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EP 0 067 526 B1

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Description

Background of the invention

This invention relates to steam turbines, and especially is concerned with a superhigh temperature and pressure steam turbine operating at a steam temperature of 600° to 650°C and under a steam pressure of 4000 to 5000 psi. (about 28000 to 35000 kPa.

Owing to petroleum shortage and a consequent rise in the price of petroleum, it has been thought desirable to raise the temperature and pressure of steam used in generating plants so as to increase efficiency. In steam generating plants, it is now usual practice to operate the turbine at a steam temperature of 538°C, and the rotor shaft is formed of a low alloy steel such as a Cr-Mo-V steel. However, in a high temperature and pressure generating plant in which the turbine is operated at a steam temperature of over 600°C and under a pressure of over 4000 psi, it has been revealed that the material now used is not fit for use from the point of view of strength because it is markedly low in creep rupture strength, thermal fatigue strength and strength at high temperatures. Meanwhile steam turbines operating at a main steam temperature of over 600°C are disclosed in "Transaction of the ASME", October 1960, for example. However, this type of steam turbines are low in thermal fatigue resistant property, and not suitable for use as steam turbines that only operate at peak load. More specifically, when a steam turbine is repeatedly started up and shut down, the starting-up and shutting-down impose severe thermal fatigue conditions. Thus it becomes necessary to provide a steam turbine having a rotor and rotor blades having a fatigue life that can withstand this transitory condition, high strength at elevated temperature and high ductility at elevated temperature. Austenitic alloys are considered suitable from the point of view of strength as materials for the rotor and rotor blades operating under conditions of a steam pressure exceeding 4000 psi (about 28000 kPa) and a steam temperature in the range over 600°C. However, these alloys carry the risks that, since various deposit phases are precipitated at high temperature, embrittlement thereof may be accelerated and their strength may be markedly lowered at elevated temperature.

GB—A—912,814 discloses an austenitic nickel-chromium-iron-base alloy which is said to be useful for gas turbine applications; mentioned are turbine wheels, buckets and bolting and structural applications (but not rotor shafts and blades). The amount of manganese present is below 0.50%, and use of high manganese content is discouraged. The 1000 hour creep rupture strength of the material is not given.

Summary of the invention

An object of this invention is to provide a steam turbine including a turbine shaft having superior thermal fatigue resistant property at a main steam temperature of 600° to 650°C.

Another object is to provide a superhigh temperature and pressure steam turbine of high reliability, particularly a steam turbine of the type described including a rotor shaft formed of austenitic forged steel having high strength at elevated temperature, high ductility at elevated temperature and low thermal embrittlement under the steam condition of a temperature range between 600°C and 650°C.

Still another object is to provide a steam turbine having a rotor shaft and rotor blades of high thermal fatigue resistant property under the steam condition of a temperature range between 600° and 650°C.

The invention is set out in claim 1.

The blade sections of the rotor shaft are preferably equidistantly located axially of the rotor shaft in the casing and are preferably unsymmetrical at the center position with respect to the axial direction.

The outer casing member preferably has a substantially spherical external shape, and is preferably formed of Cr-Ni austenitic cast steel or Cr-Mo-V cast steel having a bainite structure.

The inner casing member is preferably formed of Cr-Ni austenitic cast steel.

The rotor shaft is preferably formed of an alloy which consists of

Carbon	0.015	to	0.03%
Silicon	0.3	to	1.0%
Manganese	0.5	to	1.5%
Chromium	14	to	17%
Nickel	24	to	28%
Molybdenum	1	to	2%
Titanium	1.5	to	2.5%
Aluminium	0.15	to	0.4%
Vanadium	0.05	to	0.5%
Boron	0.004	to	0.008%

the balance being iron plus unavoidable impurities. More preferably 0.2—0.3% vanadium is present.

The rotor blades of the superhigh temperature and pressure turbine according to the invention are preferably made of an alloy consisting by weight of the following elements in amounts within the following respective ranges

C	0.01	to	0.1%
Si	0	to	1.5%
Mn	0	to	2%
Cr	10	to	20%
Ni	20	to	30%
Mo	0.5	to	3%
Ti	1.5	to	3%
Al	0.1	to	0.5%
B	0.002	to	0.01%
optionally V	0.05	to	0.5%

the balance Fe and unavoidable impurities, and having a microstructure in which a γ' phase is precipitated in austenite matrix. Most preferably the alloy of the rotor blades consists by weight of

Carbon	0.04	to	0.07%
Silicon	0.3	to	1%
Manganese	0.5	to	1.5%
Chromium	14	to	17%
Nickel	24	to	28%
Molybdenum	1	to	2%
Titanium	1.5	to	2.5%
Aluminium	0.15	to	0.4%
Vanadium	0.05	to	0.5%
Boron	0.004	to	0.008%

the balance being iron plus unavoidable impurities. More preferably 0.2—0.3% vanadium is present.

The alloy of the rotor shaft and also the alloy of the rotor blades of the superhigh temperature and pressure steam turbine according to the invention preferably have, calculated from the aforesaid chemical compositions, the following Ni equivalent and Cr equivalent:

$$\text{Ni equivalent } [Ni(\%)+((30\times C(\%)))+((0.5\times Mn(\%)))] \text{ in the range of 23 to 29}$$

$$\text{Cr equivalent } [Cr(\%)+Mo(\%)+((1.5\times Si(\%)))] \text{ in the range of 12 to } [(\frac{7}{10}\times \text{Ni equivalent})+9]$$

The reasons why the elements are limited to the respective ranges of values in the aforesaid composition are as follows:

Carbon

Strength at elevated temperature increases with an increase in C but a reduction in toughness and thermal embrittlement are accelerated with an increase in C. The carbon content in the material of the rotor shaft that increases strength at elevated temeprature without reducing toughness nor thermal embrittlement is less than 0.04%, preferably 0.015 to 0.03%, and the carbon content of the preferred material of the rotor blades is 0.01—0.1%, preferably between 0.04 and 0.07%.

Manganese

This element is the most important deoxidizing component the manufacture of the alloy. However, when the element is present in too large an amount, toughness and oxidation resistance are adversely affected. Thus the amount in the alloy for the rotor shaft 67 the amount is chosen in the range 0,5 to 1,50%, while for the rotor blades it is preferably not more than 2%, more preferably between 0.5 and 1.5%.

Nickel

This element is an important component for improving high temperature mechanical strength of the steel according to the invention, particularly creep rupture strength and thermal fatigue life thereof. More specifically, this element is conducive to formation of a stable austenite structure and increased high temperature strength. However, in view of its high cost and the reduction of high temperature ductility and yield strength of the steel when its amount is too high, the amount of Ni present is limited between 20 and 30%, preferably between 24 and 28%.

Silicon

Similarly to manganese, this element is a deoxidizing component necessary for production. When the amount present is over 1.5%, however, the forgeability of the steel diminishes and its high temperature toughness is reduced. Thus the upper limit for the preferred blade alloy is 1.5 wt%, the amount being preferably between 0.3 and 1%. For the rotor shaft alloy, the range 0.3 to 1% is chosen.

Molybdenum

This element improves creep rupture strength by strengthening the austenite matrix and forming a carbide. However, the steel has its high temperature ductility reduced and its workability deteriorates when the amount of the element is too large. Thus the amount is limited to between 0.5 and 3 wt%, preferably between 1 and 2%.

Chromium

This element is an important component for improving the high temperature oxidization property of the material according to the invention. As shown in Fig. 1(b), no satisfactory effect is achieved if the amount is under 10%. When the amount is over 20%, embrittlement increases after prolonged holding at elevated temperature. The amount is thus preferably between 14 and 17%.

Titanium

Besides being used as a deoxidizing agent, this element has the effect of hardening alloys by the precipitation thereof. However, when its amount exceeds 3%, the element reduces the ductility and toughness of the steel and accelerates notch deterioration. To increase the high temperature strength of the steel, it is necessary that the amount of this element be not less than 0.5%. Preferably the amount is between 1.5 and 2.5%. For the alloy of the blades, the range preferred is 1.5 to 3%.

Aluminium

Aluminium is added as a deoxidizing agent in the amount of 0.1—0.5%, preferably in the amount of 0.15—0.4%. This element is combined with titanium to cause precipitation of an intermetallic compound, to thereby increase high temperature strength. However, when the amount is too great, it tends to reduce strength. Thus the amount thereof is limited to between 0.1—0.5%.

Boron

This element has the effects of markedly strengthening grain boundary and providing high temperature ductility. However, workabilit, deteriorates when the amount is too large. Thus in the preferred alloy of the blades the amount is limited to between 0.002 and 0.01%, preferably between 0.004 and 0.008%. In the alloy of the rotor shaft, the amount is in the range 0.004 to 0.008%.

Vanadium

This element is optionally added to improve creep strength. When the amount thereof is below 0.05%, no satisfactory effect is achieved. When the amount is over 0.5%, however, ductility and toughness are both adversely affected. The amount is preferably between 0.2 and 0.3%.

Nickel equivalent

Fig. 7a shows the relation between the nickel equivalent (% Ni+30×% C+0.5×% Mn) regarding an

austenite heat resisting steel now in use and the 1000 hour creep rupture strength obtained at 650°C. As shown, an increase in the nickel equivalent is accompanied by an improvement in creep rupture strength. This relation tends to be saturated when the nickel equivalent is about 35%. If the 1000 hour creep rupture strength at 650°C used as a target is made to agree with the scatter band value of 26—34 kg/mm^2 (1000 hour creep rupture strength at 550°C) now used for the currently used material, then the optimum value of the nickel equivalent is between 23 and 29%. (1 kg/mm^2 can be written as 1 MPa).

Chromium equivalent

Chromium equivalent is $12 \leqq (\%\ \text{Cr equivalent}) \leqq \frac{7}{10} \times (\%\ \text{nickel equivalent}) + 9.$

Fig. 7b shows the relation between the chromium content and the increment for high temperature oxidation. Fig. 7b shows that the chromium content should be not less than 12% if its addition is to have any effect in oxidization resistance at elevated temperature. Fig. 7c shows the result obtained with the relation shown in Fig. 7b and the optimum range of the aforesaid nickel equivalent as inserted in Shefla's diagram. In Fig. 7c, if the upper limit of the chromium equivalent is set corresponding to the nickel equivalent on condition that a stable austenite structure is obtained, then the value is 7/10×(% nickel equivalent)+9 (see Fig. 7c) according to Shefla's diagram. The hatching in Fig. 7c indicates the nickel equivalent and chromium equivalent as limited by the invention.

For producing the alloys used in the present invention, it is preferred to effect melting by use of argon-oxygen blowing decarburization process or vacuum decarburization process.

Particular embodiments of the invention are given by way of example in the description set forth hereinafter, in conjunction with the accompanying drawings.

Brief description of the drawings

Fig. 1 is a schematic view of one example of the superhigh temperature and pressure steam turbine according to the invention;

Fig. 2 is a diagrammatic representation showing a relation between absorbed energy and the amount of C in the material according to the invention;

Fig. 3 is a diagrammatic representation of the working stress caused in the rotor shaft according to the operation pattern of the steam turbine now in use;

Fig. 4a is a notional view of the turbine rotor surface temperature-strain (stress) pattern of the steam turbine in actual use;

Fig. 4b is a view showing a model of a high temperature low cycle fatigue pattern simulating the thermal fatigue operation pattern of Fig. 4a which model is a strain-holding type;

Fig. 4c is another model of the fatigue pattern similar to Fig. 4b in which model the strain-holding time is removed;

Fig. 5 is a diagrammatic representation of the results of high temperature low cycle fatigue tests conducted on the materials according to the invention and material of the prior art;

Fig. 6 is a diagrammatic representation of the influence of boron exerted on the high temperature low cycle fatigue life;

Fig. 7a is a diagrammatic representation of the relation between creep rupture strength and nickel equivalent;

Fig. 7b is a diagrammatic representation of the relation between high temperature oxidization increment and the amount of chromium,

Fig. 7c is a Shefla's diagram;

Fig. 8 is a diagram showing the 10^3 hour creep rupture strength of the material according to the invention and the material of the prior art as extrapolated by the Rallson-Mirror process;

Fig. 9 is a diagrammatic representation of the high temperature low cycle fatigue of the material according to the invention and the material of the prior art; and

Fig. 10 is a microscopic photograph of a specimen structure.

Fig. 1 is a sectional view showing the essential portions of one embodiment of the superhigh temperature and pressure steam turbine in conformity with the invention, in which steam is introduced through a main steam line 1 into the turbine and is jetted in a predetermined direction by static blades 3 attached to an inner casing member 2 to thereby rotate rotor blades 5 mounted on a rotor shaft 4. After doing work, the steam flows through a gap between an outer casing member 6 and the inner casing member 2 and then is exhausted through a cooled steam outlet port 7, an exhaust outlet port 8 and an auxiliary exhaust outlet port 9. The exhausted steam is forwarded to a next steam turbine operating at a lower temperature. 10 is the center of each bearing of the rotor shaft 4. 11 and 12 are a gland and a intermediate gland leak outlet port respectively. 13 is a nozzle box. Arrows indicate the direction of flow of the steam.

The embodiment of the invention of the aforesaid construction will be described in detail with

reference to the inner casing member 2 and the outer casing member 6 formed of Cr-Ni austenite cast steel and Cr-Ni austenite forged steel, respectively.

Example 1

Table 1 shows the chemical composition of alloys for rotor shafts used in experiments. Alloys 1 to 3 are according to the present invention. Alloy 4 is for comparison. Each of the alloys 1 to 4 is prepared by the steps of vacuum arc melting, forging, solid-solution treatment of holding it at 980°C for one hour with water-cooling thereafter, and aging treatment of holding at 720°C for 16 hours with air-cooling being effected thereafter, while each of the conventional alloys for comparison is prepared by the steps of vacuum are melting, forging and succeeding necessary treatments shown hereinbelow. The alloys used in the present invention have a microstructure in which a γ' phase is precipitated in the austenite matrix.

Conventional Cr-Mo-V steel serving as a comparative material was cooled by air-blowing after heating at 970°C for 15 hours, and then reheated at 670°C for 48 hours before being cooled in the furnace.

12 Cr steel also serving as another comparative material was cooled by spraying of water in atomized particles after heating at 1050°C for 24 hours, and then subjected to tempering at 650°C for 20 hours. These steel alloys were subjected to V-notch Charpy impact tests and creep rupture tests.

Fig. 2 shows the results of impact tests conducted on the influences of the amount of C on thermal embrittlement by heating at 650°C for 1500 hours. The results show that, whereas absorbed energy of non-heat treated blanks is substantially constant irrespective of the amount of C, the absorbed energy reduces as the amount of C increases in the material heated at 650°C for 1500 hours, showing a marked thermal embrittlement in material with high C content. The values of absorbed energy at 20°C for 12 Cr steel and Cr-Mo-V steel which are now in use for producing rotors are specified as not less than 1.1 kg—m and 0.69 kg—m respectively. Particularly, in a case where the amount of C is not more than 0.03% by weight, the absorbed energy becomes not less than 1.5 kg—m, that is, superior thermal embrittlement resistant property can be obtained. The thermal embrittlement resistant property of each alloy used as a material of the turbine shaft in the present invention is larger than those of the conventional 12% Cr steel and Cr-Mo-V steel.

In the rotor shaft material for the steam turbine according to the invention, intracrystalline rupture (white triangle) prevailed when the material had a C content of below 0.04 wt%. However, when the C content had a value not less than 0.04 wt%, the rupture form has transferred to a grain boundary rupture type (black triangle). The results of analysis have shown that this change in rupture type is accounted for as the deterioration of the grain boundary due to precipitation of the Mx type and $M_{23}C_6$ type carbides in the grain boundary. Thus, to diminish thermal embrittlement in interrelation with thermal fatigue, the amount of C is limited to be less than 0.04%, preferably between 0.015 and 0.03% in the material for rotor shafts according to the invention.

TABLE 1

		Chemical composition (wt%)										
		C	Si	Mn	Ni	Cr	Mo	V	Al	Ti	B	Fe
Inventive steels	No. 1	0.03	0.70	1.35	25.78	15.92	1.23	0.29	0.14	2.23	0.0055	Bal.
	No. 2	0.02	0.69	1.21	25.98	15.79	1.24	0.31	0.14	2.36	0.0057	"
	No. 3	0.040	0.70	1.30	25.80	15.84	1.25	0.30	0.15	2.11	0.0049	"
Composition Steel	No. 4	0.048	0.74	1.35	25.75	15.86	1.25	0.32	0.15	2.98	0.0050	"
Steels of The Prior Art	12 Cr	0.16	0.40	0.71	0.51	10.73	0.98	0.21	—	—	—	"
	Cr-Mo-V	0.28	0.26	0.73	0.36	1.16	1.30	0.25	—	—	—	"

0 067 526

Fig. 3 shows an operation pattern of the severest condition for conventional steam turbines that is operated at a steam temperature of 566°C, that is, the starting-up and shutting-down of the turbine is repeated every 12 hours. When this operation pattern is applied to the superhigh temperature and pressure steam turbine according to the invention, the rotor shaft would be subjected to harsh low cycle fatigue due to high stress at the time of startup and shutdown as shown. Particularly the rotor shaft, unlike the rotor blades, would be subjected to harsh low cycle fatigue due to the combined actions of thermal stress and centrifugal stress because a rise in temperature is gradual in the rotor. Also, it would be subjected to creep due to centrifugal forces in steady-state operation. In Fig. 3, working stress represents thermal stress combined with centrifugal stress.

Table 2 shows the results of tension tests, creep rupture tests and low cycle fatigue tests conducted on the steels 1 and 2 according to the invention at 650°C and on the steel of the prior art at 550°C. It will be seen in the table that the material according to the invention is equal to or higher than the Cr-Mo-V steel in tensile strength and 0.2% yield strength, and that the rate of elongation thereof is 1.5 to 1.75 times as high as in the Cr-Mo-V steel. Creep rupture strength is 1.1 to 1.20 times as high in the material according to the invention as in the Cr-Mo-V steel and low cycle fatigue is equal to or slightly higher in the former than in the latter in spite of the difference of test temperature.

The results of the low cycle fatigue tests represent the number of repetitions continued until rupture occurs at a strain rate of 0.1%/second and strain amounts of 1.0% and 0.65% without holding of the strain (that is, in the same manner as shown in Fig. 4c).

TABLE 2

Items of tests		Prior art		Invention		
		Cr-Mo-V steel	12 Cr steel	No. 1	No. 2	Relative to Cr-Mo-V steel
High temperature tension	Tensile strength (kg/mm^2)	77.3	—	85.5	85.4	1.11
	Yield strength (kg/mm^2)	59.8	—	66.6	68.1	1.11—1.14
	Elongation (%)	14	—	24.5	21.2	1.51—1.75
	Reduction pf Area (%)	40	—	43.0	32.2	0.81—1.08
Creep rupture	1000 Hour strength (kg/mm^2)	27	32	32	34	1.19—1.26
	2000 Hour strength (kg/mm^2)	25	330	29	30	1.16—1.20
High temperature low cycle fatigue	=1.0% (Times)	700	—	770	639	0.91—1.10
	=0.65% (Times)	3000	—	3520	3110	1.04—1.17

0 067 526

From the foregoing, it will be appreciated that the material according to the invention which is used at a temperature of 650°C meets requirements regarding mechanical strength required in the case of Cr-Mo-V steel used at present at 550°C. Thus it has been made clear that the material described above can be used for forming the rotor shaft of a superhigh temperature and pressure steam turbine operating at a steam temperature of 600°—650°C and under a steam pressure of 4000—5000 psi (about 28000 to 35000 kPa).

Example 2

Table 3 shows the chemical composition (wt%) of various materials for forming the rotor blades used in the super-high temperature and pressure steam turbine according to the preferred aspect of the invention. Each material has been obtained by performing vacuum arc melting, forging, and grain size regulation into a range of ASTM G.S. 2.5—4 was effected by holding it at 1050°C for 3 hours. Then each material was water-cooled to room temperature in the same manner as in Example 1 after having been subjected to solid solution treatment of holding it at 899°C for 2 hours. Then, each material was subjected to aging treatments of two steps, i.e. the first step was holding it at 760°C for 16 hours with air-cooling thereafter and the second step was holding it at 718°C for 6 hours with air-cooling thereafter. The blanks obtained by processing the materials through the aforesaid treatments had a microstructure having a γ' phase precipitated in the austenite matrix. The blanks were machined to provide test pieces of predetermined dimensions.

Generally, the blades of a steam turbine are directly subjected to the jetted steam, so that their temperatures rise relatively quickly after the steam turbine is actuated and becomes equal to that of steam. Meanwhile a rise in the temperature of the rotor shaft is relatively slow after the commencement of actuation of the steam turbine because it has a high thermal capacity and austenite steel has a low thermal conductivity. Thus the relatively large difference in temperature between the blades and rotor shaft exists for a substantial period of time. Because of this, the working stress (thermal stress plus centrifugal stress) to which the rotor shaft is subjected is very high at starting-up or shutting-down (or when transferring to idling) as shown in Fig. 3, with the result that the thermal fatigue suffered by the blades is relatively lower than that suffered by the rotor shaft. Thus the amount of C for the materials of the blades is limited between 0.01 and 0.1%, preferably between 0.04 and 0.07%.

Fig. 4a shows a conceptual pattern of a relation between temperature and strain (stress) to which the surface of the rotor is subjected in the steam turbine of superhigh temperature and pressure according to the invention having the blades referred to hereinabove and the rotor shaft described by referring to Example 1. To assess the thermal fatigue life experimenarily, the operation conditions shown in Fig. 4a were converted to a trapezoidal strain cycle (constant temperature) with strain-holding (Fig. 4b), and a tension-compression triangular wave form (constant temperature) without strain-holding (Fig. 4c). The aforesaid conversion has been made on the basis of the hypothesis that there is a correlation between a thermal fatigue phenomenon (varying temperature) and a low cycle fatigue phenomenon (constant temperature).

Fig. 5 is a diagram showing the results of high temperature low cycle fatigue tests conducted by controlling strain or gauge length at a strain rate of 0.1%/sec and at a temperature of 650°C. As shown, it will be seen that the materials of the invention containing 0.002—0.008% B has about twice as long service life as a conventional material containing no B in a low cycle region having a strain range of 0.5—1.2%, indicating that the addition of B has the effect of improving thermal fatigue resistant property.

TABLE 3

	Specimen No.	Chemical composition (%)										
		C	Si	Mn	Cr	Ni	Mo	V	Ti	B	Fe	Al
Present invention	5	0.04	0.44	1.41	15.05	25.56	1.37	0.31	2.17	0.008	Bal.	0.30
	6	0.04	0.44	1.41	15.05	25.56	1.37	0.31	2.17	0.006	"	0.30
	7	0.04	0.44	1.41	15.05	25.56	1.37	0.31	2.17	0.004	"	0.30
	8	0.04	0.52	1.29	14.64	26.44	1.40	0.28	2.11	0.010	"	0.26
	9	0.05	0.84	1.50	14.70	26.30	1.10	0.12	2.11	0.028	"	0.30
Prior art		0.058	0.23	1.27	14.70	26.12	1.40	0.32	2.23	—	"	0.27

Fig. 6 shows the influences exerted on fatigue life by the amount of B at a temperature of 650°C and a strain speed of 0.1%/sec in a strain region of 1.0%. As shown, fatigue life has a peak in the vicinity of 0.006% B and is about twice as long as that of material containing B in an amount outside the range of the invention.

From the foregoing, it will be appreciated that the preferred blades for steam turbines according to the invention have a superior thermal fatigue resistant property and a prolonged service life.

Example 3

Materials were subjected to vacuum arc melting, forging and regulating of grain size under the same conditions as described in Example 1. Then the materials were subjected to solid solution treatment and aging under the heat treating conditions shown in Table 7, and machined to produce materials for blades and rotor shafts. The materials Nos. 10 and 11 are those used for producing blades, and the materials Nos. 12 and 13 are those used for producing rotor shafts. Tables 5 and 6 show the results of creep rupture tests and fatigue tests conducted on these materials, respectively.

As shown in Fig. 4a, the strain and stress caused on the surfaces of the rotors when a steam turbine is operated are such that transient compressive (tensile) strain due to thermal stress occurs at the time of starting-up (or shutting down) at a point A, and tensile stress acts due to centrifugal forces when the operation proceeds to a steadystate condition, so that the operation condition becomes such state as shown at points B and C in the steadystate. Therefore the strain cycle was assessed experimentally from low cycle fatigue based on the triangular waves for the starting-up and shutting-down and from creep rupture strength for steadystate operation.

TABLE 4

Tested materials		Chemical composition (wt%)										
		C	Si	Mn	Cr	Ni	Mo	V	Ti	B	Fe	Al
Invention	No. 10	0.04	0.44	1.41	15.05	25.56	1.37	0.31	2.17	0.004	Bal.	0.30
	No. 11	0.042	0.70	1.45	15.92	25.78	1.23	0.29	2.23	0.0055	"	0.14
	No. 12	0.020	0.69	1.21	15.79	25.98	1.24	0.31	2.36	0.0057	"	0.14
	No. 13	0.036	0.55	1.15	14.15	26.22	1.32	0.31	2.32	0.0082	"	0.25
Conventional material	Cr-Mo-V steel	0.28	0.26	0.73	1.16	0.36	1.30	0.25	—	—	—	—

TABLE 5

Types of steel alloy	Rallson-mirror parameter (C=20)							
	Stress 43 kg/mm^2	40	33	30	27	23	15	12
No. 11	—	37.2	38.2	—	38.9	—	—	—
No. 12	37.3	—	38.5	—	39.1	—	—	—
No. 13	—	36.9	38.1	38.4	38.3	39.4	40.6	41.7

TABLE 6

Types of steel alloy	Temperature (°C)	Strain speed % sec	Fatigue life times	
			Strain range=1.0%	0.65%
No. 11	650	0.1	800	3600
No. 12	650	0.1	700	3200

TABLE 7

Heat treatment	Solid solution treatment	Aging
A (Applied to No. 10)	900 C×2 h · W · Q	760 C×16 h · A · C +720 C×6 h · A · C
B (Applied to No. 11)	980 C×1 h · W · Q	720 C×16 h · A · C

Fig. 8 shows the results of extrapolation of the creep rupture strength of 105 hours conducted by the Rollson-Mirror process. According to this relation, the materials Nos. 10, 11, 12 and 13 covered by the claims of the invention have strength of about 133 MPa at 650°C which is similar to mean creep rupture strength of 127 MPa of the conventional Cr-Mo-V steel tested at 550°C for comparison.

Fig. 9 shows the results of high temperature low cycle fatigue tests effected by controlling strain of gauge length, at a strain rate of 0.1%/sec. The results show that in the entire strain range the fatigue life of the materials according to the present invention at 650°C is equal to or longer than that of the Cr-Mo-V steel at 550°C.

Fig. 10 is a microscopic photograph at a magnification 1000× of the No. 1 alloy having a microstructure in which a γ′ phase is precipitated in the austenite matrix.

The alloy according to the invention has high temperature strength required of the materials for the blades and the rotor operating at a steam temperature of 600°—650°C and is suitable for use as materials for the rotor blades and the rotor.

Example 4

Alloys having chemical compositions shown in Table 8 were produced in the same manner as Example 1.

TABLE 8

	C	Mn	Si	Cr	Ni	Mo	V	Fe	Ti	Al	B	S	P
No. 14	0.038	1.30	0.38	15.04	26.00	1.21	0.32	Bal	1.95	0.23	0.0039	0.0001	0.003
No. 15	0.037	1.32	0.36	15.04	26.17	1.18	0.29	Bal	1.61	0.19	0.0035	0.001	0.003
No. 16	0.021	1.30	0.36	15.13	26.08	1.19	0.31	Bal	2.01	0.22	0.0043	0.001	0.003
No. 17	0.020	1.28	0.36	14.72	25.96	1.27	0.30	Bal	1.66	0.20	0.0045	0.001	0.003

Figs. 11 and 12 show a relation between aging temperature and tensile strength and another relation between aging temperature and creep rupture time at 650°C regarding the above-described alloys of the present invention, respectively. As apparent from Figs. 11 and 12, low aging temperature not more than 740°C is preferred for obtaining improved creep rupture strength and tensile strength. Further, the amount of carbon does not have much influence regarding the enhancement of mechanical strength; however, the lower the amounts carbon and titanium in the alloys, the higher the elongation and reduction of area thereof become.

Example 5

Table 9 shows the chemical composition of another alloy for the rotor shaft 4 of a superhigh temperature and pressure steam turbine according to the invention.

Raw materials for constituting the aforesaid composition were subjected to vacuum induction melting under a vacuum of 10^{-3} to produce electrodes of about 1000 mm in diameter. The electrodes were remelted by an electro-slag-remelting process (ESR) by use of flux consisting of CaF_2 of 55%, Al_2O_3 of 35% and TiO_2 of 10% and cast into columnar ingots. The ingots were diffusion-annealed at a temperature of 1100°—1500°C and forged at a temperature below 1050°C to produce a columnar blank of 850 mm in diameter and 6000 mm in length. The blank was held for 3 hours at a temperature of 1050°C to control the grain size into a range of ASTM G.S. 2.5—4 while rotating the blank at a rate of 3 times per one minute.

TABLE 9

	C	Mn	Si	Cr	Ni	Mo	V	Ti	Al	B	S	P	Fe
No. 18	0.020	1.25	0.35	14.90	26.03	1.25	0.30	1.63	0.20	0.0043	0.001	0.003	Bal.

After subjecting the blank to solution treatment by holding same at a temperature in the range between 900 and 1000°C for 1 hour, it was water-cooled by jetting water thereagainst while vertically holding and rotating it at a rate of three revolutions per minute to room temperature. Thereafter, the blank was held at a temperature between 700° and 730°C for 16 hours to effect aging while rotating it in the same manner, to provide a microstructure in which a γ′ phase is precipitated in austenite matrix. The blank was then machined to obtain a rotor shaft having predetermined dimensions shown in Fig. 13. Specimens for experiments were taken from the left hand end of the rotor shaft as shown in Fig. 13 and were subjected to the tests of tensile strength and creep rupture strength, with the result that there were obtained strength and elongation both substantially similar to those of the specimen No. 17 described above.

Example 6

Table 10 shows the chemical composition (wt%) of a blade used in a superhigh temperature and pressure steam turbine of the present invention.

TABLE 10

	C	Si	Mn	Ni	Cr	Mo	V	Al	Ti	B	Fe
No. 19	0.031	0.65	1.25	25.81	16.00	1.21	0.30	0.15	2.25	0.0056	Bal.

Raw materials for constituting this chemical composition were subjected to vacuum induction melting to produce electrodes (600 mm in diameter) for ESR. The electrodes were remelted by ESR process by use of flux consisting of CaF_2 of 50%, CaO of 25%, TiO_2 of 15% and Al_2O_3 of 10%, and the molten metal was cast into blanks. Then, each of the blanks was heat-treated to control the grain size thereof. After that, there was effected the solution heat treatment of holding the blank at 899°C for 2 hours and of water-cooling thereafter. Then, the blank was subjected to aging treatment of two steps, i.e., in the first step the blank was held at 760°C for 16 hours with air-cooling being effected thereafter and in the second step the blank was held at 718°C for 6 hours with air-cooling thereafter, so that there was obtained a blank having microstructure in which γ′ phase is precipitated in austenite matrix. Such blank was subjected to mechanical working to obtain the blade having predetermined dimensions. Specimens were picked from the blade, which specimens were subjected to the test of evaluating tensile strength, creep rupture strength and high temperature lower cycle fatigue resistant property, with the result that there were obtained values in a degree approximately similar to those of the specimen No. 1 shown in Table 2.

Claims

1. A superhigh temperature and pressure steam turbine comprising a casing, blades in said casing to receive steam jets to cause rotation, and a rotor shaft supporting said blades for rotation, said casing including an inner casing member having static blades secured thereto for guiding the steam jets, and an outer casing member enclosing said inner casing member, said rotor shaft being made of a forged alloy consisting by weight of the following elements in amounts within the following respective ranges

C	less than		0.04%	
Si	0.3	to	1%	
Mn	0.5	to	1.50%	
Cr	10	to	20%	
Ni	20	to	30%	
Mo	0.5	to	3%	
Ti	0.5	to	3%	
Al	0.1	to	0.5%	
B	0.004	to	0.008%	
optionally, V	0.05	to	0.5%	

the balance being Fe and unavoidable impurities, and having a microstructure in which a γ′ phase is precipitated in austenite matrix, the alloy having been subjected to an aging treatment, whereby the rotor has 1000 hour creep rupture strength of 26—34 kg/mm^2 at 650°C.

2. A steam turbine according to claim 1 wherein said blades are made of an alloy consisting by weight of the following elements in amounts within the following respective ranges

C	0.01	to	0.1%
Si	0	to	1.5%
Mn	0	to	2%
Cr	10	to	20%
Ni	20	to	30%
Mo	0.5	to	3%
Ti	1.5	to	3%
Al	0.1	to	0.5%
B	0.002	to	0.01%
optionally, V	0.05	to	0.5%

the balance being Fe and unavoidable impurities, and having a microstructure in which a γ' phase is precipitated in austenite matrix.

3. A steam turbine according to claim 1 or claim 2 wherein the Ni equivalent of said alloy of which the rotor shaft is made is in the range 23 to 29, the Ni equivalent being

$$[Ni(\%)+\{(30\times C(\%)+(0.5\times Mn(\%)\}]$$

and the Cr equivalent thereof given by

$$[Cr(\%)+Mo(\%)+\{1.5\times Si(\%)\}]$$

is in the range

$$12 \text{ to } [(\frac{7}{10}\times \text{Ni equivalent})+9]$$

this alloy being an austenitic forged steel.

4. A steam turbine as claimed in any of Claims 1 to 3 wherein said alloy of said rotor shaft consists by weight of

Carbon	0.015	to	0.03%
Silicon	0.3	to	1.0%
Manganese	0.5	to	1.5%
Chromium	14	to	17%
Nickel	24	to	28%
Molybdenum	1	to	2%
Titanium	1.5	to	2.5%
Aluminium	0.15	to	0.4%
Vanadium	0.05	to	0.5%
Boron	0.004	to	0.008%

the balance being iron plus unavoidable impurities.

5. A steam turbine as claimed in Claims 2 or 3 wherein said alloy of said blades consists by weight of

Carbon	0.04	to	0.07%
Silicon	0.3	to	1%
Manganese	0.5	to	1.5%
Chromium	14	to	17%
Nickel	24	to	28%
Molybdenum	1	to	2%
Titanium	1.5	to	2.5%
Aluminium	0.15	to	0.4%
Vanadium	0.05	to	0.5%
Boron	0.004	to	0.008%

the balance being iron plus unavoidable impurities.

Patentansprüche

1. Dampfturbine für extrem hohe Temperaturen und Drücke, umfassend ein Gehäuse, Schaufeln in dem Gehäuse, denen Dampfströme zugeführt werden, um Drehung hervorzurufen, und eine die Schaufeln drehbar tragende Rotorwelle, wobei das Gehäuse ein Innengehäuseteil mit daran befestigten statischen Schaufeln zur Führung der Dampfströme und ein das Innengehäuseteil umschließendes Außengehäuseteil aufweist, wobei die Rotorwelle aus einer geschmiedeten Legierung hergestellt ist, die aus den folgenden Elementen in Gewichtsmengen innerhalb der nachstehend jeweils angegebenen Bereiche besteht:

C	weniger	als		0,04%
Si	0,3	bis		1%
Mn	0,5	bis		1,50%
Cr	10	bis		20%
Ni	20	bis		30%
Mo	0,5	bis		3%
Ti	0,5	bis		3%
Al	0,1	bis		0,5%
B	0,004	bis		0,008%
wahlweise V	0,05	bis		0,5%

Rest Fe und umvermeidbare Verunreinigungen, mit einer Mikrostruktur, bei der eine γ'-Phase in austenitischer Matrix niedergeschlagen ist, wobei die Legierung einer Alterungsbehandlung unterworfen worden ist, so daß der Rotor eine 1000 h-Zeitstandfestigkeit von 26 bis 34 kg/mm^2 bei 650°C aufweist.

2. Dampfturbine nach Anspruch 1, wobei die Schaufeln aus einer Legierung hergestellt sind, die aus den folgenden Elementen in Gewichtsmengen innerhalb der nachstehend jeweils angegebenen Bereiche besteht:

C	0,01	bis		0,1%
Si	0	bis		1,5%
Mn	0	bis		2%
Cr	10	bis		20%
Ni	20	bis		30%
Mo	0,5	bis		3%
Ti	1,5	bis		3%
Al	0,1	bis		0,5%
B	0,002	bis		0,01%
wahlweise V	0,05	bis		0,5%

Rest Fe und unvermeidbare Verunreinigungen, mit einer Mikrostruktur, bei der eine γ'-Phase in austenitischer Matrix niedergeschlagen ist.

3. Dampfturbine nach Anspruch 1 oder 2, wobei das Ni-Äquivalent der Legierung, aus der die Rotorwelle hergestellt ist, im Bereich von 23 bis 29 liegt, wobei das Ni-Äquivalent

$$[Ni(\%)+\{(30\times C(\%)+(0,5\times Mn(\%)\}]$$

ist und das Cr-Äquivalent davon, das durch

[Cr(%)+Mo(%)+{1,5×Si(%)}]

gegeben ist, im Bereich von

$$12 \text{ bis } [(\frac{7}{10} \times \text{Ni-Äquivalent}) + 9]$$

liegt, wobei diese Legierung ein austenitischer geschmiedeter Stahl ist.

4. Dampfturbine nach einem der Ansprüche 1 bis 3, wobei die Legierung der Rotorwelle gewichtsmäßig besteht aus

Kohlenstoff	0,015	bis	0,03%
Silicium	0,3	bis	1,0%
Mangan	0,5	bis	1,5%
Chrom	14	bis	17%
Nickel	24	bis	28%
Molybdän	1	bis	2%
Titan	1,5	bis	2,5%
Aluminium	0,15	bis	0,4%
Vanadium	0,05	bis	0,5%
Bor	0,004	bis	0,008%

Rest Eisen plus unvermeidbare Verunreinigungen.

5. Dampfturbine nach Anspruch 2 oder 3, wobei die Legierung der Schaufeln gewichtsmäßig besteht aus

Kohlenstoff	0,04	bis	0,07%
Silicium	0,3	bis	1%
Mangan	0,5	bis	1,5%
Chrom	14	bis	17%
Nickel	24	bis	28%
Molybdän	1	bis	2%
Titan	1,5	bis	2,5%
Aluminium	0,15	bis	0,4%
Vanadium	0,05	bis	0,5%
Bor	0,004	bis	0,008%

Rest Eisen plus unvermeidbare Verunreinigungen.

Revendications

1. Turbine à vapeur fonctionnant à températures et pressions très élevées comprenant un carter, des aubes dans ledit carter pour recevoir des jets de vapeur pour provoquer la rotation, et un arbre de rotor supportant lesdites aubes en vue de leur rotation, ledit carter comportant un élément de carter interne muni d'aubes fixes fixées à lui pour guider les jets de vapeur et un élément de carter externe enfermant

l'élément de carter interne, ledit arbre de rotor étant constitué d'un alliage forgé formé des éléments suivants dont les quantités exprimées en poids se trouvent dans les plages suivantes:

	C	inférieur	à	0,04%
	Si	0,3	à	1%
	Mn	0,5	à	1,5%
	Cr	10	à	20%
	Ni	20	à	30%
	Mo	0,5	à	3%
	Ti	0,5	à	3%
	Al	0,1	à	0,5%
	B	0,004	à	0,008%
facultativement	V	0,05	à	0,5%

le reste étant du Fe et des impuretés inévitables, et possédant une microstructure dans laquelle une phase γ' est précipitée dans une matrice d'austénite, l'alliage ayant été soumis à un traitement de vieillissement, ce par quoi le rotor possède une résistance au fluage en 1000 heures de 26—34 kg/mm² à 650°C.

2. Turbine à vapeur selon la revendication 1, dans laquelle lesdites aubes sont constituées d'un alliage comprenant les éléments suivants dont les quantités exprimées en poids se trouvent dans les plages suivantes:

	C	0,01	à	0,1%
	Si	0	à	1,5%
	Mn	0	à	2%
	Cr	10	à	20%
	Ni	20	à	30%
	Mo	0,5	à	3%
	Ti	1,5	à	3%
	Al	0,1	à	0,5%
	B	0,002	à	0,01%
facultativement	V	0,05	à	0,5%

le reste étant du Fe et des impuretés inévitables et possédant une microstructure dans laquelle une phase γ' est précipitée dans une matrice d'austénite.

3. Turbine à vapeur selon la revendication 1 ou la revendication 2 dans laquelle l'équivalent Ni dudit alliage dont est constitué l'arbre de rotor est compris entre 23 et 29, l'équivalent Ni étant

$$[Ni(\%)+\{(30\times C(\%)+(0,5\times Mn(\%)\}]$$

et son équivalent Cr donné par

$$[Cr(\%)+Mo(\%)+\{1,5\times Si(\%)\}]$$

est compris entre

$$12 \text{ à } [(\frac{7}{10}\times \text{équivalent Ni})+9]$$

l'alliage étant un acier austénitique forgé.

4. Turbine à vapeur selon l'une quelconque des revendications 1 à 3, dans laquelle ledit alliage dudit arbre de rotor est constitué en poids de:

Carbone	0,015	à	0,03%
Silicium	0,3	à	1,0%
Manganèse	0,5	à	1,5%
Chrome	14	à	17%
Nickel	24	à	28%
Molybdène	1	à	2%
Titane	1,5	à	2,5%
Aluminium	0,15	à	0,4%
Vanadium	0,05	à	0,5%
Bore	0,004	à	0,008%

le reste étant du fer plus des impuretés inévitables.

5. Turbine à vapeur selon les revendications 2 ou 3, dans laquelle ledit alliage desdites aubes est constitué en poids de:

Carbone	0,04	à	0,07%
Silicium	0,3	à	1%
Manganèse	0,5	à	1,5%
Chrome	14	à	17%
Nickel	24	à	28%
Molybdène	1	à	2%
Titane	1,5	à	2,5%
Aluminium	0,15	à	0,4%
Vanadium	0,05	à	0,5%
Bore	0,004	à	0,008%

le reste étant du fer plus des impuretés inévitables.

FIG. 1

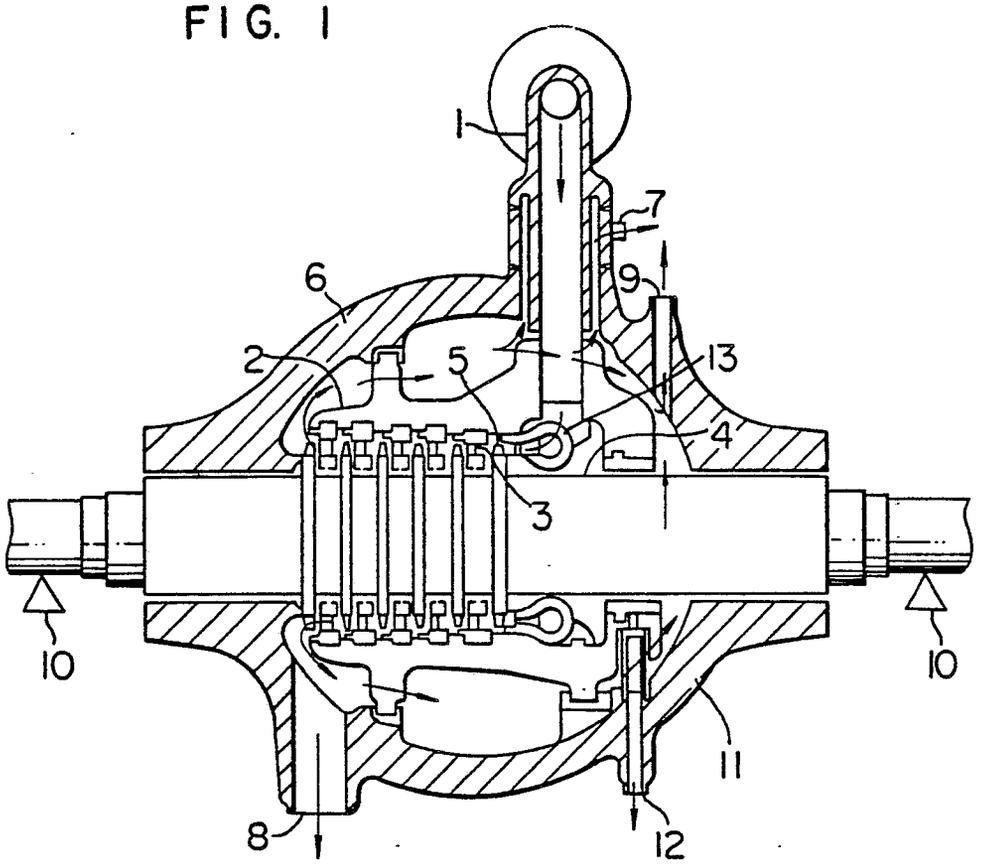

1
7
6
9
2
5
13
4
3
10
10
11
8
12

FIG. 2

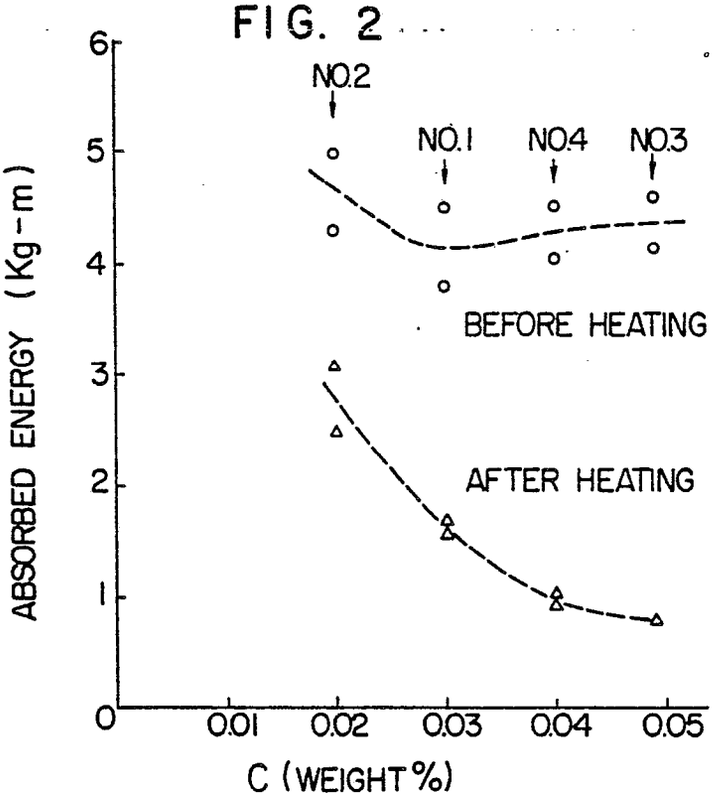

NO.2
NO.1
NO.4
NO.3
BEFORE HEATING
AFTER HEATING
ABSORBED ENERGY (Kg-m)
0
1
2
3
4
5
6
0.01
0.02
0.03
0.04
0.05
C (WEIGHT %)

FIG. 3

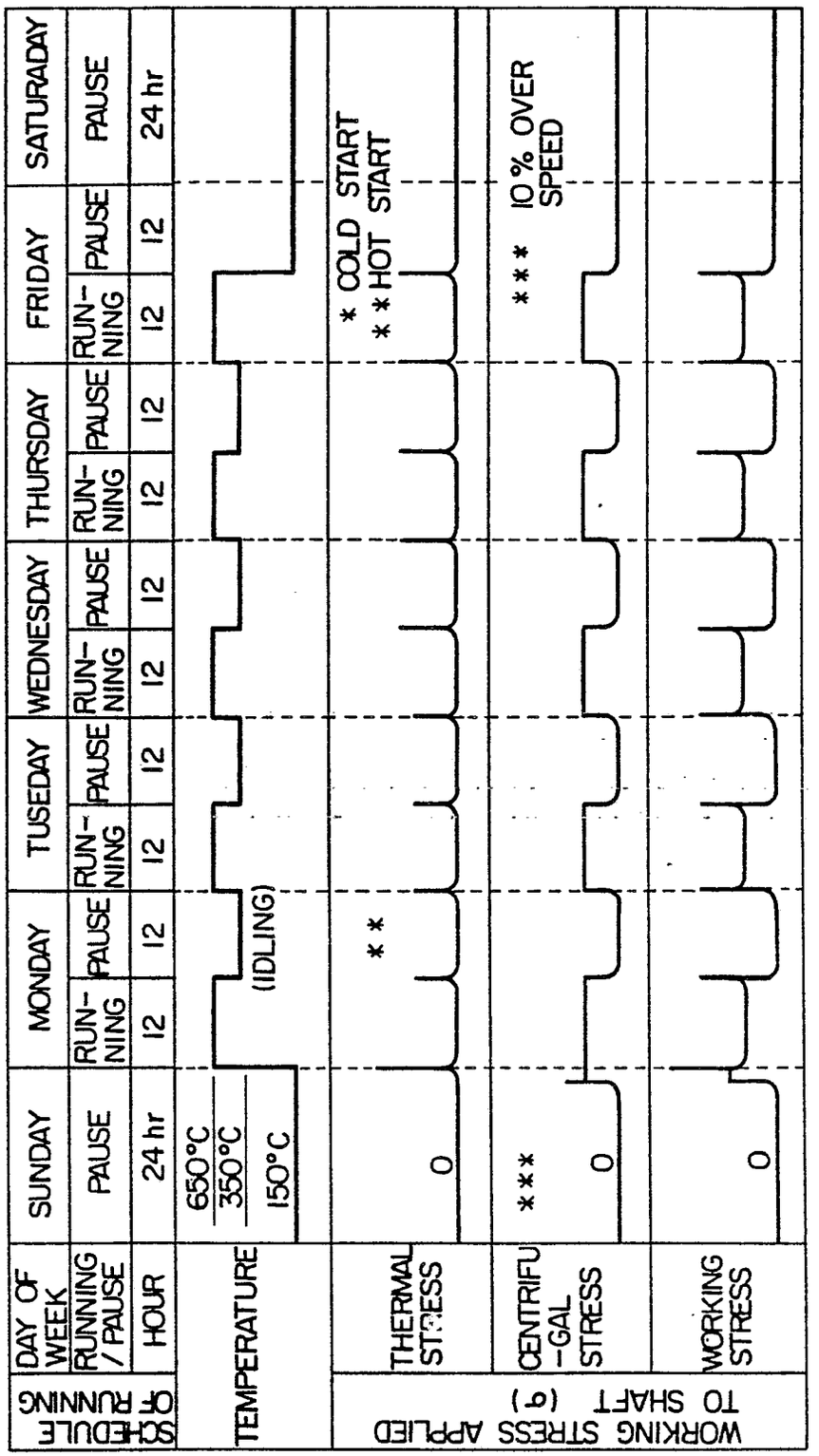

SCHEDULE OF RUNNING
DAY OF WEEK
RUNNING / PAUSE
HOUR
SUNDAY
PAUSE
24 hr
MONDAY
RUN-NING
PAUSE
12
12
TUESDAY
RUN-NING
PAUSE
12
12
WEDNESDAY
RUN-NING
PAUSE
12
12
THURSDAY
RUN-NING
PAUSE
12
12
FRIDAY
RUN-NING
PAUSE
12
12
SATURADAY
PAUSE
24 hr
TEMPERATURE
650°C
350°C
150°C
(IDLING)
WORKING STRESS APPLIED TO SHAFT (g)
THERMAL STRESS
0
**
* COLD START
** HOT START
CENTRIFU-GAL STRESS
***
0
*** 10% OVER SPEED
WORKING STRESS
0

FIG. 4a

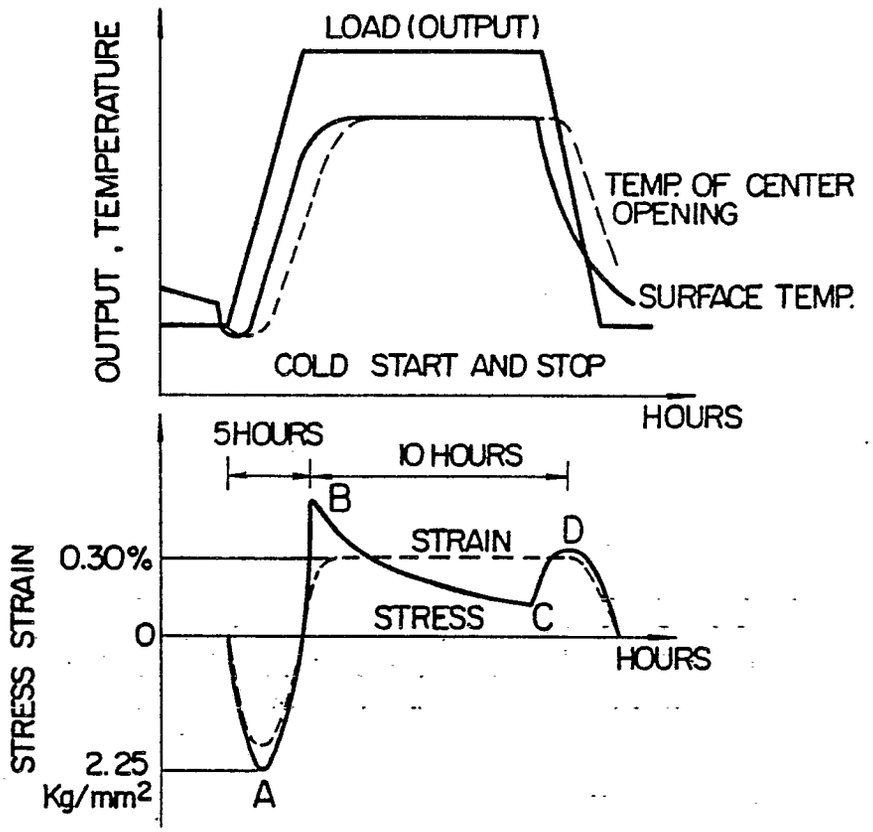


FIG. 4b

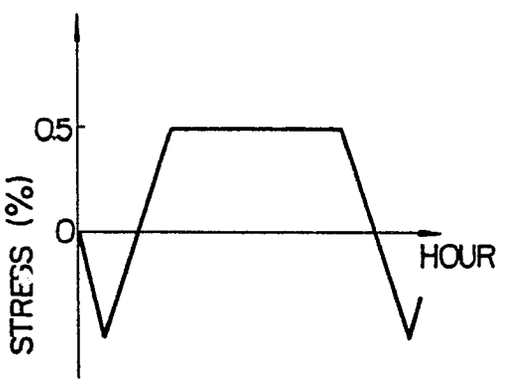


FIG. 4c

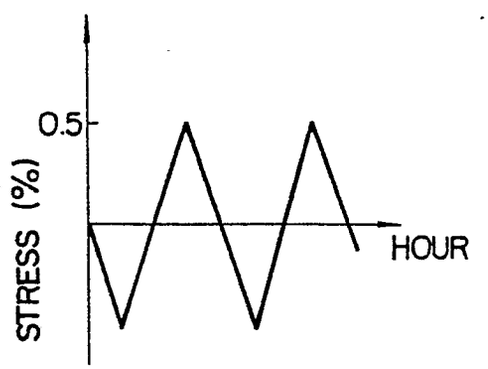


FIG. 5

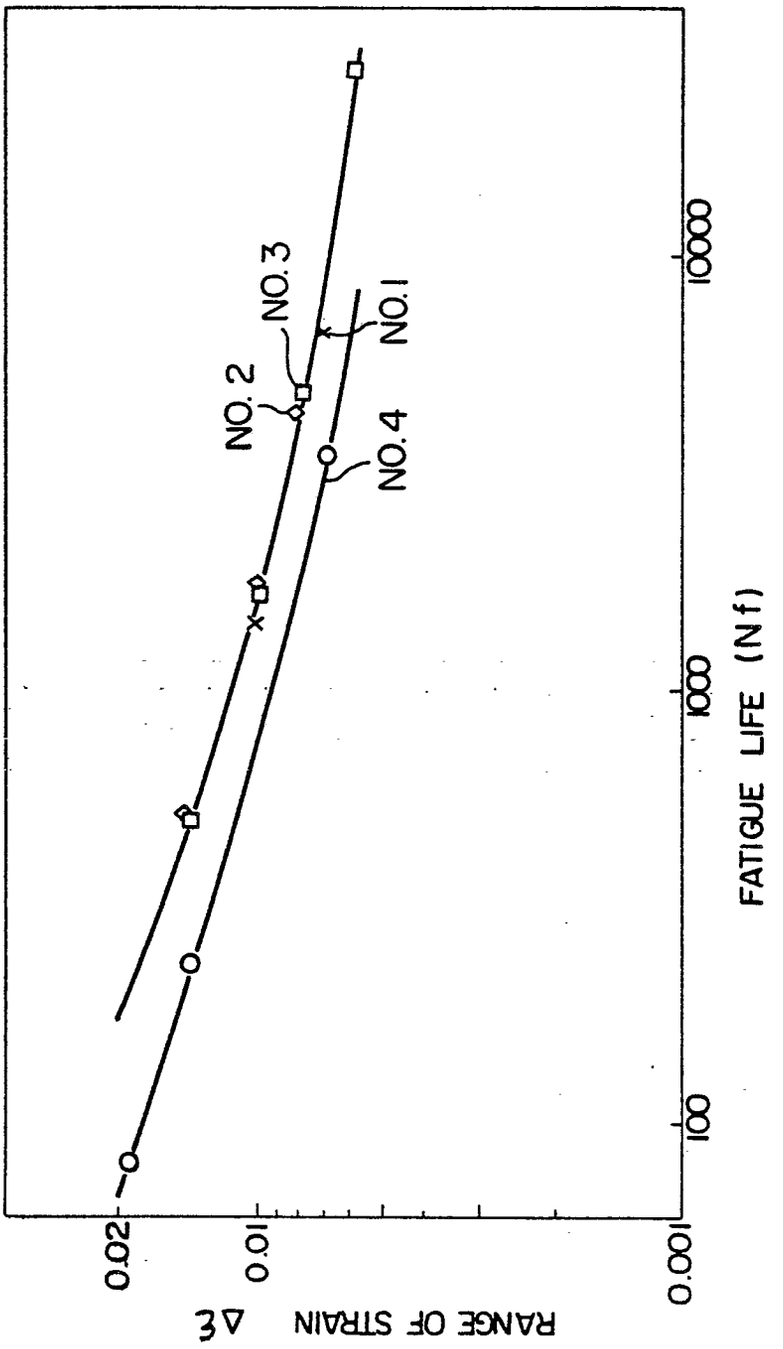
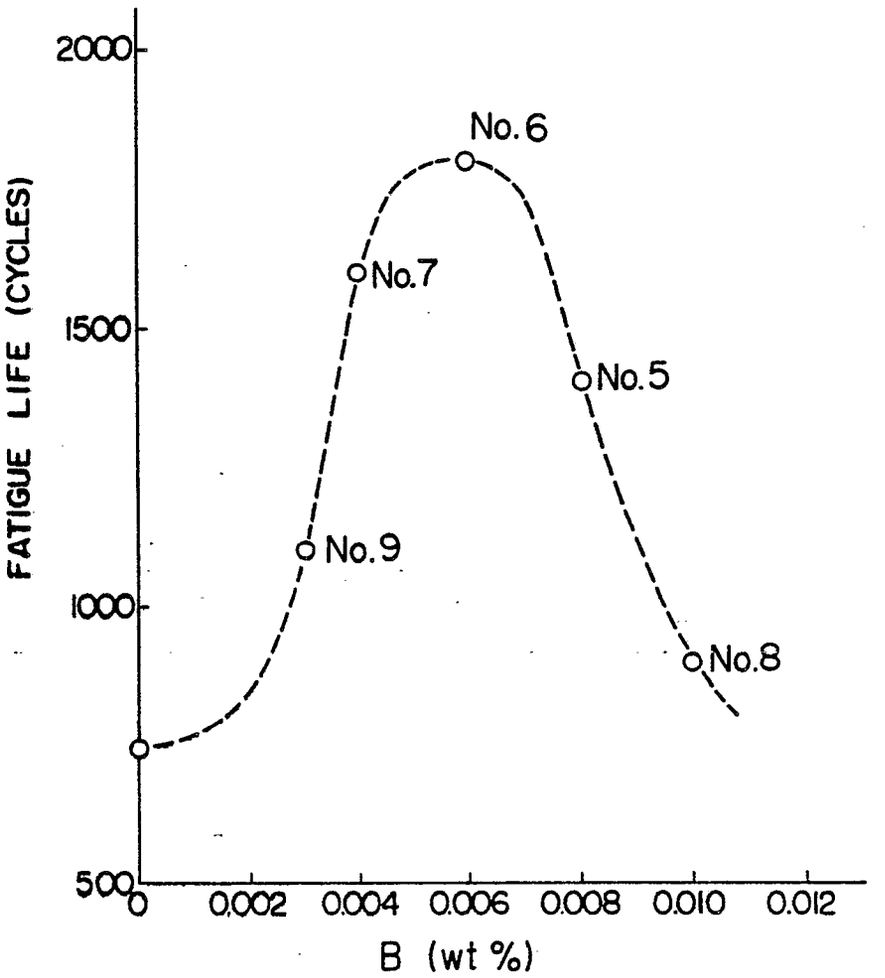

NO.1
NO.2
NO.3
NO.4
0.02
0.01
0.001
100
1000
10000
RANGE OF STRAIN Δε
FATIGUE LIFE (Nf)

FIG. 6


2000
1500
1000
500
FATIGUE LIFE (CYCLES)
No.6
No.7
No.5
No.9
No.8
0
0.002
0.004
0.006
0.008
0.010
0.012
B (wt %)

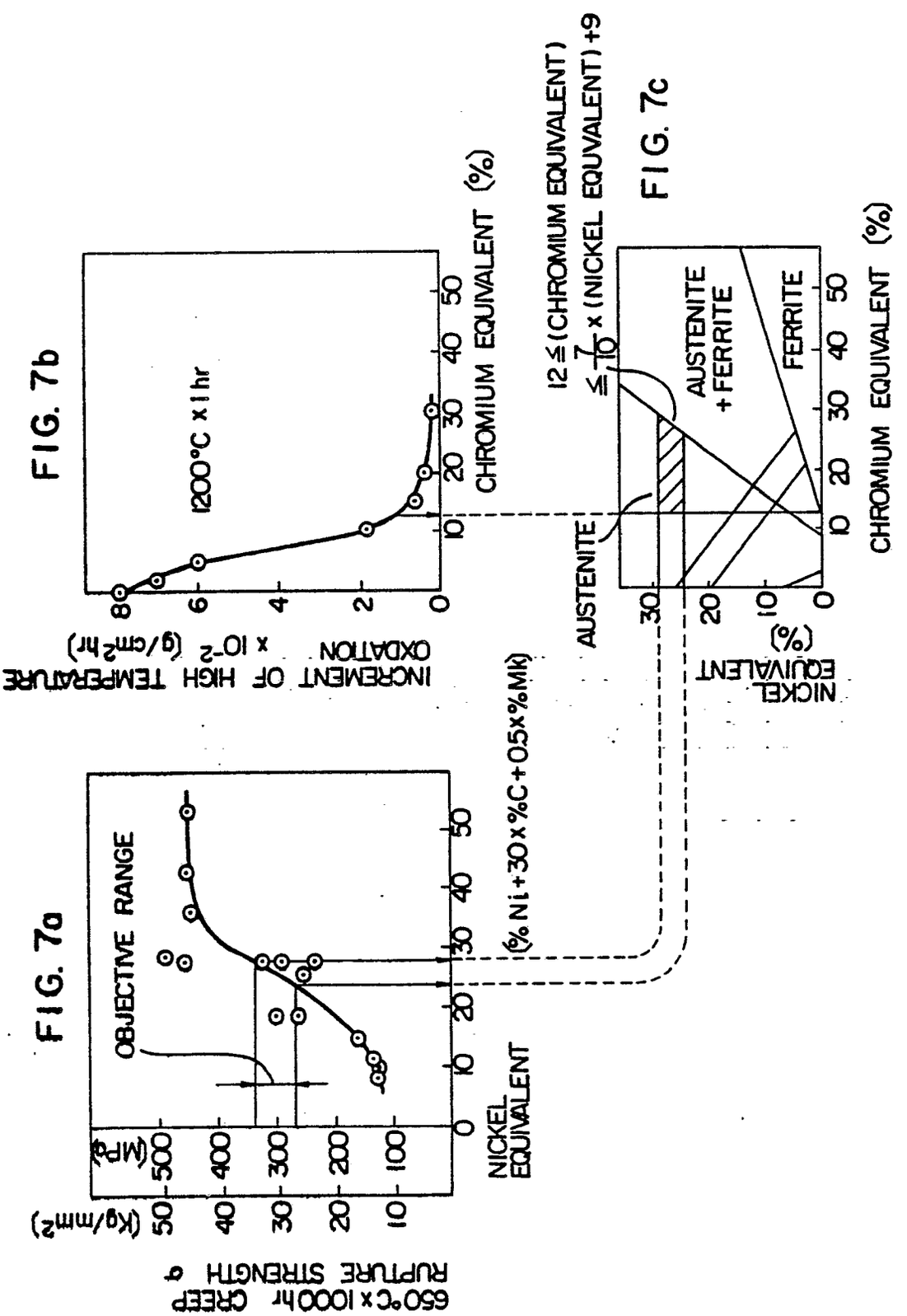

FIG. 7a
650°C x 10000hr CREEP RUPTURE STRENGTH σ
(Kg/mm²)
(MPa)
OBJECTIVE RANGE
NICKEL EQUIVALENT
(%Ni+30x%C+0.5x%Mk)
FIG. 7b
INCREMENT OF HIGH TEMPERATURE OXDATION x 10⁻² (g/cm²hr)
1200°C x 1hr
CHROMIUM EQUIVALENT (%)
FIG. 7c
12 ≦ (CHROMIUM EQUIVALENT) ≦ 7/10 x (NICKEL EQUVALENT)+9
AUSTENITE
AUSTENITE +FERRITE
FERRITE
NICKEL EQUIVALENT (%)
CHROMIUM EQUIVALENT (%)

FIG. 8

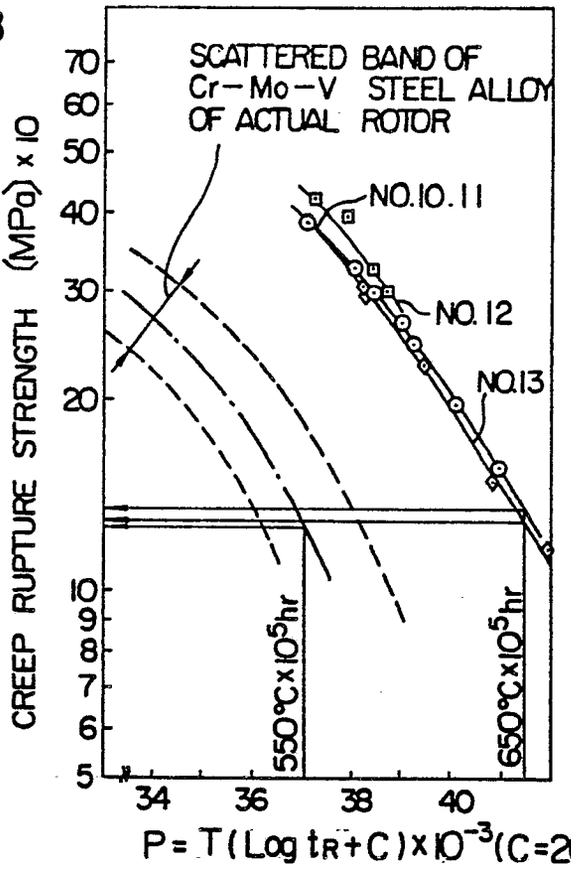

SCATTERED BAND OF
Cr-Mo-V STEEL ALLOY
OF ACTUAL ROTOR
NO.10.11
NO.12
NO.13
CREEP RUPTURE STRENGTH (MPa) x 10
70
60
50
40
30
20
10
9
8
7
6
5
550°C x 10^5hr
650°C x 10^5hr
34
36
38
40
P = T(Log tR + C) x 10^-3 (C=20)

FIG. 9

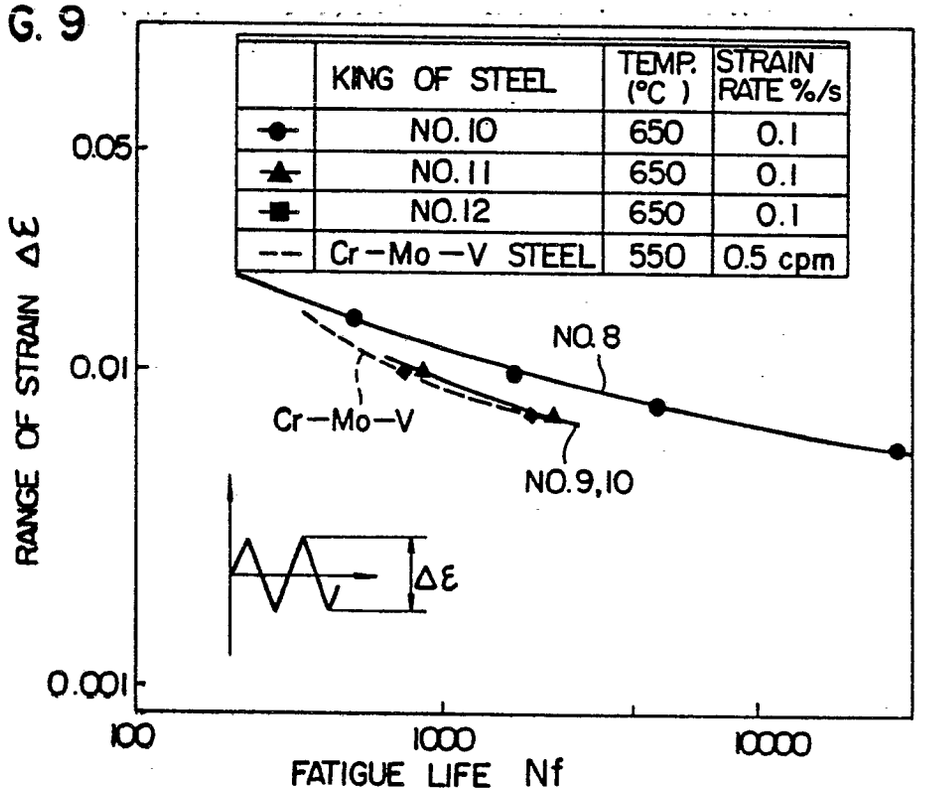
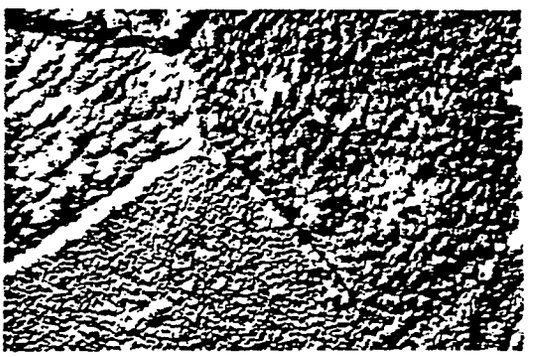

KING OF STEEL
TEMP. (°C)
STRAIN RATE %/s
NO.10
650
0.1
NO.11
650
0.1
NO.12
650
0.1
Cr-Mo-V STEEL
550
0.5 cpm
NO.8
Cr-Mo-V
NO.9,10
Δε
RANGE OF STRAIN Δε
0.05
0.01
0.001
100
1000
10000
FATIGUE LIFE Nf

FIG. 10



x 5000

FIG. 11

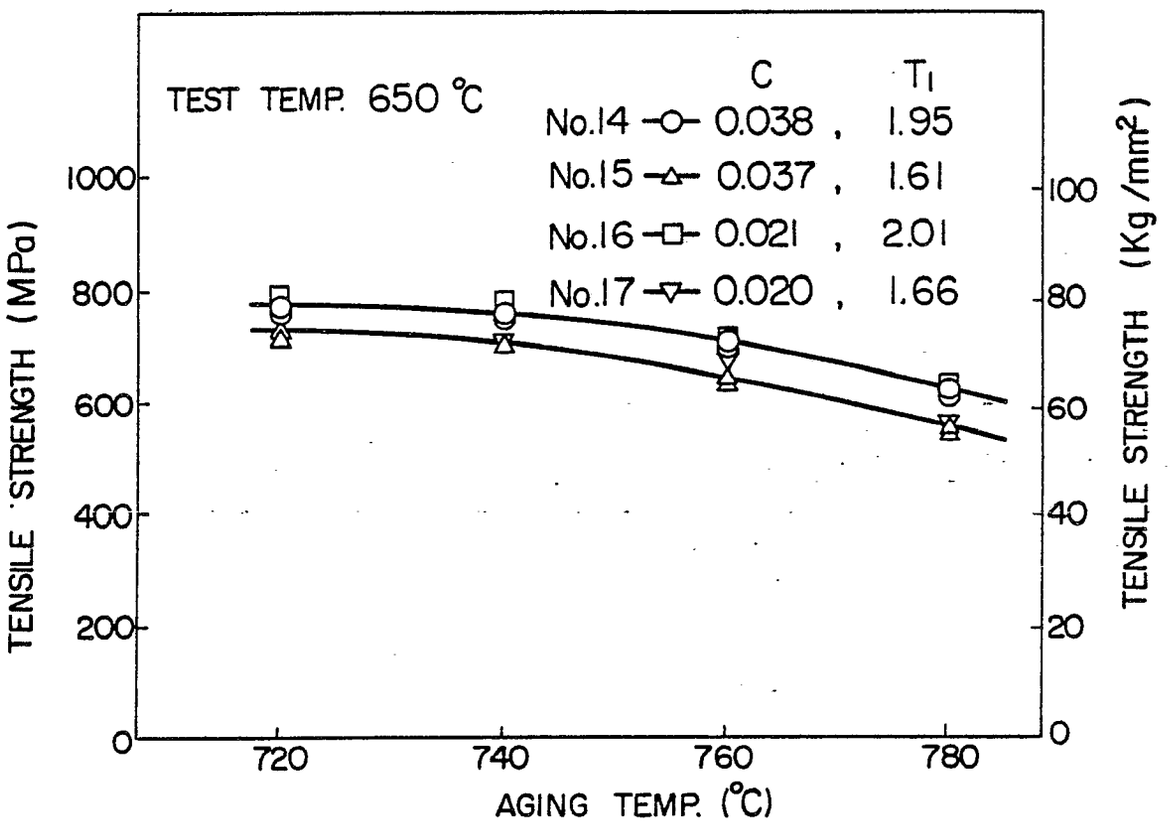

TEST TEMP. 650 °C
C
T_1
No.14 0.038 , 1.95
No.15 0.037 , 1.61
No.16 0.021 , 2.01
No.17 0.020 , 1.66
TENSILE STRENGTH (MPa)
TENSILE STRENGTH (Kg/mm²)
1000
800
600
400
200
0
100
80
60
40
20
0
720
740
760
780
AGING TEMP. (°C)

FIG. 12

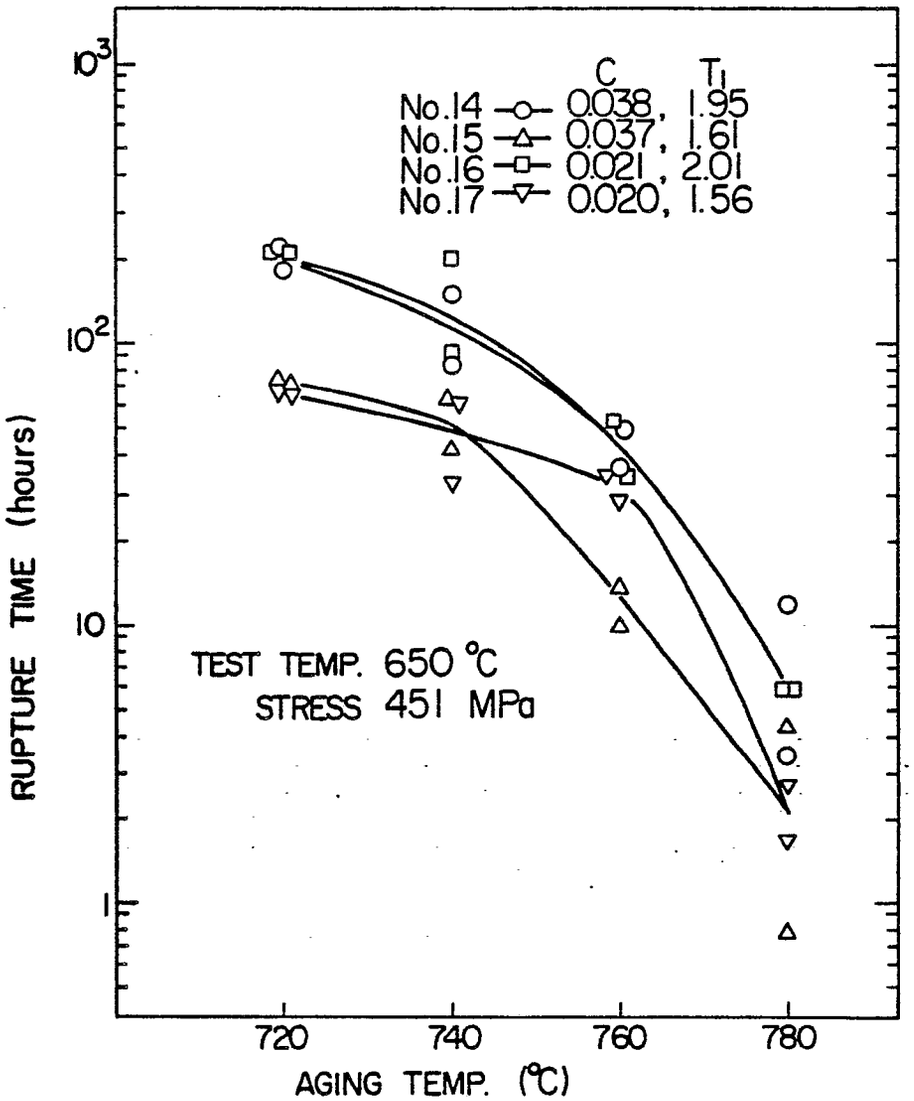
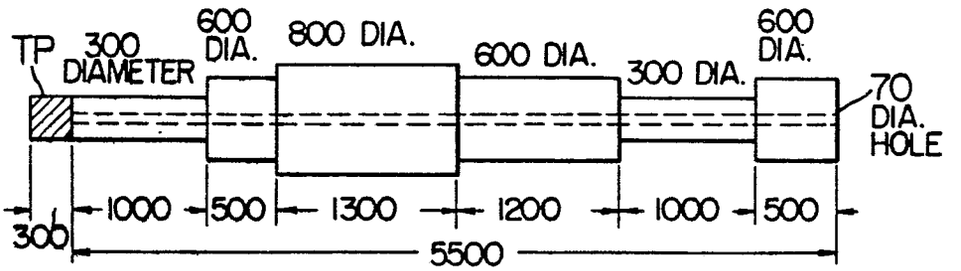

C T1
No.14 0.038, 1.95
No.15 0.037, 1.61
No.16 0.021, 2.01
No.17 0.020, 1.56
TEST TEMP. 650 °C
STRESS 451 MPa
RUPTURE TIME (hours)
10³
10²
10
1
720
740
760
780
AGING TEMP. (°C)

FIG. 13


TP
300 DIAMETER
600 DIA.
800 DIA.
600 DIA.
300 DIA.
600 DIA.
70 DIA. HOLE
300
1000
500
1300
1200
1000
500
5500