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STUDIES AND TESTS CONCERNING SMELTING CASTING ALLOY OF UNIVERSAL ALUMINIUM

Abstract:
The direct consequence use on a major scale a silumin alloy type at casting pieces which working at high temperatures, e.g. thermal engine pistons, is: finding, smelting an alloy that possess good mechanical and technological properties, with preservation of those properties for a long time. The most recommended are Al-Si alloys type, allied with addition agent (Ea) such as Cu, Mg, Mn, B, Ti, Cr, etc. Large used alloys are $ATSi_9Cu_2MgB$ and $ATSi_{10}MgMn$ but with the shortcoming technology and mechanical properties, has been necessary to find a "universal" cast alloy that has proven to be $ATSi_9Cu_3MgMnB$, which is studied in this paper and responded well to the demands of constructors and technologists requirements of the non-ferrous foundries.

Keywords:
alloy, aluminium, thermal engine, pistons, casting

INTRODUCTION

In automotive industry is good to have a cast aluminium alloy (AAT) which correspond to the requirements of foundries technology, to have high mechanical strength both at ambient temperature and at high temperatures (resistance duration), dimensional stability over time, regardless of operating conditions, etc.

For developing (creating) a new alloy (made of AAT), with high technological and mechanical proprieties called UNIVERSAL DESTINATION Alloy, the basis for it had in mind the following considerations [1]....[4]:

1. Si content should be around 8-10%, to ensure high casting properties, which allow casting of PT by any method (in FAF, in shell, with lost models (fusible), centrifugal, under low or high pressure, etc.) without appear casting cracks.
2. Solid solution α must have a high quality of supersaturate in Ea, to allow during the aging process to obtain a high density of ultra-dispersed particles (micro-heterogeneousness) inside s.s.α grains. This allows achieving high flow limits and mechanical resistance.
3. Alloy components -Ea- and inclusion, in the alloy crystallization should not form particles occur on the separation of the s.s.α. [5]… [7].

Following research [1], [2] developed a new alloy cast in AAT, which possesses a chemical composition (8-11%Si; 3-4%Cu; 0.13-0.35%Mg; 0.1-0.3%Mn; 0.01-0.1B; 0.4%Fe, rest aluminium) and notes ATSi9,5Cu3MgMnB. Alloy has received a fast use at casting the various PT. In figure 1 (2.78) is shown typical PT, cast of this alloy, pieces that once was made by plastic deformation with very high costs.

To note that the proportion of Mg and Cu content in this alloy is other compared to $ATSi_9Cu_2MgB$. That alloy has characteristics of resistance much higher, both at ambient temperature and at higher temperatures. The complex configuration of PT and walls thickness, and the operation of the PT must be considered on load composition of the alloy development. As before, the highest is Si content (but Si<12%), the greater is eutectic quantity, this means that has highest properties of casting and PT hermetically. However, with increasing Si content of alloy increases the tendency towards abortion gas, therefore PT is formed in a high porosity. For casting the complexity PT (complex configuration), with wall thickness 3-4mm is sufficient for the load to be 9% Si.

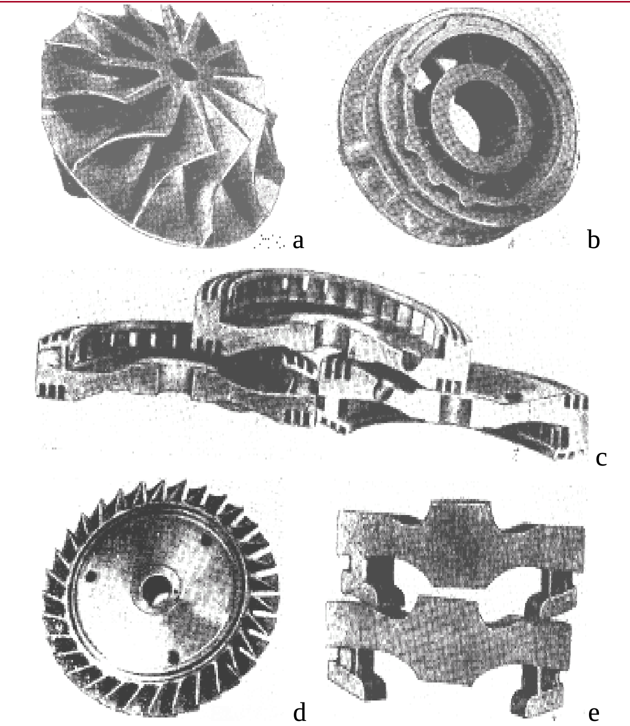


Fig.1 Pieces of great complexity cast of polynar alloy $ATSi_{10}Cu_3MgMnB$: a- centrifugal; b- precision; c-section of the turbine body; d- Precision; e- Shell (microstructure)

With increasing content of Cu, refractory alloy increases, but plasticity (A) decreases at ambient temperature. If the PT in this alloy is intended to work long time at high temperatures the content that should be at upper limit, and if working at ambient temperature, then at lower limit.

The increasing of Mg content leads to increase refractory alloy, but decreases plasticity at ambient temperature. To increase plasticity, in PT, the content of Mg should be kept at the lower limit, same for Fe content. In this situation the content of Mn should be 0.8% and the content of Fe = 0.2-0.3%.

If PT is accomplished by stamping of semi liquid state, then permitted Fe inclusion even 0.4%. This has a great technical and economic meaning, because in the use of Al load (waste) technical with high content of Fe.

Ti and B are introduced in the alloy as modifier.

$ATSi_9Cu_3MgMnB$ ALLOY STRUCTURE

The alloy $ATSi_9Cu_3MgMnB$ as you can see possesses a complex chemical composition. Depending on the chemical composition fluctuation, Fe content in the PT and the rate of crystallization, phases of the alloy composition may change strongly. Alloy structure in a cast (especially in parts of the massive and high content of Cu) may have the following phases: α, Si, Mg_2Si, $CuAl_2$, AlSiMnFe (figure 2, (2.79)). In case of very slow crystallization (at $T \approx ct$, in equilibrium) can form $W(Al_xMg_5Si_4Cu_4)$ phase which is seen in figure 3 (2.80).

Taking into account that in the $ATSi_9Cu_3MgBTi$ alloy can be eutectic with different melting temperatures (because of the complexity phase component, which depends on the rate of crystallization) and also considering the differential thermal analysis (figure 4 (2.81)) for PT is recommended two regimes of hardening:

- heating at 500°C/4h + 515°C/10h followed by cooling in water temperature 20-30°C;
- heating at 490°C/4h + 500°C/4h + 515°C/6h and cooling in water temperature 20-30°C.

The first hardening regime is recommended to PT in the shell or PT with thin walls cast in FAT (when the cooling rate of casting is great - $v_{cr}>>>0$). The second hardening regime is recommended for large size PT with thermal node (joints massifs).

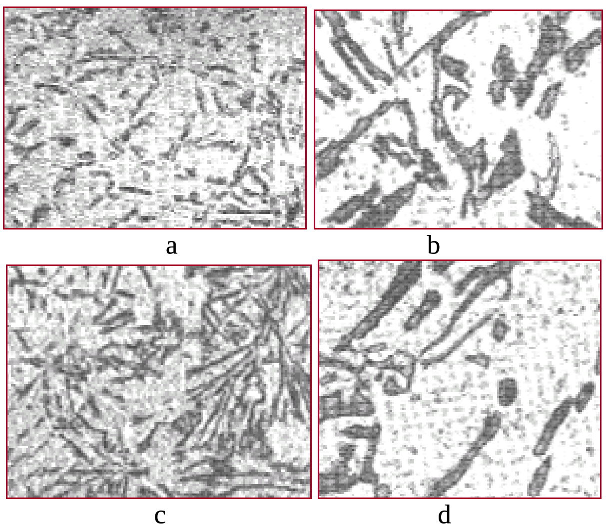


Fig.2 $ATSi_9Cu_3MgMnB$ alloy structure: a- cast (100:1); b- cast (500:1); over TT; c- 100:1; d- 500:1

Fig.3 Type of crystallization phase W($Al_xMg_5Cu_4Si_4$) la 500:1

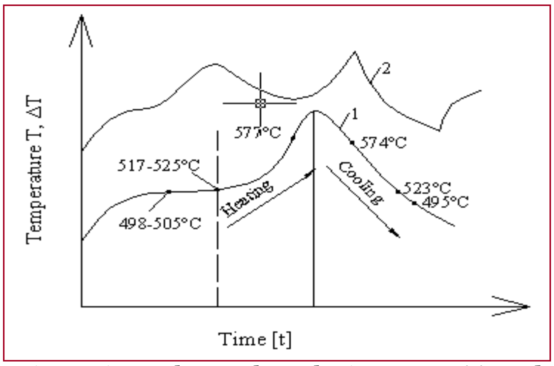


Fig.4 Direct thermal analysis curves (1) and differential (2) of the $ATSi_9Cu_3MgBTi$

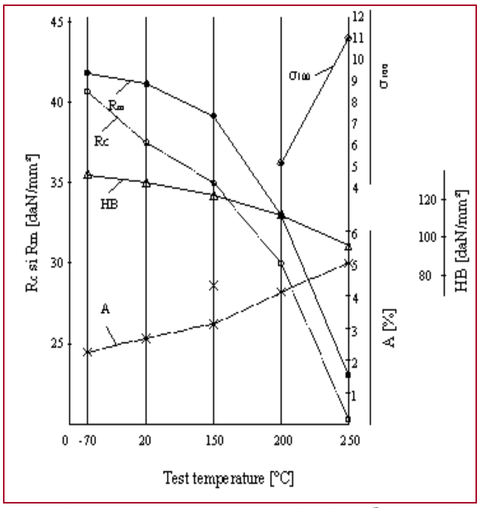


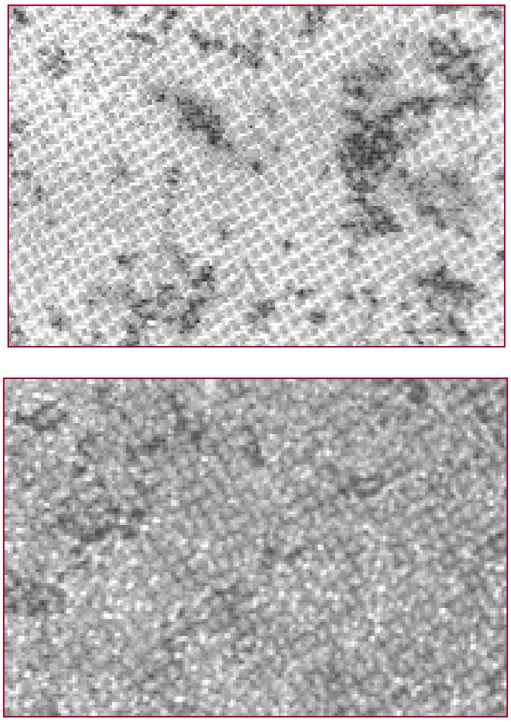
Fig.5 $ATSi_9Cu_3MgMnB$ proprieties determined on sample of Φ12mm cast in shell

Mechanical properties of $ATSi_9Cu_3MgMnB$ hardening after the first regime and the aging after the regime 165°C/22h or 175°C/7h, with air cooling are shown in figure 5 (2.82). At 20°C has HB=120daN/mm^2 and decrease with test temperature increase.

CONCLUSIONS

The physical properties of the alloy $ATSi_9Cu_3MgMnB$ practically are the same as those of the alloy $ATSi_9Cu_2MgB$, and the casting of the alloy as $ATSi_{10}MgMn$. The cutting process is better [5], than the two mentioned alloys. The welding is good.

To confirm the correctness of the choice of aging regime is shown in figure 6 (2.83) s.s.α alloy structure ATSi9Cu3MgMnB in hardening state and cast in forms of mixture formation.



a

b

Fig. 6 $ATSi_9Cu_3MgMnB$ structure alloy in hardening state after casting in: a- FAT and b- shell (10000:1)

Note that the cooling was the first stage of aging, this mean that, they have managed to form ZGP and agglomerations of Si ultra-dispersed particles what were separately from s.s.α. In comparing these pictures, you can see that ZGP density in solid solution alloy $ATSi_9Cu_3MgMnB$, cast in FT, is much reduced, comparative to s.s.α of the alloy cast in shell. This can be explained by the fact that: supersaturate degree of s.s.α in the last sample, in all probability, is higher, and the degree of distortion R_{cr} of s.s.α also higher, which accelerates the process of form ZGP, and ultra-dispersed particle of elementary Si.

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