

## ROLE OF ALUMINIUM CONCENTRATION ON THE STRUCTURE BEHAVIOUR OF Cu-ALLOYS

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### Abstract

*Effect of Al (5, 10 and 20) wt % on the structure behaviour of Cu-Al alloys was investigated by both microhardness measurements and optical microscopic investigations. Pure Cu was used for comparison. The analysis of the hardness - temperature curve shows a type of dependence which obeys an empirical exponential law, but consists of several distinguishable regions. For Cu-20% Al, one region is observed, and two regions for pure Cu, while for alloys of 5 and 10% Al concentration three regions were observed. The activation energy varies between 0.03 and 0.9 eV for these regions, depending on Al-concentration as well as the annealing temperature. The recrystallization temperature was found to increase with increasing Al-concentration. It was also observed that Cu-20% Al alloy is very hard and brittle owing to the formation of  $\gamma_2$  phase (as indicated from the phase diagram) and also to phase separation as being detected from optical microscopic investigations.*

### Introduction

Copper aluminium alloys have increasing applications in many industrial purposes. The proper selection of heat treatment parameter will produce material with significant mechanical properties, often comparable with those of low carbon steels. The mechanism of anneal hardening in dilute Cu-alloys was reported by many authors for alloys annealed below 623 K [1-6].

The phase diagram of Cu-Al bronze is shown in Fig. (1). The

Cu-Al system exhibits an eutectoid reaction.  $T_E$  and  $C_E$  are the eutectoid temperature and eutectoid composition, respectively. All the phases involved in the eutectoid reaction are solid, either solid solution or intermediate phases (in the case of Cu-Al  $\beta$  and  $\gamma$  are intermediate phases and  $\alpha$  is a solid solution).

Mechanical behaviour of crystalline materials, yield stress, hardness, flow stress, as a function of temperature was formulated by Borisenko and Krashchenko [7]. It was shown that in order to describe the mechanical behaviour of crystalline materials in a broad temperature range one can use an expression of the type

$$\alpha = \sqrt[n]{\beta E' G^P T \exp(U/kT)} \dots \dots \dots (1)$$

for the temperature dependence of proof stress, flow stress, and hardness [8-10] where  $E'$  is the plastic strain rate,  $T$  stands for the absolute temperature,

$k$  is the Boltzman constant,

$U$  is the activation energy,

$G$  is the shear modulus,

$\beta$  is a function of material parameter, which contains the entropy term  $\exp(-S/k)$ ,

$P$  is a numerical constant equal to 1 or 2 [11], and

$n$  is a constant equal to 3 in the particular case in consideration.

The activation energy of plastic deformation was determined by using the formula

$$U = 3K \left[ \frac{\delta \ln H}{\delta \ln 1/T} \right] \quad (2)$$

The aim of the present work is to determine the effect of Al on the structure behaviour of Cu-Al alloys from microhardness measurement, and metallographic investigations.

### Experimental Procedure :

Identified microstructures, mounted in polyester resin of diameter

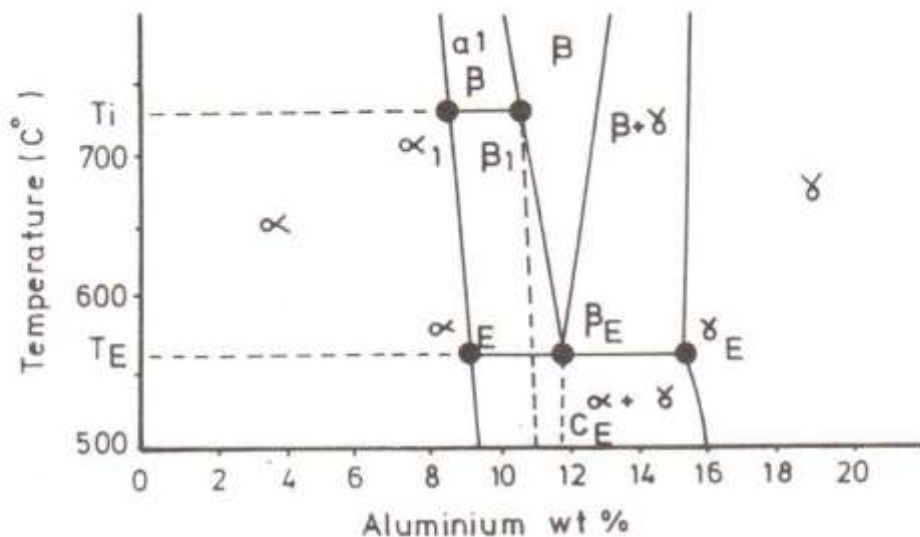


Fig. (1): Part of the equilibrium phase diagram of the Cu-Al system.

8 mm were used. Alloys of 5, 10 and 20% Al concentration were supplied by Griffin, England, together with a cast copper. The chemical composition of these materials is shown in Table (1).

Specimens were annealed in special designed oven at temperature ranging from (323 to 973 k) for 2 hours. The oven temperature was automatically controlled and measured to an accuracy of  $\pm 2^\circ\text{C}$ .

Microhardness measurements were carried out by using a Shimadzu microhardness tester (Japan), with an applied load of 50 g for 5 seconds. Sometimes a load of 100 g was used. More than 10 readings were taken for each specimen and standard deviation was calculated.

The etchant solution consists of 120 ml distilled water with 10 g amonical amonica copper chloride. The specimens were carefully dried and optical microscope type Leitz Laborlux 12 was used for investigation.

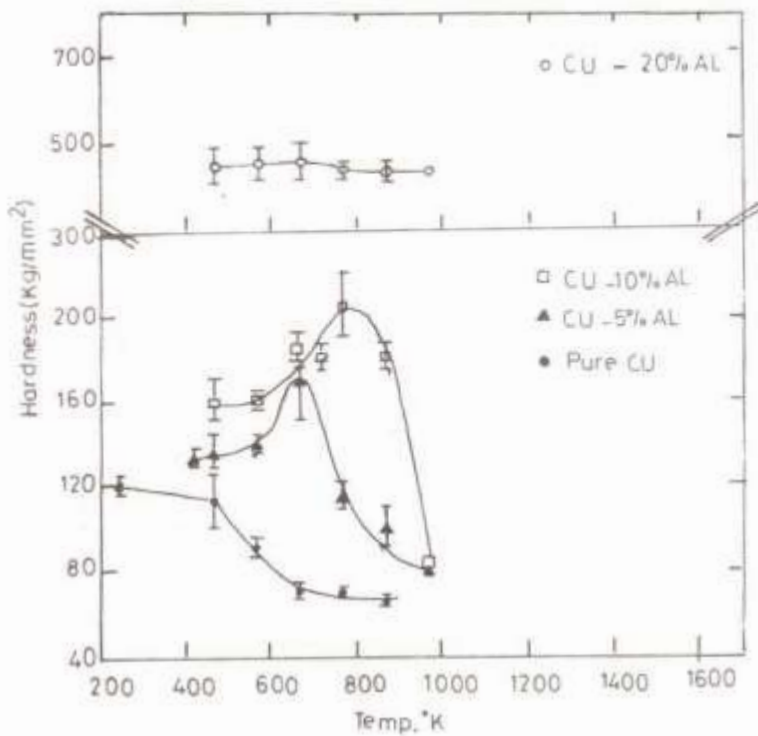


Fig. (2): Isochronal annealing of Al-bronze and Copper.

**Table (1)**  
Chemical Composition of the materials used

Material	Weight Wt %							
	Cu	Fe	Si	Mg	Mn	Al	Pb	Ni
Cu-5 wt% Al	balance	0.5	<0.003	0.00066	<0.003	5	0.082	0.18
Cu-10 wt% Al	"	0.5	<0.003	<0.003	0.5	10	0.1	0.8
Cu-20 wt% Al	"	0.5	<0.003	<0.003	1.5	20	<0.002	1.2

**Results :**

Fig. (2) represents the isochronal annealing results of the different samples studied. It indicates that :

- a) The V. H. of pure Cu starts decreasing at temperatures higher than 473 K, reaching to approximately constant values over the range 673-873 K.
- b) On the other hand, the V.H. of Al-bronze alloys starts increasing at temperatures higher than 573K, and progressively increases up to a maximum value depending on the Al-content, beyond which it decreases again with increasing temperature.
- c) The V.H. values of Cu-20% Al are nearly the same over the whole temperature range studied.

Optical microscopic investigations of pure Cu annealed at 473 k (plate A) indicated that recrystallization started at this temperature, however complete recrystallization was achieved at 673 K (Plate B).

On the other hand, in the 5% Al-bronze, nucleation started at 573K (Plate C), while complete recrystallization was achieved at 773 K (Plate D).

In the case of 10% Al-bronze, relatively higher temperatures were needed to start recrystallization (Plate E). Phase separation between Cu and Al was observed in 20 % Al-bronze annealed at 773 K (Plate F).

Analysis of the hardness - temperature curve according to Borisenko and Krashchenko [7] is represented in Fig. 3. In this figure, one can observe a type of dependence which obeys this exponential law (Eq. 2), but it consists of several distinguishable regions. For 20% Al-bronze one region is observed. At the same time pure Cu is characterized by two regions, and the other Al-bronze concentrations are characterized by three regions.

These different characterizing regions are reflected on the relation between the activation energy versus  $(1/T)$ , calculated according to



Plate A: The recrystallized grains of cast pure copper at 473 K.

Plate B: The recrystallized grains of cast pure copper at 763 K.



Plate C: The recrystallized grains of Cu-5 wt% Al at 573 K.

Plate D: The recrystallized grains of Cu-5 wt% Al at 773 K.



Plate E: The recrystallized grains of Cu-10 wt% Al at 873 K.

Plate F: Phase separation of Cu-20 wt% Al

Eq. (2) and represented in Fig. 4, together with the previously determined values of Borisenko [12].

### Discussion

The analysis represented in Fig. 4 for Al-bronze alloys indicates a temperature dependence of the activation energy over the temperature range 323-923 K. According to Ashby [11], for moderate strain rates and applied stresses of  $\geq 10^{-4}G$ , glide mechanism is dominating under  $0.5 T_m$ , while, dislocation creep mechanism is dominating at higher temperatures. The instigation of b.c.c. metals [12,13] and gold-copper-silver alloy [14] shows the activation energy as being determined from proof stress, flow stress, and hardness data for different temperature regions, has nearly the same value. Also Borisenko et al. [12], found that the temperature dependence of the activation energy of plastic deformation for Cu, Ag, and Au, are nearly the same as the well known dependence of Al, and crystalline NaCl [13-15], as well as for tungsten and molybdenum [9, 10]. It corresponds to the activation energy spectra of the elementary atomic rebuilding in metals [16]. This implies that the activation energy is independent upon deformation, and agrees with our results shown in Fig. 4.

There may be several explanations for the behaviour of commercially pure metals in the moderate temperature range ( $0.5 T_m$ ). It was suggested that under  $0.5 T_m$ , the plastic flow stress in close-packed lattice metals could be associated with impeded displacement of impurity atoms atmospheres interacting with dislocations [17]. Other possible explanations of the change of flow stress in the moderate temperature region are associated with the part played by vacancies [18], and interstitial atoms [19] formed in the course of plastic deformation.

The third region observed in Cu-5% Al, and Cu-10% Al, which disappeared in cu-20% Al alloys (see Fig. 4), may be attributed to the

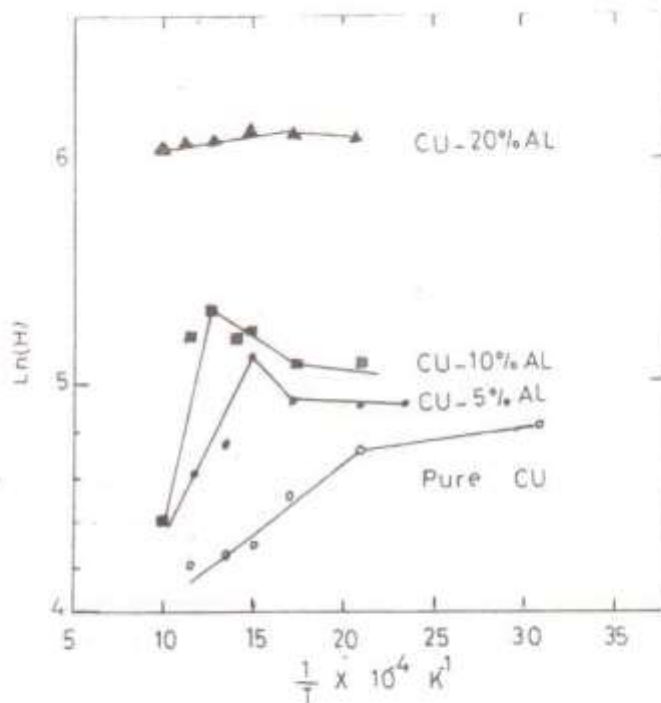


Fig. (3): The relationship between  $\ln(H)$  and  $1/T$  for Al-bronze and pure copper.

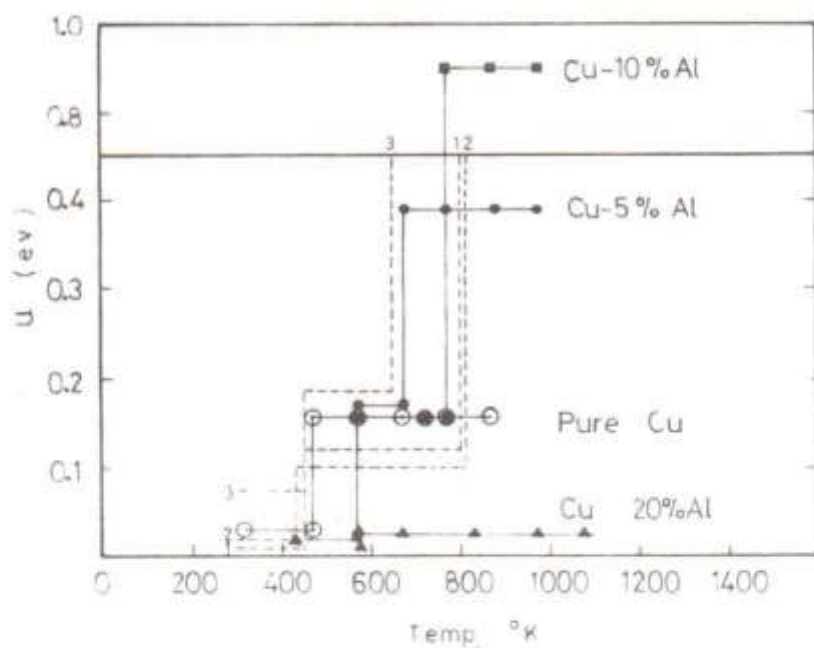


Fig. (4): The relationship between the activation energy ( $U$ .eV) and absolute temperature (Temp. K) for Al-bronze and pure copper.



difference in nucleation at temperatures higher than 673 K. The observed hardening effect in Al-bronze alloys is due to anneal hardening, however, the mechanism responsible for this hardening effect is incompletely understood [2], although some previous [3-5] observations of this effect had interpreted it as an indication of the main responsibility of atomic ordering. However, in a detailed investigation of anneal hardening of Cu-Al alloys by Bader et al. [6], it was concluded that solute segregation to dislocation gives rise to the predominant hardening mechanism.

Further evidence for the effect of solute segregation in Cu-base solid solutions was provided by Durrsclinabel and Lederer [20] as well as by Nelmes and Wilshire [21], using creep data. Some property changes other than hardening response during annealing may be considered in support of solute segregation to dislocations, as it had been suggested for the major decrease in electrical resistivity on annealing Cu-Al alloys. Anneal hardening is not due to short range ordering, and so segregation to dislocation was considered to be the main response for this decrease.

Similarly, Opplewell and Crane [23] from DTA found that the exothermic heat of deformed Cu-Al alloys specimens were considerably higher than that expected from short range ordering [6]. Moreover, exothermic heat effects were found even in 4 and 6 at % Al specimens which should not show any significant ordering.

### Conclusions :

- 1) The increase in hardness for Cu-5% Al, and Cu-10% Al over the temperature range 623-773 K, may be due to solute segregation to dislocations.
- 2) Ordering effects are not a primary cause of anneal hardening, but may be partly contribute to it.
- 3) Anneal hardening may be considered as a genuine hardening

mechanism in analogy to other hardening mechanisms, such as work, grain size, solid solution, and/or dispersion hardening.

- 4) At temperatures higher than 700 K, for Cu-5% Al-bronze, nucleation was started and complete recrystallization was achieved at 773 K as indicated from the decrease of hardness in Fig. (2) and the optical observation in Plate D. However, for Cu-10% Al-bronze, complete recrystallization was not achieved, instead nucleation was observed at temperature 773 and 873 K plate E. This may be the main cause of the observed decrease in hardness at temperatures higher than 773 K.
- 5) For Cu-20% Al bronze, which appeared very hard and brittle, this may be due to the formation of  $\gamma_2$  phase [22], and partly to the phase separation observed in the microscopic investigation (Plate F).

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