Development of Microstructure in Silicon-Aluminum-Bronze

J. Iqbal[1], F. Ahmed[1] and F Hasan[1]

[1] *Metallurgical and Materials Engineering Department University of Engineering & Technology, Lahore, Pakistan*

Abstract

The aim of present study was to determine the sequence of micro-structural development during continuous cooling from temperatures approaching the solidus, in an as cast silicon- aluminum bronze alloy having composition Cu-6 % Al-2% Si-0.6% Fe. This alloy has good resistance to sea-water corrosion and low magnetic permeability. This alloy is used in marine and chemical applications where corrosion resistance is a prime requirement.

A study of the development of microstructure during continuous cooling from the solidus temperature showed that at high temperatures the alloy consisted of β and α phases in which the α-phase had a Widmanstatten morphology. As the temperature is lowered Fe_5Si_3 particles formed throughout the microstructure and this is followed by the formation of Fe_3Si_2 precipitates in the α-grains. On further cooling the β phase is transformed into $γ_2$.

Keywords: Microstructure; α-grains; β-phase; $γ_2$-regions; intermetallic particles

1. Introduction

Aluminum Bronzes have a wide range of such applications where exceptional corrosion resistance in aggressive environments is a primary requirement. The silicon-aluminum-bronze alloy offers an exceptional resistance to a large variety of corrosive agents and is thus widely used in marine and chemical environments.

The additions of silicon into binary aluminum bronze are primarily made to improve its machinability [1]. Silicon addition may be considered as a further addition of aluminum as it has a tendency to form β and causes some degree of hardening.

A vertical section of Cu-Al-Si ternary diagram [2] is shown in Figure 1. The as-cast microstructure of silicon aluminum bronze consists of α, $γ_2$, twin-like plates at α-$γ_2$ grain boundaries, and intermetallic particles based on Fe-Si [2,3]. However, the evolution of the microstructure during cooling as a result of the decomposition of the high temperature β-phase is poorly understood. Previous studies had concluded that either the β-phase was retained at room temperature [4,5], or decomposed to mixture of α + $γ_2$ phases [6,7] or α + κ phases [8].

2. Experimental

The composition of the alloy studied during present research is given in Table 1. The specimens, 15 mm x 10 mm x 1 mm, were cut from the as-cast alloy and suspended circumferentially around, and equi-distant from, a thermocouple located centrally in a vertical tube furnace. The specimens were heated to 965°C and kept at this temperature for 30 minutes. The furnace was then switched off and the temperature allowed to gradually fall. The thermocouple was connected to a chart-recorder which registered the drop of temperature with time, Figure 2. Specimens were taken from the furnace at regular intervals as the temperature fell and quenched into water.

For optical microscopy the samples were prepared by using standard techniques. The polished samples were etched in a solution of 10% ferric nitrate in water.

Table 1: Chemical composition of the alloy.

Cu wt %	Al wt %	Si wt %	Mn wt %	Fe wt %
90.98	6.04	2.32	0.06	0.6

Corresponding Author: F. Hasan (drfaiz@uet.edu.pk)

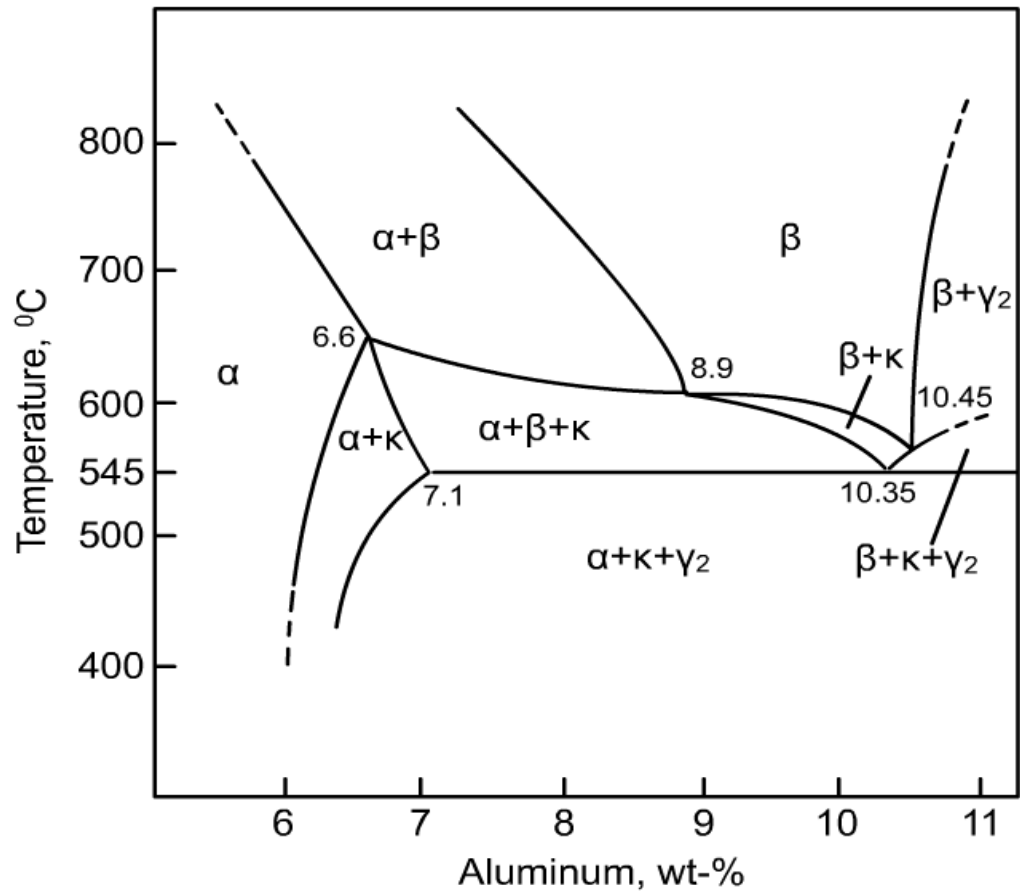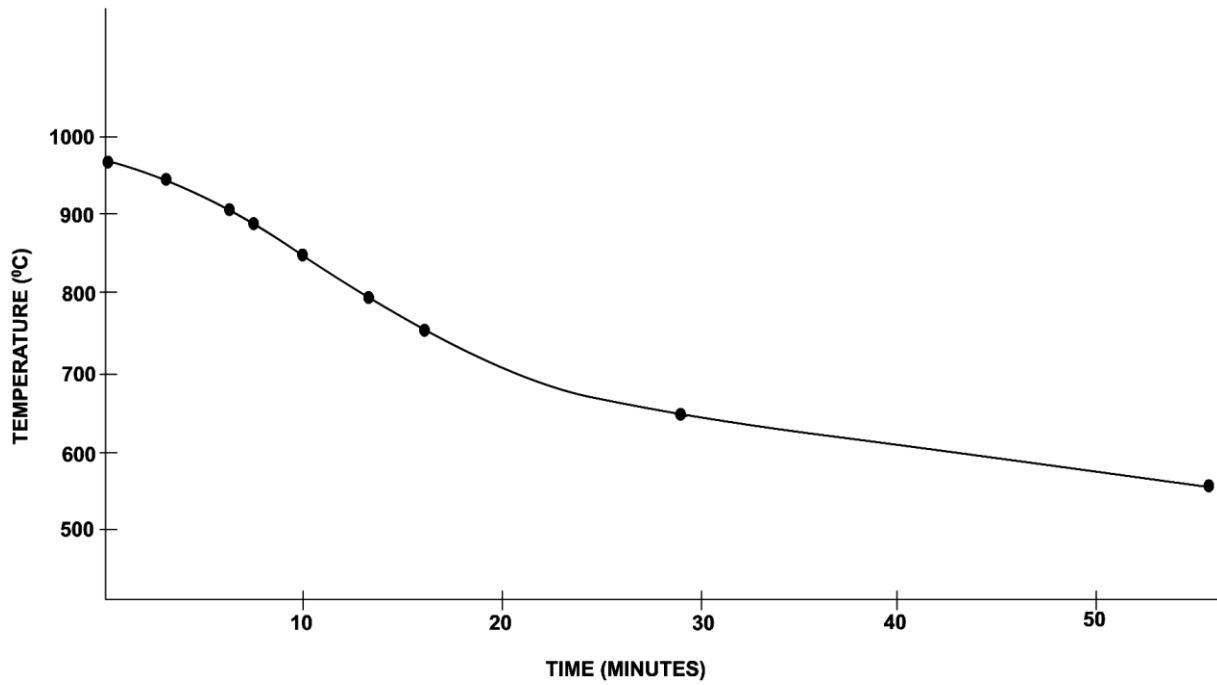**Figure1:** Vertical section of ternary diagram of Cu-Al-2 wt % Si [6].

Figure 2: Cooling curve recorded, showing cooling rate ~8^{0}C/min.

Table 2: Structure and composition of the phases.

Phases	Structure	Composition
α	FCC	Cu-rich (containing ~ 12%Al, 5%Si)
γ_2	CPH	Cu-rich (containing ~ 12%Al, 3%Si)
High-temp phase β	BCC	
Lath-shaped particles	B2	Fe-rich (containing ~38%Si)
Irregular-shaped particles	Hexagonal	Fe-rich (containing ~30%Si, 16%Cu)
Twin-like plates at α-γ_2 boundaries	CPH & FCC	Cu-rich (containing ~15%Al, 8%Si)

3. RESULTS

The microstructure of the as-cast silicon-aluminum-bronze (SAB) alloy is shown in Figure 3. It consists of dark etching areas of α-phase, light-etching areas of γ_2-phase and intermetallic precipitates of different morphologies. The inter-metallic precipitates, which are rich in iron, are of two types: irregularly-shaped particles and lath-shape particles. In addition, the formation of twin-like plates at α-γ_2 grain-boundaries is also observed. The details of these phases which have been published elsewhere [9] are summarized in Table 2.

The microstructures of the alloy quenched from various temperatures during slow cooling are shown in Figures (4-9). The cooling rate between 965ºC and 550ºC (the temperature range in which most of the microstructural changes occurred) was about 8ºC per minute, Figure 2. The microstructure of the sample quenched from 965ºC, Figure 4, consisted of a small amount of α, at the grain boundaries in mainly the β matrix. The β-phase has transformed to martensite upon quenching.

The next sample quenched from 940ºC (Figure 5), showed that the α-grains grew in size and the precipitates of Fe_5Si_3 began to appear at the α-β boundaries,. Further growth of α and nucleation of more Fe_5Si_3 particles continued with decreasing temperature (Figure 6).

As the temperature approached 790ºC, a high density of fine precipitates (Fe_3Si_2) were observed inside the α-grains, but there was a precipitate free zone along the grain-boundaries, Figure 7. It can also be seen from Figure 7, that all the entrapped β-regions between the α-grains started transforming to γ_2-phase.
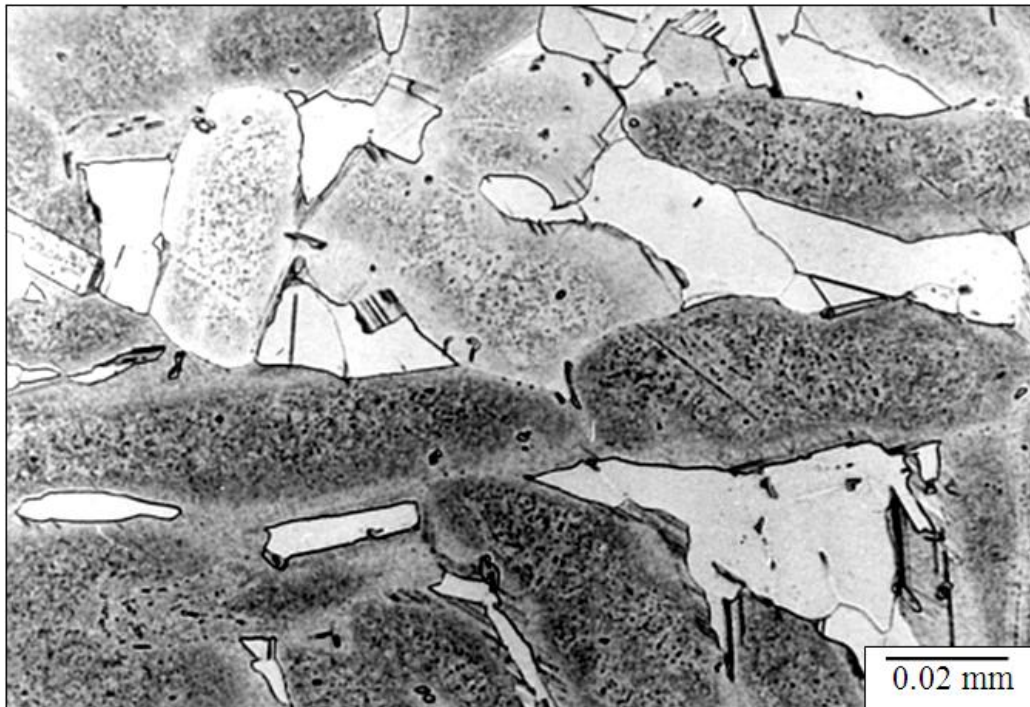

Figure 3: Microstructure of the as-cast alloy showing the α , γ_2 and intermetallic precipitates of different morphologies.
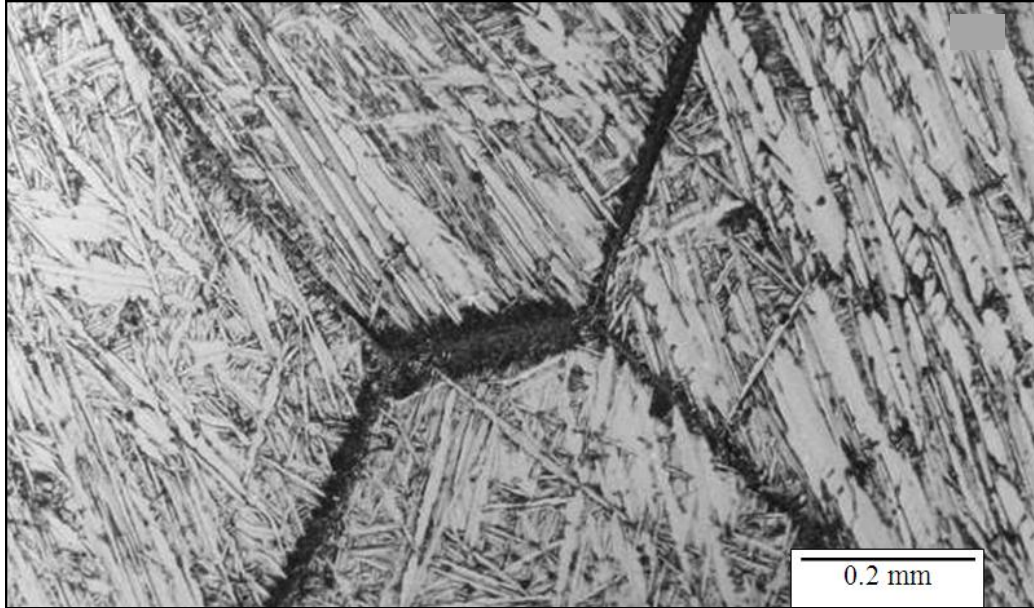
Figure 4: Microstructure of sample quenched from 965^{0}C, showing martensitic (β-phase) and some α-phase at the grain boundaries.

Figure 5: Microstructure showing the widmanstatten growth of α-phase in the β matrix.

The twin-like structure along the α-γ$_2$ boundaries was first observed in samples quenched from 650ºC, Figure 8. The volume fraction occupied by these regions increased with decreasing temperature. The microstructure of the sample quenched from 550ºC, Figure 9, was similar to that of the as-cast alloy, which indicated that no significant change in the microstructure took place below 550ºC.

4. Discussion

The sequence of microstructural development in the SAB alloys involves the precipitation of α-phase from β. As the α forms at very high temperature, its iron and silicon content are likely to be very high, and this consequently results in the observed precipitation of lath shaped precipitates at high temperatures. Moreover

the centers of α-grains, i.e., the first regions to form from β, contain the highest amount of iron and silicon in solution, and therefore during cooling are the first regions to reach their limit of solubility for iron. This variation in iron and silicon–content across the α-grains is reflected in the existence of precipitate-free zone, as shown in Figure 7. The peripheral regions of α, which are the last to form from β are richer in aluminum and silicon. With reference to phase diagram shown in Figure 1, it can be seen that the aluminum (and silicon) content of α gradually increases with decreasing temperature, while the maximum solubility occurs at 650°C.

Considering that the present alloy contains only 0.06 wt % iron and yet the iron-rich precipitates are formed in both α and β, indicates that the solubility of iron in both α and β is reduced due to the presence of silicon. In aluminum bronzes of AB1 [10] and AB2 [11] types, which do not contain silicon, iron concentrations as high as 2.5 wt % have been reported to present in the α-solid solution [10, 11].
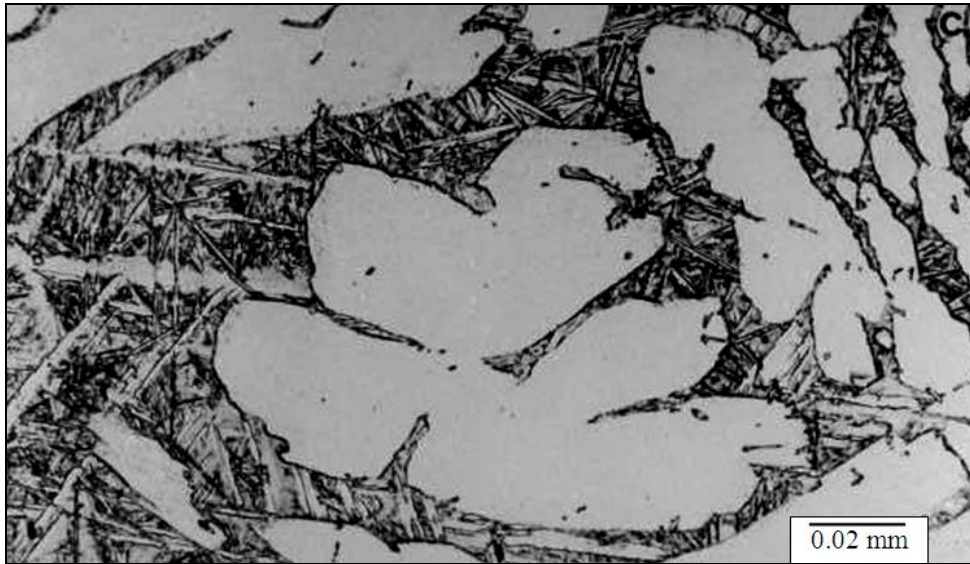

0.02 mm

Figure 6: Microstructure of the sample quenched from 940^{0}C, showing nucleation of irregular-shaped particles throughout the microstructure.
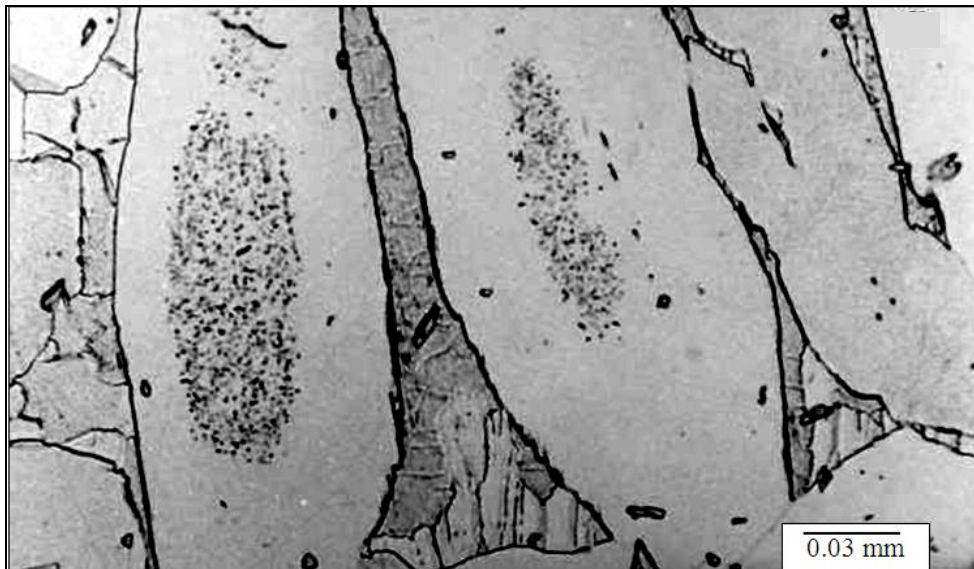

0.03 mm

Figure 7: Microstructure of the sample quenched from 790^{0}C, showing the high density of fine precipitates along with irregular-shaped particles in α-grains.

Figure 8: Microstructure of the sample quenched from 650^0C, showing the α, γ_2 and twin-like structure along α / γ_2 grain boundaries.

Figure 9: Microstructure of the sample quenched from 550^0C, showing a microstructure similar to the as-cast alloy.

The present experiments show that the β-phase gradually transforms into α–phase in the temperature range of 790-650 ℃. Samples quenched from temperatures greater than 650℃, showed the presence of a martensitic structure where the high temperature β-phase had pre-existed. The absence of a martensitic structure in samples quenched from temperatures less than 650℃ indicates that the β-phase remaining after the formation of the α-phase had completely transformed to γ_2 below this temperature. These results are consistent with those of Cairns et al [3].

The twin-like structure along α-γ_2 boundaries was first observed in samples quenched from 650℃. A schematic diagram of the development of microstructure is shown in Figure 10.

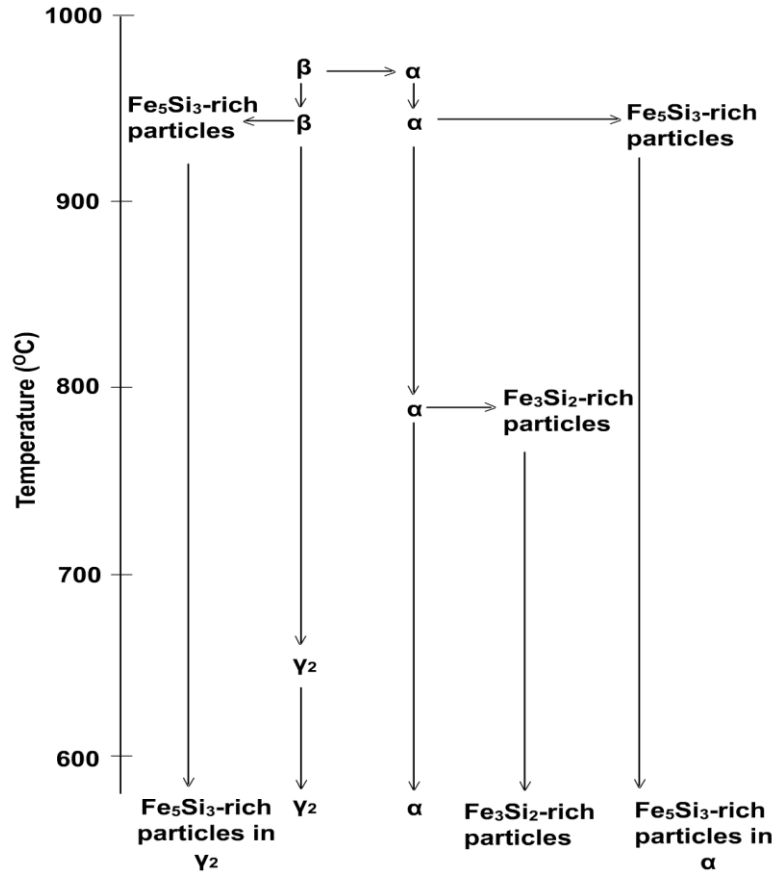**Figure 10:** Schematic illustration of the development of microstructure of the as-cast alloy

5. Conclusions

The sequence of microstructural development during continuous cooling from the temperatures approaching solidus showed that at higher temperatures the alloy consists of β plus α. The α-phase exhibits a Widmanstatten morphology. As the temperature drops, irregularly-shaped particles (Fe_5Si_3) are nucleated. At lower temperatures, the solubility of iron in α is exceeded and small lath-shaped particles (Fe_3Si_2) form in the α-phase. On further cooling the β-phase is transformed into the γ_2-phase.

REFERENCES

[1] Macken, P. J. and Smith, A. A.; *The Aluminum Bronzes*, CDA, London, 1966.

[2] Lioyd, B. A. and Pyemont J. W.; *Met. Tech.*, 1(1974) 534.

[3] Cairns, W. J., Goodwin, R. J. and Stephens D. M.; *J. Inst. Met.*, (1970) 244.

[4] Upton, B.; *Corrosion*, 19(1963) 204.

[5] Bradley, J. N.; *Inst. Met. Rev.*, 17(1972) 88-92.

[6] Dies, K.; *Kupfer and Kupfer Legierungen in der Technik*, Berlin, Springer-Verlag, 1967.

[7] Newcombe, G.; Copper Conf., Cleveland, Ohio, New York, CDA, 1972.

[8] Wilson, F. H.; *Trans. AIME*, 175(1948) 262.

[9] Iqbal, J.; *PhD Thesis*, University of Manchester, 1985.

[10] Hasan, F., Iqbal, J. and Ridlay, N.; *J. Materials Sci. Tech.*, 1(1985) 312.

[11] Hasan, F., Jahanfrooz, A., Lorimer, G. W. and Ridley, N.; *Met. Trans.*, 13A(1982) 1337.