MECHANICAL AND MICROSTRUCTURAL PROPERTIES OF 8090 AL-LI ALLOY WELDED JOINTS

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Abstract

The interest on the class of Al-Li alloys is related to the fact that Li content decreases the density and increases the Young modulus. These two features make the Al-Li alloys an attractive alternative to the other more usual conventional Al-alloys and composite materials. In this paper, microstructure and tensile properties of the 8090 Al-Li alloy, welded using a 5% Mg filler metal with a gas tungsten arc welding (GTAW) were studied. Experimental work included tensile tests, optical metallography and Scanning Electron Microscopy (SEM) fractography on both welded and unwelded tensile specimens. The values of mechanical properties are compared with those of unwelded specimens.

Riassunto

INTRODUCTION

Al-Li alloys represents a class of high performance, lightweight alloys that can be utilized for welded constructions, in all applications where reduction of weight is a crucial point in the selection of structural materials [1]. As well known, the effectiveness of Li is related to the fact that every 1% in weight of Li in Al-alloys decreases the density by about 3% and increases the Young modulus by about 6% [2]. These two features make the Al-Li alloys an attractive alternative to the other more usual conventional Al-alloys. Moreover, reduced density and increase of Young modulus can make these alloys economically competitive also in comparison with composite materials, which on the other hand exhibit more convenient weight, but higher production costs.

Initially, most of Al-Li alloys were conceived for riveted or bolted joining. However welded joints give rise to weight savings and less manpower during manufacture. In order to extend the use of welding during fabrication, weldability studies are being performed and more easily weldable variants of alloy are developed. Weld solidification cracking is an issue common to many Al-alloys, depending on relationship between solidification temperature range, alloy composition and nature of eutectic constituent formed during the last stages of solidification [3]. Generally cracks originate along grain boundaries; here low melting liquid films, containing impurities, may crack under thermal stresses due to large solidification range and high thermal expansion coefficient [4]. As the solute approaches the maximum solubility limit, the solidification temperature range and the portion of eutectic constituent at the grain boundaries increase. Under non-equilibrium, such as during welding, the maximum combined effect occurs below the maximum solubility limit (~4% wt). For instance, crack susceptibility for a binary Al-Li alloy is evaluated to be maximum, when Li is around 2.5 % wt [2]. When Li is over the solubility limit, the solidification temperature range becomes shorter and more amount of eutectic constituent is available in the last stages of solidification to fill and "heal" all cracks formed, if any. In such a way, in both the cases of low and high Li content (away from the most critical point of 2.5 %), the risk of forming cracks should be less. Unfortunately many Al-Li

alloys, such as the alloy 8090 here considered, are designed with the primary purpose of enhancing mechanical properties, thus nominal composition can result in a peak in cracking susceptibility.
Weld porosity constitutes another particular problem for Al-Li alloys, because these alloys are more susceptible than other Al-alloys to weld soldification cracks due to the hygroscopic nature of Li-containing aluminium oxides. Thus a decrease of tensile strength can be attributed to the presence of these flaws. The edges to be welded should be submitted to an appropriate cleaning procedure or, better, to a mechanical milling, by removing a layer of about 0.1 mm. This practise can reduce remarkably the presence of porosity [2].
In this paper, microstructure and tensile properties of the 8090 Al-Li alloy in the condition T3 (solution heat treated, strain-hardened and naturally aged), welded using ER-5556 (5% Mg) filler metal with a gas tungsten arc welding (GTAW), were studied. This alloy is a typical one utilized for aviation welded constructions.

MATERIALS AND METHODS

The material utilized in the welding trial is one version of Al-Li 8090 alloy, in the condition T3, solution heat treated, strain-hardened and naturally aged. The nominal composition of base metal is shown in Table 1. The material was supplied as a plate, 4 mm thick. Two plates of 150x500 mm were butt welded with 60° V-groove edge preparation, 1 mm shoulder, 1.5 mm gap.
Manual gas tungsten arc welding (GTAW) process in a single pass with variable polarity. Since the best results are often achieved by utilization of filler metals series 5000 (Mg-alloys), a 5356 filler metal (5%Mg) was used. The nominal composition of filler metal is shown in Table 2. The other welding parameters were: current 120 A; voltage 21 VAC; with a welding speed of about 7 mm/s.
Specimens for tensile strength tests, hardness survey and optical micrography were prepared by cutting weldment normally to welding direction. The specimens were ground and polished by usual techniques by abrasive papers and alumina slurry. Hardness and macrographic specimens were etched using Keller's reagent. Microstructure of WM and HAZ were observed through optical microscopy.
Mechanical properties were determined through tensile tests on both welded and unwelded samples. Tensile specimens were cut of rectangular section of about 12.5x4 mm and reference length of 50 mm. After tensile tests, broken edges were submitted to fractographic examination by Scanning Electron Microscope (SEM) type JEOL JSM 35 CF, with an accelerating voltage of 15 kV.

TABLE 1. NOMINAL COMPOSITION OF 8090 –T3 AL-LI ALLOY ELEMENT CONTENT WT %

Si	Fe	Cu	Mg	Li	Cr	Zn	Zr	Mn
0.4..0.8	0.30	1.0..1.6	0.6..1.3	2.2..2.7	0.10	0.25	0.04..0.16	0.10

TABLE 2. NOMINAL COMPOSITION OF 5556 FILLER METAL AL-MG ALLOY (AWS A5.10-69).

ELEMENTI CONTENT WT %

Si+Fe	Mn	Cu	Mg	Cr	Zn	Ti
0.40	0.50..1.00	0.10	4.70..5.50	0.05..0.20	0.25	0.05..0.20

TABLE 3. UNWELDED SPECIMENS TENSILE TESTS (BASE METAL)

Specimen N.	Reference Area (mm^2)	Reference length (mm)	Elongation %	Yield Strength (MPa)	Tensile Strength (MPa)
1	52.1	50	18.3	209.6	324.7
2	53.1	50	19.9	210.3	340.5
3	52.7	50	18.4	209.9	338.0

RESULTS

Tensile tests results on a number of both base metal and welded specimens are reported on Table 3 and 4 respectively. No significant differences were revealed on tensile properties between specimens broken after 30 or 60 days after welding.
Values of tensile properties on welded specimens are quite lower than those of unwelded specimens, since both tensile strength and yield strength decreased down to about 2/3 of initial values. Percent elongation underwent a very remarkable reduction (from ~19% down to ~5%). All of welded specimens broke of the WM.
A representative profile of hardness survey across the welded section is reported on Fig.2. Moving from fusion boundary, HAZ hardness first decreases gradually from 85 HV at the interface down to 68 HV at 4 mm of distance. This minimum

value is kept along ~2 mm; from here along 3 mm hardness increases to reach initial value of base metal (105 HV). The total width of HAZ is equal to ~ 9 mm. Hardness of weld zone is quite variable between a minimum of 61 HV near the weld center line and 92 close to fusion boundary. Minimum value is ~ 58% of base metal value.
The microstructure of WM and HAZ are shown at variable magnification in Fig. 3a,b,c.
As usually observed in WM of such alloy, non-equilibrium weld solidification results in segregation of some eutectic at the grain boundaries. Thus, precipitation aging is limited, since most of elements needed for precipitation reactions were taken away by eutectic phase. Due to the hygroscopic nature of the Li-alloy, a noticeable amount of fine sized (~10 mm) porosity is present.
In fig. 4 are reported some micrographs obtained by SEM observation of fracture surfaces of tensile specimens. On unwelded specimens, the overall aspect (Fig.4a-b) show a mixed brittle+ductile areas, exhibiting cleavage and dimples-shaped appearance respectively. Cleavage aspect, with sharp planar facets characterizes low-energy (brittle) fracture, occured along specific crystallographic planes. On the contrary, the presence of coalescence of microvoids, formed around inclusions, which after failure give rise to the formation of semi-cavities or dimples.

TABLE 4. WELDED SPECIMENS TENSILE TESTS

Specimen N.	Reference Area (mm^2)	Reference length (mm)	Elongation %	Yield Strength (MPa)	Tensile Strength (MPa)
1	50.81	50	4.3	140	232
2	50.05	50	4.3	140	219
3	50.55	50	5.1	138	225
4	50.79	50	4.9	142	243
5	50.60	50	5.0	141	230

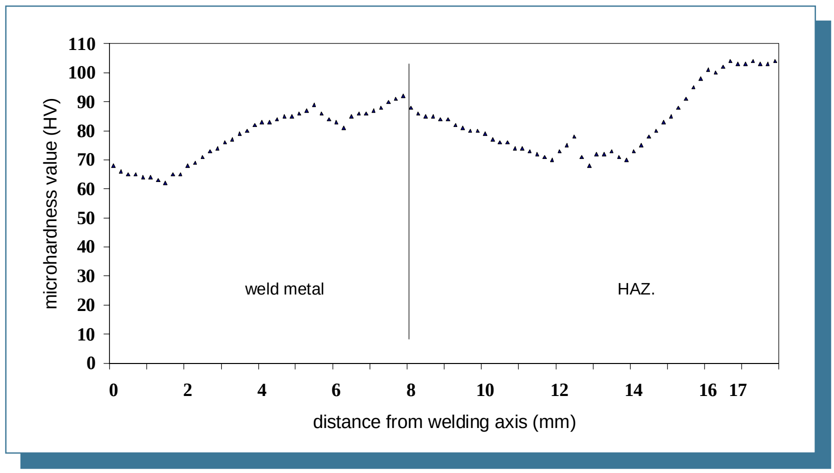

Fig. 2: Hardness profile on welded section

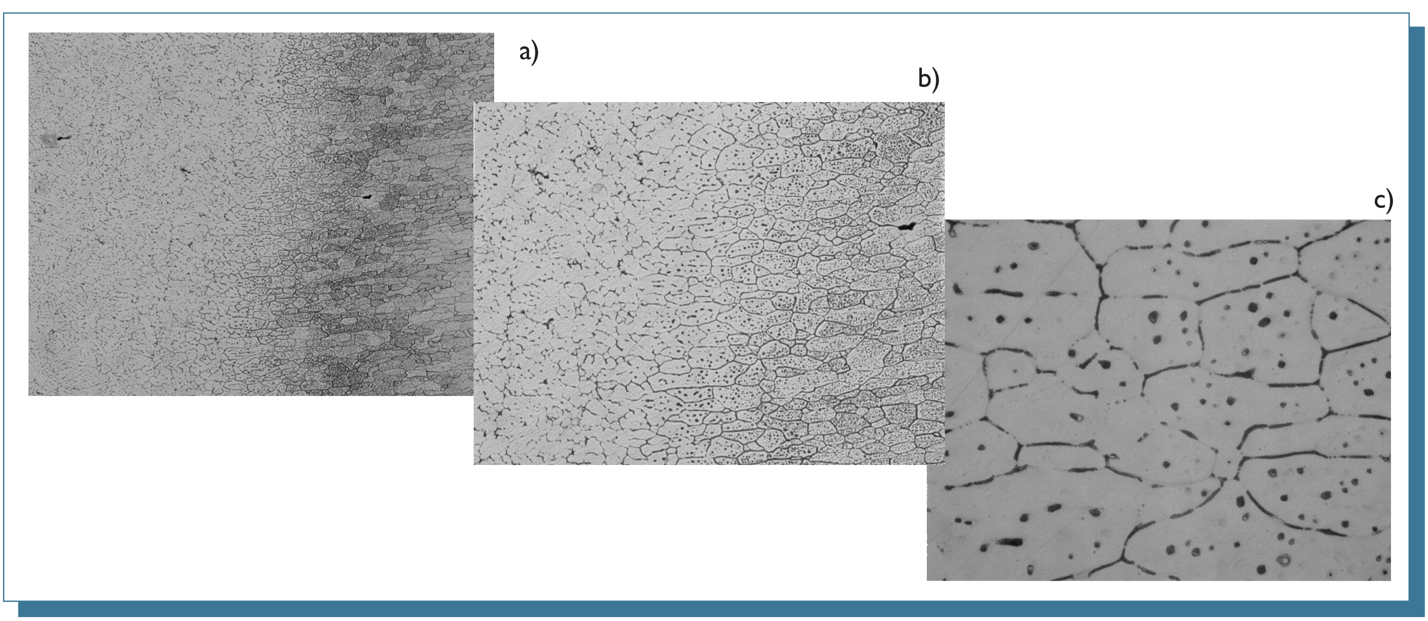

Fig. 3: – Microstructure of weld: a) WM/HAZ overall view, 50x, b) WM/HAZ interface, 100x
c) Detail of WM near the interface, 500x

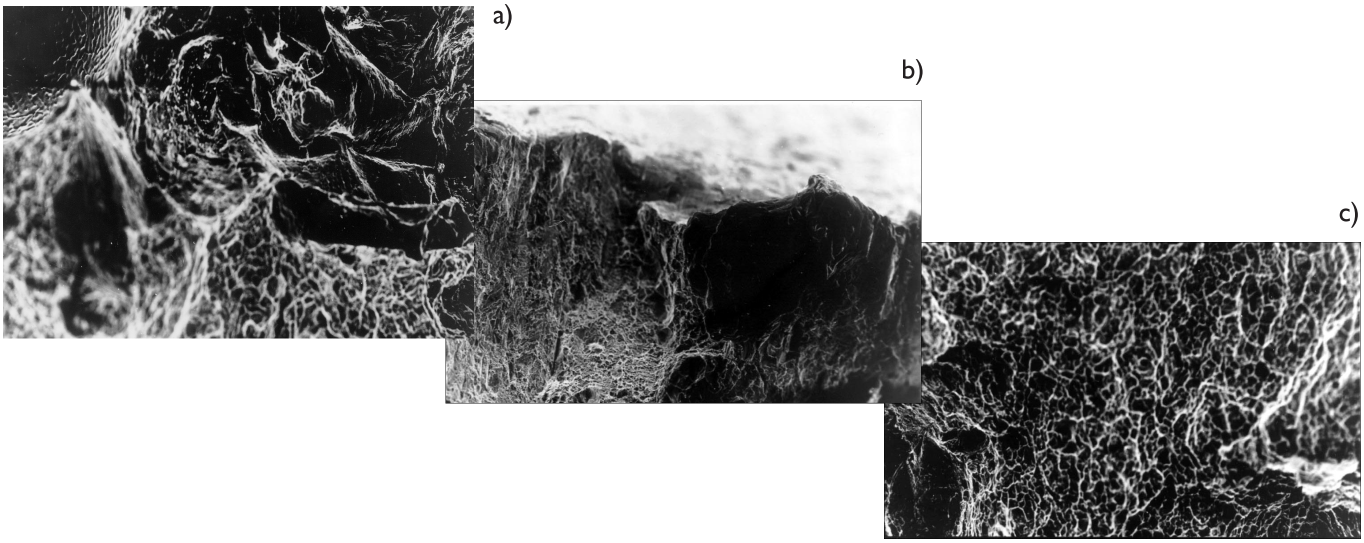

Fig. 4: SEM appearance of fractured surfaces: a) Base metal, 30x,central region of fractured section;
b) Base metal, 30x,at the boundary of fractured section; c) WM on welded specimen, 30x

Welded specimens, the fractured surface inside the WM (Fig.4c) have an appearance similar to that observed on unwelded specimens, but the ductility of the fusion zone is quite lower (see elongation values), also due to the presence of porosity and some microcracks.

DISCUSSION

Let us consider first the base metal. As results from Al-Li diagram (Fig.1)[2], strengthening of this alloy during fabrication is possible, due to the limited solid solubility of Li in Al (4% at 600°C, less than 1% at room temperature). The primary strengthening constituent is the metastable d¢ phase, having a composition Al_3Li, which is coherent with the matrix phase (the lattice parameter of d¢ is very close to that of matrix and low misfit strain arises). During aging after a solution treatment, during aging after a solution treatment, δ′ precipitates in form of spherical particles. If overaged (at temperature over about 300°C) δ′ dissolves and subsequently the stable phase d (Al-Li) precipitates. The precipitation sequence is likely to occur as follows:

$$\alpha_{solution} \rightarrow \alpha + \delta' + \delta \rightarrow \alpha + \delta$$

Since the alloy 8090 contains also Cu and Mg, other possible strengthening phases are δ′ (Al_2CuMg) and T_1 (Al_2CuLi) [5].
The complete precipitation reactions can be described as follows:

$$\begin{aligned} &\rightarrow \alpha + \delta' + \delta \\ \alpha_{solution} &\rightarrow \alpha + S'(Al_2CuMg) \\ &\rightarrow \alpha + T_1(Al_2CuLi) \end{aligned}$$

The precipitation sequence $\mathrm{GP\ (Guinier-Preston)\ zones} \rightarrow \theta'' \rightarrow \theta'$ $\rightarrow \theta$ typical of alloys Al-Cu [6], is unlikely to occur in the 8090 alloy, due to low ratio Cu/Mg [7]. In the T3 condition, straining before aging of alloys provides new heterogeneous nucleation sites and promotes T_1 and δ′ precipitation.
After welding we have to consider the fusion zone, where melting and resolidification occur, and a zone deteriorated by the heat of welding process. In the fusion zone the weld metal formation is primarily depending on solidification conditions, that is solidification growth rate, temperature gradient and diffusion. Precipitation aging is limited by the formation of an eutectic constituent, at the end of solidification, which takes away of the matrix most of elements needed for precipitation reactions.
In the HAZ, the microstructure is mainly governed by two types of solid state reactions:
- coarsening of particles located in zones submitted to lower peak temperatures
- precipitate solution in zone submitted to higher peak temperatures.

This results in a quite complex evolution of structure, when in a same portion of material some precipitates can be subject to dissolution, some others to coalescence. In particular, with regard to the distance from fusion boundary, the dissolution of precipitates in the zone near the fusion boundary, reaching more elevated peak temperatures, gives rise to an increase of tensile strength, both for solid solution strengthening and precipitation during cooling or natural aging. In the portion where the peak temperatures are lower, the dissolution effects are reduced; instead, a reduction of tensile strength due to coarsening is likely to occur.
According to the results given by theoretical

models [8], the microstructure of HAZ, starting from fusion boundary interface, is likely to be constituted as follows:

1) A 1st zone where dissolution of precipitates (such as δ′ phase) and partial or complete resolubilization of elements occurred, ~4 mm wide, with estimated peak temperature between 300 and 550 °C; due to the effects of both solid solution strengthening and natural precipitation, hardness reaches values higher than in the next zone, more distant from interface;
2) A 2nd zone where no resolubilization occurs and the effects of coarsening (growing of δ′, S', T_1 phases, etc.) are most pronounced, ~2 mm wide, with estimated peak temperature between 200 and 300 °C ; here minimum hardness (68 HV) is measured at ~ 5 mm from interface;
3) A 3rd zone , ~3 mm wide, where alloy, having reached an estimated peak temperatures in the range of 120°-200°, is overaged and microstructure underwent a coarsening, but the degree of weakening reduces with increasingly less coarsening of phases, as the peak temperature decreases. The hardness rises monotonically up to the value of unaffected metal (105 HV).

Thus thermal cycles in the HAZ give rise to undesired metallurgical modifications. In this zone a final post-weld heat treatment (solution treatment+aging) could be useful to recover the initial properties, but is not easily practicable on large components. Moreover the minimum mechanical resistance is obtained in the WM, where the joint efficiency is mainly determined by the metallurgical constitution and solidification conditions.

If compared to common values achieved in the as-welded condition (60%), the overall joint efficiency (ratio of weld strength to base metal strength) appear satisfactory, since both tensile strength and yield strength decreased down to about 2/3 of initial value.

CONCLUSIONS

The weldability of the 8090 T3 Al-Li alloy, as usual for many Al-alloys, is associated with the issue of reduced tensile strength and ductility in the HAZ, and by the tendency to porosity formation and susceptibility to solidfication cracking in the WM. The joint efficiency, assessed as ratio of weld strength to base metal strength, is resulted equal to about 0.67 on as-welded specimens. As described above, thermal cycles in the HAZ give rise to undesired metallurgical modifications. A post-weld heat treatment of solution treatment+aging, which could recover partly the initial properties, is not easily practicable.

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