Influence of fine-grained structure and superplastic deformation on the strength of aluminium alloys

Part II *The physical nature of the influence of fine-grained structure on the strength of aluminium alloys*

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Samples of 1560 (Al–Mg–Mn) and 1960 (Al–Zn–Mg–Cu) alloys have been used to investigate the nature of the effect of grain size and superplastic treatment on the strength of aluminium alloys. The observed increase in the work needed for crack formation with the transition from coarse-grained (CG) to fine-grained (FG) structure is connected to a greater homogeneity of the plastic deformation in the material volume. This leads to a reduction in local stress concentrations at the sites of preferential crack initiation. The easier crack growth in FG alloys is mainly caused by a reduction in the energy for plastic deformation at the head of a long crack and also for the formation of free fracture surfaces.

1. Introduction

The data discussed in Part I [1] of the present series of papers have shown the formation of fine-grained (FG) structure instead of a coarse-grained (CG) one in aluminium alloys affects the mechanical properties. The beneficial effects are an increase in static strength and ductility, high-cycle fatigue life and limit, and a decrease in anisotropy of some properties. In addition it results in a higher resistance to crack nucleation. The main drawback is a decrease in the crack growth resistance, which can lead to a reduction in the alloy low-cycle fatigue endurance and reliability, static and impact toughness, and to the increase of sensitivity to sharp stress concentrators.

The present paper deals with the physical nature of grain size effect on the mechanical behaviour of the alloys.

2. Experimental procedure

The aluminium alloys 1560 and 1960 were used in this investigation. The composition, structure and properties produced in the alloys after different treatment modes are described in Part I [1].

The effect of grain size on static crack initiation was studied on flat wedge-shaped specimens (Fig. 1) of longitudinal and transverse (LT and TL) orientations. The specimens with polished surfaces on which two rows of bench points were marked with an interval of 500 μm, were extended to failure by tension at room temperature. The strain in different sections of the specimens and the micro-crack density on them were determined optically at a magnification of 800.

The size of the plastic deformation zone (PDZ) at the tip of the static and fatigue cracks (λ) was estimated on flat specimens of LT orientation after their rupture. The depth of the plastic deformation penetration for static cracks was determined by the change of microhardness depending on the distance from the crack surface. For fatigue cracks it was established from the observed broadening of X-ray peaks. The distance from the crack surface at which the mentioned parameters achieved the levels corresponding to non-deformed material was taken as λ.

The extent of the crack surface asperities (l) was estimated on longitudinal templets of ruptured specimens by the measurement of fracture surface relief by means of the optical microscope.

3. Results and discussion

3.1. The influence of grain size on static strength and ductility

In various alloys, including those based on aluminium, the strengthening caused by grain refinement depends on the composition and structure of the material. In all alloys the Hall–Petch effect [2, 3] contributes to the strengthening by the relationship:

$$\sigma_e = \sigma_i + k_g d_g^{-m} \qquad (1)$$

where σ_e is the flow stress, σ_i is the stress characterizing the resistance to dislocation movement within the

† Deceased.

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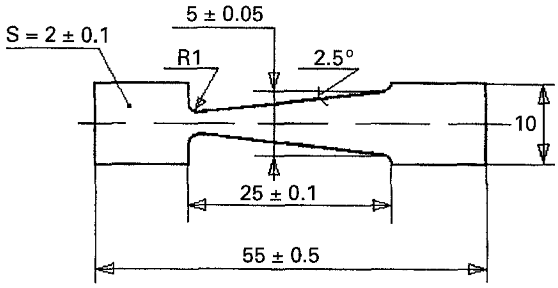

Figure 1 A wedge-shaped tensile specimen.

grain, k_g is a constant depending on the mechanism of slip transition through the grain boundary, d_g is the grain size and m is an exponent approximately equal to 0.5 depending on the alloy nature [4, 5].

Equation 1 is true when CG and FG alloys are in the same state, that is after complete recrystallization. When this condition is not fulfilled the strength of a CG material can be higher. This is connected with a substructural strengthening effect [6–8]. In this case the alloy strength is determined by [9];

$$\sigma_y = \sigma_i + k_{sg} d_{sg}^{-1/2} + \alpha G b d_{sg}^{-1} \qquad (2)$$

where σ_y is the yield stress, d_{sg} is the size of subgrains, G is the shear modulus, b is the Burger's vector and α is a coefficient.

Thus when the subgrain size in CG material is less than the grain size in a FG material the strength of the CG alloy may be higher. This was the reason for the observed higher strength characteristics of conventionally treated 1960 alloy (CT1) in comparison with superplastic processed (SPT) [1].

A grain size reduction increases the ductility of aluminium alloys since it provides greater homogeneity of microplastic deformation in the material volume [4, 10–12]. The latter is caused by the fact that more grains are similarly oriented in a unit volume. This results in a simultaneous appearance of a greater number of slip bands. As the length of these bands is small (restricted by the grain size), a lower level of local stress concentration occurs at the boundary regions of a FG material. Thus, the possibility of crack initiation is reduced (discussed later) and consequently, the greater polycrystal ductility can be exploited.

The higher ductility of the alloys after superplastic treatment may be explained by the uniformity and refinement of grain structure. In heat hardenable alloys (1960), which are distinguished by low ductility after conventional treatment, the elimination of substructural strengthening additionally contributes to the increase of plasticity after SPT. The higher ductility of superplastic treated alloys may also be caused by changes of banded structure, as a specific feature of superplastic flow is broadening (spreading) of bands of excess phases [13, 14].

3.2. The influence of grain size on crack formation

It is established in aluminium base alloys of different composition that grain refinement increases the crack nucleation resistance (CNR) irrespective of loading conditions (see Part I). The same situation has been observed in alloys based on other metals [15, 16].

Such mechanical behaviour is caused by the following physical premises.

The formation of a crack by any dislocational (stress) mechanism requires the fulfilment of the equation (4):

$$n\tau \geqslant \text{Const} \simeq 0.7G \qquad (3)$$

where n is the number of dislocations in pile-up(s) at an obstacle (interphase or grain boundary) and τ is the applied shear stress.

For equal relative deformations of specimens with different grain sizes the absolute deformation of grains in a FG material is lower. Consequently, n, which in turn is proportional to the grain size, must be lower in such a structure. Since τ is proportional to $d^{-1/2}$ (Equation 1), $n\tau$ in an FG material is less than in a CG one and it is necessary to increase the value of τ in order to obtain the critical value of stress concentrations (Equation 3). This produces the higher CNR in fine-grained materials.

If the crack initiation is thermally activated, then at a low stress concentration in fine grains a higher activation energy is needed [17]. This means that in a FG material at a given τ the period of crack formation increases.

In such a material the critical stress value (σ_c) at which the crack initiates without activation will also be higher [17],

$$\sigma_c = \frac{1.84\,G}{nb2\pi(1-\nu)} \qquad (4)$$

where ν is Poisson's ratio.

The larger amount of work needed for crack initiation in a FG material is also connected with an increase of the homogeneity of microplastic deformation within the grain. According to the von Mizes criteria [15] with a decrease in the grain size the deformation by multiple sliding spreads over a larger volume of the grain.

The experimental confirmation of a higher CNR in a FG material is shown in Fig. 2. On polished surfaces of LT specimens (at this orientation the grain size is more pronounced due to the features of the studied structure) of the 1560 alloy whose elongation does not exceed 3 %, the number of microcracks is less in a fine-grained (FG-1) state. The absence of such a distinction in TL specimens is caused by the dominant influence of bands of excess phases on crack initiation.

In the case of cycle loading conditions the preferential sites for crack initiation are surface intrusions or extrusions. The reduction in the number of dislocations in slip bands caused by grain refinement must lead to a decrease of the damage depth of specimens or surfaces. Thus the nucleation of a fatigue crack in a FG material needs a longer time or a higher applied stress. The latter was clearly shown by experimental data (Part I) [1].

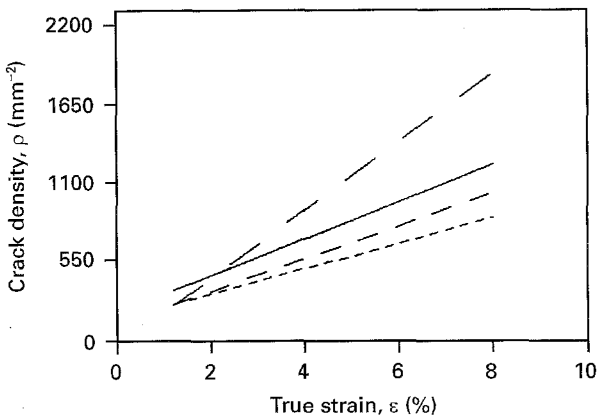

Figure 2 Microcrack density versus tensile strain on surface of specimens made of coarse- (——— LT and ---- TL) and fine-grained (FG-1) (— — LT and - - - TL) 1560 alloy.

3.3. The influence of grain size on crack growth resistance

The growth of cracks requires the expenditure of energy according to the equation

$$A_g = A_d + A_s - A_r \quad (5)$$

where A_g, A_d, A_s and A_r are the energies expended on crack growth, plastic deformation of the material at the crack head, formation of free surfaces and relaxation respectively. It may be admitted that A_d is proportional to λ^3 (λ is the linear size of PDZ) and A_s is proportional to l^2 (l is the crack length).

It has been established (Table I) that with grain refinement the size of the PDZ for static and fatigue cracks in alloys decreases. The analogous results are obtained in reference [18] for a 1141 alloy.

These data can be visually explained with the help of Fig. 3. At the same tension force (P) the stress state in the notch section of the specimen is grain size independent, but the size of the PDZ at the tip of a crack is less in a FG material corresponding to a higher level of its yield strength.

The observed decrease of the PDZ size is in good agreement with known equation [19]

$$\lambda_c = \frac{1}{\alpha\pi}\left(\frac{K_c}{\sigma_y}\right) \quad (6)$$

where λ_c is the radius of the zone at the tip of a crack of the critical length and α is a coefficient equal to 2 or

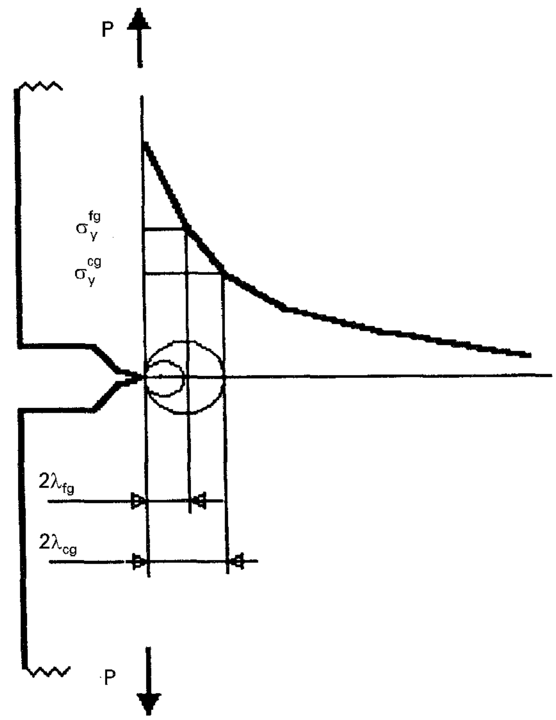

Figure 3 The schematic diagram explaining the difference in the size of the plastic deformation zone (λ) at the tip of a growing crack in materials with coarse- and fine-grained structures.

6 for plane stress or plane strain states respectively. As the grain size is reduced σ_y increases in line with the Hall–Petch equation whilst K_{1C} for aluminium alloys generally decreases, [18, 20].

It is also established (Table I) that the length of a crack path in its growth perpendicular to the bands is less in FG alloys under both static and cyclic loading. This path reduction testifies to a smaller area of fracture surfaces and consequently to less work needed for crack growth. The decrease of l is caused by the change of the character of the fracture relief: the reduction in the height (h) and the width (t) of surface asperities (Fig. 4). This situation is schematically shown in Fig. 5. The decrease in h and t is connected with the fact that in a FG material the density of microcracks in the PDZ is higher than in a CG material. The latter may be illustrated by the data in Fig. 2. With increasing strains greater than 3% in LT and TL specimens the microcrack density becomes higher in the FG 1560 alloy. This is evidently caused by a higher homogeneity of microplastic deformation in fine-grained material. Grain refinement results in an

TABLE I Parameters characterizing the transverse crack growth resistance in the 1560 and 1960 alloys

Alloy	Structure (state)	Static crack		Fatigue crack*	
		λ, mm	l, mm	λ, mm	l, mm
	CG	6.8 ± 0.5	1.43 ± 0.03	0.92 ± 0.12	1.24 ± 0.02
1560	FG-1	4.2 ± 0.3	1.34 ± 0.03	0.65 ± 0.11	1.20 ± 0.02
	FG-2	3.5 ± 0.3	1.34 ± 0.02	0.41 ± 0.11	1.19 ± 0.02
1960	CT-1	5.3 ± 0.4	1.59 ± 0.03	0.84 ± 0.84	1.46 ± 0.03
	SPT	3.1 ± 0.3	1.46 ± 0.03	0.32 ± 0.12	1.26 ± 0.03

l – an average summarized path of a crack on one mm of its length.
λ – the size of plastic deformation zone at the tip of the crack;
*at $\Delta K = 11$ MPa·m$^{1/2}$.

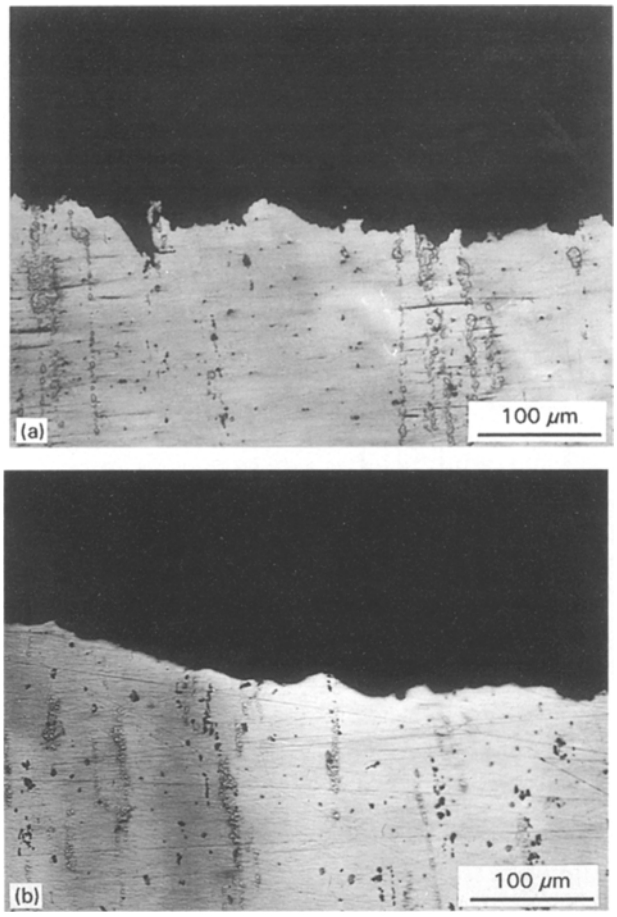

Figure 4 The surface asperities of static fractures in specimens with coarse- (a) and fine-grained structures made of 1560 alloy.

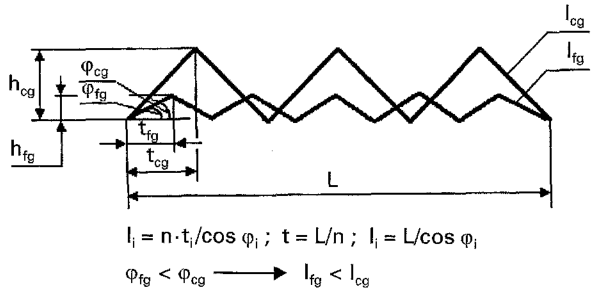

Figure 5 A schematic diagram explaining the difference in the surface relief of the crack in coarse- and fine-grained materials.

increasing number of grains where the local deformation causing crack initiation is likely to occur. Due to the increase of microcrack density the lower h and t correspond to the path of the main crack moving by breaking of the ligaments between the microcracks [21].

It is also necessary to take into account that microcracks initiate in the volume limited by the size of the PDZ. Due to its smaller size the maximum deviation of the long crack path is less in FG materials.

Other factors of grain size influence on crack growth resistance may be effective under fatigue.

With grain refinement, the deformation is accompanied by an increase in strain hardening [4] and an accumulation of crystalline structural defects. These defects provide the possibility of a higher rate of crack growth at the constant level of maximum stresses in the cycle [10, 15, 22].

A difference in the grain structure may also influence fatigue crack blunting. According to reference [23] the crack tip opening displacement (Δ) is determined by the equation:

$$\Delta = \frac{\Delta K^2}{4\sigma_y E} \qquad (7)$$

where ΔK is the range of stress intensity ratio and E is the Young's modulus. So due to higher yield stress if all other conditions are equal an FG material is characterized by smaller blunting of fatigue cracks, which also promotes an increase in the crack growth rate.

During the reverse movement of the dislocations during the unloading part of cycle, the size of the reverse PDZ is usually higher than the grain size in FG material. This leads to the formation of pile-ups near grain boundaries and the nucleation of cracks, this is the reason for the easier propagation of long ones [10, 15].

There are some other physical premises of grain size effect on A_d and A_s [10, 15].

According to the Hall–Petch relation (1) deformation and rupture of a FG material in an elastic–plastic region takes place in a higher external stress field, therefore the relaxation energy must also be higher.

Thus an analysis of Equation 5 shows that grain refinement leads to a negative effect on crack growth resistance.

3.4. The influence of grain size on the characteristics of high-cycle fatigue

An increase of the alloy fatigue limit (σ_R) and endurance under high-cycle fatigue (HCF) often accompanies the transition from CG to FG structure. This was clearly established in alloys on aluminium and other bases [1, 24–26]. The dependence of the fatigue limit on grain size is usually described by an equation similar to the Hall–Petch relation (1)

$$\sigma_R = \sigma_{iR} + k_R d_g^{-1/2} \qquad (8)$$

However, it is necessary to note, that whereas the physical nature of the Hall–Petch relationship is determined by the mechanism of strain hardening due to grain refinement, the rise of σ_R is caused by an increase in the crack initiation resistance. The increase of the HCF strength is due to the fact that the time needed for long crack formation is of the order of 60–90% of the total alloy endurance [24].

4. Conclusions

The nature of the observed effect that a fine grain structure has on the strength of aluminium alloys is determined by its influence on crack resistance characteristics.

The effect of grain size on crack resistance consists of an increase in the crack nucleation resistance and a decrease in crack growth resistance with grain refinement. This effect is similar under static, impact and cycle loading conditions, testifying to its common nature.

The influence of grain size on the main strength characteristics, such as fracture and impact toughness, fatigue limit, low- and high-cycle fatigue endurance and sensitivity to stress concentrators is determined by its effect on crack initiation and crack growth processes.

The grain size effect reported in the present work is not observed in some other investigations. Conflicting results are obtained especially in estimating the fracture toughness [18, 20, 27, 28] and fatigue strength [18, 30–32]. In addition, some reported data on the same materials are different [29, 31]. This discrepancy is due to simultaneous changes, not only in an alloy grain size but also in other structural parameters. The latter may be caused by variation in material composition (impurity contents), methods of ingot processing and further thermomechanical treatment. The structural differences could include the nature, size, distribution and volume fraction of inclusion and precipitates, the presence and parameters of substructure and crystallographic texture, etc. The influence of these parameters may be to reduce or even surpass the grain size effect.

Professor Rabinovich Meer Khaimovich passed away on 1 April 1996, in Ufa, Russia. Born on 26 January 1920, he graduated in 1944 from the Moscow Aviation-Technological Institute, in the Department of Hot Treatment of Metals, where he was also employed as a lecturer. Professor Rabinovich also received a Ph.D in 1950. For the next 45 years he worked at Ufa Aviation Institute, in the Department of Material Science and Technology. He was an author for more than 100 scientific publications inclulding 4 monographs. The book "Strength and superstrength of metals" was printed in Russia, Poland and Japan.

Professor Rabinovich was one of the leading Russian scientists in materials science, especially in methods of thermo-mechanical treatment, structure properties analysis and superplasticity of aluminium alloys.

His memories will be cherished by his wife (Ludmila), family, friends, co-workers and many generations of students.

References

1. M. Kh. RABINOVICH and M. V. MARKUSHEV, *J. Mater. Sci.* **30** (1995) 4692.
2. E. T. HALL, *Proc. Phys. Soc.* **B64** (1951) 747.
3. N. J. PETCH, *J. Iron and Steel Inst.* **174** (1953) 25.
4. R. HONECOMBE, "Plastic deformation of metals" (Mir, Moscow, 1972) (in Russian).
5. R. W. ARMSTRONG, *Trans. Inst. Met.* **39** (1986) 85.
6. L. K. GORDIENKO, "Substructural strengthening of metals and alloys" (Nauka, Moscow, 1973) (in Russian).
7. A. W. THOMPSON, *Met. Trans. A* **8A** (1977) 833.
8. V. I. ELAGIN, "Alloying the deformable aluminium alloys by transition metals" (Metallurgia, Moscow, 1975) (in Russian)
9. YU. M. VAINBLAT, Ph.D. thesis, Moscow, 1977 (in Russian).
10. J. C. WILLIAMS and E. A. STARKE, in Deformation, processing and structure. Proc. ASM Mat. Sci. Sem., St. Louis, 1984, edited by G. Krauss (1984) p. 279.
11. M. E. DRITZ, YU. P. GUK and L. P. GERASIMOVA, "The fracture of aluminium alloys" (Nauka, Moscow, 1980) (in Russian).
12. G. I. BATURIN, P. E. PANFILOV and M. A. BOCKMAN, *FMM* **4** (1987) 827 (in Russian).
13. I. N. FRIDLYANDER, E. V. EKHINA, T. M. KUNYAVSKAYA and V. L. LIKIN, *MiTOM* **2** (1985) 62 (in Russian).
14. M. Kh. RABINOVICH, M. V. MARKUSHEV and R. R. KHABIBULLINA, in "Improvement of materials reliability and durability on the basis of new methods of heat and chemical-heat treatment", Proceedings of the Materials Science Conference, USSR, 1988 (USSR Academy of Sciences, Khmelnitsky, 1988) p. 63.
15. A. LASALMONIE and J. L. STRUDEL, *J. Mater. Sci.* **21** (1986) 1837.
16. P. A. DULNEV and P. I. KOTOV, "Thermal fatigue of metals", (Mashinostroenie, Moscow, 1980) (in Russian).
17. V. I. VLADIMIROV and A. N. ORLOV, *FTT* **2** (1969) 370 (in Russian).
18. V. V. TELESHOV, YU. K. SHTOVBA, V. I. SMOLENTZEV and O. M. SIROTKINA, *MiTOM* **7** (1983) 29 (in Russian).
19. V. T. TROSHENKO and L. A. SOSNOVSKY, "Fatigue strength of metals and alloys" (Naukova dumka, Kiev, 1987) (in Russian).
20. V. G. KUDRYASHOV and V. I. SMOLENTZEV, "Fracture toughness of aluminium alloys" (Metallurgia, Moscow, 1976) (in Russian).
21. V. I. VLADIMIROV, "Physical nature of metals rupture" (Metallurgia, Moscow, 1984) (in Russian).
22. E. HORNBOGEN and J. STANIEC, *J. Mater. Sci.* **9** (1974) 879.
23. M. SRINIVAS and G. MALAKONDIAH, *Scripta Metall.* **22** (1986) 689.
24. V. S. IVANOVA and V. Ph. TERENTIEV, "The nature of metals fatigue" (Metallurgia, Moscow, 1975) (in Russian).
25. R. W. ARMSTRONG, *Met. Trans. A* **1** (1970) 1169.
26. A. W. THOMPSON, *Scripta Mettall.* **5** (1971) 859.
27. E. D. RUSSO, M. CONSERWA, M. BURATTI and F. GATTO, *Mat. Sci. Engng.* **14** (1974) 23.
28. G. T. HAHN and A. R. ROSENFIELD, *Metal. Trans. A* **6** (1975) 653.
29. G. LUTJERING and A. GYSLER, in "Ermudungs verhalten metallische Werkstoffe, edited by D. Munz (Oberrursel, 1985) p. 39.
30. D. SHONG, S. SHIGEOKI and H. SHIGENORI, *J. Jap. Inst. Light Met.* **36** (1986) 434.
31. C. C. BAMPTON and J. W. EDINGTON, *J. Eng. Mat. Techn.* **105** (1984) 55.
32. N. E. PATON, C. H. HAMILTON, J. WERT and M. MAHONEY, *J. Metals* **8** (1982) 21.

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