INFLUENCE OF HEAT TREATMENT PROCESSES ON MECHANICAL PROPERTIES OF A Cu-Ni-Al-Ti ALLOY①

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ABSTRACT

The influence of heat treatment processes on the strength, hardness and fatigue crack propagating rate of a Cu-Ni-Al-Ti alloy was studied. The results show that the cooling rate after solution heat treatment has great influence on the properties of the alloy; namely, the larger the cooling rate, the higher the strength and the lower the plasticity. But the alloy can retain relatively high strength even under air-cooling. With regard to the strength, it is advisable to adopt the following process: solution heat treatment + water cooling + cold deformation + aging. However, with regard to the combination of the strength, plasticity and fatigue crack propagating rate, it is advisable to adopt the following process: solution heat treatment + air-cooling + cold deformation + aging.

Key words: Cu-Ni-Al-Ti alloy heat treatment strength plasticity fatigue crack propagating rate cold deformation

1 INTRODUCTION

The Cu-Ni-Al alloys are typical aging hardening alloys based on Spinodal decomposition, which can obtain very high strength and elastic properties through proper heat treatment. However, these alloys are prone to take place discontinuous precipitation, which will affect their plasticity and toughness. Adding proper amounts of titanium into the Cu-Ni-Al alloys can inhibit the discontinuous precipitation, thereby improve the properties of the alloys[1,2], and make them the good substitute materials for the beryllium bronze. Our recent work[3] demonstrated that the Cu-Ni-Al-Ti alloys added with minor amounts of other alloying elements have excellent high temperature properties and are especially suitable for manufacturing high temperature elastic components.

Heat treatment can greatly change the properties of the Cu-Ni-Al-Ti alloys. 中山丰[4] *et al* studied the influence of multi-stage aging on the properties of the Cu-7Ni-1Al alloy, and pointed out that the multistage aging can not only inhibit the discontinuous precipitation, but also improve the mechanical properties of the alloy. They also studied the fatigue properties of the Cu-7Ni-1Al alloy[5]. However, up till now, few works have reported the influence of heat treatment on properties of high nickel, high alumium Cu-Ni-Al-Ti alloys. The purpose of this paper is to study the influence of heat treatment processes on the properties of a high nickel, high alumium Cu-Ni-Al-Ti alloy so as to find the suitable heat treatment processes which can ensure its good comprehensive properties, thus provide theoretical foundation for its wide use.

2 EXPERIMENTAL

2.1 *Determination of Heat Treatment Processes*

In selecting heat treatment processes, first, we took into account of the influence of the cooling

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rate after solution heat treatment on the properties. For this we adopted water-, oil- and air-cooling in the experiment. Second, we took into account of aging and pre-aging cold deformation on the properties. For this we adopted such processes as quenching + direct aging and quenching + cold working + aging. In view of the above considerations, we studied the following heat treatment processes:

1# —Solution heat treatment + water-cooling.

2# —Solution heat treatment + oil-cooling.

3# —Solution heat treatment + air-cooling.

4# —Solution heat treatment + water-cooling + 500 ℃, 2 h aging.

5# —Solution heat treatment + oil-cooling + 550 ℃, 2 h aging.

6# —Solution heat treatment + water-cooling + 50% cold deformation + aging.

7# —Solution heat treatment + oil-cooling + 50% cold deformation + aging.

8# —Solution heat treatment + air-cooling + 50% cold deformation + aging.

9# —Solution heat treatment + oil-cooling + 650 ℃ aging + cold deformation + 550 ℃ aging.

It should be noted that all samples for determination of fatigue crack propagating rates were oil-cooled or air-cooled.

2.2 *Experimental Method*

The composition of the studied alloy is as follows: Ni = 12.7%, Al = 2.48%, titanium and other alloying elements < 1.5%, Cu balance.

Commercially pure copper, nickel and aluminium were used as raw materials to prepare the alloy. The alloy was melted in a medium frequency induction furnace in atmosphere and cast in iron moulds. After homogenization, hot rolling, cold rolling and heat treatment having been conducted for the ingot castings, the samples were prepared.

The hardness data were measured by a HVA-10A type Vicker's tester (load = 9.8 N, holding time = 15 s), and an average of ten measurements was taken as the final hardness for each sample.

The tensile tests were conducted on an Instron-8032 type materials tester. The fatigue crack propagating rates were measured according to GB6398-86 using CCT samples. After the prefabrication of cracks of 0.2 mm in length on both sides of the slot had been completed on the Instron-8032 type materials tester (conditions: triangle wave, cycling frequency = 10 Hz, σ_{max} = 290 MPa, σ_{min} = 87 MPa), the crack propagating rates were measured with a 20-fold reading microscope to determine their length (σ_{max} = 240 MPa, σ_{min} = 73 MPa, R = 0.303). If $\triangle a \leqslant 0.6$ mm after a certain number of cycling, the tester was stopped for observation, but the interval was not more than 2 min. Three samples of each treatment state were used for measurement.

3 RESULTS AND DISCUSSION

3.1 *Mechanical Properties*

Table 1 lists the HV data of the samples aged at 550 ℃ for different times.

Table 1 HV data of samples aged at 550 ℃ for different times

State before aging	aging time/h							
	0	0.5	1	2	4	8	16	24
SA+AC	263	322	328	317	317	311	317	296
SA+WC	211	313	315	340	325	283	275	302
SA+AC +50%CD	287	333	338	339	340	338	324	334
SA+WC +50%CD	266	317	337	341	357	357	331	318
SA+WC +650 ℃, 0.5h +50%CD	308	365	352	334	322	330	333	340

* SA—solution heat treatment; AC—air-cooling; WC—water-cooling; CD—cold deformation.

It is clear from Table 1 that the hardness of the alloy is very stable; previous work[3] also showed that the alloy decreases very little in hardness after 128 h aging at 500 ℃ and correspondingly the strength of the alloy is also stable, which means that the alloy is an ideal material for making high temperature elastic components. In addition, preaging before cold deformation can contribute very little to the improvement of hardness.

Table 2 shows the tensile test results of the alloy. It can be seen from Table 2 that the cold deformation before aging can greatly raise the strength, but reduce the plasticity obviously, which is due to that the cold deformation can increase the dislocation density, then produce cold working hardening. Morever, the densified dislocations can promote the sequent precipitation. Thus, the combination of these two effects raises the strength and reduces the plasticity. Meanwhile, the cold deformation before aging can raise the yield ratio ($\sigma_{0.2}/\sigma_b$) of the alloy, which is beneficial to the elastic alloys.

The cooling rate after solution heat treatment

Table 2 Room temperature tensile properties under different heat treatment states

Sample number	$\sigma_{0.01}$/MPa	$\sigma_{0.2}$/MPa	σ_b/MPa	δ/%
2	302.0	318.7	650.2	25.10
5	390.2	517.8	739.4	5.62
6	643.3	1218.9	1256.2	1.17
7	576.6	917.9	1016.9	3.47
8	536.0	823.8	937.5	7.02
9	545.2	919.9	1026.8	3.15

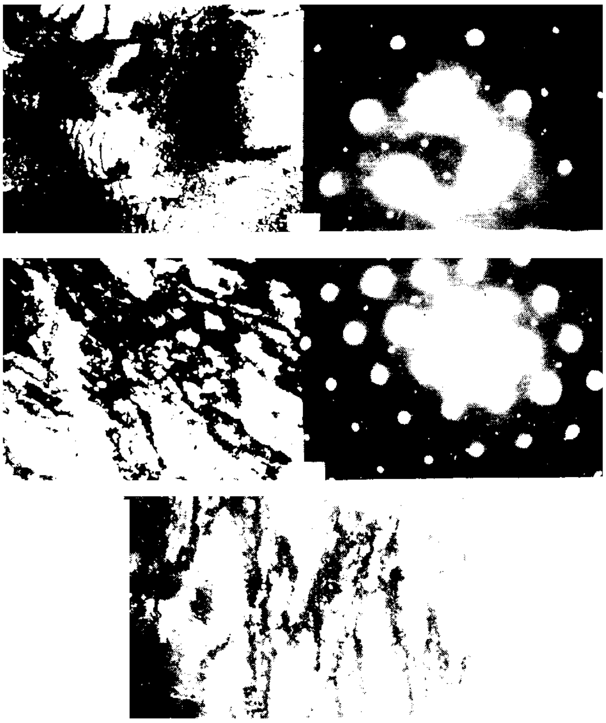

Fig. 1 TEM morphologies and diffraction patterns
(a) Sample No. 2; (b) Sample No. 7; (c) Sample No. 8

has great influence on the strength of the alloy as mentioned above, precipitation rate of the alloy is fast. It can be seen from Fig. 1 (a) that there obviously occured granular γ' phase in the oil-cooled samples. It can also be confirmed that there must be more and coaser granular γ' phase in the air-cooled samples. The self-strengthening effects of those second phase particles precipitated at the high temperatures are small. Furthermore, the precipitation will reduce the supersaturability of the alloy, then reduce the volume ratio and dispersity of the strengthening phases after aging; thereby the total strengthening effect will weaken. Table 2 also indicates that the pre-aging before the cold working has little effect on the strength. The reason for this may be improper processes; but needs to be further studied.

3.2 *Fatigue Crack Propagating Rate*

Using the seven-point progressive increase polynominal method to process the experimental data groups(a, N) with a computer, the double logarithmic curves $\lg(da/dN) - \lg\triangle K$ in different states can be obtained. From these curves the values of c, m of the material can be calculated, as shown in Table 3.

Table 3 Calculated crack propagating parameters of samples in different states

Parameter	Sample No.				
	2	5	7	8	9
C	2.257×10^{-10}	2.089×10^{-10}	3.890×10^{-13}	8.909×10^{-11}	3.82×10^{-13}
m	3.08	3.43	5.75	3.80	5.85
R	0.303	0.303	0.303	0.303	0.303

Fig. 2 (a), (b), (c) respectively show the $\lg \frac{da}{dN} - \lg\triangle K$ curves of samples No. 2, 7 and 8.

According to Paris formula: $da/dN = C(\triangle K)^m$ and Table 3, the crack propagating rates of the samples in different states at the second stage can be experessed by the following equations:

for No. 2: $\frac{da}{dN} = 2.257 \times (\triangle K)^{3.08} \times 10^{-10}$

for No. 5: $\frac{da}{dN} = 2.089 \times (\triangle K)^{3.43} \times 10^{-10}$

for No. 7: $\frac{da}{dN} = 3.890 \times (\triangle K)^{5.75} \times 10^{-13}$

for No. 8: $\frac{da}{dN} = 8.909 \times (\triangle K)^{3.80} \times 10^{-11}$

for No. 9: $\frac{da}{dN} = 3.82 \times (\triangle K)^{5.85} \times 10^{-13}$

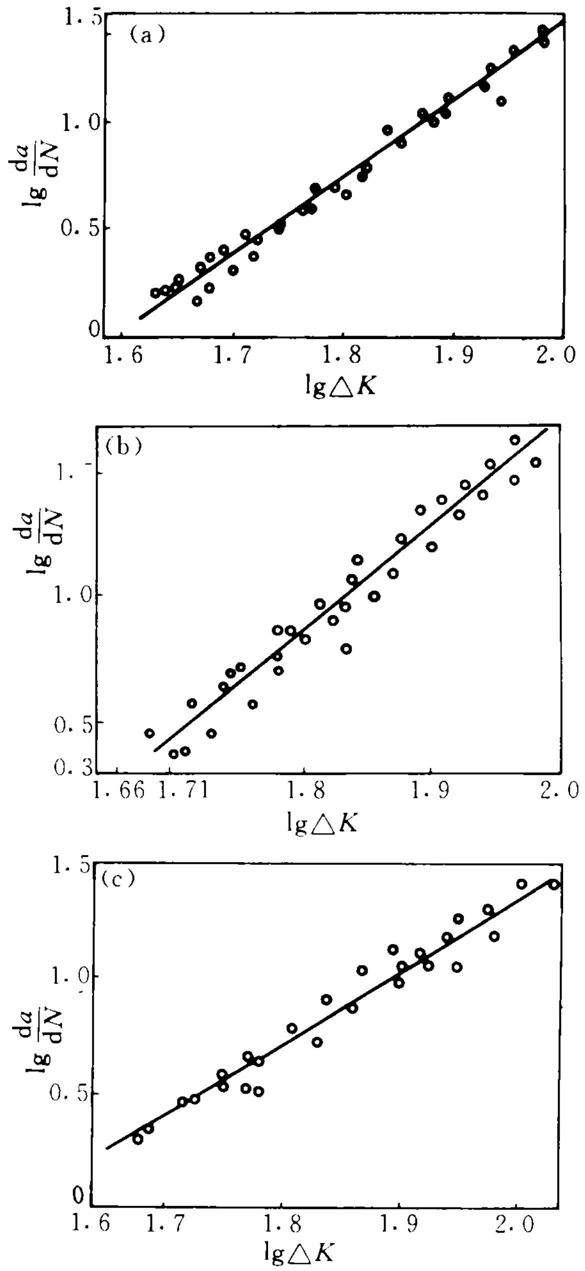

Fig. 2 $\lg \frac{da}{dN} - \lg\triangle K$ curves

(a) No. 2; (b) No. 7; (c) No. 8

The experimental results show that the crack propagating rates of samples No. 7 and No. 9 rank the fastest, that of Sample No. 2 ranks the slowest, and that of sample No. 8 ranks between them. By combining this fact with the data in Table 2, it can be seen that with regard to the comprehensive properties of strength, plasticity and fatigue crack propagating rate, it is advisable to adopt the following heat treatment process: solution heat treatment + air-cooling + cold working + aging (Sample No. 8).

The influence of the microstructure on the crack propagating rate is very complicated and can not be explaned using a simple mode.

Fig. 1 shows the TEM morphologies and diffraction patterns of fatigue damage samples in three different states, from which it can be seen that the dislocation configuration of sample No. 2 is typical of resident slip bands structure, that of sample No. 7 (similar to sample No. 9) also presents band distribution but with branches connected. Furthermore, the dislocation density of sample No. 7 ranks the highest; that of sample No. 8 ranks the next; and that of sample No. 2 ranks the lowest. Due to the band distribution of the dislocations, there form "double phase" structures, namely dislocations densified and dislocation-poor zones. With increasing dislocation bands, the mechanical properties of the grains tend to be unstable. For the Cu-Ni-Al-Ti alloy, it precipitates LI_2 type coherent ordered γ' phase[2]. According to the viewpoint of Calabress *et al* [6], the reciprocating movement of the dislocations in the slip bands makes the ordered structure to be disordered, thus weakens or even eliminates the strengthening effect caused by the reverse domains, consequently reduces the mechanical properties.

The difference in crack propagating rates may be related to the boundary precipitates free zones. TEM observations for samples No. 7 and No. 9 indicate that there existed wide PFZs, but no PFZs were found in other samples. PFZs are weak bands. Under the actions of reciprocating stresses, the dislocation pile-ups at the boundaries may bring about micro-cracks.

The fatigue properties of Cu-9Ni-6Sn alloy were studied and it was found that this alloy has longer fatigue life when solution heat treated in the double phase range and then cold worked and aged than when solution heat treated in the single phase range. This may be due to the fact that there exist part of equilibrium second phase particles and part of the cold worked structures, thus the microstructure is more complex[7]. Air-cooled sample No. 8 should have similar effects. For sample No. 8 there precipitates part of γ' phase particles during air-cooling and a large amount of fine and homogeneous γ' phase particles. This complex microstructure may be one of the reasons why the crack propagating rate of sample No. 8 is lower than those of samples No. 7 and No. 9.

4 CONCLUSIONS

(1) The cooling rate after solution heat treatment has great influence on the properties of the studied Cu-Ni-Al-Ti alloy; namely, the larger the cooling rate, the higher the strength and the lower the plasticity. But this alloy can retain relatively high strength even under air-cooling.

(2) With regard to the strength, it is advisable to adopt the following process for the studied alloy: solution heat treatment + water cooling + cold deformation + aging. However, with regard to the combination of the strength, the plasticity and the fatigue crack propagating rate, it is advisable to adopt the following process: solution heat treatment + air-cooling + cold deformation + aging.

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