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Research Paper

Effects of ripening on the mechanical properties of the laminated alloy $AlCu_3Si_2Mg$

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ABSTRACT

Pure aluminium has poor mechanical properties. For this purpose, the objective of this work is to study principally the influence of the incorporation of 3% copper, 2% silicon and less than one percent (1%) magnesium, which represent very important agents for improving the mechanical properties associated with the laminating process. In order to further increase the characteristics of this wrought alloy $AlCu_3Si_2Mg$, we have carried out heat treatments in a further step by varying the maturation time in two-hour increments over 24 hours. These different ripenings reveal a finer structure with several precipitates that prevent the movement of the dislocations. The selection of this alloy was motivated by the large number of mechanical applications in which it is widely used. The results obtained in this study show that the twelve-hour hardened state M12h is the best compromise between strength and maximum ductility.

1 Introduction

The objective assigned to this paper is to study mainly, in a first step, the influence of the addition of 3% copper 2% silicon and < 1% magnesium, which respectively facilitate the machining, casting and setting of heat treatments. These three elements added to aluminum are, in a second stage, very important agents for improving the mechanical properties associated in a third stage with the industrial rolling process. The alloy is thus hardened with the presence of initial dislocations.

In a fourth stage, to further increase the mechanical properties of this $AlCu_3Si_2Mg$ alloy, it will be subjected to heat treatments that generate important microstructural modifications, generally hardening.

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The cycle industry and transport companies are very interested in this alloy, whose strength is equivalent to that of ordinary steel and whose weight is a third of that of steel. Since then, new alloys with different properties and with or without heat treatments have been constantly developed and found. The use of these new alloys has led to unexpected developments in aluminium production [1-6].

2 Issue

Pure aluminium in its pure state has low mechanical characteristics making it impossible to use it in the construction of mechanical structures. However, the addition of additive elements and the application of heat treatments cause certain phases to precipitate in the aluminium matrix. The alloy is then generally stronger but also less ductile than pure aluminium. The low density, inalterability, low melting point, indefinite recyclability, abundance in nature, aesthetics, good reflectivity, good heat resistance and food compatibility of aluminium alloys combined with their good mechanical characteristics are of interest to the rail, naval, road and aeronautical transport industry [7-11].

3 Material studied

In all our experiments, we used Aluminium copper alloy containing 3% copper, 2% silicon, in mass percentage, an amount of magnesium less than 1% (by mass) and some traces of impurities, of chemical composition AlCu₃Si₂Mg. It is supplied free of charge by the Technology Hall of Oued-Aïssi of the University Mouloud Mammeri of Tizi-Ouzou Algeria [1-6]. The analysis of specimens and samples shows the following chemical composition

Table.3 - Results of the chemical analysis after control on specimens

Chemical elements	Cu	Si	Mg	Mn	Fe
Percentage (%)	2,97	1,60	0,8	0,72	0,41

4 Elaboration of the wrought alloy AlCu₃Si₂Mg

In our study we machined 45 tensile specimens, 45 impact specimens and 60 samples from AlCu₃Si₂Mg bars in cylindrical form with a diameter of 20 mm and a length of 6 m delivered by Pechiney.

This alloy is produced by an industrial rolling process and will be categorised as a wrought alloy [12-19].

5 Experimental procedure

This part is devoted to the presentation of the different methods and experimental techniques used. We mainly used seven techniques, namely

Tensile strength to identify the different stresses. Initial dislocations occur when a load is applied: in general, the generation of dislocations in the same alloy under the effect of stresses are generally of the same sign, which repel each other. It will be necessary to develop an additional constraint to make the dislocations move in the vicinity of the others to enable them to overcome the various obstacles encountered and to move from one obstacle to another,

- The different deformation-free energies that provide information on the mode of rupture,
- The stress intensity factor $< K_s >$ which quantitatively characterizes the resistance of a material to crack propagation,
- Brinell hardness $< HB >$ and microhardness $< H_{v0,1} >$ to determine the stress field strength,
- $< K_{cv} >$ resilience to assess impact resistance,
- The microstructure to recognise the different structures.

We will then describe and present in detail the main mechanical characteristics obtained from the AlCu₃Si₂Mg alloy which is the subject of this study [20-37].

6 Results and discussion

The values of the mechanical characteristics are shown in figures 1 to 26 below and are obtained by measuring three

identical specimens of the AlCu3Si2Mg alloy in a series of three measurements.

The different states of treatment of the alloy are noted with: a - as-cast: < F >, b - hardened: < T > and matured: c - < M0h >, d - < M2h >, e - < M4h >, f - < M6h >, g - < M8h >, h - < M10h >, i - < M12h >, j - < M14h >, k - < M16h >, l - < M18h >, m - < M20h >, n - < M22h >, o - < M24h > and (p) - comparison group of the 15 corresponding states.

6.1 Average tensile strength curves of AlCu3Si2Mg alloy

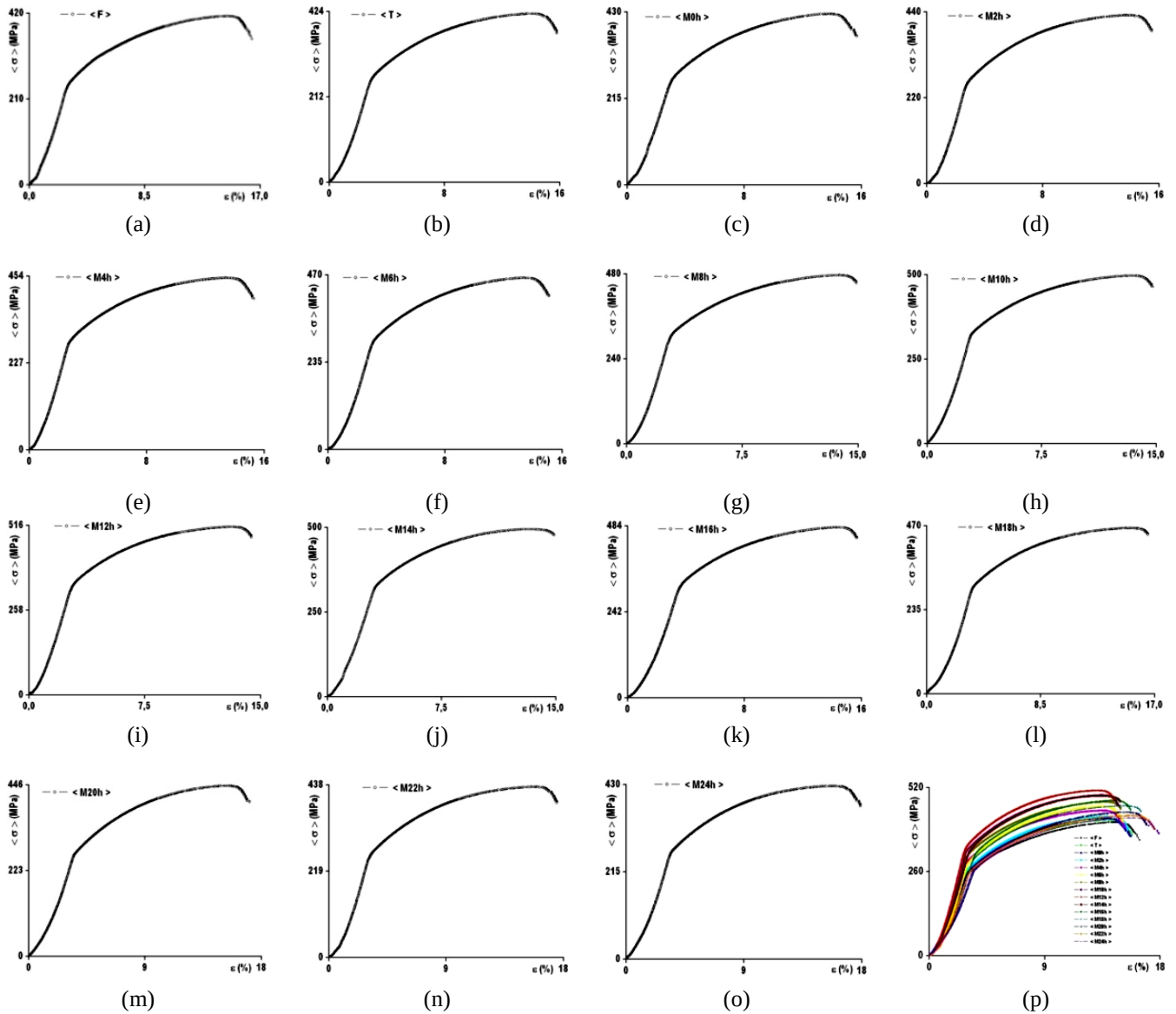


Fig. 1 – Tensile stress versus strain curves for AlCu3Si2Mg alloy

The strength and ductility properties, the stress intensity factor K_s and the different fracture energies of the AlCu3Si2Mg alloy are very important parameters for the design of a mechanism consisting of an assembly of various parts subjected to various external stresses and which must not exceed the elastic range. All the curves (Fig.1) obtained show almost the same appearance, which can be distinguished by three different regions:

- A large linear elastic part of the equation $\sigma_e = E\epsilon_e$: this part can be explained by the movement of atoms bound by electrostatic forces.
- A very large homogeneous plastic part with a homogeneous shape $\sigma = K\epsilon^n$: before the elastic deformation has reached its limit, a second part appears, which is the plastic deformation; it is mainly carried out by a sliding

movement which is interpreted by the movement of the initial dislocations or those created during the deformation and their interactions with various obstacles present in the material.

- And a second plastic part, this time heterogeneous, with a ductile fracture: determined by the elongation $\langle A\% \rangle$, the stress coefficient $\langle Z\% \rangle$, the resilience $\langle K_{cv} \rangle$ and the stress concentration coefficient $\langle K_s \rangle$. It goes through two stages, the first is the birth of the cracks and their coalescence with the appearance of the striction and the second, their propagation through the whole section of the specimen.

Figure.1 shows respectively the grouping of the applied stress versus strain dependence curves and their study has allowed us to determine the mechanical characteristics of each state of the alloy.

Figure.1(p) shows the comparison grouping of the average curves (stress-strain) of the fifteen states of the AlCu3Si2Mg alloy. Figures 3 to 26 show the different results obtained during our study of the different mechanical properties of our alloy.

6.2 Average deformation energy curves of AlCu3Si2Mg alloy

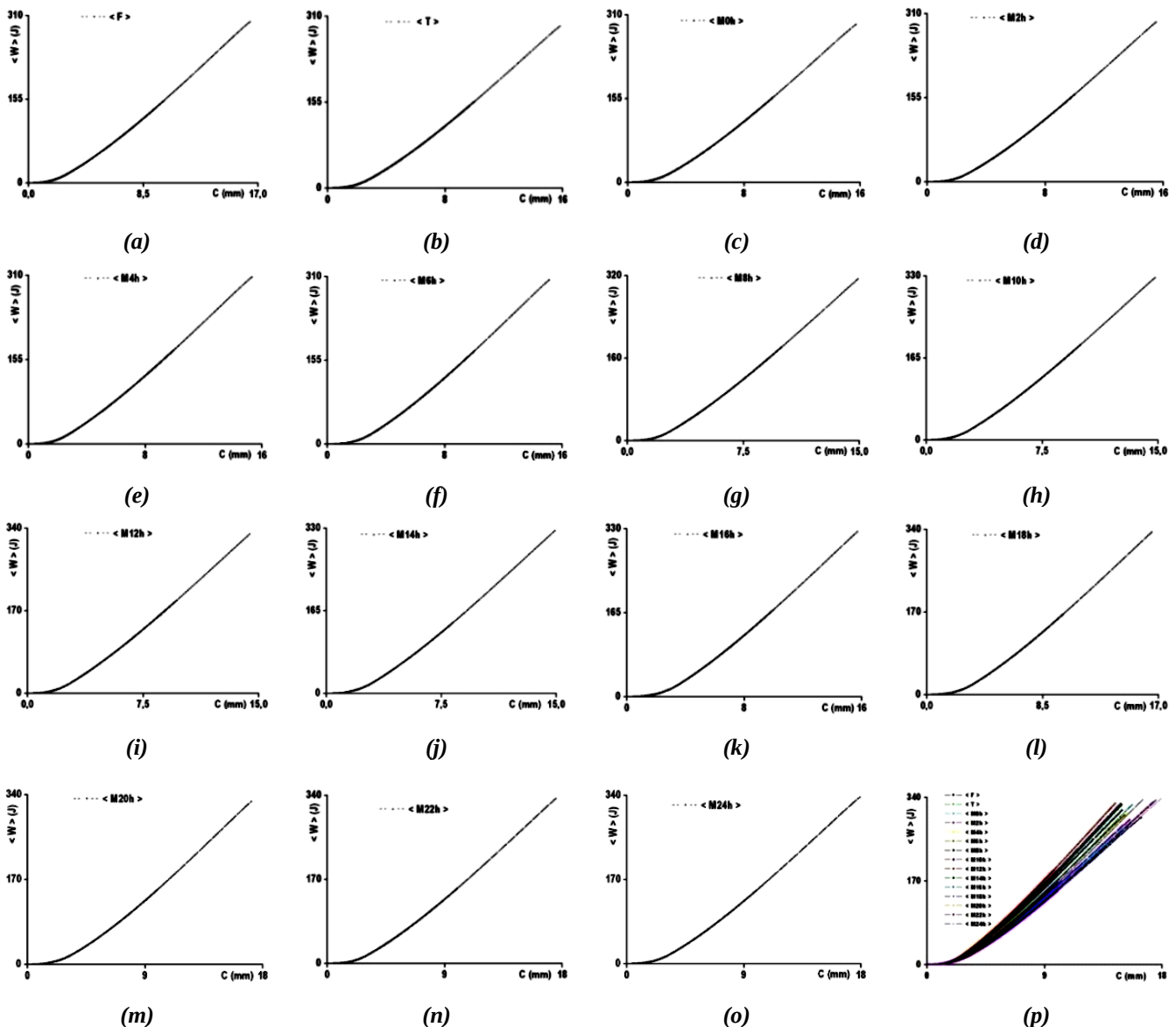


Fig. 2 – Deformation free energy curves versus of the AlCu3Si2Mg alloy stroke

Figure 2 shows the average energy curves free of deformation ($\langle W(J) \rangle - C(mm)$) and their studies have enabled us to determine the different energies Figures 21 to 25 of each state of the AlCu3Si2Mg alloy. Figure.2(p) shows the comparison grouping of the average curves (Energy - Stroke) of the fifteen states of the AlCu3Si2Mg alloy.

6.3 Influence of the maturation time on the characteristics of

6.3.1 Medium resistances

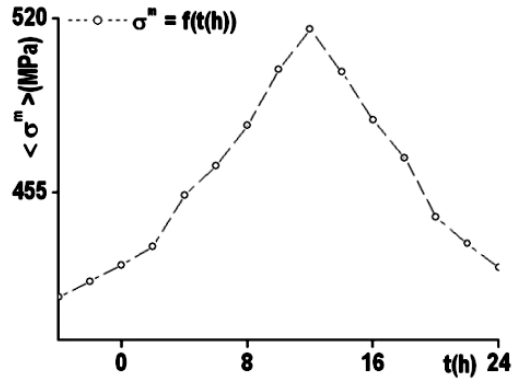


Fig.3 - Maximum stress versus of the ageing time of the AlCu3Si2Mg alloy

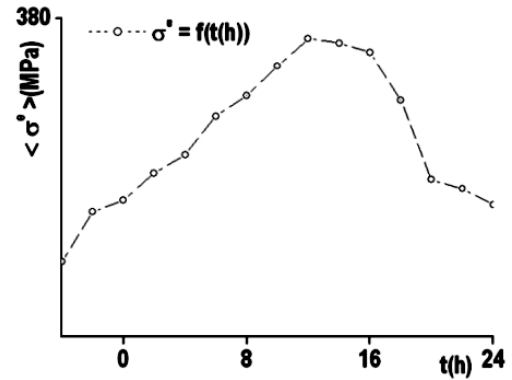


Fig.4 - Elasticity stress versus of the ageing time of the AlCu3Si2Mg alloy

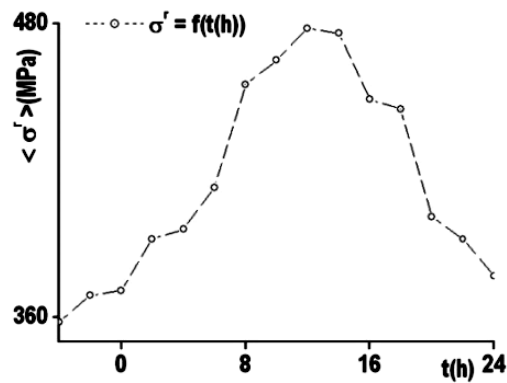


Fig.5 - Breaking stress versus of the ageing time of the AlCu3Si2Mg alloy

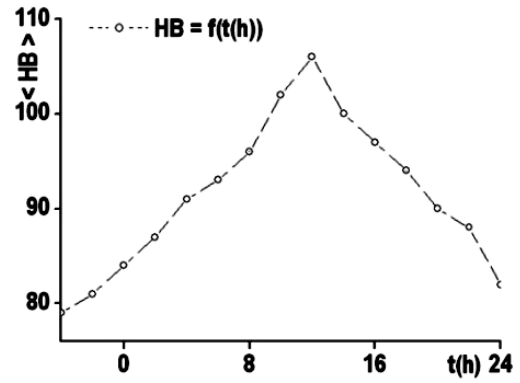


Fig.6 - Brinell hardness versus of the ageing time of the AlCu3Si2Mg alloy

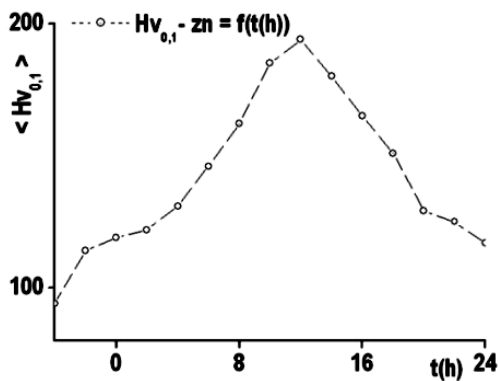


Fig.7 - Vickers microhardness of the black zone versus of the maturation time of the AlCu3Si2Mg alloy

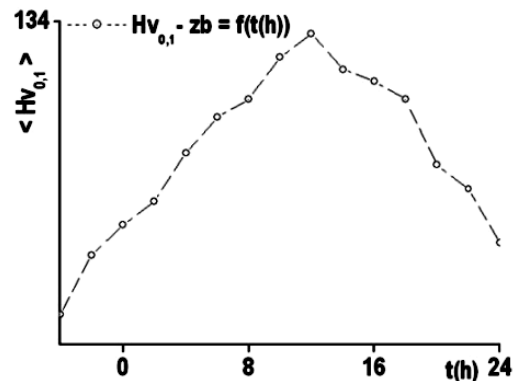


Fig.8 - Microhardness of the white zone versus of the maturation time of the AlCu3Si2Mg alloy

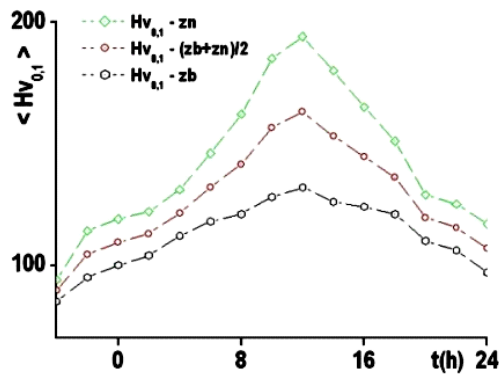


Fig.9 –Average Vickers phardness of the black and white zones versus of the maturation time of the AlCu3Si2Mg alloy

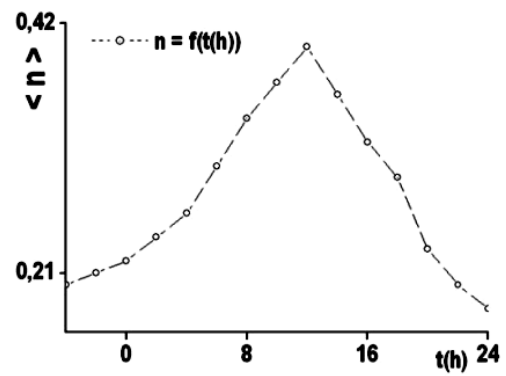


Fig.10 - Hardening coefficient versus of the ageing time of the AlCu3Si2Mg alloy

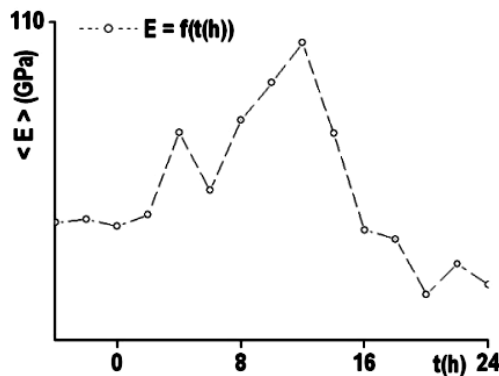


Fig.11 - Young's modulus versus of the ageing time of the AlCu3Si2Mg alloy

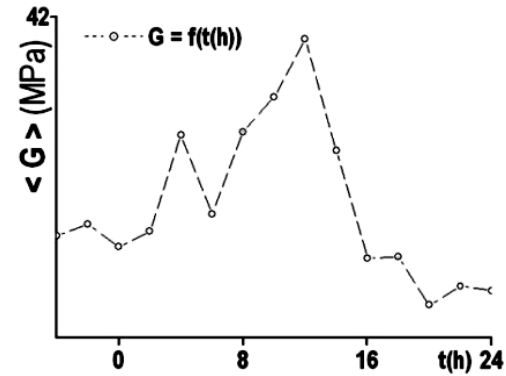


Fig.12 - Shear modulus versus of the ageing time of the AlCu3Si2Mg alloy

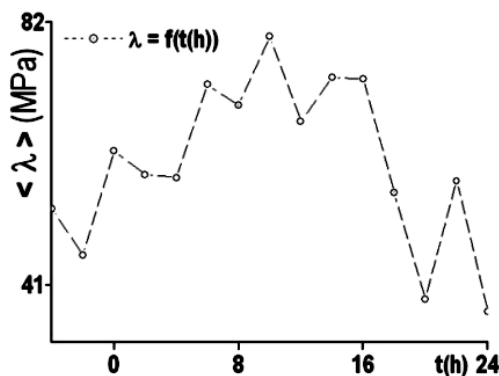


Fig.13 - Coefficient of lamination versus of the maturation time of the AlCu3Si2Mg alloy

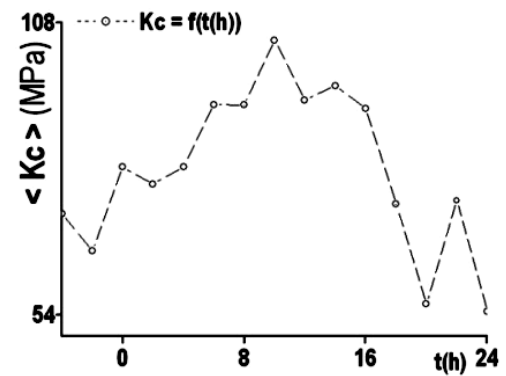


Fig.14 - Hydrostatic compression module versus of maturation time

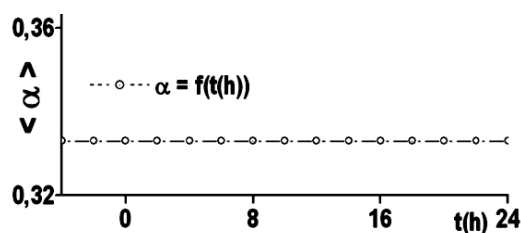


Fig.15 - Coefficient of linear thermal expansion versus of the maturation time of AlCu3Si2Mg alloy

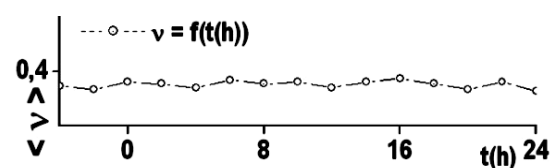


Fig.16 - Poisson coefficient versus of the maturation time of AlCu3Si2Mg alloy

6.3.2 Average ductility

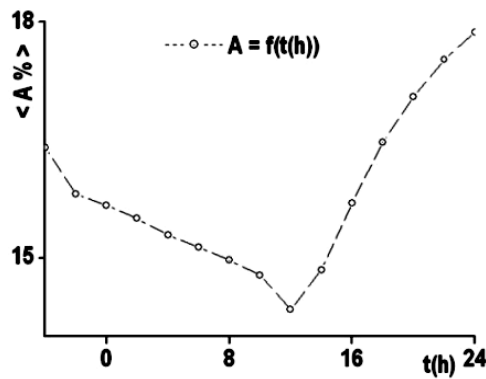


Fig.17 - Percentage elongation versus of the ageing time of the AlCu₃Si₂Mg alloy

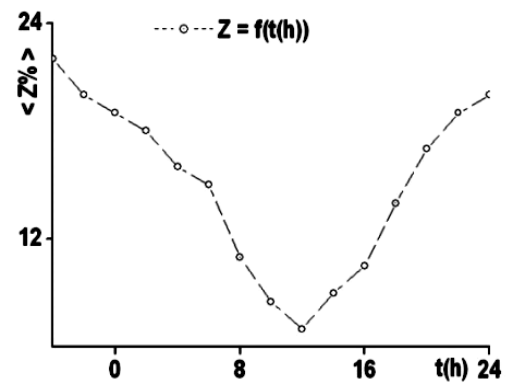


Fig.18 - Shrinkage coefficient versus of the maturation time of the AlCu₃Si₂Mg alloy

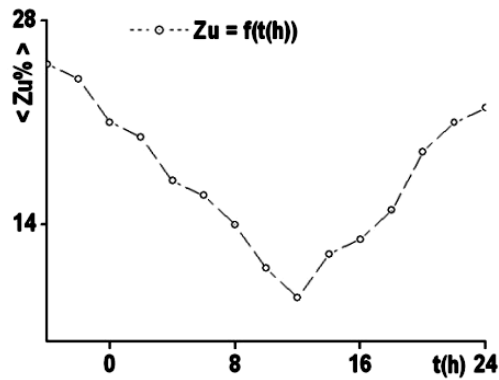


Fig.19 - Stress elongation versus of the ageing time of the AlCu₃Si₂Mg alloy

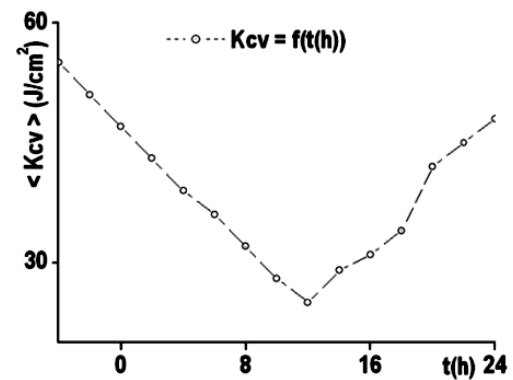


Fig.20 - Impact strength versus of the ageing time of the AlCu₃Si₂Mg alloy

6.3.3 The different average energies free of deformation

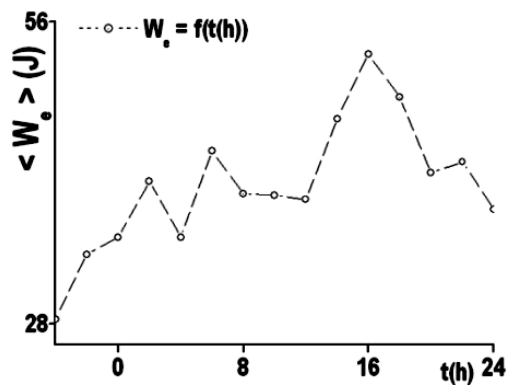


Fig.21 - Elastic deformation energy versus of the ageing time of the AlCu₃Si₂Mg alloy

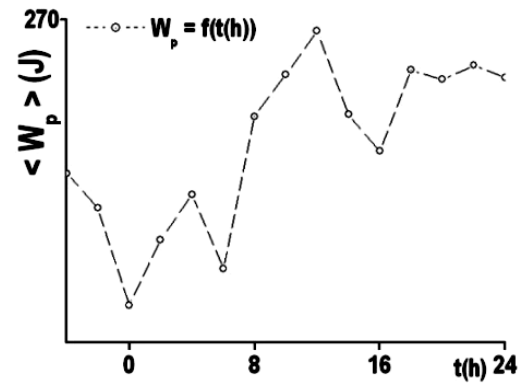


Fig.22 - Plastic deformation energy versus of the ageing time of the AlCu₃Si₂Mg alloy

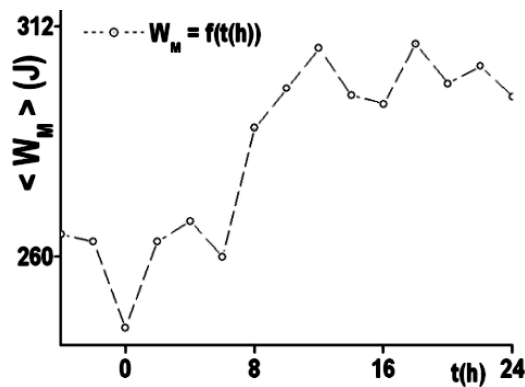


Fig.23 - Maximum deformation energy versus of the ageing time of the AlCu3Si2Mg alloy

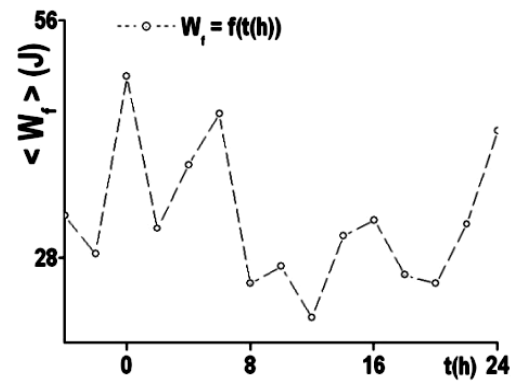


Fig.24 - Crack deformation energy versus of the curing time

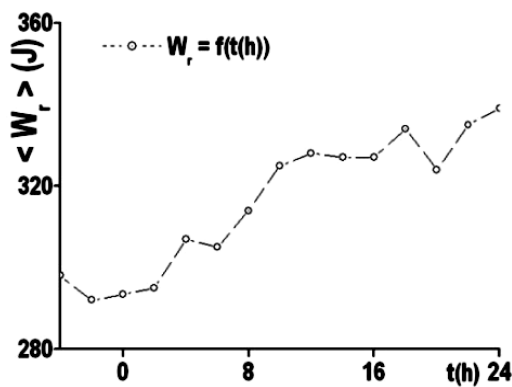


Fig.25 - Fracture strain energy versus of the ageing time of the AlCu3Si2Mg alloy

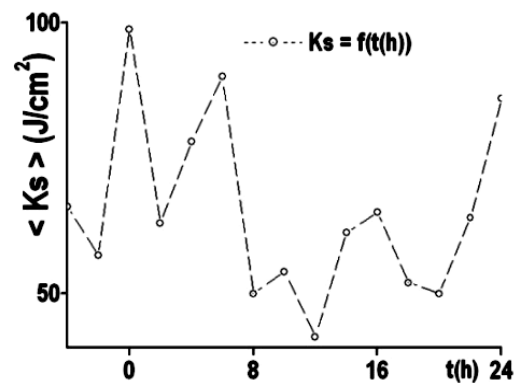


Fig.26 - Stress intensity factor versus of the ageing time of the AlCu3Si2Mg alloy

6.3.4 Discussion

In the as-cast state: F, the AlCu3Si2Mg alloy is produced by an industrial rolling process. This alloy has therefore undergone a strain-hardening process which consists of deforming the grains in the preferred rolling direction, creating initial dislocations. Under the effect of tensile stress, the sliding of the atomic planes of the work-hardened specimen is hindered by the presence of the defects generated, and therefore more energy is required to deform it. This justifies the significant increase in its strength characteristics to the detriment of ductility.

In order to further increase the strength properties of the as-cast state F, the movement of dislocations in the entire mass of the material must be hindered by causing the formation of finely dispersed precipitates (Al_2Cu , Al_2CuMg , Mg_2Si etc.) and a fine-grained structure, this leads to the AlCu3Si2Mg alloy being subjected to a structural hardening treatment. The addition of magnesium is necessary to make the alloy sensitive to this treatment, which is carried out in the following four steps:

First step: Heating, solution and homogenisation at 500°C for 10 hours in order to completely dissolve and homogeneously distribute the elements added to the aluminium. This temperature allows rapid diffusion of the solute elements (Cu, Si, Mg) in the Al matrix. However, it is important to avoid reaching the burning temperature. As for the choice of the holding time at this temperature, it must be sufficient for the copper and silicon present in the as-cast state to diffuse evenly towards the interior of the material. The correct choice of these two parameters has a great influence on the mechanical characteristics obtained.

Second stage: Immediately followed by quenching with water in the ambient environment, a highly supersaturated solid solution is obtained, the quenched state being noted: T. This quenching makes it possible on the one hand to maintain the existing homogenised structure at 500°C, and on the other hand to obtain a homogenous microstructure which becomes finer with the precipitation of Al_2Cu very close together at the joint and inside the grain because the free enthalpy of precipitation is high and the germination rate is also high. Mobile dislocations are difficult to overcome and the alloy is thus more resistant. This quenching creates a residual stress field within the alloy consisting of a balanced system of surface compression and core tension.

For a more homogeneous distribution of this field and in order to obtain a uniform structure to considerably increase the properties of the alloy, it is subjected to two successive ageing processes as follows

Third stage: Curing treatments at room temperature by varying the curing time in 2-hour increments over a 24-hour day with 13 designated states: M (0, 2, 4, ..., 24) h.

Fourth step: immediately followed by tempering at 165°C for 6 hours in an air-cooled oven for each of the 13 respective states to reduce the brittleness due to hardening by softening and eliminating internal stresses.

The different maturations followed by the same considered tempering create other precipitates of types (Mg_2Si , Al_2CuMg etc.) which are arranged inside the grains of the studied alloy and act as a block to the movement of the dislocations, thus increasing the resistance characteristics.

The heat treatments applied to the alloy cause:

- Hardening by refinement of the grain size with high obstacles to the movement of dislocations.
- Formation of precipitates (Al_2Cu , Al_2CuMg , Mg_2Si etc.) out of solid solution: coherent, semi-coherent or incoherent with the aluminium matrix.

In order to achieve high levels of mechanical resistance of the alloy, the movement of dislocations throughout the mass of the material must be impeded.

During loading, the structural hardening of this alloy is due mainly to the following three main factors which generate the blocking of dislocation movement.

- 1) The individual impurity and lattice elements have atoms of smaller or larger radii, can be in insertion or substitution solid solution respectively and create lattice distortions with a stress field due to the size difference between the solute and solvent atoms.
 - The interaction between this stress field and the dislocations makes them difficult to move,
 - Dislocations anchored by these atoms of impurity and network.
 - Two mechanisms can then occur during the passage of a dislocation at these atoms:
 - i. Shearing or bypassing of these atoms respectively by dislocation and creation of defects.
 - ii. Interaction between existing and created dislocations of the same crystal
- 2) Grain joints can be strong traps:
 - The dislocation disappears in the disturbed area and the plastic deformation is interrupted,
 - The grain seal can stop the first dislocation that occurs; the following dislocations that are unable to continue their movement will pile up behind it and the sliding will be interrupted. The movement of the dislocations will be all the more hindered and therefore the hardening obtained will be all the more important as the dislocations will encounter a greater number of grain joints.

The formation of different precipitates (Al_2Cu , Al_2CuMg , Mg_2Si etc.).

In order to cross these precipitates, the moving dislocation can then either shear them or by pass them by creating defects.

Once the heat treatments have been carried out, the material will be subjected to the various low load and speed tests and observations at room temperature.

The (Fig.1(p)) shows us that the curve of the state < M12h > is above all the other curves whatever the states considered. The properties of extrinsic resistances are as follows: the maximum tensile strength < σ_m >, the yield strength < $\sigma_{\text{strain-hardening}}$ >, factor < n >, the Young's modulus < E >, the shear modulus < G >, the Lamé coefficient < λ > and the hydrostatic compressive modulus < Kc > increase respectively from the as-cast condition: < F > to minimum strength properties in the matured state < M0h > with incoherent precipitates through M6h with semi-consistent precipitates to reach the inflection point < M12h > with coherent precipitates and maximum compromise of strength and ductility properties to then decrease to < M18h > with semi-consistent precipitates to finally reach the final state < M24h > with incoherent precipitates and minimum strength properties Fig. 3 to 14 to the detriment of the extrinsic ductility properties which are < A% > -- percent elongation, < Z% > -- shrinkage coefficient, < Zu% > -- shrinkage elongation, < Kcv > -- resilience and the various free energies of deformation which decrease respectively from the as-cast state: < F > at maximum ductility up to the state < M12h > at minimum ductility and finally increasing up to < M24h > at maximum ductility Fig. 17 to 25, however, the rate of growth or decay differs from one property to another. On the other hand, the coefficient of linear thermal expansion < α > and the Poisson coefficient < ν > remain almost invariant Figs. 15 and 16.

It can be seen that the strength characteristics (Figs. 3 to 14) of the alloy in the aged state < M12h > are higher than those of the other states at the expense of ductility (Figs. 17 to 25). Analysis of these results shows us that the compromise is the < M12h > state for strength parts in all states, as the natural ageing process was completed after 12 hours of ageing followed by complete tempering at 165°C for 6 hours.

In figures 7 to 9, the graph of the average microhardness of the black area is clearly above that of the average microhardness of the white area in all states.

The black zone is probably the solid solution of aluminium, silicon and magnesium in copper, while the white zone is the solid solution of copper, silicon and magnesium in aluminium.

The tensile (stress-strain) curves (Fig.1) of the $\text{AlCu}_3\text{Si}_2\text{Mg}$ alloy show us that:

Deformations at break (Fig.1) as well as the various free energies of deformation Fig. 21 to 25 are very high with the presence of the heterogeneous plastic domain because the damage occurred in a progressive way; moreover the section of the curve showing the passage from the maximum average stress < σ_m > to that of the average breaking stress < σ_r > is a curve of approximate parabolic shape: progressive fall from < σ_m > to Z >>> 0.1 (Fig. 18), the impact strength < Kcv > is very high (Fig. 20), the tensile and impact fracture surfaces (Fig. 40 and Fig. 41) are cup-shaped and have a matt, grainy appearance, the stress intensity factor < Ks > (Fig. 26) which quantitatively characterises the resistance of the alloy to progressive crack propagation is very high and the fracture energy of the crack is very high (Fig.24).

When the specimen is loaded, the fracture surface of the specimen is approximately 45° (Fig. 42). In this case, we probably have the superimposition of the Type II and Type III fracture modes with instantaneous displacements of the two straight and screw slides, since the shear stress is at its maximum on planes inclined at 45° to the line of the applied tensile force. These eight criteria and observations show us that the fracture is substantially represented by a topographic facies characteristic of a ductile cup-shaped fracture highly dependent on the morphological structure of the $\text{AlCu}_3\text{Si}_2\text{Mg}$ alloy grains.

6.4 - Determination of the average elastic properties of the wrought alloy AlCu3Si2Mg by ultrasound

6.4.1 Recorded signals

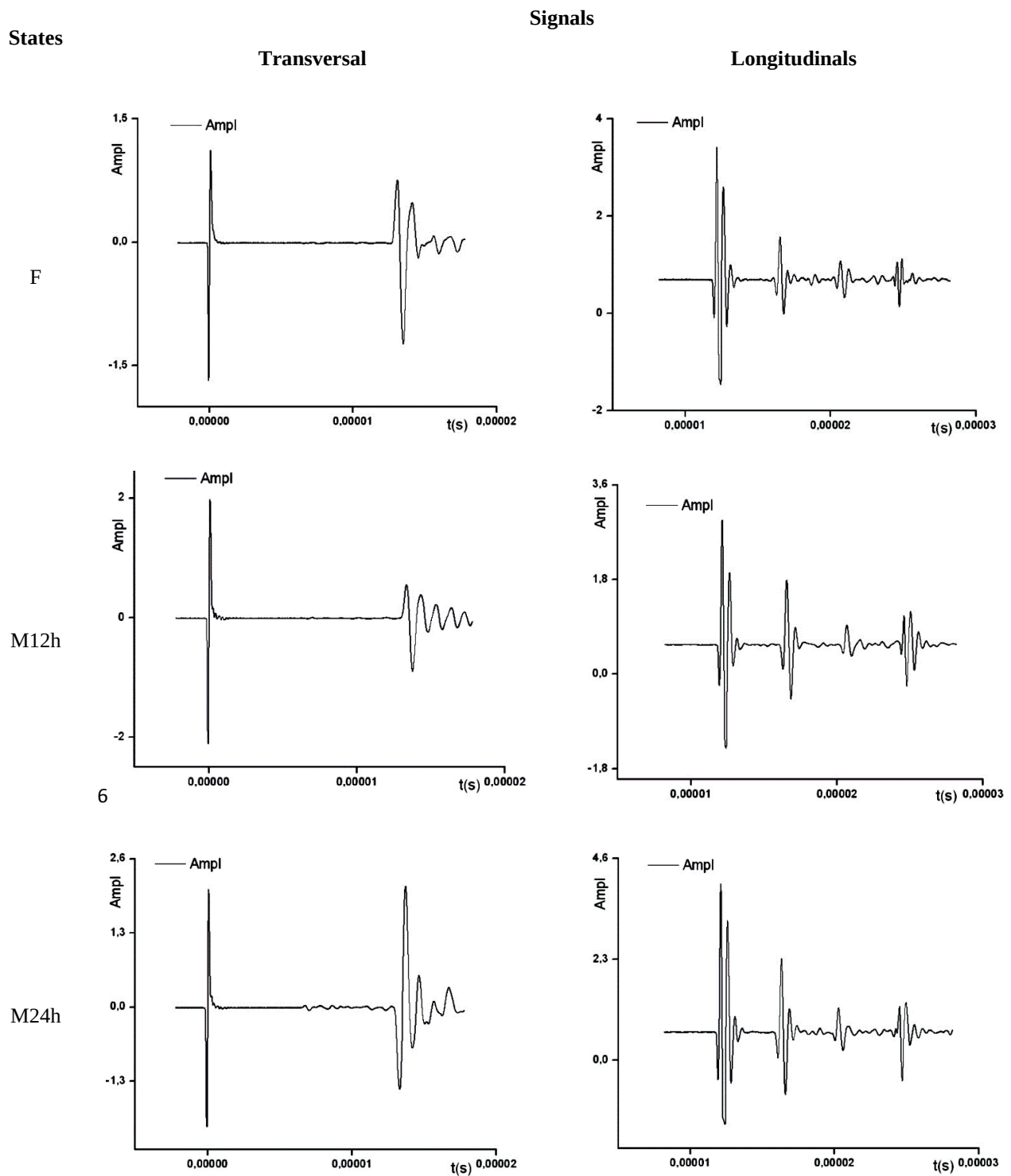


Fig .27 – Signals recorded on 40 mm wide prismatic ultrasonic specimens of the wrought alloy AlCu3Si2Mg in the as-cast condition: *F* and matured: < M12h > and < M24h >.

6.4.2 Elastic properties

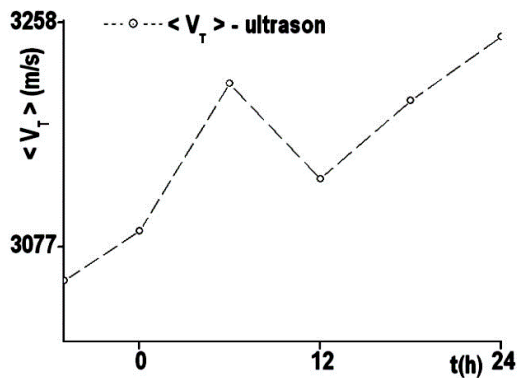


Fig. 28 - Transverse speed versus of maturation time

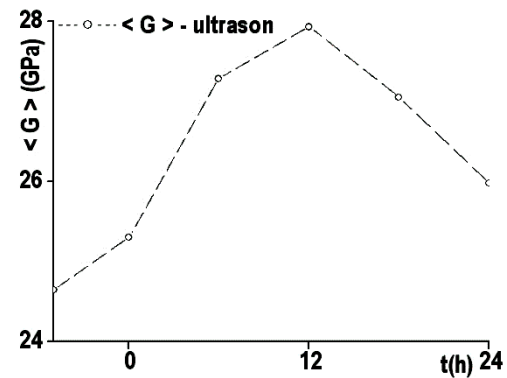


Fig.29 - Shear modulus versus of maturation time

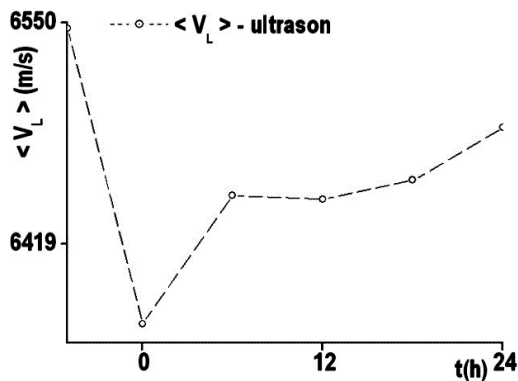


Fig.30 - Longitudinal speed versus of maturation time

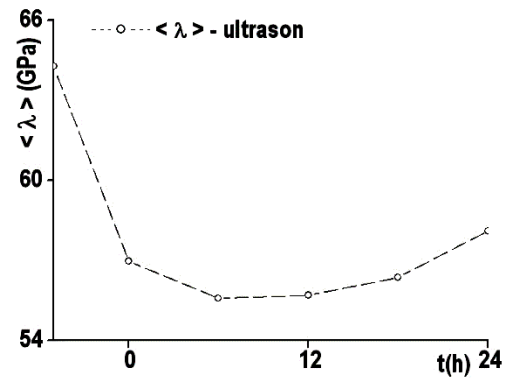


Fig. 31 - Lamé coefficient versus of maturation time

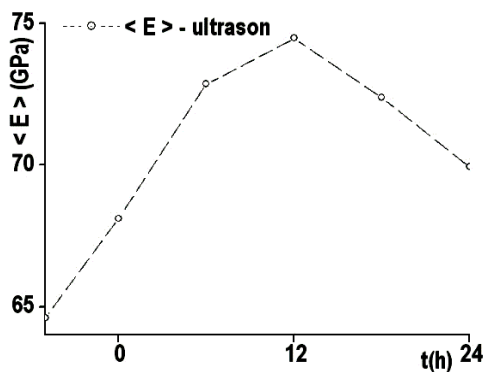


Fig. 32 - Young's modulus versus of maturation time

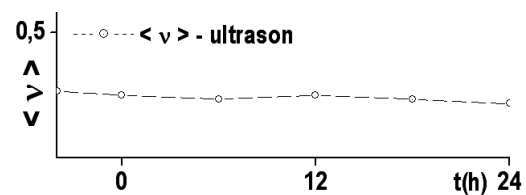


Fig. 33 - Poisson coefficient versus of maturation time

Figures 27 to 33 show the development of the six elastic properties determined by the non-destructive ultrasound method versus of the maturation time. In particular, the two most important elastic strength properties are shown: the Young's and shear moduli increase with increasing maturation time, from the as-cast condition in the $\langle M12h \rangle$ state with maximum properties to decrease to $\langle M24h \rangle$, although the rate of increase or decrease varies differently depending on the elastic characteristic and the condition under consideration. The extrinsic values of the Young's modulus and shear modulus of maturation $\langle M12h \rangle$ are higher than the other states considered, while keeping Poisson's ratio constant. Therefore, the elastic properties are maximum to the detriment of ductility whatever the states considered at maturation $\langle M12h \rangle$.

The evolution of the elastic strength characteristics given by the non-destructive ultrasonic method are on the one hand more reliable and on the other hand verify those given by the destructive tensile method. All the data obtained by ultrasound comply with the AFNOR standard.

6.5 Comparative study of the average elastic tensile and ultrasonic properties of the wrought alloy AlCu3Si2Mg

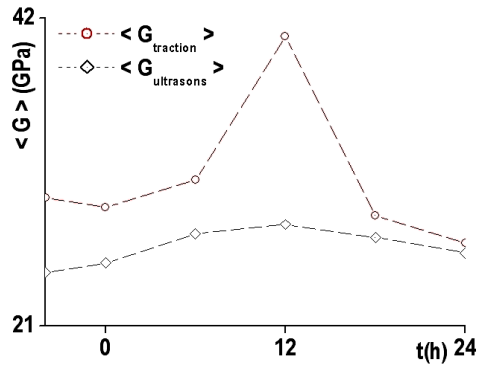


Fig.34 - Shear modulus versus of maturation time

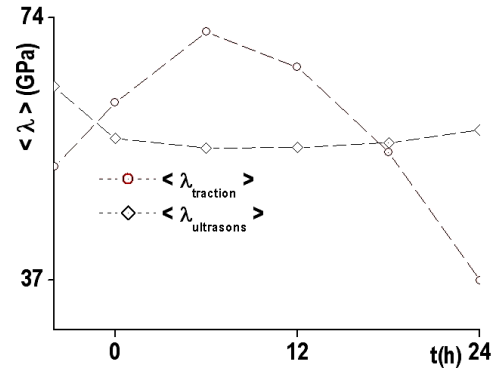


Fig.35 - Lamé coefficient versus of ripening time

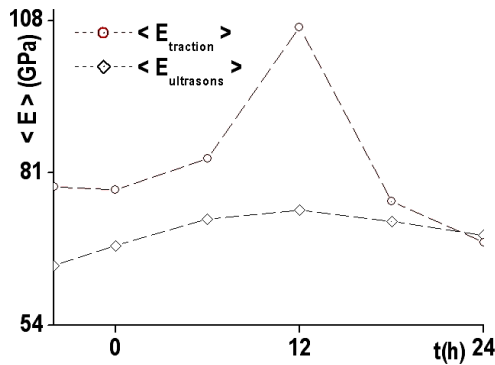


Fig.36 - Young's modulus versus of maturation time

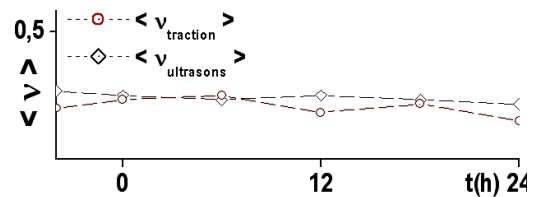


Fig.37 - Poisson coefficient versus of maturation time

A comparative study of the four common elastic properties considered in Figs.(34-37): Young's and shear moduli, Lamé and Poisson coefficients obtained by ultrasound and tensile tests versus of the maturation time increase from the as-cast state (at minimum properties) to the < M12h > maturation (at maximum properties) and then decrease to the < M24h > state (at minimum properties), however, the rate of increase or decrease varies differently depending on the elastic characteristic and the state under consideration. However, the Poisson's ratio remains almost invariant.

The graphs of the properties obtained by traction are far below those of ultrasound.

Therefore, the elastic properties are maximum to the detriment of ductility whatever the state and the two traction or ultrasound methods considered at maturation < M12h >. Regardless of the curing time considered, the elastic tensile characteristics are considerably greater than those obtained by ultrasound, which are in line with those given in the literature and the standard. The results of the common elastic properties obtained by ultrasound are reliable and conform to the AFNOR standard. On the other hand, the imprecise tensile method gives us erroneous results, but the latter has a great advantage because it allows the complete characterisation of materials: elastic, homogeneous and heterogeneous plastics, stress- and breakage properties.

7 Microstructure of the wrought alloy AlCu3Si2Mg

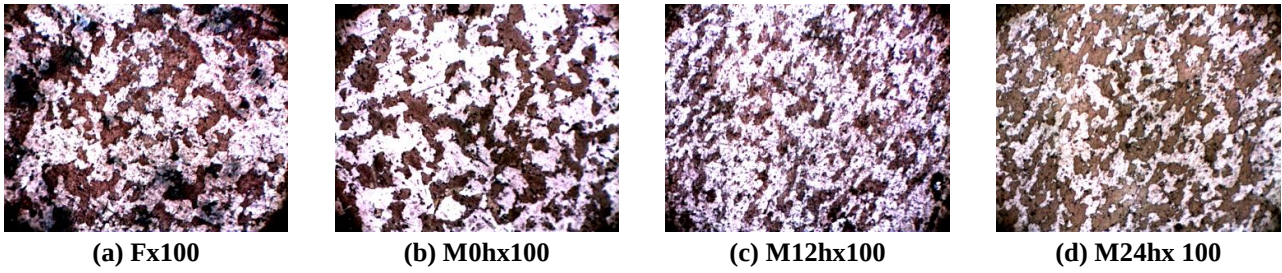


Fig. 38 – Micrographic structures of the rolling alloy AlCu3Si2Mg - (a) - $< F >$: as-cast, matured: (b) - $< M0h >$, (c) - $< M12h >$ and (d) - $< M24h >$.

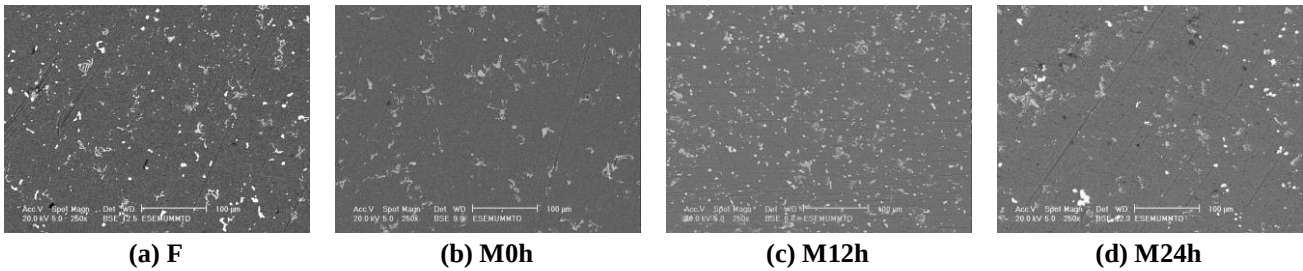


Fig. 39 – SEM microstructures of the rolling alloy AlCu3Si2Mg - (a) $< F >$ - as-cast, matured (b) $< M0h >$, (c) $< M12h >$ and (d) $< M24h >$.

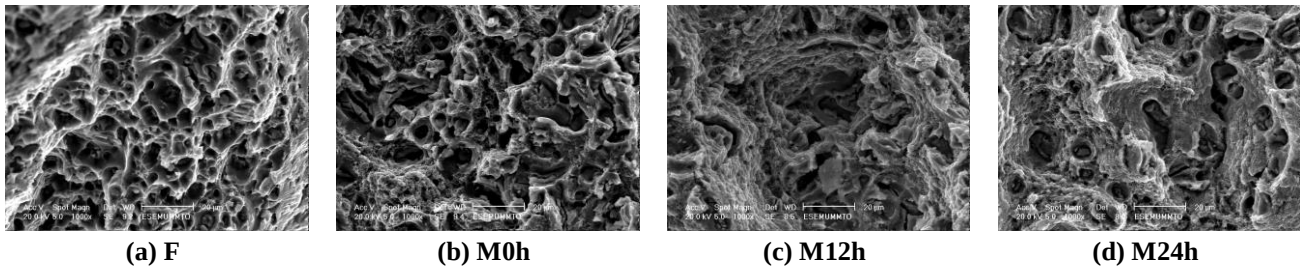


Fig. 40 – SEM tensile fracture toughness of the rolling alloy AlCu3Si2Mg - (a) $< F >$ - as-cast, matured (b) $< M0h >$, (c) $< M12h >$ and (d) $< M24h >$.

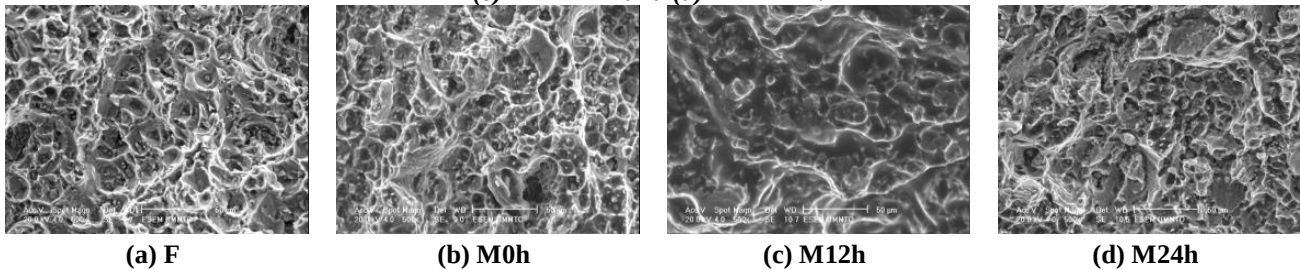


Fig. 41 - SEM impact fracture toughness fracture toughness of the rolling alloy AlCu3Si2Mg - (a) $< F >$ - as-cast, matured (b) $< M0h >$, (c) $< M12h >$ and (d) $< M24h >$.

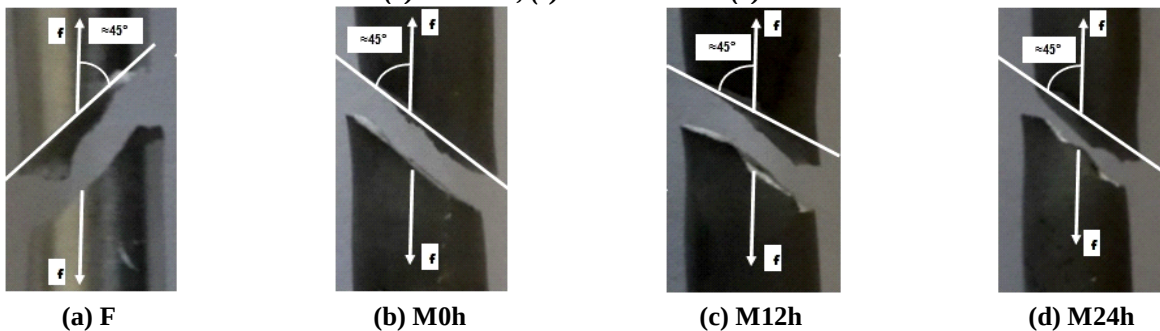


Fig. 42 – Ductile fractures at 45° to the tensile axis of the AlCu3Si2Mg alloy - (a) $< F >$ - as-cast, matured (b) $< M0h >$, (c) $< M12h >$ and (d) $< M24h >$ with f - tensile stress

When analysing the microstructures obtained by microscopy, we noticed that the grain size obtained by the rolling process decreases from the as-cast state: < F > (with larger heterogeneous grains) to the mature state < M12h > Figs. 38 to 39 ((a), (b), (c) and (d)), which explains the results obtained in the tensile, hardness, microhardness and impact test Figs. 3 to 26.

In the as-cast state, the alloy under study AlCu3Si2Mg generally contains large, randomly distributed grains formed after solidification in which the distribution of the content of the various additive elements (Cu, Si and Mg) is non-uniform; this leads to the formation of structural and local heterogeneities which generate a stress field that varies from one point to another in the material. On the other hand, their distribution in the matrix in mature and tempered states is homogeneous. In order to improve the mechanical characteristics of the alloy under study, it is necessary to hinder the movement of dislocations throughout the mass of the material, causing the formation of a fine structure and finely dispersed precipitates (Al₂Cu, Al₂CuMg, Mg₂Si, Al, Cu, Si, Mg etc.), for which we have subjected the alloy to a structural hardening treatment.

8 Conclusion

The mechanical characterisation of the wrought alloy AlCu3Si2Mg obtained by laminating, subjected to specific structural hardening, suggests the choice of the twelve-hour matured state denoted M12h >, which represents a compromise of strength and ductility regardless of the states and the tensile or ultrasonic methods considered.

The choice of this alloy is mainly due to its lightness combined with high mechanical characteristics; moreover, it was chosen because it is widely used in a very wide range of manufacturing processes. In recent years, its use has grown considerably in the automotive and rail industry in general and in high-tech sectors such as aeronautics and naval aviation in particular. The application of heat treatments, resulting in a finer structure with many grain joints as well as precipitates. The latter form strong obstacles to the displacement of dislocations, allowing a significant increase in the mechanical properties which are intimately related to the microstructure.

Thus AlCu3Si2Mg is a heat-treated lamination alloy with sufficient castability due to the presence of 2% silicon by weight.

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