Application Number  
EP 10 17 6784

DOCUMENTS CONSIDERED TO BE RELEVANT			CLASSIFICATION OF THE APPLICATION (IPC)
Category	Citation of document with indication, where appropriate, of relevant passages	Relevant to claim	
X	EP 0 726 333 A2 (GEN ELECTRIC [US]) 14 August 1996 (1996-08-14) * the whole document *	1-13	INV. C22C19/03 C22F1/10
X	US 5 547 523 A (BLANKENSHIP JR CHARLES P [US] ET AL) 20 August 1996 (1996-08-20) * the whole document *	1-13	
X	GB 2 225 790 A (GEN ELECTRIC [US]) 13 June 1990 (1990-06-13) * the whole document *	1-13	
X	FR 2 703 608 A1 (ALUMINUM CO OF AMERICA [US]) 14 October 1994 (1994-10-14) * the whole document *	1-13	
X	US 5 529 643 A (YOON DUK Y [US] ET AL) 25 June 1996 (1996-06-25) * the whole document *	1-13	
X	US 5 374 323 A (KUHLMAN G WILLIAM [US] ET AL) 20 December 1994 (1994-12-20) * the whole document *	1-13	TECHNICAL FIELDS SEARCHED (IPC)
X	US 5 891 272 A (RAYMOND EDWARD L [US] ET AL) 6 April 1999 (1999-04-06) * the whole document *	1-13	C22C C22F
X	US 5 413 752 A (KISSINGER ROBERT D [US] ET AL) 9 May 1995 (1995-05-09) * the whole document *	1-13	
X	US 5 120 373 A (MILLER JOHN A [US] ET AL) 9 June 1992 (1992-06-09) * the whole document *	1-13	
The present search report has been drawn up for all claims			
1	Place of search Munich	Date of completion of the search 23 February 2011	Examiner Patton, Guy
CATEGORY OF CITED DOCUMENTS X : particularly relevant if taken alone Y : particularly relevant if combined with another document of the same category A : technological background O : non-written disclosure P : intermediate document			
T : theory or principle underlying the invention E : earlier patent document, but published on, or after the filing date D : document cited in the application L : document cited for other reasons ..... & : member of the same patent family, corresponding document			

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ON EUROPEAN PATENT APPLICATION NO.

EP 10 17 6784

This annex lists the patent family members relating to the patent documents cited in the above-mentioned European search report. The members are as contained in the European Patent Office EDP file on The European Patent Office is in no way liable for these particulars which are merely given for the purpose of information.

23-02-2011

Patent document cited in search report		Publication date		Patent family member(s)	Publication date
EP 0726333	A2	14-08-1996	DE	69513138 D1	09-12-1999
			DE	69513138 T2	17-08-2000
			US	5593519 A	14-01-1997
US 5547523	A	20-08-1996	NONE		
GB 2225790	A	13-06-1990	CA	1334345 C	14-02-1995
			DE	3926289 A1	21-06-1990
			FR	2640285 A1	15-06-1990
			IT	1232780 B	05-03-1992
			JP	2166260 A	26-06-1990
			JP	3010050 B2	14-02-2000
			US	4957567 A	18-09-1990
FR 2703608	A1	14-10-1994	DE	4412031 A1	13-10-1994
US 5529643	A	25-06-1996	NONE		
US 5374323	A	20-12-1994	US	5360496 A	01-11-1994
US 5891272	A	06-04-1999	US	5584947 A	17-12-1996
US 5413752	A	09-05-1995	NONE		
US 5120373	A	09-06-1992	DE	69220311 D1	17-07-1997
			DE	69220311 T2	02-10-1997
			EP	0533918 A1	31-03-1993
			JP	3058915 B2	04-07-2000
			JP	5508194 T	18-11-1993
			KR	100232758 B1	01-12-1999
			WO	9218660 A1	29-10-1992