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EFFECT OF THERMOMECHANICAL TREATMENTS ON AGING BEHAVIOR AND FRACTURE TOUGHNESS OF AL-CU-MG ALLOY PLATES USED IN AIRCRAFT STRUCTURE

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ABSTRACT

The effects of variations in the thermomechanical treatments parameters on the microstructure and mechanical properties of Al-Cu-Mg alloy plates used in aircraft structure subjected to a post heat treatments have been conducted. It was confirmed that a large gain in fracture toughness combined with a good mechanical properties could be achieved by choice of the correct initial condition, amount and temperature of deformation, and the time temperature of the final age hardening sequence. In general, it was found that the material should be age hardened to the peak hardness conditions before deformation at an elevated temperature (warm rolling at 200°C) in order to prompt homogeneous slip. Room temperature deformation resulted in planar slip at the peak hardness conditions non uniform response on subsequent aging and give lower fracture toughness.

الخلاصة

تم في هذا ألبحث دراسة ألتأثيرات ألناجمة من تغيير بعض معاملات ألتعامل ألحراري-ألميكانيكي على استجابة سبيكة ألألمنيوم-نحاس-مغنيسيوم ألمستخدمة في هياكل ألطائرات للتقادم ألاصطناعي أللاحق ومتانة ألكسر. وقد ثبت بأن من ألممكن ألحصول على استجابة جيدة ومتانة كسر عالية عن طريق ألاختيار ألصحيح للظروف ألابتدائية كدرجة حرارة وزمن ألتقادم ألأولي،نسبة ودرجة حرارة ألتشويه بالدرفلة وكذلك درجة حرارة وزمن ألتقادم أللاحق. وبصورة عامة فقد وجد بأن ألواح هذه ألسبيكة يجب أن تصلد حتى ألوصول ألىأعلى صلادة ممكنة قبل أن تدرفل بدرجة حرارة (٢٠٠°م) وذلك لتشجيع حصول عملية انزلاق منتظمة لكل ألبنية ألمجهرية. و ثبت أيضا بأن ألعينات ألمدرفلة بدرجة حرارة ألغرفة في ظروف قمة ألصلادة لاتستجيب جيدا لعملية ألتقادم أللاحق وتعطي متانة كسر منخفضة.

Keywords: thermomechanical treatments, microstructure, fracture Toughness, hardness, aging.

The Duralumin has a dominant in the structure of aircraft to some extent where most of the plates and rivets are manufactured from the 2024 and 2016 aluminum alloys.

The plates and shells describes above subjected to many processes which is either plastic deformation to get the required thickness, heat treatments or both of the two techniques to improve the mechanical properties as well as corrosion resistance. This technique found a great acceptance in many world companies that looking to improve this type of aluminum alloys. An improving of 2024 aluminum plates through the combination between rolling technique and the precipitation hardening is not so easy, and a try and errors is very spread in solving such a problem.

The technique described above known as a thermo-mechanical treatment (TMT) which is a combination between a heat treatment processes and a plastic deformation processes to obtain a good mechanical and physical properties [Y.Lakhtin,1983] .The most common heat treatment of the aluminum alloys is the precipitation hardening heat treatment. The basic requirements of a precipitation hardening alloy is that, the solid solubility limit should decrease with decreasing temperature as shown in fig. (1) for aluminum- copper system of alloys [V. Ozoline, 2001]. In this figure, this alloy at high temperature exists as a homogeneous k-solid solution, but on cooling, it becomes saturated with respect to a second phase θ, lattice hardening often results when θ-phase separates out at lower temperature [3].

During this procedure, the alloy is first solution-treated at high temperature and then rapidly cooled by quenching into water or some other cooling medium. The rapid cooling suppresses the separation of the θ-phase, so that the alloy exists at the low temperature in an unstable super-saturated state [R.B. Nicholson, 1958] after quenching process, the alloy is allowed to age for a sufficient time, the second phase precipitates out. The precipitation-hardening heat treatment process occurs by a nucleation and growth process, the solute atoms concentration fluctuations providing small clusters of atoms in the lattice, which act as nuclei for the precipitates.

The rate of growth of these nuclei is controlled by the migration rate of atoms. For this reason the precipitation increases with the increasing of aging temperature since migration or diffusion is a function of temperature [R.B. Nicholson, 1958]. However, the size of precipitate becomes finer as the temperature at which precipitation occurs is lowered, and an extensive hardening of the alloy is associated with a critical dispersion of the precipitate.

These precipitates included the Gunier-Preston (G.P) zones rich in the solute elements. The morphology of the zones depends mainly on the difference in atomic diameter of the solute and solvent elements. With the aluminum-copper system, where the copper atom is of smaller diameter than aluminum, plate-like zones initially one atom of copper in thickness are found on <100> planes of the aluminum matrix [R.E.Smallman, 1985]. Now because of the difference in atomic size coherency strains develop around the zones and in some cases these may be so large that they modify the formation of the zones.

In many alloy systems, there may be an appreciable temperature interval between the solvus and solidus. Varying the solution treatment temperature within this range could have a marked effect on the subsequent mode of the decomposition of the supersaturated solid solution produced on quenching. The solvus temperature is usually above the recrystallization temperature: consequently, grain growth can occur during solution-treatment.

A further possibility is that, the vacancy concentration will be increasing as the solution treatment temperature is raised. Thus on quenching greater vacancy super-saturation will be obtained [Marcco, Fontecchio, 2002]. Moreover, this will lead to an increase in the rate of decomposition of the super-saturated solid solution particularly during the early stages of zone formation.

Now since the solubility and diffusion rate are both increases with temperature, it is desirable usually to use the high treatment temperature that will not cause re-melting. When the temperature attained by the parts or pieces being heat-treated are appreciably below the normal range, solution is incomplete, and strength somewhat lower than normal is expected.

If the eutectic melting discernible in the micro-structure occurs as a result of over-heating, then strength, ductility and fracture toughness may be lowered [R.E.Smallman, 1985]. The time at the nominal solution heat-treating required affecting a satisfactory degree of solution of the un-dissolved or precipitated soluble phase constituents and to achieve good homogeneity of the solid solution is a function of microstructure before heat treatment and alloying elements.

In aluminum-copper system of alloys, the quenching rate and quenching temperature affects the G. P. zones formation rate, the excess vacancies caused a rapid diffusion of copper atoms to join the zones, these vacancies retained from the solution temperature by rapid quench [Marcco, Fontecchio, 2002].

Now, if the quench is interrupted for few seconds at high temperature to eliminate these vacancies, the zones then form much less rapidly. The decrease in quenching rate reduces the yield strength. However, the rate at which yield strength is reduced varies with alloy and temper.

Influence Of Deformation Amount & Temperature On TMT

After solution, treatment and quenching highly super-saturated solid solutions are obtained. The super-saturated solid solution may show interesting effects due to segregation of the thermally produced vacancies onto dislocations, probably introduced by the plastic deformation and/or by quenching processes [P.P. Muller, 1989].

During the subsequent aging treatment for a relatively short time at a temperature up to the G.P. zone formation temperature as shown in fig. (1), the dislocations act as nucleating sites for the precipitates and poor mechanical properties are obtained [R.E.Smallman, 1985]. If the quenching strains can be avoided, the precipitates are nucleated homogeneously and a uniform structure is obtained with improved mechanical properties.

It is very difficult to avoid the quenching effects or inadvertent strain during the removal of material from the quenching bath before aging. It may be possible to neutralize the effects of the dislocation if alloying elements are present to segregate to dislocation at relatively low temperature [G. Thomas, 1955]. The preferential nucleation of intermediate or equilibrium precipitates is especially important in the production of the aluminum-copper alloy sheet.

Such a material is normally distorted by the internal stresses generated during quenching after solution treatment. If it is deformed plastically, by a few percent before aging in order to produce an even uniformly flat sheet, then under these conditions, the deformation (strain) may vary in different parts of the product. A further important consideration is that, because of difficulties of organizing a smooth uniform flow of material through the production schedule, varying times may elapse between the different processes, for example between the solution quenching and deformation or between deformation and aging are out of our scope of work in this paper.

These factors all influence the aging response, as will be seen in this work. It may be concluded therefore, that it is not sufficient to specify simply the final aging treatment for attainment of given properties and ignore the possible changes which may already have occurred as a result of room temperature prior aging or deformation. In addition, the prior deformation may influence the kinetics of zones formation since, the vacancy concentration will be raised, but the morphology of zones is not affected greatly.

Now, if the aluminum – copper alloy is quenched and aged to harden it with G. P zones at room temperature, then heated for a few seconds at a temperature above the solvus as shown in the fig. (1) [V. Ozoline, 2001]. The alloy is softened again because the copper in the G.P. zones dissolves at that temperature above the G.P. zones solubility line. This is called retrogression.

Influence Of Aging Practices On TMT

The super-saturated solid solution obtained from the quenching process is unstable and suffers from decomposition on aging, which involves several stages depending upon the aging temperature and holding time. No decomposition of this solid solution was observed on natural aging, since there is no precipitation. In fact natural aging results in the formation of G.P. zones.

When the copper content is high and the aging temperature is low, the sequence of intermediate stages followed, are (G.P. 1, G.P. 2, θ´, θ) . When the alloy is aged artificially at (100-150ºC), an intermediate phase precipitate at the places where there is an increased concentration of solute atoms, the θ´-phase, which does not differ in composition from the stable CuAl2 – phase but has distorted tetragonal lattice [R.B. Nicholson, 1958].

If the alloy is held for long time at the above temperature or if the temperature is raised to 250ºC and higher, coagulation of the particles formed at the centers of decomposition was observed and the intermediate phase becomes stable [4]. This is called over-aging which leads to the transformation of the intermediate phase θ´ into the stable phase CuAl2 having a tetragonal lattice [R.B. Nicholson, 1958].

The aims of present work can be summarized as follows:

It is hoping to combine the toughness advantages of the homogeneous alloys with the high tensile and yield strengths of TMT alloys.

Investigation of the pre-deformation aging, amount of deformation, and final aging practices.

It is therefore, the above findings represent an attempt to improve these deficiencies in our knowledge to the response of the 2024 aluminum alloy plates to thermomechanical treatments.

EXPERIMENTAL PROCEDURE

The chemical composition of the material employed in this work (commercial 2024 aluminum alloy) is given in table (1). All commercial alloys were received in the form of 24mm plate. A homogenization practice of (12 hrs.) at (500±5°C) is performed on all set of samples. The homogenized plates were pre-heated to the specified warm rolling temperature and reduced by rolling to desired reduction rates, keeping the finishing rolling temperature always at above 50ºC. The rolling was performed on 250mm diameter rolls driven by 30 hp motor. The solution heat treatment was performed in an electric furnace by heating specimens to 490±5°C for two hrs. Then water-cooled (quenched), pre-aging and post-aging was done at various temperatures for various time as can be seen in table (2.2).

For preliminary study, four levels of warm rolling as well as rolling at room temperature namely (5, 10, 15, 20% WR) were employed following the pre-aging treatments plus a control run with no deformation. Final aging for all rolled stock (post aging) was conducted at (150 and 190ºC) for times between (0-100 hrs.).

Aging Treatments

After the solution treatment and quenching process, some specimens aged in the room temperature for a long period up to (100 hrs.) for hardening the alloy naturally and to obtain good mechanical properties.

To make the aging process faster than natural aging, the aging process is done at an elevated temperature range (130, 150, 170, 190, 210ºC) for different periods up to (100 hrs.).

Thermo-mechanical Treatments

The general plan for heat treatments and rolling practices is given in table (2):

Mechanical Properties Testing

Evaluation of which the best practice that followed to give the better mechanical properties occurs by examination a pre selected samples in hardness, tensile and fracture toughness as follows.

Hardness Testing

After the grinding and polishing of the specimens under appropriate conditions, the hardness test is done by using Brinall hardness test, where the load are (25kgf.) and (10mm) diameter steel ball.

For purpose of accurate readings, an average of three readings for each point on the curve.

Tensile Testing

Round bar miniature tensile specimen's with a 2.5mm diameter, 20mm gauge length was used for all tensile strength determination. The maximum specimen's length obtainable from the short transverse plate thickness employed for this work.

Fracture Toughness Test

Fracture toughness is a critical input parameter for fracture-mechanics based fitness-for-purpose assessments. Although fracture toughness can sometimes be obtained from the literature or materials properties databases, it is preferable to determine this by experiment for the particular material and joint being assessed.
Various measures of 'toughness' exist, including the widely used but qualitative Charpy impact test. Although it is possible to correlate Charpy energy with fracture toughness, a large degree of uncertainty is associated with correlations. It is preferable to determine fracture toughness in a rigorous fashion, in terms of K (stress intensity factor), CTOD (crack tip opening displacement), or J (the J integral); Standards exist for performing fracture mechanics tests, with the most common specimen configuration shown in Fig.(2) (the single-edge notch bend SENB specimen). A sharp fatigue notch is inserted in the specimen, which is loaded to failure. The crack driving force is calculated for the failure condition, giving the fracture toughness.

The most widely used fracture toughness test configurations are the single edge notch bend (SENB or three-point bend). It is generally recommended that these specimens be of full section thickness, to ensure that lower-bound fracture toughness values for the structure under consideration are measured.

Generally, the notch depth is positioned within the range 45-70% of the specimen width, W, giving a lower-bound estimate of fracture toughness, associated with high levels of constraint.

A notch is machined into the fracture toughness specimen, following which a fatigue crack is grown by applying cyclic loading to the specimen. The cyclic load has been applied by means of a servo-hydraulic machine.

Various parameters are measured during fracture toughness testing, including the load, ram displacement and loading rate (using load cells and displacement transducers).

In addition, the crack-mouth opening is measured using clip-gauges attached to knife edges at the crack mouth. Fracture toughness tests are performed in universal hydraulic test machines, generally using displacement control.

Typically, fracture toughness in terms of KIC, can be determined from a single fracture toughness specimen: K (stress intensity factor) is a stress-based measure, derived from a function which depends on the load (stress) at failure (or at the point where the test is stopped, or at the maximum load). K also depends on geometry (the flaw depth, together with a geometric function, which is given in test standards for each test specimen geometry) according to the following equation:

$$KIC = PQ \times L / (BW)^{3/2} \left(2.9 (a/W)^{1/2} - 4.6 (a/W)^{3/2} + 21.8(a/W)^{5/2} - 37.6(a/W)^{7/2} + 38.7(a/W)^{9/2}\right) \quad (1)$$

Where,

PQ: represent the yield bending load, L represent the distance between the two supports, a & W are the characteristics dimensions indicated in fig.(2), and B is the thickness of the specimen measured in meter length units.

The load-displacement diagram is recorded automatically, the scale of the coordinate axes of the recorder should be such that the tangent of the linear portion of the diagram that be within 0.7-1.5 [C.D. Beacham,1965]. The initial non-linearity that may be present at the beginning of the test is due to the friction between the loading pins and the grips.

Now, according to the above formula, and reported data from the test a sample of calculations can be reported as shown in table (3):

Microstructure Examinations

A micro structure usually reflects the origin and history of the product and frequently used to predict or explain characteristics and performance. The microstructural characteristics of the specimens were based on metallographic examinations. The specimens were taken in the rolling direction. All specimens were ground and polished by using a grinding paper in different degrees (120, 320, 500, and 1000 micron). Rough and final polishing were performed by using a Nylon cloth for rough polishing and using Sevyt Cloth for final polishing, each step in polishing is accomplished by using a diamond paste and alumina solution respectively. The specimens were etched using (0.5% HF + 99.5% H_2O)solution Standard metallographic examinations using optical microscopy (Neophot-1) type was used to reveal the alloys structures.

RESULTS AND DISCUSSION

TMT is applied to various samples in order to produce high strength allied to high fracture toughness products, it is based on metallurgical processes which exploit to the full the strengthening mechanisms such as precipitation hardening, dissolution hardening and grain refinement by means of defined deformation and temperature conditions in order to ensure the best properties of the plate in the as rolled conditions.

Recently, the major improvement in the high strength 2000 series aircraft aluminum alloys has been in the fracture toughness. Also increased stress corrosion resistance has been achieved by over aging with consequent loss in strength. It is generally agreed that TMT offers a means of achieving high strength with good fracture toughness.

In the present study, it has been aimed to combine the toughness advantages of the homogeneous alloys with the high strength of TMT process. Realistic property goals for material up to 24 mm in thickness where equal or greater than 495.4 Mpa for yield strength and equal or greater than 528.9 Mpa for ultimate strength and plain fracture toughness of equal or greater than 32.3 Mpa.√m as can be seen in fig. (10).

Influence Of Preaging Temperature On TMT

Although the aging response of the alloy used in this paper is well known and has been reported previously, in many works, it was found necessary to have an overall hardness testing survey of the alloy in order first to report our own results from which the best aging temperature and time can be derived. Fig. (8) shows the accumulative results for a range of aging temperatures are (25, 130, 150,170, 190 and 210ºC). The results as expected showed the very long times needed for the lower aging temperature due to the limited diffusion and hence limited Gunier-Preston zones precipitates.

Within the high aging temperature i.e. above 150ºC, diffusion is higher which resulted in more extensive second phase precipitate Mg2Cu6Al5 as the x-ray diffraction analysis shows in fig. (4) for an alloys at the peak hardness conditions where aged at 150°C and hence resulted in higher peak hardness with shorter aging time. However aging at 210 ºC considered being a high aging temperature resulted in lower peak hardness at very short aging time with a clear over-aging response of the alloy. This is so expected since at this aging temperature an accelerated second phase precipitates takes place due to the high diffusion rate which consequently accelerated coarsens the precipitates.

The as homogenized microstructure of alloy was examined optically by using the recommended etching solution remembered above. The structure consist of light gray particles of insoluble CuFeMnAl6 , and undissolved CuMgAl2 that precipitate during homogenization as can be seen in fig. (5)

Influence Of Cold Rolling At Peak Hardness On TMT (TMT1-Practices)

The post aging response of the 2024 commercial aluminum alloy that rolled at room temperature after preaging to peak hardness is shown in fig. (6). This figure and fig. (10) indicate that these conditions resulted in fairly high hardness, yield strength and tensile strength, but noticeably low fracture toughness value. Fig. (7) also shows the microstructure of this alloy in these conditions which reveals a massive planar slip.

This observation is compatible with the fact that when rolling this alloy in these conditions, dislocations density is concentrated at the active slip planes which resulted in a non homogeneous planar slip illustrated in fig. (7). This effect has been observed by other workers [10] for Al-Zn alloys. The planar slip resulted from the cutting of coherent and semi coherent precipitates which is present in the peak hardness conditions. Therefore, this rolling scheme, although improves mechanical properties; and it markedly reduces the fracture toughness because of the planar non homogeneous slips which form an easy crack propagations path. Accordingly, there are no advantages gained from this condition. Unless the fracture toughness of the alloy is not considered essential in the design and applications of the alloy which is not related with our applications in aircraft structure.

Influence Of Warm Rolling At Peak Hardness Conditions On TMT (TMT2 - Practices)

As mentioned earlier, the warm rolling scheme was chosen to be at 200ºC which is an appropriate temperature since; it is higher than the solvus line of the G.P zones and much lower than the recrystallization temperature 420ºC of 2024 aluminum alloys [Alan Cottrel, 1980]. Another point is that from fig. (3) one can appropriately choose any of the preaging temperature in the range of 150-190ºC to obtain peak hardness condition at relatively short aging time. However, preaging temperatures of 150ºC was selected on the basis that at this temperature mostly coherent G.P zones or semi coherent precipitate is predominantly present at this condition, and there is little possibility of incoherent precipitates to be present [A. Deshamps, Y.Brechet,1999]. This is needed for faster partial re-dissolution of the fine coherent precipitates when reheated for warm rolling.

The reasons behind choosing the TMT2 conditions where described in table (2) based on the fact that, the initial warm rolling step must impart a homogeneous distribution of dislocation density which must yield better mechanical properties and better fracture toughness. This is what the result of fig. (10) indicates. These results illustrate that correctly chosen thermomechanical processing can developed yield and tensile strength equal or better than the peak hardness, yield and tensile strength with improved fracture toughness (comparing data in fig. (10) for various conditions.

A survey of the amount of warm rolling was performed as shown in fig. (8) to obtain the best amount (percentage) of rolling. The response of post aging for (5, 10, 15 and 20%) warm rolling as illustrated in figures above respectively clearly shows that warm rolling increase the kinetics of precipitation during post aging treatments as indicated by the early obtaining of peak hardness at less aging times.

Further more, the higher dislocation densities as the deformation increases will enhanced the resulted peak hardness, also the aging time to reach peak hardness for the standard aging specimen (without rolling) to about (8 hrs.) for 5% rolling, (3 hrs.) for 10% rolling, (1.5 hrs.) for 15% rolling and (1 hrs.) for 20% rolling percentages.

The sequence of increase of peak hardness is according to the following: (130 BHN for 0% rolling), (136 BHN for 5% rolling), (140 BHN for 10% rolling), (148 BHN for 15% rolling) and (150 BHN for 20% rolling). It is interesting to note that the fracture toughness values for all the specimens in figure (3.6) except for 20% rolling where above (25.5 Mpa√m) with the highest value of (32 Mpa√m) associated with 15% rolling. The lowest value of (25.5 Mpa√m) was super singly obtained with 20% rolling which was neglected therefore the warm rolling scheme at 15% level was found to be generally successful for the alloy studied in this work. In this context more investigations is needed to explain the low fracture toughness associated with rolling at 20% or may be above.

The origin of these improved properties was sought in terms of the micro-structure responsible for the observed mechanical properties as shown in fig. (9). The above figure shows the optical microstructures of the alloy, an extensive elongation of grains in the rolling direction is evident without planar slip which is inferior to fracture toughness, that figure are compared with the fig. (5) which shows the annealed conditions of the alloy. From fig. (5) for (15% rolling), it can be expected that a uniform distribution of dislocations exists in the microstructure as the mechanical properties implies. The precipitate remaining from the prior aging treatments seems to be sufficient to pin the dislocation generated during the rolling step in the subsequent aging (post aging). This prevents the recovery and loss of dislocations density required to impart good properties to the TMT products.

The choice of (200ºC) as the rolling temperature seem to be correct since it is above the GP zones solvus line were a transitions from a planar to random slip occurs. However, more detailed investigations about the range of temperature of rolling could be conducted in the future to gain more information's about the possible effects of rolling temperature above or below (200ºC) on mechanical properties and fracture toughness.

The primary advantages of the TMT process at an elevated temperature is that it resulted in homogeneous distribution of dislocations and also, it greatly increased the kinetic of the age hardening process, therefore a decreased time in the post aging step is needed. The TMT condition then is seen to consist of microstructure with a proper precipitate size with super imposed high dislocation density uniformly distributed throughout the microstructure.

CONCLUSIONS

Good combinations of fracture toughness, strength and hardness for a homogeneous alloy could be obtained by applying a set of TMT practices to the commercial 2024 aluminum alloys that used in aircraft structure.

The choice of correct initial conditions, such as preaging temperature 150ºC, amount and temperature of deformations 15% at 200ºC, temperature and time of post aging (150ºC for 3 hrs.) sequence was found to be crucial in producing the plate of 2024 aluminum alloy with a high mechanical properties and fracture toughness.

The room temperature deformation at the peak hardness (TMT2) resulted in a planar slip and non-uniform response on the subsequent post aging (yield stress = 448.1 Mpa, ultimate stress = 480.2 Mpa, elongation = 9, KIC = 24.2 Mpa √m).

The planar slip that mentioned above could be avoided and could obtained a homogeneous deformation by deforming the alloy at elevated temperature 200ºC in the peak hardness conditions.

Deformations the alloy at an elevated temperature at the peak hardness(TMT2) conditions was found to be generally successful and it is expected to produced a uniform distribution of dislocation with better mechanical properties (yield stress = 455.4 Mpa, ultimate strength = 500 Mpa, elongation = 11, KIC = 32.3 Mpa√m).

In the TMT23 practices, it seems that, the precipitates which remaining from the prior aging treatments are sufficient to pin the dislocations during subsequent post aging treatments.

The recommended temperature for the deformation treatment has been found to be at or just above the GP- solvus temperature (about 200ºC), were a transition from planar to random slip occurs.

The amount for warm rolling has been found to be 15%.

Table (1) Chemical Analysis of commercial 2024-Aluminum alloy

Alloy designation	Cu	Mg	Mn	Fe	Si	Cr	Al
2024	4.5	1.25	0.58	0.53	0.37	0.02	Balance

Table (2) Plan of thermo-mechanical Treatments

Practice No.	**Alloy Designation**	**Details***	
1	T6	T61	25-100
		T62	130-100
		T63	150-60
		T64	170-50
		T65	190-40
		T66	210-10
2	TMT_1	TMT_{11}	150-20-CR5-190-30
		TMT_{12}	150-20-CR10-190-30
		TMT_{13}	150-20-CR15-190-30
		TMT_{14}	150-20-CR20-190-30
3	TMT_2	TMT_{21}	150-40-WR5-190-30
		TMT_{22}	150-40-WR10-190-20
		TMT_{23}	150-40-WR15-190-3
		TMT_{24}	150-40-WR20-190-2

A code was used to distinguish each process in the table above, this code, included the samples group number, the preaging temperature in degree centigrade – the preaging time in hours-the method and percent of working-post aging temperature in degree centigrade – postaging time in hours.

Table (3) Fracture toughness values (sample of calculations)

Parameter	a (m)	W (m)	B (m)	H (m)	L (m)	P_Q (N)	K_{IC}(Mpa m½ (calculated)
Value	0.01	0.04	0.02	0.176	0.15	23120	28.2

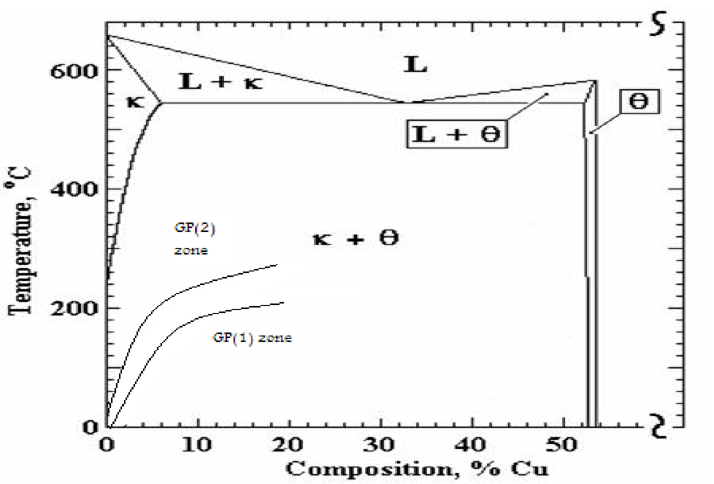

Fig.(1):Part of equilibrium diagram for Al-Cu alloys shows the Meta-stable phase boundaries for G.P. zone