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Abstract

A preliminary experimental comparison of the behaviour of aluminium and magnesium alloys subjected to Liquid Hot Isostatic Pressing (LHIP) is proposed. The two metals melt at approximately the same temperature. However, as a consequence of a larger deformability of magnesium at elevated temperatures, the choice of LHIP parameters – and especially the temperature at which the pressure is applied – in the present exploratory case was constrained to values far smaller than those one would like to select in order to improve the ultimate tensile stress and the elongation to fracture.

Riassunto

In questo lavoro viene proposto un confronto sperimentale tra i comportamenti di leghe di alluminio e di magnesio sottoposte a Liquid Hot Isostatic Pressing (LHIP). I due materiali presentano temperature di fusione vicine. Comunque, a causa della maggiore deformabilità ad alta temperatura del magnesio, in queste prove preliminari la scelta della temperatura di applicazione della pressione per LHIP è limitata a valori molto più bassi di quelli che si desidererebbero scegliere al fine di incrementare lo sforzo a rottura (UTS) e il corrispondente allungamento.

OBJECTIVE

The objective of this paper is to explore the possibility of applying Liquid Hot Isostatic Pressing (LHIP) – a technology recently developed by Teksid Aluminum – to magnesium alloys: in fact the LHIP turns out to be very effective for the improvement of the mechanical behaviour of aluminium

alloys [1] and, in principle, should be even more appealing for magnesium alloys, usually characterized by a much higher hydrogen concentration, resulting from water vapour dissolved in the ingot.

MECHANICAL TESTS

A set of 8 sand cast samples of the magnesium alloy QE22A was heat treated and then subjected to standard tensile uniaxial tests (Fig. 1a). The samples, reproduced in Fig. 1b, had a diameter of 10mm and 14mm in the clamping zones. All the specimens underwent a solubilization stage, at 530 °C for 6 hours, then quenching in water at 65 °C and ageing for 10 hours at 204 °C. Two specimens were then subjected to LHIP, at the temperature of 390 °C and at the hydrostatic pressure of 101 MPa for 40 seconds, while other two specimens were subjected to LHIP at 412 °C and 105 MPa for 40 seconds. For the 4 remaining specimens the treatment ended with the ageing stage and was not followed by LHIP. Thereafter all the specimens underwent a tensile test, in which the Young modulus E and the ultimate stress σ_{max} were determined. The results of the tests are presented in Table 1.

TABLE 1. RESULTS OF THE TENSILE TESTS.

QE22A specimen	E [GPa]	σ_{max} [MPa]
non LHIPPed	45	269
LHIP (390 °C)	46	262
LHIP (412 °C)	45	270

In Table 1 the elastic modulus E is evaluated by measuring the strains with a biaxial extensometer (see Fig. 1a). The failure mode, as evidenced by the shape of broken samples (see Fig. 1b), is the same for all of them.

Within the experimental uncertainties and the intrinsic dispersion of material properties, the results obtained from the three sets of specimens are identical. Also the stress-strain curves can be almost perfectly superposed (see Fig. 2) and, accordingly, the $\sigma_{0.2}$ proof stresses (~ 185 MPa) are almost identical; the elongations to fracture ($\sim 9.4\%$) are also very similar.

In other words, with the adopted parameters (temperature of 390 °C or 412 °C, pressure of 101 MPa or 105 MPa, for 40 seconds) the LHIP treatment turns out to be ineffective.

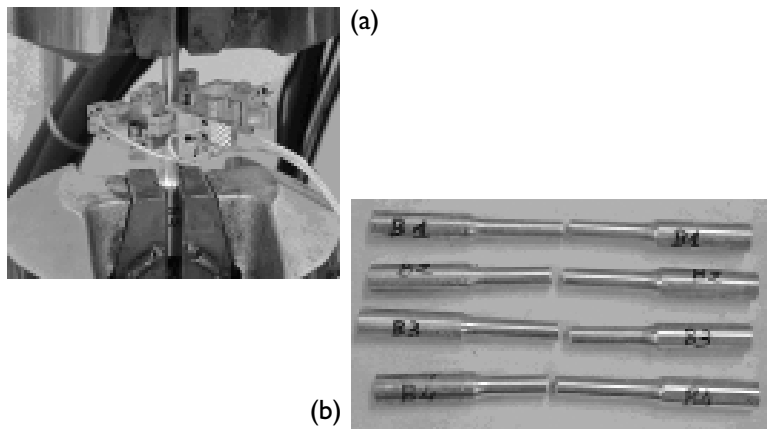


Fig. 1: (a) Tensile test with biaxial extensometer; (b) failure mechanism.

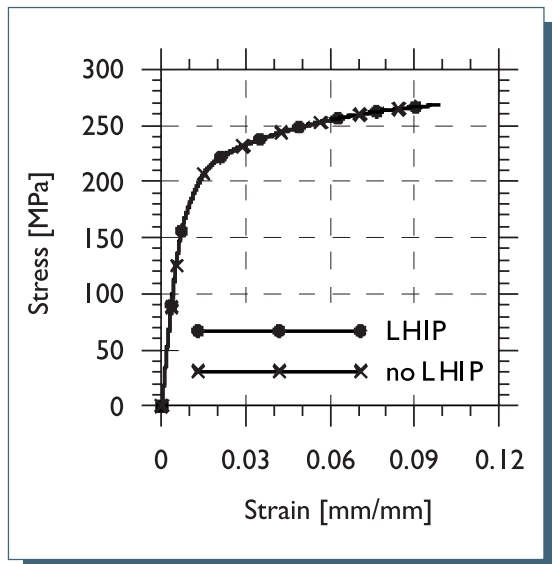


Fig. 2: Typical stress-strain curves for a LHIP and no-LHIP sample.

ANALYSIS OF RESULTS

The outcome of the mechanical tests is analyzed here, following the scheme adopted to analyze the LHIP treatment in the case of aluminum alloys, in which LHIP turned out to be effective [1].

The relevant parameters of the two metals and of the LHIP process are summarized in Table 2. The self diffusion coefficient D_v plays a significant role. Beside a pre-exponential coefficient D_0 , it exhibits an exponential dependence on the activation energy for vacancy diffusion E_a , and on the thermal energy kT , where k is Boltzmann's constant: $D_v = D_0 \exp(-E_a/kT)$. In the case of the aluminum alloys it was proposed that the LHIP process occurs in two stages: in stage I the pores shrink promptly by plastic deformation. In stage II the self diffusion of metal atoms causes some further

TABLE 2. PROPERTIES OF THE ALUMINIUM [1] AND MAGNESIUM [THIS WORK] ALLOYS

Property	Al (AA356)	Mg (QE22A)
Ideal density [kg/m^3]	2700	1740
Melting temperature T_m [K]	933	922
$\sigma_{0.2}$ proof stress at room temperature [MPa]	240	185
Vacancy volume Ω [10^{-30} m^3]	12	17
LHIP temperature T_{LHIP} [°C]	520	390 / 412
LHIP pressure [bar]	115	101 / 105
LHIP time [s]	35	40
T_{LHIP}/T_m	0.85	0.72 / 0.74
Coefficient D_0 (see text) [$10^{-4} \text{ m}^2/\text{s}$] [2]	2.25	1.75
Activation energy E_a (see text) [kJ/mole] [2]	144	138
Self-diffusion coefficient D_v at T_{LHIP} [$10^{-15} \text{ m}^2/\text{s}$]	68 *	5

* In Ref. [1] a value of $6 \cdot 10^{-13} \text{ m}^2/\text{s}$ was erroneously reported. However, the main conclusions – also justified by footnote (2) at p.10 – remain practically unaltered.

pore shrinking, and more importantly, a sort of welding of the folded surface of collapsed pores. In stage I the local stress, determined by the applied external pressure (~100 MPa) turns out to be significantly larger than both the yield stress at LHIP temperature (below 5 MPa) and the estimated stress needed to cause yielding of the supposedly spherical shell surrounding each pore (about three times the yield stress, for the density observed at the end of the yield stage [1]). In stage II the drift velocity for atoms diffusing from a grain to an adjacent collapsed pore, and causing a partial 'healing', is (see [1])

$$\langle v \rangle = \left[14 \frac{\sigma \Omega}{k T} \right] \left[\frac{D_v}{G} \right]$$

As noted above, the kinetics of densification depends exponentially on temperature, through the diffusion coefficient D_v . With the values of table II, in aluminum $\langle v \rangle \sim 2 \cdot 10^{-9}$ m/s, implying, for a 35 s duration of stage II, a migration length of the order of 0.07 μm , accounting for the disappearance of the dendritic appearance of the fracture surfaces. In the case of magnesium, at both LHIP temperatures (390 °C and 412 °C) the yield stress is low enough to allow for the plastic pore shrinkage. The estimates of the parameters controlling Stage II are shown in Table II for

the temperature of 412 °C, the more favourable one. They imply a migration velocity $\langle v \rangle$ of about $2.5 \cdot 10^{-10}$ m/s, meaning, for a 40s duration of stage II, a migration length of the order of 0.01 μm , quite inadequate to cause a significant healing of squashed pores. A migration length comparable to that achieved in aluminium would require either a LHIP time longer by an order of magnitude, which might be unacceptable for industrial operation, or a LHIP temperature around 480 °C. The compatibility of such a temperature with the geometrical stability of the sample undergoing LHIP has to be experimentally assessed: the applied pressure field for technical reasons will never be ideally isostatic or isotropic.

In conclusion, this analysis confirms that, due to the strong temperature dependence of diffusion phenomena, the efficiency of the LHIP treatment strongly depends on the LHIP temperature, and that the achievement of good results with economically viable treatment times requires LHIP temperatures closer to the melting temperature.

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