Some Results of Examination of Fractures of Aluminium Cast Alloys, Magnesium Alloyed by Means of Fractography

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In order to achieve satisfactory strength and ductility in cast aluminium - magnesium alloys, alloyed with 5 % ma= gnesium, it is necessary to avoid the formation of larger β' precipitates, by adopting the appropriate procedure of heat treatment. The intrinsic strength of the β' structure is more than that of the aluminium matrix. However with the growth of certain larger precipitate particles at the ex= pense of neighboring smaller particles, upon continued ov= eraging, the number of sites interfering dislocation moti= on decreases to the possible extreme, that the hardness of a completely overaged alloy can be less than the hardness of the supersaturated solid solution, which existed prior to the aging heat treatment.

The general sequence of a typical precipitation - harde = ning system :

supersaturated solid solution $\rightarrow$ transition structure $\rightarrow$ $\rightarrow$ equlibrium phase

can be presented as follows :

supersaturated solid solution $\rightarrow$ GP zones $\rightarrow \beta' \rightarrow \beta$.

As already known (1) the β' structure is similar to the structure of GP zones and develops directly from it through the formation of dislocation loops around the precipitate, which destroys the coherency with the matrix and therefore eliminates most of the long - range elastic forces. Thus the formation of the β' structure leads to a softening of

the material.
The β phase is coherent with the matrix and contributes to matrix strain, the magnitude of which depends on the re= version time, where the β' phase is incoherent with the ma= trix, producing thus the lowering in hardness.
After heating the alloy to a temperature above the solvus line and quench, the alloy can be stored, before the ageing begins. The time of storage of the alloy between the quench and ageing also influences the final effect of heat treat= ment. FASHLEY with coworkers (2 - 3) investigated the effect of storage time between 1 hour and 88 days on sam= ples held at various temperatures from -20 °C, 20 °C and 50 °C. The hardness of the samples was increased, neverthe= theless the fact, that the presence of zones could not be detected.
In the present work the scanning electron microscope, besi= des the transmission type of electron microscope, were used to examine the effect of the microstructure on the frac= ture modes of cast Al - Mg alloys, containing the β phase and mixtures of β and β' phases, that have been obtai = ned by aging treatment.

E x p e r i m e n t a l   p r o c e d u r e

Al - Mg alloys with the commercial name Hy - 511, correspon= ding mostly to the alloy G - Al Mg 5 after DIN 1725, con= taining between 4,3 to 5,5 % Mg, up to 0,6 % manganese and up to 1,5 % silicon, with small amounts of cuprum, were cast in induction melting furnace. The induction melting furnace of the Junkers Comp. allowed to the melt to be he= ated up to 700 °C. After slowly heating up to 700 °C, the melt was conducted to a pre - furnace, where the temperatu= re of the melt was increased up to 780 °C and the cleaning procedure of the melt, with following stages was performed :

- degasation of the melt ;
- addition of the nucleants to the melt ;
- desoxydation of the melt ;
- correction of the chemical compostion .

After the correct chemical composition of the melt is obta=

ined, the alloy is stored in the pre - furnace for a period of 12 hours. After that period the chill - casting procedu= re of cylinder heads for diesel engines begins, the alloy Hy - 511 being aimed principally for the production of hi= gh duty castings[1]. After the casting of the cylinder heads was finished, the castings were heat treated with the follo= wing schedule of heat treatment :

SHT 520 °C / 8 hours, WQ
ageing 170 °C / 18 hours.

After finished heat treatment, the following mechanical pro= perties are required :

tensile strength ...... 19 to 22 kp/mm^2
elongation .... min. 3 %
Brinell hardness .... 75 to 85 kp/mm^2.

Tensile test specimens were cut off from a sample casting, the intervals of sampling being 50 castings. Tensile tests were performed with Mohr - Federhaff testing machine on nor= mally tensile test specimens, having a diameter of d = 10 mm with a gauge length of 5 d. All tensile tests were conducted at room temperature.
The fractured surfaces of the tensile test specimens were examined in a Jeol Scanning Microscope - JSM - type U 3, u = sing an accelerating potential of 20 Kv.
The results of chemical analysis of the investigated specim= ens are shown in T a b l e I and the results of mechani = cal testings are presented in T a b l e II.

TABLE I. Chemical composition of the investigated Al-Mg all.

Melt No.	Elements in %				
	Mg	Mn	Si	Cu	Al
5	4,85	0,38	1,85	0,43	rest
22-0	4,80	0,36	1,17	0,41	rest
116	5,80	o,35	1,32	0,51	rest
119	5,90	0,39	1,38	0,45	rest
139	4,67	0,36	1,34	0,34	rest
141	4,88	0,38	1,30	0,34	rest

1) Factory of diesel engines Torpedo in Rijeka

TABLE II. Results of mechanical testings of the investi= gated Al - Mg alloys.

Melt No.	Results of tensile tests Tensile strength kp/mm^2	Elongation %	Hardness Brinell BHN
5	21,4	2,85	81,3
22-0	21,2	2,71	85,8
116	20,0	1,42	84,4
119	19,7	1,42	85,8
139	19,9	2,4	70,6
141	23,5	1,85	90,2

The results of chemical analysis and mechanical testing of the investigated aluminium - magnesium alloys clear= ly indicate that excess content of magnesium in two melts 116 and 119 resp. caused the lowest values in elongation. Independent of both melts all the other melts showed un= satisfactory results of elongation.

R e s u l t s

With the heat treatment procedure the grain size of prima= ry α crystals could not be influenced, because different grain sizes were measured, where the increase in grain si= ze, caused a decrease in elongation. The α phase was sour= rounded by small amounts of binary eutectic mixture $\alpha \succ Mg_2Si$ superimposed by the trenary eutectic $\alpha \succ Si / Mg_2Si$. After the so= lution heat treatment procedure β and β' were formed, and proportional to longer heating periods more β' was present.

Fractographic examinations of the broken surfaces of the tensile test specimens showed, that all the heat treated specimens exhibited a ductile mode of fracture. Only at one sample, the fracture test specimen of the melt 116, we ob = served mixed ductile and brittle fracture modes. Elongated dimples, observed on some regions of the fracture surface showed, that there has been considerable local plastic de= formation.

As known, ductile mode of fracture occurs after extensive

deformation and is characterized by slow crack propagation resulting from the formation and coalescence of voids and a ductile fracture surface has a characteristic fibrous appearance (5). Comparisions of the fractographic results obtained showed the same result as obtained by CHANDRASE= KARAN and coworkers (6) that it is difficult to corre = late the network of fine dimples on the fracture surface with the size spacing of the aged α phase in the matrix . In the case of melt 116 we obtained slip lines appearing only in the grains adjacent to the fractured surface, indi= cating that the localized plastic deformation has occured prior to fracture.

Further investigations were devoted to the question of for= mation of transition structures after the following pro = cedure of heat treatment :

SHT 520 ^{o}C / 18 hours , WQ and then ageing 180 ^{o}C / 18 h.

These investigations are in progress (7) and on the re = sults obtained will be reported later.

Thanks are due to the Board of Directors of the Torpedo motor works company in Rijeka, providing material for this study and giving permission for publication of the results. The author wish to thank to Dr.G.Petzow from Max - Planck - Institute in Stuttgart, W.Germany for his help in using the Scanning electron microscope facilities at the Institute.

R e f e r e n c e s

1. Guard, R.W. " Mechanics of Fine Particles Strengthe= ning " in Strengthening Mechanisms in Solids, A.S.M. Cleveland, 1962.
2. Lorimer, G. and R.Nicholson, Acta Met. 14, 1966, 1006.
3. Pashley, D., Rhodes, J. and A.Sendorek, J.Inst.Metals, 94, 1966, 41.
4. Pashley, D., Jacobs, M. and J.Vietz, Phil.Mag., 8, 1967, 51.
5. Hayden, H.W., Moffatt, W.G. and J.Wulff in The Structu= re and Properties of Materials, Vol.III. John Wiley & Sons, Inc. New York, London, Sydney, 1965.
6. Chandrasekaran, V., Taggart, R. and D.H.Polonis, Metallo= graphy, 5, 1972, 235.
7. Pavlin, M. Habilitation work, Rijeka, 1972.