THE STRENGTHENING EFFECT OF NIOBIUM ON Ni-Cr-Ti TYPE WROUGHT SUPERALLOY

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In present work based on the measured distribution of niobium in γ, γ' and carbide (roughly in 5:3:1 proportion) and the effect of niobium in increasing the lattice spacings of γ and γ', as well as the long range order parameter of γ', the contribution of niobium on solid solution hardening and age hardening were estimated separately theoretically.

1. INTRODUCTION

The effect of niobium in superalloys has been studied extensively. However, most of the research works had been limited on its alloying behaviour, and the detailed study on strengthening effect of niobium from various aspects of strengthening mechanisms had scarcely been seen. The present work was proceeded in order to clarify the cause of alloy strengthening by niobium.

2. EXPERIMENTAL PROCEDURE AND RESULTS

The matrix alloy was vacuum induction melted with the following composition (in wt%) C 0.04, Al 0.95, Ti 2.9, Cr 20, B 0.01, Ce 0.01, balance by Ni, a group of experimental alloys were then obtained by vacuum induction remelting the matrix alloy with the addition of various proportion of niobium and a little quantity of C, Ti and Mn. The percentage niobium content of the experimental alloys is shown in table 1. The ingots were hot rolled to rods, which were solution treated at 1080°C for 8 hours, A.C., aged at 750°C for 16 hours, A.C. Specimens were machined from the rods. Room temperature tensile properties are given in table 3.

In order to study the phase constitution of the alloys and the partitioning of niobium in γ' and γ phases, two regimes of electrolysis for quantitative extraction of γ' phase and other phases were chosen: (1) 0.5% HNO_3 + 5% $C_4H_6O_6$ water solution, density of current 0.03 A/cm^2, room temperature. (2) 3.6% $ZnCl_2$ + 5% HCl + 2% $C_4H_6O_6$ + CH_3OH solution, density of current 0.1 A/cm^2, temperature -7°C to -5°C. The results of phase analysis of extracted residues were shown in table 1, from which the distributions of niobium in γ, γ' and carbide were obtained and shown in Fig. 1. Furthermore, the lattice spacings of γ and γ', the γ' particle size and its long range order parameter for alloys of various niobium content were measured by X-ray diffraction technics and were shown in table 2 and Fig.1. The long range order parameter was measured according to R. Mihalish (1). Integrated intensities of X-ray diffraction lines were measured from powder compact of extracted residue. Specific gravity and shear modulus of the alloys were also measured and shown in table 2. Thin foils of the alloys were examined in a JSEM-200B scanning transmission electron microscope, cubic form of γ' phase in various experimental alloys was observed, furthermore, superlattice dislocation pairs were also observed as shown in Fig. 2.

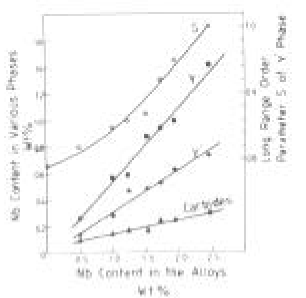

Fig.1.Partitioning of Nb and variation of long range order parameter

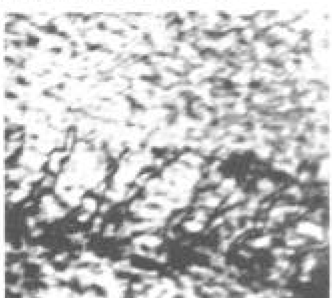

Fig.2. Superlattice dislocation pairs in alloy No.5, bright field. 100,000X

Table 1. Nb Content in the Alloys and Results of Phase Analysis

No.	Nb Content in the alloys wt%	γ' content wt%						Carbide content wt%				
		Ni	Cr	Nb	Al	Ti	Total	Ni	Ti	Nb	Cr	Total
1	--	9.46	0.222	--	0.666	1.387	11.74	<0.01	0.023	--	0.520	0.543
2	0.51	9.40	0.467	0.138	0.710	1.404	12.12	"	0.066	0.103	0.168	0.337
3	1.00	10.40	0.532	0.291	0.710	1.427	13.36	"	0.063	0.140	0.153	0.356
4	1.24	10.75	0.554	0.469	0.702	1.475	13.95	"	0.055	0.181	0.126	0.362
5	1.53	10.80	0.551	0.485	0.705	1.476	14.02	"	0.054	0.170	0.129	0.352
6	1.72	10.90	0.547	0.529	0.700	1.481	14.15	"	0.069	0.246	0.128	0.443
7	1.94	11.70	0.608	0.628	0.710	1.489	15.14	"	0.061	o.246	0.122	0.429
8	2.46	11.80	0.580	0.739	0.745	1.522	15.39	"	0.058	0.301	0.122	0.481

Table 2.

No.	Lattice spacing a		Mismatch ε %	particle radius r Å	Long range order parameter S	Volume fraction of γ' %	Specific gravity	Shear modulus G N/mm^2
	γ'	γ						
1	3.5906	3.5634	0.76	84.5	0.79	12.56	8.22	81.7×10^3
2	3.5933	3.5666	0.75	85.5	0.82	12.98	8.22	82.4×10^3
3	3.5951	3.5671	0.78	98.5	0.85	14.18	8.22	82.8×10^3
4	3.5965	3.5674	0.81	101.5	0.86	14.85	8.29	83.8×10^3
5	3.5967	3.5682	0.80	98.5	0.87	14.84	8.24	83.8×10^3
6	3.5970	3.5701	0.75	98.5*	0.92	14.95	8.25	84.2×10^3
7	3.5976	3.5701	0.77	105	0.95	15.91	8.24	--
8	3.5987	3.5713	0.77	113	1.01	16.21	8.27	85.1×10^3

* Estimated by interpolation.

3. DISCUSSION

From Fig. 1 it is evident that in all alloys with various Nb content, niobium dissolves in solid solution, γ' and carbide with about the same proportion: 5:3:1. Thus, it is naturally to assume that niobium strengthens the alloy essentially through strengthening the solid solution and γ' phase.

3.1 Lattice Distortion of γ Matrix and Solid Solution Strengthening of γ by Niobium

Atom diameter of Nb differs 18% from that of Ni, the large niobium atom, when dissolving (as solute) into γ solid solution, should cause lattice distortion of the matrix. According to the work of Pelloux (2), for every 0.01 KX increment of lattice spacing, the increment of yield strength is ΔYS=56 N/mm^2, since the electron vacancy number of niobium is 5.66. Thus, from the measured Δa values caused by various niobium content in solution, the increment of the yield strength may be roughly estimated as shown in table 3. It is evident that the solid solution strengthening by niobium contributes about 1/3-1/2 increment of yield strength of the alloys.

3.2 Strengthening of γ' Phase by Niobium

As was shown by phase analysis, about 30% of niobium content in alloy dissolves in γ', however, the volume fraction of γ' in the alloys is only 10%, the concentration of niobium in γ' phase is about 3 times of that in matrix. Since Nb in γ' occupies the position of Al and Ti, it replaces Al and Ti in γ', the replaced Al and Ti form again new γ'. Furthermore, Nb in γ matrix lowers the solubility of Al and Ti, and increases further the volume fraction of γ' phase.

Using the measured specific gravity it is not difficult to estimate the volume fraction of γ' in the alloy (see table 2), the result shows that the volume fraction of γ' in the alloys increases with niobium content of the alloys.

Based on dislocation theory, at the presence of dispersed second phase the gliding of moving dislocations can be blocked and the alloy is strengthened. There are two mechanisms for dislocations to get through the second phase: the Orowan bowing mechanism and shear-through mechanism. Since superlattice dislocation pairs were observed in thin foils under TEM examination, it seems that the dislocations should shear through the γ'.

3.2.1 The contribution of niobium to the coherent strain field is rather small It is evident from table 2 that niobium simultaneous by increases the lattice spacing of γ and γ', with the increase of niobium content in the alloys, calculated mismatch of γ $-\gamma'$ phase varies within the range of 0.75 -0.80%, which is basically at the same level of the alloy without niobium addition.

Brown had proposed the following formula for estimating the yield strength (increment) caused by (mismatch) strain field (3):

$$\tau \cong 0.7 G f^{1/2} (|\varepsilon| b^3 / r^3)^{1/4}; \quad b/4|\varepsilon| < r < 2b/3|\varepsilon|$$

where b is the length of Burgers vector (b=2.52Å);
f is the volume fraction of the γ' particle;
r is the average radius of the γ' particle;
ε is γ $-\gamma'$ mismatch; G is the shear modulus.

The results of calculation show that the yield strength (increment) due to coherent strain is large but in the same level for all alloys with and without niobium. Thus, the addition of niobium does not raise the strengthening by coherent strain.

3.2.2 Niobium has the effect of raising the APB energy (antiphase boundary energy) The shear-through of order precipitate by dislocations during deformation of the alloy changes the initial order arrangement and causes the formation of a new interface, on which the most stable atomic arrangement had been shifted. Thus, the formation of APB during shearing accompanies with increase of energy and higher applied stress is required for the dislocations to get through, this means alloy strengthening by anti-phase boundaries (of the γ').

The APB energy for the (111) slip plane is given by

$$\gamma_0 = 1.41 k T_c S^2 / a^2$$

where k - Boltzmann constant = $1.38 \times 10^{-23} J/^{o}K$;
a - lattice parameter;
S - long range order parameter;
T_c - critical ordering temperature.

The value of long range order parameter appeared in the table 2 and Fig. 1 shows that the long range order of γ' increases with increasing niobium content of the alloys until the full order is reached. R.W. Guard had pointed out that at least under $1000^{o}C$ Ni_3Al remains ordered (4). However γ' in the alloys dissolves in temperature range $980^{o}C$-$1000^{o}C$. In order to calculate APB energy T_c can be taken to be about $1000^{o}C$. The calculated values of APB energy of γ' for the experimental alloys were shown in table 3. It is evident that APB energy increases with increasing Nb content.

Based on experimental observation on the motion of superlattice dislocation pairs, a theory on the interaction between the dislocation pair and the coherent, stress-free, ordered particles had been proposed by Gleiter (5,6), in which the yield strength of the alloy is given by

$$\tau = A \gamma_0^{3/2} r_s^{1/2} f^{1/2} T^{-1/2} b^{-1} - f \gamma_0 / 2b$$

Where A - a constant (generally equal to 0.5);
γ_0 - APB energy;
r_s - mean particle radius cut by dislocation, $r_s = (2/3)^{1/2} r_0$ (where r_0 is the mean particle radius);
T - line tension of dislocation; $T \doteq \frac{1}{2} G b^2$.

The yield strength increments caused by the addition of Nb were thus calculated (table 3). It appears that Nb strengthens the alloys by raising the APB energy of the γ' particle, and the effect increases with increasing Nb content.

In the above discussion, the contributions of niobium to the yield strength increment of the alloys

Table 3

No.	R.T.strength (mean)			calculated value			
	σ_b N/mm^2	$\sigma_{0.2}$ N/mm^2	$\Delta\sigma_{0.2}$ N/mm^2	Y.S.increment of γ matrix by Nb $\Delta\tau_1$ N/mm^2	APB energy J/m^2	Y.S. increment by APB $\Delta\tau_2$ N/mm^2	$\Delta\tau_1+\Delta\tau_2$ N/mm^2
1	1062	743	--	--	119.9×10^{-3}	--	--
2	1175	783	40	18	129.0×10^{-3}	4	22
3	1206	784	41	21	138.5×10^{-3}	12	33
4	1203	816	73	23	141.6×10^{-3}	14	37
5	1216	816	73	26	144.9×10^{-3}	14	40
6	1229	830	87	37	162.0×10^{-3}	23	60
7	1242	834	91	37	172.7×10^{-3}	31	68
8	1268	826	83	44	191.3×10^{-3}	46	90

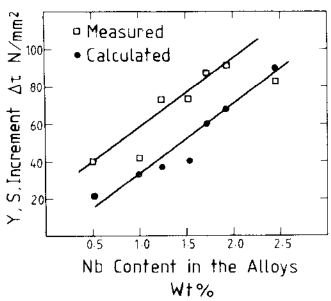

Fig. 3 Comparison of measured and calculated Y.S. increment

by various strengthening mechanisms have been estimated theoretically. Of course, the strengthening mechanism of superalloy is very complicated, there are some other strengthening mechanisms, for instance, grain boundary strengthening, that have not been considered. Rather good agreement between the trends of the theoretical and experimental values can be observed (Fig. 3) when superposition of γ' strengthening and solid solution strengthening by Nb is made.

4. CONCLUSIONS

1. About 50% of the Nb content in the alloys dissolves in γ matrix and 30% dissolves in γ', thus it increases the γ and γ' lattice parameter and the volume fraction of γ' and long range order parameter of γ' increases with increasing Nb content.

2. The essential strengthening mechanism of niobium is solid solution hardening and γ' hardening

by increasing its long range order parameter. In the strengthening mechanism by coherent strain field Nb does not play a major role.

It is interesting to show that the calculated yield strength increments by the superposition of solid solution hardening and APB hardening of γ' are in rather good agreement with the measured values.

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