Effect of precipitates on static and fatigue strength of a severely forged aluminum alloy 1570C

To cite this article: E V Avtokratova *et al* 2018 *IOP Conf. Ser.: Mater. Sci. Eng.* **447** 012038

IOP Conf. Series: Materials Science and Engineering **447** (2018) 012038 doi:10.1088/1757-899X/447/1/012038

Effect of precipitates on static and fatigue strength of a severely forged aluminum alloy 1570C

E V Avtokratova[1], O Sh Sitdikov[1], O E Latypova[1], M V Markushev[1], M L Linderov[2], D L Merson[2] and A Yu Vinogradov[2,3]

[1] Institute for Metals Superplasticity Problems, Russian Academy of Sciences, 39 Khalturin st., Ufa 450001, Russia
[2] Institute of Advanced Technologies, Togliatti State University, 14 Belorusskaya st., Togliatti 445020, Russia
[3] Department of Mechanical and Industrial Engineering, Norwegian University of Science and Technology—NTNU, N-7491 Trondheim, Norway

E-mail: mvmark@imsp.ru

Abstract. Bulk billets from commercial ingots of the alloy 1570C with different size of aluminides of transition metals were subjected to severe deformation up to the equivalent strain e~24 via multi-step isothermal forging (MIF) with an inter-step temperature decrease from 325 to 175 °C. The high potential of MIF to produce UFG billets with enhanced balance of static and fatigue properties has been demonstrated. The role of precipitates of aluminides of transition metals and regimes of the MIF in the structure of the alloy and for the control of properties is discussed.

1. Introduction

Non-heat-treatable Al-Mg alloys with complex additions of transition metals (TM) belong to a modern class of materials for a wide range of commercial applications due to the unique combination of service properties which they possess [1,2]. However, the range of properties, which can be controlled during industrial processing, is limited by two main tempers of semi-finished products, namely, annealed and work-hardened ones. The primary strengthening mechanism involved in the processing of these alloys is dislocation hardening which has an obviously limited capacity. An alternative popular (though admittedly less studied) approach to the improvement of properties of the alloy is based on grain boundary hardening, assuming the formation of ultrafine grain (UFG) and nanocrystalline (NC) structures [3-5]. This approach is facilitated by a fairly high content of TM, which are sufficient to form high densities of aluminides during homogenization and stabilize the resultant deformation microstructures [6]. Such precipitates also play an active role as strengthening agents, providing remarkable hardening potential to the alloys [1]. It has been recently demonstrated that the most attractive method towards extreme enhancement of properties in complex alloys is severe plastic deformation (SPD) (with $e > 1$) combining all specified strengthening mechanisms in a single framework [3-5]. Besides the control of geometrical parameters of the microstructure (grain/subgrain size and shape, volume fraction of the recrystallized grains and etc.), SPD is capable of varying the distribution of second phases and their dispersion in the matrix as on the whole as well as in individual grains. The multi-step isothermal forging (MIF) is one of the modern metal processing techniques that meets the specified requirements along with high efficiency to process UFG/NC bulk billets [3,7].

The present study is aimed at exploring the efficiency of MIF for the structuring and strengthening of the Al-Mg-TM alloys and clarifying the role of aluminides and their distribution in microstructure and mechanical behavior.

2. Material and procedure

In order to change the size of secondary TM aluminides, an ingot of the alloy 1570C of a standard chemical composition (Al-5Mg-0.18Mn-0.20Sc-0.08Zr, wt. %) was subjected to one (at 360 °C, 6h) and two-step (360 °C, 6h + 520 °C, 1h) homogenization processes. In this way, the aluminum solid solution abnormally supersaturated by TM was decomposed and coherent to the matrix. $Al_3(Sc,Zr)$ precipitates (dispersoids) were formed and grown. Further MIF of billets ∅80×150 mm was carried out in three stages with an inter-step decrease in temperature. Each step included several deformation cycles after which the shape and dimensions of the billet were maintained as much as possible. For this purpose, in each cycle, several consecutive settings and drawings were implemented along all axes of the billet [3]. In the first stage, consisting of 12 cycles at 325 °C, the cumulative equivalent strain reached e ~12. In the second and third stages, conducted at 250 and 175 °C, respectively, e was about 6 at each stage. The microstructure of the alloy was studied by means of transmission and scanning electron microscopy (TEM and SEM) and X-ray analysis. Most of the parameters of the aluminum matrix were determined using SEM-EBSD data. The size and density of dispersoids were detected by computer analysis of TEM images of at least 2000 precipitates. Samples for the studies were prepared by mechanical grinding and polishing and electropolishing at -28 °C in 20% solution of HNO_3 in CH_3OH. The microhardness of the alloy was determined at 15 seconds loading of 1 N. Tensile and fatigue tests were carried out at ambient temperature on mechanically pre-polished flat dog-bone and corset-shaped samples with a gage section of 1.5×3×6 mm^3 and a cross-section of 1×0.9 mm^2, consequently.

3. Results and discussion

In contrast to the one-stage homogenization, after which the modal size of the dispersoids and their number density were about 10 nm and 1×10^4 μm^{-3}, respectively, after two-stage annealing their size increased more than two times, with a corresponding reduction in their density to 5×10^3 μm^{-3}. However, the microstructure of the aluminum matrix was almost the same, with the mean size of equiaxed grains of 25 µm. In both billets, new fine subgrains and grains, surrounded by low and high-angle boundaries (HABs) were formed during the very first cycles of the first stage of MIF near the boundaries of initial grains along the free zones of precipitates formed during homogenization.

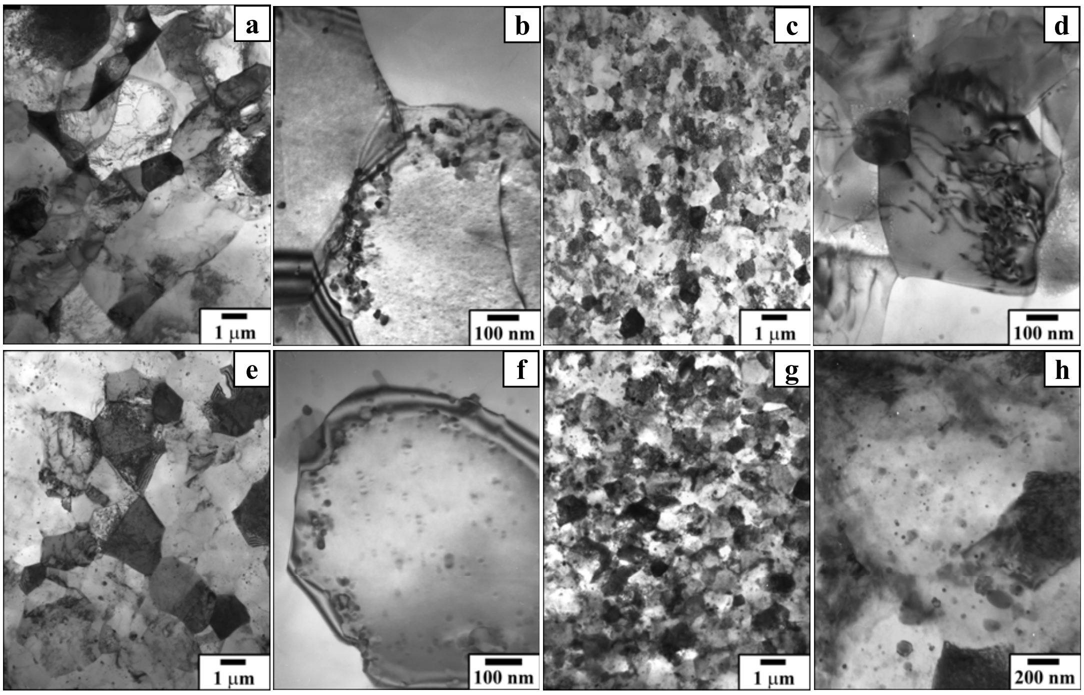

Figure 1. TEM structures of the alloy 1570C after the 1st and the 3rd stages of MIF at 325 (a,b,e,f) and 175 °C (c,d,g,h), consequently. (a-d) fine precipitates, (e-h) coarse precipitates.

With increasing strain, the fraction and misorientation of these grains increased, which led to a gradual replacement of the coarse-grained structure by the fine-grained one, regardless of the initial size of the aluminides (figure 1). The main mechanism for grain refinement was continuous dynamic recrystallization, controlled by the interaction of lattice dislocations and (sub)grain boundaries with dispersoids that prevent migration of boundaries. This process is inevitably accompanied by a rearrangement of dislocation and recovery. The effect of precipitate size on the evolution of microstructure was rather small: in both states of the alloy, the structure was partially recrystallized with almost equal subgrains (in the range of 1.8-2.0 μm) and slightly recrystallized with almost equal subgrains (in the range of 1.8-2.0 μm) and slightly recrystallized with smaller grains in the alloy with smaller precipitates (2.2 and 2.7 μm).

Analysis of TEM and SEM showed that after the third stage of MIF, the structure of both states of the alloy became more refined: the size of the subgrains was almost equal and decreased to ~1 μm, while the grain size of 1.1 and 1.3 μm was found in the alloy with fine and coarse precipitates, respectively. Simultaneously, the structure became less homogeneous and equilibrium. The majority of (sub)grains were surrounded by boundaries with the uneven TEM extinction contrast and contained increased dislocations density (X-ray data indicate an increase by two orders of magnitude). Due to the higher suppression of the dynamic recovery caused by the decrease in the deformation temperature, the structural heterogeneity of the alloy with smaller precipitates was also slightly larger. Besides, when processing at temperatures below the solvus point (~250 °C), rather coarse (50–250 nm) precipitates of the β-phase (Al-Mg) were formed in the crystallites in addition to TM aluminides (figure 1).

Measurements of alloy hardness and tensile strength parameters showed their good correlation with changes in the microstructure discussed above. Thus, coagulation of dispersoids in the second stage of homogenization of the alloy resulted in a decrease in hardness and yield stress from 110 to 90 Hv and from 240 to 180 MPa, respectively, due to a partial loss of the dispersion hardening. A similar difference in the hardness of the alloy and its tensile behavior was observed after the first stage of MIF (table 1). This was due to the fact that high-temperature forging did not lead to the strengthening of the alloy, despite the remarkable grain refinement. The latter can be explained by the compromise of strengthening the grain boundaries by partial softening of the alloy due to a decrease in the hardening of the dispersion caused by coagulation of the precipitates, which was continued during high-temperature straining and inter-cycle annealing. This difference is due to the formation of an almost equilibrium recrystallized structure with low dislocation densities in the alloy states, giving a small work hardening effect, and resulting also in higher ductility of the alloy.

After the third stage of the MIF, the precipitate size effect was accompanied by a noticeable increase in strength, especially yield stress, and an insignificant decrease in ductility (table 1). This behavior was reasonably expected from the processing of the nonequilibrium UFG structure with a smaller number of grains/subgrains, higher dislocation density and β-precipitates.

Table 1. Properties of the alloy 1570C at ambient temperature.

Condition	Vickers Hardness, HV	Yield Stress, MPa	Ultimate Tensile Strength, UTS, MPa	Elongation to failure, El, %
Ingot	110/90[a]	240/180	355/340	28/37
MIF 1st stage	105/95	230/215	355/345	38/36
MIF 3rd stage	125/110	315/285	385/370	32/32

[a]- fine/coarse dispersoids

Thus, based on these observations, it could be concluded that: (i) the grain refinement with a decrease in the grain size by more than an order of magnitude down to the UFG range, caused only a slight hardening effect on the parameters of the static strength, which was simultaneously compensated by a decrease in dispersion strengthening due to TM aluminides coarsening in the high-temperature stage of MIF processing; (ii) MIF, resulted in work-hardened mixtured/bimodal structure, consisting of ultrafine grains and subgrains with increased dislocation density (and virtually without a strong metallographic and crystallographic texture, as frequently observed in conventional hot pressing and rolling), gave a rather high impact in the strengthening of the alloy, especially in yield stress, and (iii) the higher strength of the alloy with finer precipitates is due to the formation of a less regular and less developed UFG structure under the MIF, demonstrating the indirect influence of the dispersion of the strengthening phases. Thus, if we take into account the absolute values of hardness and tensile strength parameters, their increase with decreasing temperature of the MIF is mainly conditioned by: (i) work-hardening due to an increase in dislocation density (ii) structural strengthening due to continuation of the (sub)grain refinement. As for the literature data and our estimates, the dispersion hardening due to the decomposition of the Mg-rich solid solution of aluminum with formation of β-phases is not very large and is usually able to compensate for the loss of only the aluminum solid solution strengthening [8].

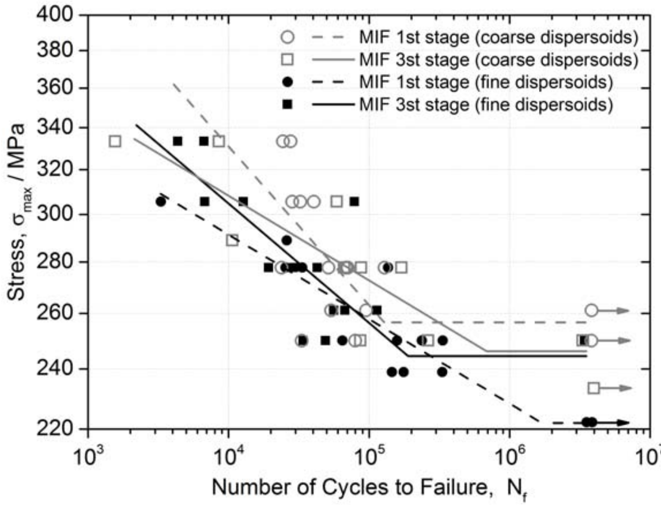

Figure 2. 1570C alloy fatigue dependences after 1st and 3rd MIF stages.

The S-N graphs of the alloy after the first stage of MIF revealed a higher fatigue resistance of the material with larger TM aluminides (figure 2) despite the lower yield stress value (table 1). Considering the conventional endurance and fatigue limit determined in 4×10^6 cycles, the behavior of the strengthened alloy with less precipitation was preferable both in low and high-cycle fatigue regions. However, after the third stage of the MIF, the effect of precipitates disappeared and the alloys in both states showed almost the same behavior.

Virtually, the fatigue behavior of the alloy has the same origin as in static tests, emphasizing the effect of dispersion of TM aluminides, which is especially noticeable in the intensity of dynamic recovery. Thus, this effect is more pronounced in fatigue testing of an alloy with an almost equilibrium fine-grained structure of a matrix, containing different densities of disperse phases. Improvement of fatigue resistance of the alloy by means of fine dispersoids after the third stage of the MIF was controlled by dislocation structures formed during low-temperature forging. However, their formation was also conditioned by the interaction of dislocations with precipitates of both the TM aluminides and the β-phase. This phenomenon should be studied in more detail in order to discuss the nature of the observed behavior of the alloy.

4. Conclusion

The role of precipitates of transition metal aluminides in the formation of UFG structures and the properties of Al-Mg alloys is rather complicated, especially when determining the alloy's resistance to fatigue. However, its origin is one and the same under different loading conditions and is controlled by the mechanisms of interaction of precipitates with moving dislocations and grain boundaries that affect the intensity and kinetics of recrystallization and matrix recovery.

Acknowledgments
The work was supported by the grant No. 16-19-10152 of Russian Science Foundation on the basis of the Shared Service Center of IMSP RAS «Structural and Physical-Mechanical Studies of Materials».

References

[1] Filatov Y, Elagin V and Zakharov V 2000 *Mater. Sci. Eng. A* **280** 97
[2] Avtokratova E, Sitdikov O, Markushev M and Mulyukov R 2012 *Mater. Sci. Eng. A* **538** 386

[3] Sitdikov O, Garipova R, Avtokratova E, Mukhametdinova O and Markushev M 2018 *J. Alloys Compd*. **746** 520
[4] Valiev R and Langdon T 2006 *Prog.Mater. Sci.* **51** 881
[5] Vinogradov A and Estrin Y 2013 *Acta Mater*. **61** 782
[6] Avtokratova E, Sitdikov O, Mukhametdinova O, Markushev M, Murty S V S N, Prasad M J N V and Kashyap B P 2016 *J. Alloys Compd*. **673** 182
[7] Markushev M 2011 *Letters on Mater.* **1** 36
[8] Markushev M and Murashkin M 2004 *Phys. Metals Metallogr.* **98** 221