Journal of

Research Article Open Access

.JDSPTUSVDUVSBM BOE .FDIBOJDBM 1SPQFSUJ .BOVGBDUVSFE CZ 4FMG QSPQBHBUJOH)JHI U

Amiour Y [1]*, Zemmour K [1] and Vrel D [2]

[1]Laboratoire LEREC, Département de Physique, Université de Annaba, Algeria
[2]Laboratory LIMHP, UPR 1311 CNRS, Université Paris XIII, 99 Avenue J.-B. Clément, 93430 Villetaneuse, France

*Corresponding author: Amiour Y, Laboratoire LEREC, département de physique, Université de Annaba, Algeria, Tel: 213/038752758; E-mail: amiouryacin.@gmail.com

Received August 16, 2016; Accepted October 12, 2016; Published October 20, 2016

Citation: Amiour Y, Zemmour K, Vrel D (2016) Microstructural and Mechanical

Citation: Amiour Y, Zemmour K, Vrel D (2016) Microstructural and Mechanical Properties of Cu-based Alloy Manufactured by Self-propagating High-temperature Synthesis Method. J Powder Metall Min 5: 144. doi:10.4172/2168-9806.1000144

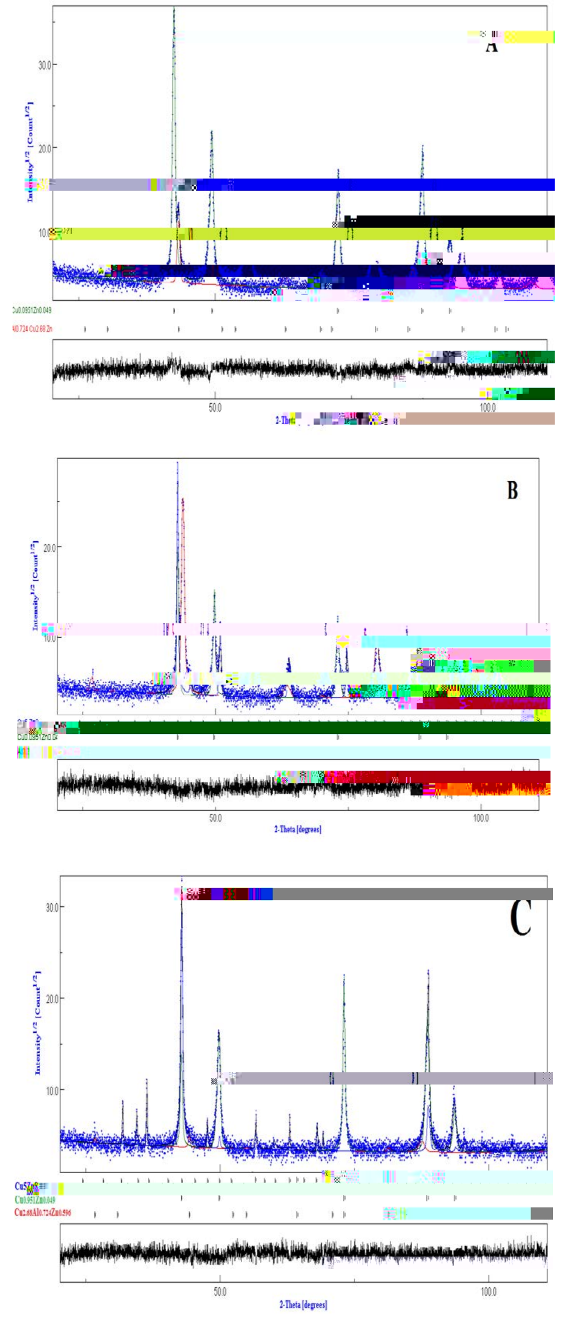

Figure 1: 5LHWYHOG UH¿QHPHQW RI WKH ;5' SDWWHUQ RI WKH VDPSOHV $ % DQG &

e relative density of the green compacts is between 71% of theoretical density.

e sample must be of a certain length and starting density in order to allow a good propagation observation: the length has to be great enough so a clear distinction between ignition zone and normal SHS propagation could be made; a natural suggestion for the starting density would be to increase it as much as possible, to facilitate nal sintering, but above 80% of the theoretical density, the thermal conductivity of the sample increases too much, and the ignition of the sample becomes di cult (in our experiment, we targeted an initial density ranging from 60 to 70%; the pressure is not measured, it is increased until the nal

Citation: Amiour Y, Zemmour K, Vrel D (2016) Microstructural and Mechanical Properties of Cu-based Alloy Manufactured by Self-propagating High-temperature Synthesis Method. J Powder Metall Min 5: 144. doi:10.4172/2168-9806.1000144

from a possible structural model. In the rst step of re nement, the global parameters, such as background and scale factors, were re ned. In the next step, the structural parameters such as lattice parameters, pro le shape and width parameters, preferred orientation, asymmetry, isothermal parameters, atomic coordinates, and site occupancies were re ned in sequence. e tting quality of the experimental data is assessed by computing the parameters such as the 'goodness of t, ^{2}, and the R factors (R_p=pro le factor, R_B=Bragg factor, and R_F=crystallographic factor) [15]. When these parameters reached their minimum value, the best t to the experimental di raction data is achieved, and the crystal structure is regarded as satisfactory [15].

Figure 4 shows the Rietveld re nement of XRD patterns Cu-Al-Zn samples at room temperature. As shown in Figure 1, in sample A, only two phases can be indexed with cubic structure. However for the samples B and C, other additional peaks appear indicating the formation of a new phase. e new phase can be identi ed as Cu_5Zn_8 phase with face-centered cubic structure. We note that the shape memory e ect (SME) observed in these alloys was always accompanied with the presence of a major CuZnAl phase. e crystal lattice parameters, grain size and amount of each phase in samples deduced from Rietveld re nement are listed in Table 2.

Mechanical analysis

e as-cast samples A, B and C have been rst subjected to tensile test in order to determine the maximum value of the unit loading to render evident the shape memory e ects and the pseudoelastic behaviour. e A, B and C samples presented similar values of the tensile strength, ranging between 220 MPa and 300 MPa. A er that, the sample A was subjected to loading-unloading cycles on the tensile testing machine at a temperature of 20°C.

Figure 5 shows the loading-unloading curves of alloys with 0.06% pre-strain at room temperature. e sample A show that for a unit loading of 300 MPa close to the run-out limit on this alloy, a permanent strains $_p$ of about 1% have been obtained. is strain is much lowers then the value of 6% which can be obtained for a Ni-Ti-Fe alloy stressed under the same conditions [17], is close to the value speci c to the Cu-Zn-Al alloys [18]. e sample B shows, a er the rst tensile loading-unloading cycle, a pseudoelastic curve with a total speci c elongation of $_t$=4% and a pseudo-elastic annealing with $_r$=3.33% and a plastic over strain $_p$-

e C sample subjected to tensile loading-unloading tests showed a characteristic curve with a plastic overstrain $_p$=1.52% (which is higher than the values obtained for the A and B alloys), as well as a non-linear elastic recoverable strain $_r$=1.48%, which is smaller than the one for the B alloy, get higher than the one for the A alloy.

A er the rst tensile loading-unloading cycle samples from A, B, and C, alloys have been subjected to a dilatometric analysis within temperature ranger 20-140°C a er having previously been tensile strained up to values of the tensile unit loading lower than the breaking strain and to values corresponding to the permanent strain ($_p$) of about 1%. Obviously, it follows that this alloy shows no shape memory e ect within the positive temperature range, but this e ect might be present at negative temperatures, its critical transition points being placed below 0°C. is hypothesis is also backed up by the aspect of the loading-unloading curve presented in Figure 5, which shows a typical pseudoelastic twinning curve usually resulting within the austenitic temperature range exceeding the critical temperature Md corresponding to the maximum temperature of stress induced martensite (SIM).

e signi cant changes done on the level of the concentration of the components (Zn, Al) as compared to the A and B alloys, lead to the dri of the critical transition points within the positive temperature range, and therefore to the occurrence of a shape memory e ect (Figure 7).

Figure 5 shows the Brinell hardness of B_1 A_1 and C_1 alloys. One observes that the plastic deformation and the hardening treatment did not result in a signi cant modi cation of the values and critical points as compared to the as-cast state, indicating the homogenisation of alloys.

Conclusion

For the studied Cu-Zn-Al alloys, a narrow-range modi cation of the stoichiometry (chemical composition) does not result in important alteration of the unit tensile loading, but it leads to signi cant di erences with respect to the pseudoelastic properties and shape memory e ect:

- Both the A and B alloys present a pseudoelastic austenitic twinning curve at the room temperature and do not show a shape memory e ect at positive temperatures.

- e increasing of Zn concentration to 13%wt and Al concentration to 12.9%wt. results in the shape memory alloy (C) within the positive temperature range (0-140°C).

References

1. Creuziger A, Crone WC (2008) Grain Boundary Fracture in CuAlNi Shape Memory Alloys. Mater Sci Eng A 498: 404-411.

2. Xiao Z, Li Z, Fang M, Xiong S, Sheng X, et al. (2008) Effect of processing of mechanical alloying and powder metallurgy on microstructure and properties of Cu-Al-Ni-Mn alloy Mater Sci Eng A 488: 266-272.

3.

Citation: Amiour Y, Zemmour K, Vrel D (2016) Microstructural and Mechanical Properties of Cu-based Alloy Manufactured by Self-propagating High-temperature Synthesis Method. J Powder Metall Min 5: 144. doi:10.4172/2168-9806.1000144