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

## Part I *The phenomenology of the influence of fine grained structure and superplastic deformation on the strength of aluminium alloys*

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The influences of fine grained (FG) structure and superplastic deformation on the mechanical properties under quasistatic, impact and cycling loading conditions have been established for the aluminium alloys 1560 (Al–Mg–Mn), 1141 (Al–Cu–Mg–Ni–Fe) and 1960 (Al–Zn–Mg–Cu–Zr). FG materials compared with recrystallized coarse grained (CG) ones improve tensile strength and ductility and high-cycle fatigue endurance, but reduce static and impact toughnesses. The effect of grain refinement on crack resistance is directly manifested in the difficulty of crack initiation and in facilitating its growth. Blanks with FG structure, after superplastic processing, are recommended instead of CG recrystallized ones for the production of principal parts whose service life is limited by first crack appearance. In the presence of cracks, higher strength can be obtained by the use of blanks made of CG commercial semifinished products and those manufactured by traditional methods of hot die forging.

## 1. Introduction

The problem of ensuring the necessary strength of alloys in parts obtained by superplastic (SP) deformation is one of the less studied aspects of practical use of structural (micrograined) superplasticity. The influence of SP deformation on the alloys' properties is determined partly by their fine grained (FG) structure, which remains after deformation, and partly by specific mechanism of SP flow in the alloy structure. The latter is manifested in void formation, some grain coarsening, formation of particle free zones, changes in grain boundary structure, etc. [1–5]. At relatively low strains and a stress state mode typical of traditional methods of die forging, structural changes promoted by SP deformation, and consequently their influence, on the properties of alloys are not significant. The present study therefore deals with the estimation of the influence of FG structure on service properties of a number of commercial aluminium alloys. The peculiarities of mechanical behaviour changes in alloys after SP treatment have been estimated in comparison with the properties of coarse grained (CG) alloys subjected to conventional treatment (isothermal forging).

## 2. Experimental procedure

The main requirements of the constructional material are a high level of plastic deformation resistance, ductility and crack resistance. According to the conditions of "safe flaw", materials with low crack growth rate and high fracture toughness are preferable. This gives the opportunity of increasing the durability of exploitation and test inspection periods of details with cracks. But, in accordance with the demands of "safe resource", the articles durability is limited to the period prior to crack formation. Thus the influence of grain size, *d*, and modes of treatment on the mechanical properties characterizing crack formation and crack growth resistance under different loading conditions, quasistatic, impact and cycling, were investigated.

Three commercial alloys 1560, 1141 and 1960 (Table I) were tested. The choice of these alloys was determined by the difference in the levels of their strength and ductility and the difference in microstructure, such as the volume fraction of inclusions of excess primary phases which provide banded structures and dispersed particles, and precipitate free zones (PFZs). Besides, the investigated alloys with FG structure are capable of SP deformation [1].

All alloys have a matrix type structure. The 1560 alloy is non-heat treatable. At room temperature it is characterized by grains of aluminium solid solution with bands of coarse primary particles of $Al_{10}Mg_2Mn$ and $Al_6Mn$ phases [6], and with secondary precipitates

 

TABLE I Chemical composition of the alloys

| Alloy | Amount (wt %) of elements | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | Zn | Mg | Cu | Zr | Mn | Ni | Fe | Si | Al |
| 1560 | – | 6.2 | – | – | 0.6 | – | 0.15 | 0.17 | Balance |
| 1141 | – | 1.7 | 2.3 | – | – | 1.2 | 1.2 | 0.20 | Balance |
| 1960 | 8.2 | 2.6 | 2.5 | 0.14 | 0.05 | – | 0.10 | 0.12 | Balance |

of $\beta(Al_3Mg_2)$ and $Al_6Mn$ phases. In the heat treatable alloy 1141 the banded structure consists of coarse particles of $FeNiAl_9$ phase and grains hardened by precipitates of $S(Al_2CuMg)$ phase, products of artificial ageing. Besides, there are boundary PFZs. The high strength 1960 alloy is the most complex one among those studied concerning their structural state after heat treatment: bands are constituted of coarse particles of two phases S and T $(Al_2Mg_3Zn_3)$; hardening dispersed phases are of two types, T′ and η′ [metastable modifications of T and η $(MgZn_2)$ phases]. Moreover there are primary (stable) and secondary (metastable) particles of zirconium aluminide, $Al_3Zr$, and boundary PFZs [7]. The specific feature of the 1960 alloy is that in blanks, processed by traditional methods of hot forging, substructural strengthening (press-effect) typical of hot pressed bars remains after quenching.

The effect of grain size on the mechanical properties was determined for the 1560 and 1141 alloys having completely recrystallized structures. The FG structure (FG-1) was obtained by annealing the samples in a salt bath at 400 °C for 30 min following the flattening of blanks made of hot pressed bars in a hydraulic press at room temperature with a strain of 40–45%. The CG state resulted from annealing fine grained (FG-1) blanks at 400 °C for 1 h, initially deformed by 8–10% at room temperature. In both cases the total degree of deformation was similar. It permitted one to obtain structural states differing from each other only by grain sizes and to neglect the influence of other structural parameters on the mechanical behaviour of the alloys. The FG-2 structure of the 1560 alloy was obtained by additional hot deformation, namely by flattening fine grained (FG-1) blanks by 50% in the optimal temperature–strain rate, $T/\dot{\varepsilon}_0$, interval of superplasticity of the alloy ($T = 410\,°C$, $\dot{\varepsilon}_0 = 2 \times 10^{-3}\,s^{-1}$).

The 1560 alloy was tested after water quenching from the temperature of annealing (for FG-2 deformation) and natural ageing for 1.5 months, and alloy 1141 was tested in the T6 condition: water cooling from 520 °C and ageing at 190 °C for 10 h.

The influence of the mode of treatment was investigated on the 1960 alloy. Three types of treatment, imitating the main methods of isothermal forging were compared. Conventional treatment one (CT-1), included 50–60% flattening of coarse grained blanks made of hot pressed bars in a hydraulic press under conditions when the alloy with the FG structure exhibited superplasticity ($T = 450\,°C$, $\dot{\varepsilon}_0 = 3 \times 10^{-3}\,s^{-1}$) [1]. At conventional treatment two (CT-2), analogous deformation of the same blanks was conducted at $T = 400\,°C$ and $\dot{\varepsilon}_0 = 10^{-1}\,s^{-1}$, i.e. in the range of the alloy's typical isothermal forging regimes. The superplastic treatment (SPT) involved two main stages. The first was the formation of FG structure by means of intermediate thermomechanical treatment (ITMT) [8, 9], including quenching of blanks from 460 °C, overageing at 400 °C for 10 h, flattening with a strain of 40–42% at 200 °C, recrystallization annealing in a salt bath at a temperature of 460 °C for 30 min; and the second stage consisted of flattening deformation by 50% under the optimum SP conditions. The final stage of every mode of treatment was T6 heat hardening: water quenching from 460 °C, ageing at 140 °C for 16 h.

The average grain size in longitudinal and transverse directions was determined by the mean linear intercept method on samples cut from the centre of a blank. Samples were etched in Keller's reagent.

The fine structure was studied in a transmission electron microscope, Tesla BS-540, on foils prepared by means of jet electropolishing in a 10% nitric acid–water solution. The fracture surfaces were studied on a scanning electron microscope, JSM-840.

Quasistatic tension tests were carried out on smooth cylindrical specimens with a diameter of 5 mm and a working part length of 25 mm and specimens with a ring notch having different radii at their tops. The critical stress intensity factor, $K_{Ic}$, was determined on compact specimens with a thickness of 20 mm. The work needed for static crack formation and growth, $a_f$ and $a_g$, was estimated by Kahn's method [10]. The impact toughness was studied on Charpy's specimens (10 × 10 × 55 mm). Fatigue tests were carried out on specimens with a minimum diameter of the polished working part equal to 5 mm. The low cycle fatigue endurance was evaluated under repeated tension conditions with a frequency, $\nu$, of 0.5 Hz and asymmetry coefficient $R = 0.1$; and in the region of high cycle fatigue, under conditions of pure bending with a frequency of 50 Hz and $R = -1$. The fatigue crack growth rate (FCGR) was determined on the specimens for $K_{Ic}$ estimation at repeated tensions with a frequency of 11 Hz and $R = 0.1$. The fatigue crack formation resistance was evaluated on the same specimens by the value of $N_1$, the number of cycles required for the formation of a crack of 1 mm length.

All types of tests were carried out in laboratory air.

As the orientation of specimens under investigation greatly influenced the mechanical characteristics of the aluminium alloys, tests were conducted on longitudinal (LT) and transverse (TL) specimens. In the first case, cracks propagated perpendicularly to the bands of excess primary phases (a transverse crack). In the second case, cracks grew along bands (a longitudinal crack).

## 3. Results and discussion
### 3.1. Microstructure of the alloys
The grain structure of the 1141 and 1560 alloys (Figs 1 and 2) is completely recrystallized. Grain sizes are given in Table II. The structure of the 1960 alloy after conventional treatment is fibrous and unrecrystallized (Fig. 3a–d). In states CT-1 and CT-2 the subgrain size is equal to 1–5 and 4–9 μm, respectively, and, judging by the diffraction contrast and worse dislocation resolving at subgrain boundaries, subgrain misorientation is stronger after CT-1. After SP treatment the 1960 alloy is practically fully recrystallized (Fig. 3e–f). The presence of separate unrecrystallized grains and variation in size of the grains is the result of non-complete recrystallization after ITMT: the specific volume of fine grains did not exceed 64%. Formation of the FG structure is completed during further SP deformation.

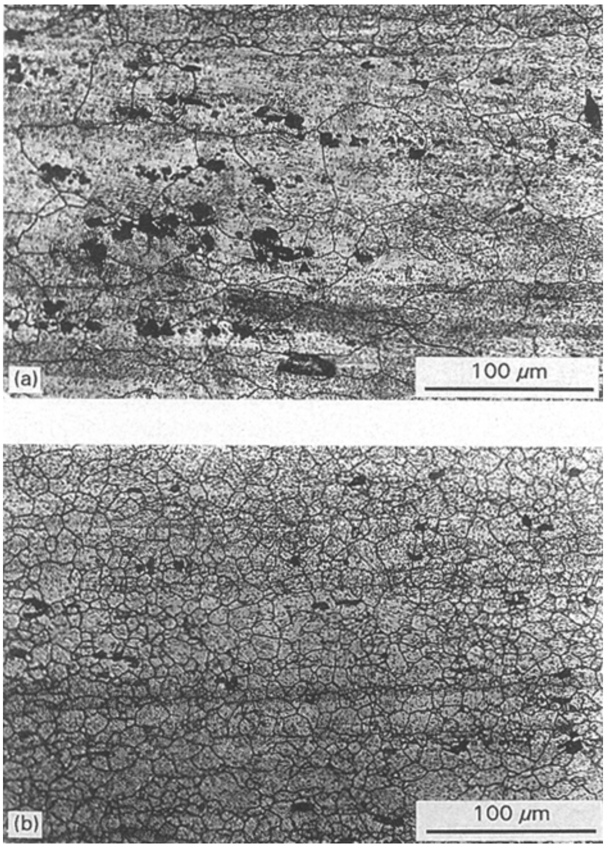


*Figure 1* (a) Coarse grained (CG) and (b) fine grained (FG-1) structures of the 1141 alloy.

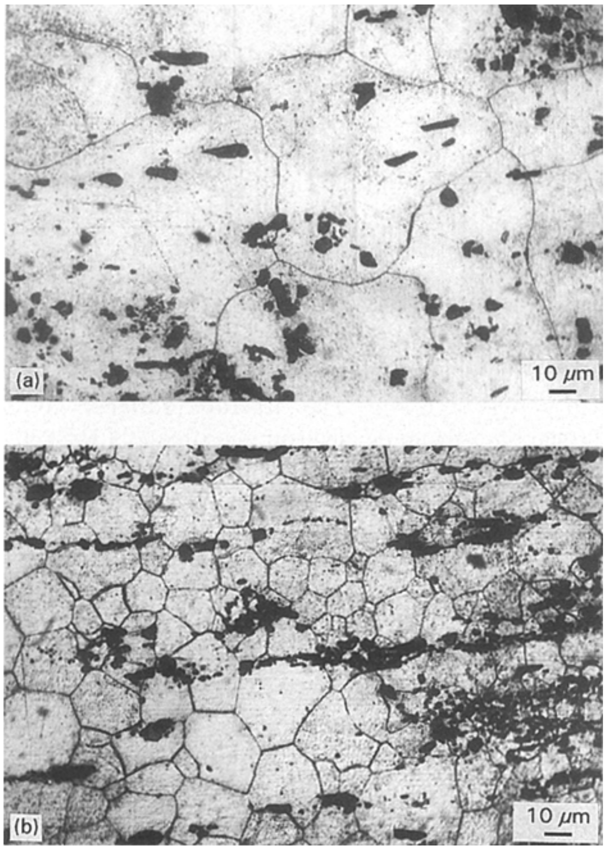


*Figure 2* (a) Coarse grained (CG) and (b) fine grained (FG-1) structures of the 1560 alloy.

### 3.2. Mechanical properties
#### *3.2.1. Grain size influence*
The mechanical properties of the 1560 and 1141 alloys established by quasistatic tension are given in Table II.

The 1560 alloy grain refinement resulted only in slight increase of strength, $\sigma_{0.2}$ (yield stress) and $\sigma_t$ (tensile strength), and plasticity, $\delta$ (specific elongation and $\varphi$ (reduction of area), determined on

TABLE II Mechanical properties of the 1560 and 1141 alloys at quasistatic loading

| Alloy | Structure (state) | Specimen orientation | $\bar{d}$ (μm) | $\sigma_{0.2}$ (MPa) | $\sigma_t$ (MPa) | $\delta$ (%) | $\varphi$ (%) | $\gamma$ (J) | $\sigma_t^{0.05}$ (MPa) | $\sigma_t^{0.30}$ (MPa) | $a$ (kJ m$^{-2}$) | $a_f$ (kJ m$^{-2}$) | $a_g$ (kJ m$^{-2}$) | $K_{Ic}$ (MPa × m$^{1/2}$) |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 1560 | CG | long. (LT) | 42 | 157 | 352 | 23.2 | 31 | 30 | 423 | 448 | 308 | 143 | 165 | 23.2 |
| | | trans. (TL) | 28 | 154 | 343 | 22.3 | 26 | 28 | – | – | 159 | 88 | 71 | 22.7 |
| | FG-1 | long. (LT) | 12 | 162 | 354 | 24.6 | 34 | 32 | 415 | 450 | 325 | 155 | 170 | 25.0 |
| | | trans. (TL) | 9 | 156 | 346 | 21.1 | 24 | 27 | – | – | 167 | 85 | 82 | 21.1 |
| | FG-2 | long. (LT) | 9 | 165 | 362 | 25.1 | 36 | 33 | – | – | 353 | 173 | 180 | 29.0 |
| | | trans. (TL) | 6 | 155 | 346 | 21.8 | 25 | 27 | – | – | 186 | 94 | 92 | 29.3 |
| 1141 | CG | long. (LT) | 48 | 352 | 405 | 7.3 | 11 | 14 | – | – | – | – | – | 31.1 |
| | | trans. (TL) | 39 | 349 | 402 | 5.6 | 7 | 10 | – | – | – | – | – | 23.5 |
| | FG-1 | long. (LT) | 14 | 360 | 426 | 9.8 | 13 | 20 | – | – | – | – | – | 30.4 |
| | | trans. (TL) | 11 | 347 | 406 | 7.2 | 9 | 14 | – | – | – | – | – | 25.4 |

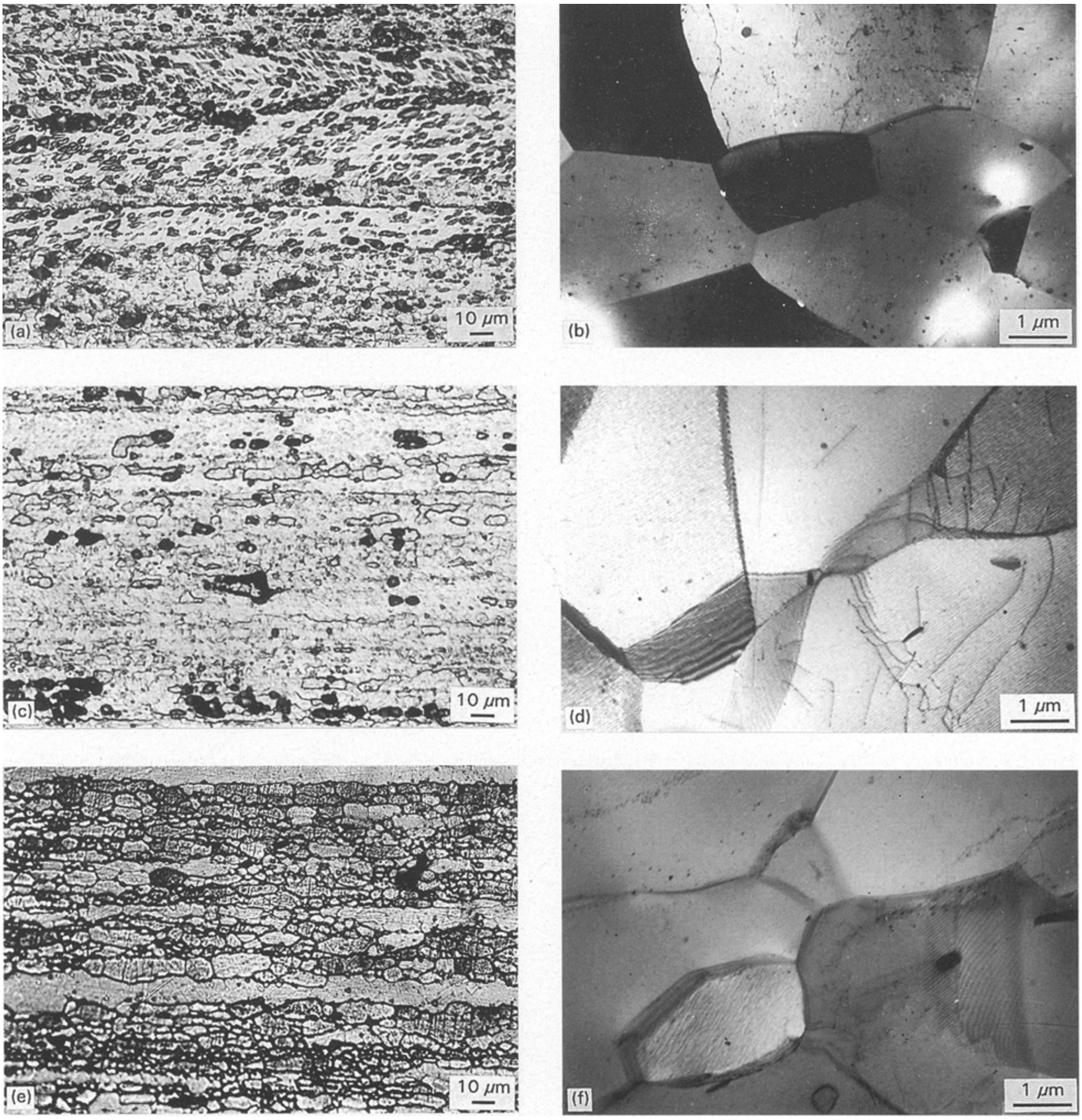


*Figure 3* Optical and transmission electron microscope structure of the 1960 alloy after (a, b) conventional treatment (CT-1) and (c, d) CT-2 and (e, f) superplastic treatment (SPT).

longitudinal specimens. In the 1141 alloy it caused a significant increase of such characteristics in the same direction. The difference in grain size effect on the alloys' strength is connected with the peculiarities of their structure. The non-heat treatable 1560 alloy contain an insignificant number of secondary precipitates of a metastable modification of β-phase. This facilitates the process of dynamic recovery, which in its turn neglects the influence of grain size on the strength properties [11, 12]. In the 1141 alloy aged to a maximum strength, the process of dynamic recovery is more retarded. Under these conditions the influence of grain size is exhibited more strongly.

The fracture surface analysis of longitudinal specimens after static tension showed that in CG and FG 1560 alloy failure had a ductile character. In general the fracture is transcrystalline, with dimple domination (Fig. 4a, c). The volume of the transgranular fracture is practically independent of grain size and constitutes 88–92% for all states studied. The main fracture difference is in the size of areas with transcrystalline quasicleavage, which are considerably smaller in a fine grained material. This testifies to the more homogeneous character of plastic deformation in an alloy with FG structure, and that was the reason of some increase of its plasticity [6, 12–14]. In the 1141 alloy the fracture is complex too (Fig. 5a, c), but with a larger volume fraction of intercrystalline mode ($22 \pm 5\%$) . It is probably caused not only by the existence of a larger number of particles of excess phases, but also by the presence of grain boundary PFZs in the 1141 alloy. Grain refinement did not lead to qualitative changes of the fracture structure. The higher plasticity of the fine grained state is likely to be

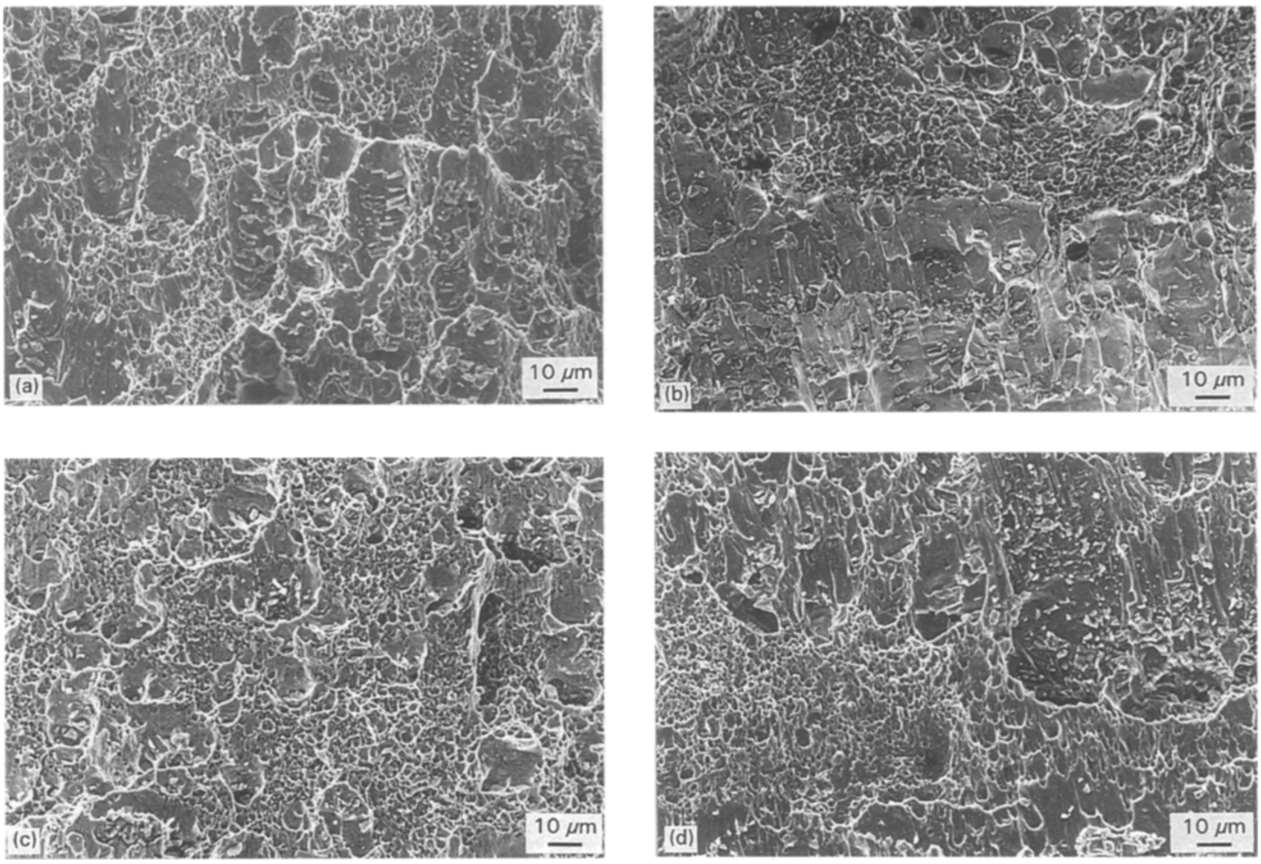


*Figure 4* Fracture surfaces of (a, c) longitudinal and (b, d) transverse tensile samples of the 1560 alloy with coarse grained (CG) (a, b) and fine grained (FG-1) (c, d) structures.

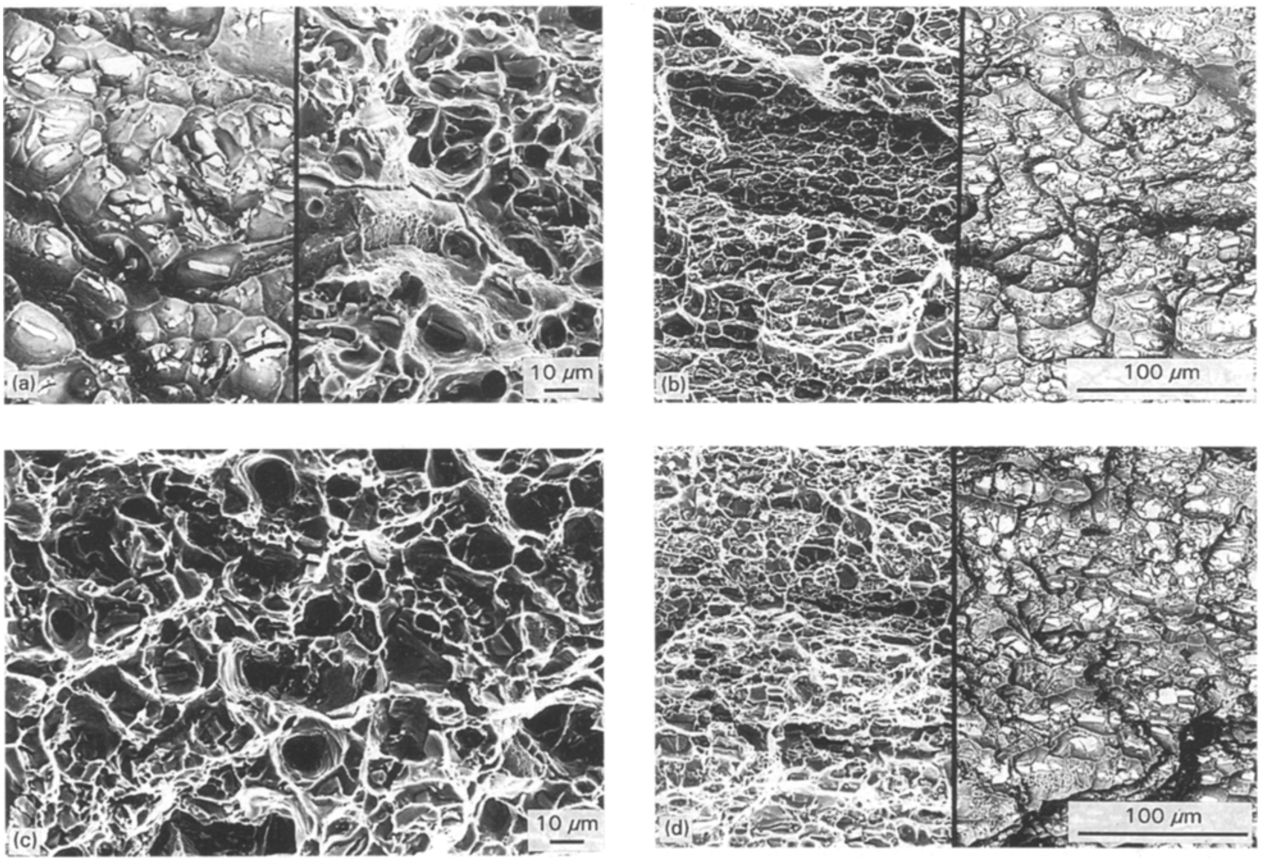


*Figure 5* Fracture surfaces of (a, c) longitudinal and (b, d) transverse tensile samples of the 1141 alloy with coarse grained (CG) (a, b) and fine grained (FG-1) (c, d) structures.

connected with greater homogeneity of plastic deformation in material volume as well.

The effect of grain size on static strength and plasticity of the 1560 alloy is not practically found in the transverse direction (Table II). This may be caused by the preferable influence of primary phases constituting the banded structure on the processes of crack formation and crack growth which exceeds the effect of grain size influence. The latter is illustrated by the type of fractures sufficiently well (Fig. 4b, d).

The analogous dependence of strength on $d$ in the transverse direction is observed for the 1141 alloy (Table II). However, in contrast to alloy 1560, with decreasing grain size, an increase of plasticity in the transverse direction is observed, though according to the mode of fracture the character of failure did not quantitatively change (Fig. 5b, d). This testifies to the fact that for the age hardenable alloy the effect of grain size is exhibited more strongly in a given direction. The cause of plasticity rising is probably the same as in the longitudinal direction.

The work needed for the static failure of cylindrical specimens, $\gamma$, (Table II) increases with decreasing grain size. For the 1141 alloy this effect is more significant and observed in longitudinal and transverse directions.

For the purpose of determining grain size effect on the work needed for crack formation and crack growth during failure, longitudinal cylindrical specimens with a circle notch of different sharpness were tested. According to the data for the 1560 alloy (Table II), at a radius of 0.05 mm in the top of the notch the fine grained material exhibited less tensile strength, but at a softer notch (0.3 mm) the grain size influence was not observed. This testifies to decreasing transverse crack growth resistance with grain refinement: as the sharpness of the notch increases, the stage of crack propagation becomes more prominent in the fracture process. Taking into account the fact that grain refinement leads to $\gamma$ increase and crack growth resistance decrease; it should be considered that the crack formation work in an FG material is higher than that in a coarse grained one.

Tests conducted according to Kahn's method on the 1560 alloy, as well as standard tension tests, showed that irrespective of specimen orientation (LT or TL) the specific work of failure, $a$, increased with decreasing grain size of the alloy (Table II). This is connected not only to the increase of crack formation work, $a_f$, but also to crack growth work, $a_g$, as well. However, in the case of longitudinal specimens $a_f$ increases to a greater extent.

At the same time grain refinement of the 1560 alloy results in increasing fracture toughness, $K_{Ic}$, (Table II). But due to high ductility of the alloy, definition of the critical stress intensity factor on specimens of relatively small thickness has not been sufficiently correct. In this case, according to [15] the rise of $K_{Ic}$ is caused by an increase of crack formation resistance, but not by its growth.

For the less plastic 1141 alloy whose $K_{Ic}$ was determined correctly, it was established that grain refinement resulted in an increase of fracture toughness of TL specimens and its decrease in LT ones. This testifies to the rise of crack growth resistance along the bands and its reduction in a direction perpendicular to the bands in the FG structure.

Thus the transition from a coarse grained structure to a fine grained one in aluminium alloys non-uniformly influences the characteristics of mechanical properties under static loading. The main effect of grain refinement lies in the increase of crack formation resistance. The work needed for crack growth may be increased or decreased depending on the direction of their propagation; along or across the bands. The reduction of crack growth resistance lead to the increase of fine grained material sensitivity to sharp stress concentrators.

Impact tests have shown that change of grain size in the studied range practically does not influence toughness, obtained in TL specimens of both alloys (Table III). Crack growth resistance characterized by the value of KCC (specimen with a fatigue crack) remains approximately constant. The work needed for crack formation, which may be estimated according to [16] by the difference between KCU (specimen with a notch, $r = 1$ mm) and KCC, does not depend on grain size. The absence of a grain size effect on impact toughness in the transverse direction is caused by the prevailing influence of the banded structure on the process of failure, as in static tests. But impact strength in the longitudinal direction decreases

TABLE III Mechanical properties of the 1560 and 1141 alloys at impact loading

| Alloy | Structure (state) | Specimen orientation | KCU ($kJ\,m^{-2}$) | $\Delta$KCU[a] ($kJ\,m^{-2}$) | KCC ($kJ\,m^{-2}$) | $\Delta$KCC[a] ($kJ\,m^{-2}$) | $a_f^b$ ($kJ\,m^{-2}$) | $\Delta a_f^a$ ($kJ\,m^{-2}$) |
|---|---|---|---|---|---|---|---|---|
| 1560 | CG | LT | 460 | – | 280 | – | 180 | – |
| | | TL | 270 | – | 120 | – | 150 | – |
| | FG-1 | LT | 430 | – 30 | 250 | – 30 | 180 | 0 |
| | | TL | 260 | – 10 | 120 | 0 | 140 | – 10 |
| | FG-2 | LT | 420 | – 40 | 210 | – 70 | 210 | + 30 |
| | | TL | 280 | + 10 | 120 | 0 | 160 | + 10 |
| 1141 | CG | LT | 195 | – | 90 | – | 105 | – |
| | | TL | 135 | – | 50 | – | 85 | – |
| | FG-1 | LT | 155 | – 40 | 65 | – 25 | 90 | – 15 |
| | | TL | 125 | – 10 | 55 | + 5 | 70 | – 15 |

[a] $\Delta$KCU, $\Delta$KCC and $\Delta a_f$ are the differences between the values of corresponding parameters for CG and FG alloys.
[b] $a_f$ = KCU – KCC.

with grain refinement and the decrease of both characteristics KCU and KCC being practically the same. This testifies to the fact that the reduction of work required for failure of the LT specimen with a notch is caused by lower crack growth resistance in fine grained alloys. Judging by KCU and KCC differences, crack formation work is nearly the same for CG and FG structures.

The decrease in impact toughness due to grain refinement of aluminium alloys has been established in [17–19, etc.] too.

On the basis of the 1560 and 1141 alloys' properties studied under two loading conditions, quasistatic and impact, one may note a common feature of the behaviour of materials with different grain size: grain refinement results in a decrease of transverse crack growth resistance. As for the difference in the influence of loading conditions on transverse and longitudinal crack formation work and the growth resistance of longitudinal ones, it is evidently caused by the difference between stressed and deformed states in tensile and impact tests, and by different sensitivity of methods of crack initiation and crack growth characteristics estimation as well.

Fig. 6 shows the results of testing the 1560 alloy in CG and FG-1 states under cycling loading. In the region of low cycle fatigue, with reducing grain size the endurance of longitudinal specimens decreases at bending (Fig. 6a) and at repeated tension tests (Fig. 6b). However, high cycle endurance of the FG alloy is higher and the fatigue limit, $\sigma_{-1}$, is consequently higher too.

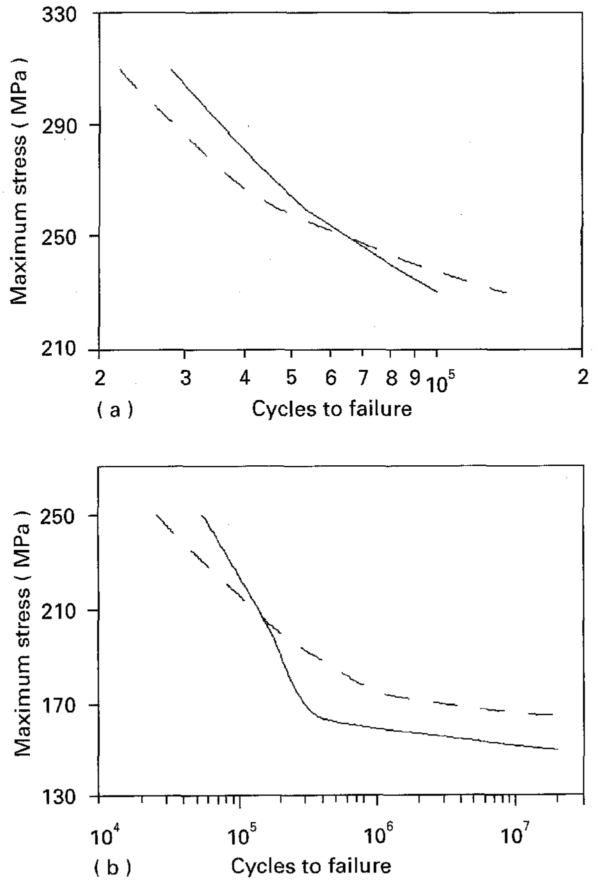


*Figure 6* Fatigue curves of CG (——) and FG-1 (– –) 1560 alloy: (a) $R = 0.1$, $\nu = 0.5$ Hz; (b) $R = -1$, $\nu = 50$ Hz.

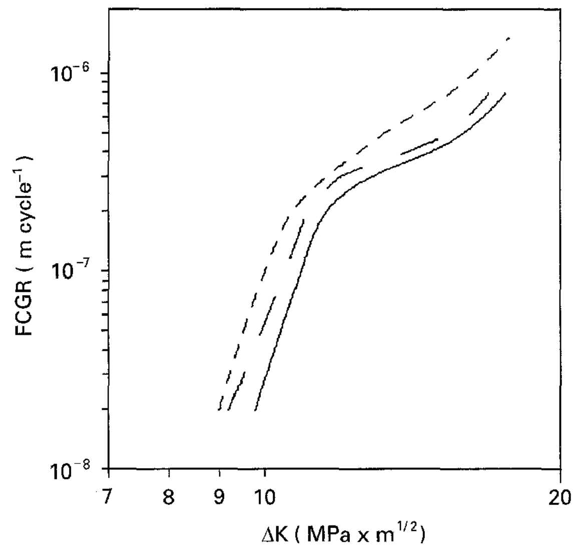


*Figure 7* Fatigue crack growth rate (FCGR) versus the ratio of stress intensity factor, $\Delta K$, for CG (—— LT and – – – TL specimens) and FG-1 (— — LT and TL specimens) 1560 alloy: $R = 0.1$, $\nu = 11$ Hz.

The grain size effect in the 1560 alloy on fatigue crack growth rate is shown at Fig. 7. The dependence of FCGR on the ratio of stress intensity factor, $\Delta K$, is plotted on the basis of the average data of seven to ten specimen tests for each state of the alloy. The tendency to increase the rate of transverse fatigue crack propagation with decreasing grain size is clearly observed. In the case of a longitudinal crack, when it grows along bands, a higher FCGR is observed in CG material. The result of this is that FCGR anisotropy is practically absent in fine grained alloy and is well expressed in a CG one. According to data in the literature the observed behaviour of the 1560 alloy is non-typical. Namely, [12, 17–22] show the rate increase of longitudinal and transverse cracks alongside the grain refinement of aluminium alloys. The latter is typical of the alloys on other bases [23–26].

It has been established that the decrease of grain size in the 1560 alloy increases the time necessary for the formation of a fatigue crack and its growth up to 1 mm long, $N_1$, in a low cycle region (Fig. 8). In FG-1 material the value of $N_1$ is 1.5–2.0 and is two to three

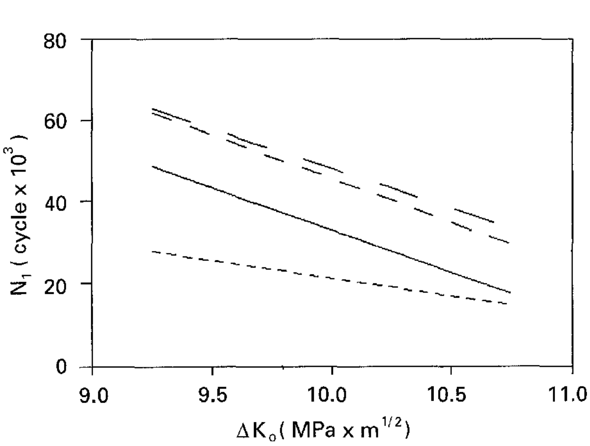


*Figure 8* Cycles required for the formation of a fatigue crack 1 mm long, $N_1$, versus the range of initial stress intensity factor, $\Delta K_o$, for CG (—— LT and – – – TL specimens) and FG-1 (— — LT and – – – TL specimens) 1560 alloy: $R = 0.1$, $\nu = 0.5$ Hz.

times higher than in CG material when crack growth is in transverse and longitudinal directions, respectively. Moreover, the fine grained alloy appears to be isotropic concerning this characteristic too. Taking into account the established increase of transverse crack growth rate due to grain refinement it is necessary to consider that the increase of $N_1$ is caused by the increase of crack formation resistance in the fine grained alloy. This result corresponds to the data for aluminium alloys [12, 27] and other alloys bases [23, 26]. On the basis of the obtained dependencies one cannot define the effect of fine grained structure on crack growth resistance of longitudinal ones as it is impossible to establish which factor, a higher formation work or a lower FCGR, has contributed to the increase of $N_1$.

The difference in the fine and coarse grained 1560 alloy fatigue limit under low and high cycle conditions depends on the change in the correlation between the duration of crack nucleation and crack growth in the process of failure. The lower endurance of the alloy with fine grained structure under low cycle fatigue is caused by the prevailing influence of FCGR increase. Under high cycle fatigue conditions according to [28], the fatigue life is mainly determined by the time necessary for crack formation. Thus the increase of endurance and fatigue limit of an alloy at the transition to FG structure may be caused only by the observed increase of crack formation resistance.

The results of fatigue strength estimation of the 1141 alloy differ from that of the 1560 alloy. With the decrease of grain size one observed not only the increase in high cycle, but also in low cycle, fatigue limit (Fig. 9). The same results were obtained for that alloy in [19, 20]. However, the behaviour of the 1141 alloy is not contrary to the data on the grain size effect in 1560 alloy. According to the present results and those of [20], grain refinement in the 1141 alloy leads to an increase in crack growth rate too. That is why a higher low, as a high cycle, fatigue strength of an FG alloy is to be considered the result of increased resistance to crack formation only. The latter may be explained by a higher ratio of yield–tensile strength, and due to that the level of applied maximum stress in a cycle does not

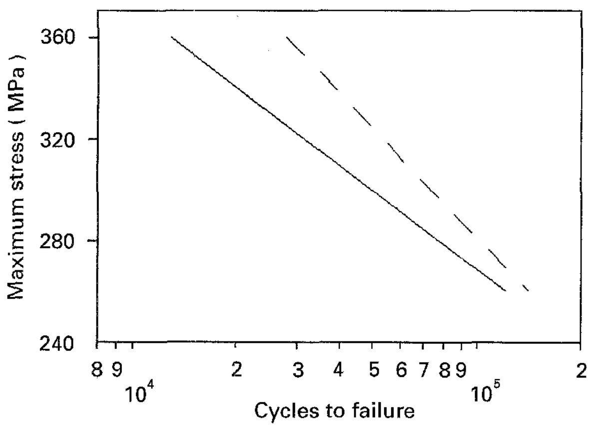


*Figure 9* Low cycle fatigue curves of CG (——) and FG-1 (– –) 1141 alloy: $R = 0.1$, $\nu = 0.5$ Hz.

exceed $\sigma_{0.2}$ in contrast to the 1560 alloy. As a result, the absence of macroplastic deformation at the beginning of loading cycles increases the period of crack formation in the total durability.

So, on the basis of these results and literature data one may conclude that irrespective of loading conditions the effect of decreasing grain size at the transition to the FG structure of aluminium alloys on crack resistance is practically identical: crack formation becomes more difficult, while crack growth is facilitated. The decrease of transverse crack growth resistance is usually observed, but the propagation of longitudinal cracks may be reduced, which depends on the alloy composition. Under similar loading conditions the peculiarities of the grain size effect on the characteristics of mechanical behaviour are determined by the relative periods of crack formation and growth in fracture duration, which in turn depends on the nature of the alloy.

### 3.2.2. The influence of treatment mode

The results of quasistatic and impact loading tests of 1960 alloy are given in Table IV.

The maximum static strength in the cross direction is obtained after conventional treatment one. After CT-2 strength characteristics are noticeably low. The decrease of deformation temperature and the increase of its rate under CT lead to higher accumulation of energy (hot strain hardening) which causes greater softening of the alloy by recovery during further heating of blanks for quenching. This results in partial removal of the press-effect and consequently in the decrease of strength. A similar strength as under CT-2 has been exhibited by the material with recrystallized FG structure (SPT). In the cross-direction the alloy possesses maximum strength after CT-1 as well. After SPT it is lower by 15–25 MPa and minimum after CT-2.

The plasticity of the FG alloy in longitudinal and transverse directions is higher than in unrecrystallized states. This may be explained by the difference in fracture mechanisms. After conventional treatments the intersubcrystalline and intercrystalline character of fractures along PFZs at low and high angle boundaries prevails (Fig. 10a, b). And after SPT the fracture mode is more ductile at prevailing dimple failure (Fig. 10c).

Alloy grain refinement leads to an increase of work needed for static rupture of cylindrical specimens, $\gamma$, and this rise is more noticeable in the transverse direction (Table IV).

Judging by the $\sigma_t^r$ and $K_{Ic}$ values static crack growth resistance is lowest in the alloy with the FG structure. This is the reason for higher sensitivity to a sharp stress concentrator. After SPT the alloy exhibits a more pronounced decrease of resistance to failure of specimens with a concentrator with increasing sharpness of the notch (Table IV). This behaviour is similar to that established in the 1560 alloy.

Upon impact bending the alloy also manifests the lowest fracture toughness after SPT. The cause is

TABLE IV Mechanical properties of the 1960 alloy at quasistatic and impact loadings

| Treatment | Specimen orientation | $\bar{d}$ (μm) | $\sigma_{0.2}$ (MPa) | $\sigma_t$ (MPa) | δ (%) | φ (%) | γ (J) | $K_{Ic}$ (MPa × $m^{1/2}$) | $\sigma_t^{0.05}$ (MPa) | $\sigma_t^{0.30}$ (MPa) | KCU (kJ $m^{-2}$) |
|---|---|---|---|---|---|---|---|---|---|---|---|
| CT-1 | long. (LT) | – | 632 | 660 | 7.0 | 22 | 23 | 26.5 | 807 | 838 | 64 |
| | trans. (TL) | – | 560 | 594 | 3.9 | 10 | 11 | 21.7 | – | – | 40 |
| CT-2 | long. (LT) | – | 585 | 615 | 7.4 | 19 | 22 | 26.6 | – | – | 146 |
| | trans. (TL) | – | 513 | 550 | 5.1 | 11 | 14 | 20.4 | – | – | 43 |
| SPT | long. (LT) | 13 | 584 | 607 | 8.5 | 22 | 25 | 21.3 | 723 | 784 | 56 |
| | trans. (TL) | 9 | 545 | 571 | 5.6 | 15 | 16 | 17.2 | – | – | 41 |

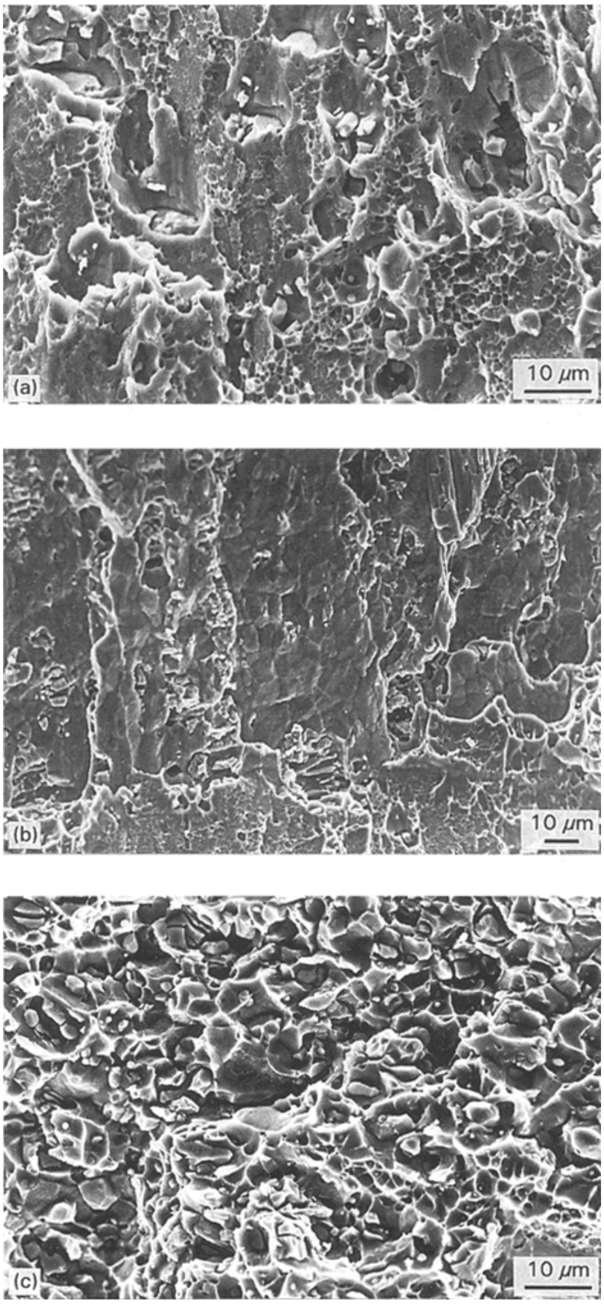


*Figure 10* Fracture surfaces of longitudinal tensile samples of the 1960 alloy after (a) CT-1, (b) CT-2 and (c) SPT.

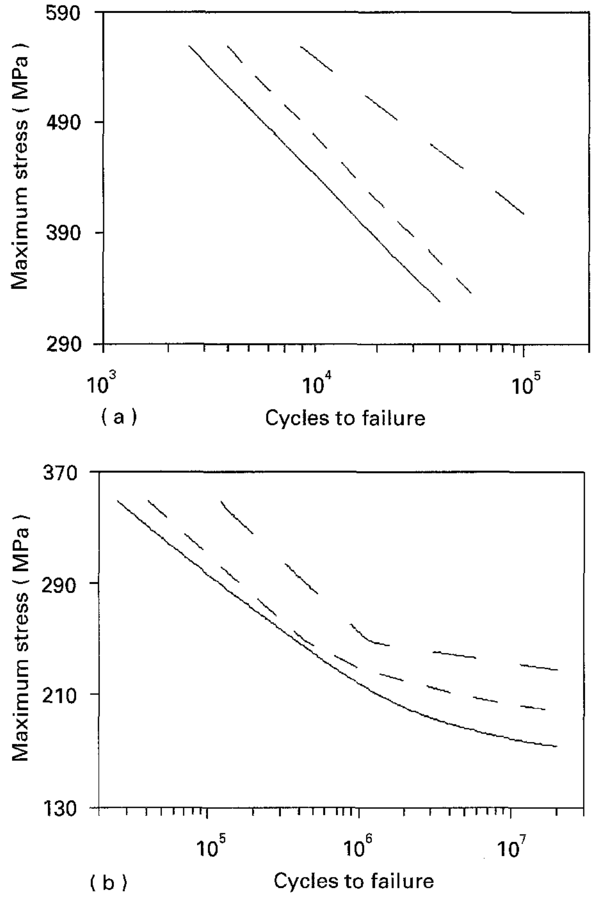


*Figure 11* Fatigue curves of CT-1 (– –), CT-2 (—) and SPT (- - -) states of 1960 alloy: (a) $R = 0.1$, $\nu = 0.5$ Hz; (b) $R = -1$, $\nu = 50$ Hz.

evidently the same as in 1560 and 1141 alloys, i.e. decrease of crack growth resistance in the material with fine grained structure. However, in SPT state the alloy is more isotropic concerning impact toughness. Beside the influence of grain structure this may be caused by change in the character of the banded structure: the spreading of bands which is observed due to the specific mechanism of SP flow [29].

The resistance to low and high cycle fatigue of the fine grained alloy 1960 is higher than that of the coarse grained alloy with the partially lost press-effect (CT-2); however, it is lower than in a coarse grained state when the press-effect is preserved (CT-1) (Fig. 11). The first is evidently connected with increasing crack formation resistance caused by grain refinement. It is similar to those effects observed in the 1560 and 1141 alloys. The lower fatigue limit after SPT in comparison with CT-1 can be explained by the fact that the influence of a subgrain structure on crack formation prevails over that of a grain structure: the size of subgrains with large angle subboundary misorientations is sufficiently smaller than the size of recrystallized grains after SPT.

The given statement is confirmed fully by the results of $N_1$ and FCGR estimation. FG structure formation in the 1960 alloy leads to an increase of fatigue crack initiation resistance (Fig. 12) and crack growth rate (Fig. 13). The difference in subgrain structure in CT-1 and CT-2 states insignificantly influences FCGR. On the basis of this, one may conclude that the greatest resistance to fatigue in the alloy after CT-1 is caused by the greatest work needed for crack formation. The latter can be explained in the following way only: the effect of high misorientated subgrains on crack formation is analogous to that of recrystallized fine grains and due to smaller subgrain size it is higher.

Judging by the results of low cycle fatigue tests of longitudinal specimens with a notch (Table V), after SPT the 1960 alloy is more sensitive to the sharp concentrator at values of maximum tension stresses in the cycle close to yield stress. At lower maximum stress in the cycle, $\sigma_{max}$, the fine grained alloy becomes less sensitive to the sharpness of the notch. Such behaviour is evidently caused by increasing the time necessary for crack formation in the total life at low stresses. This time is higher in the FG state.

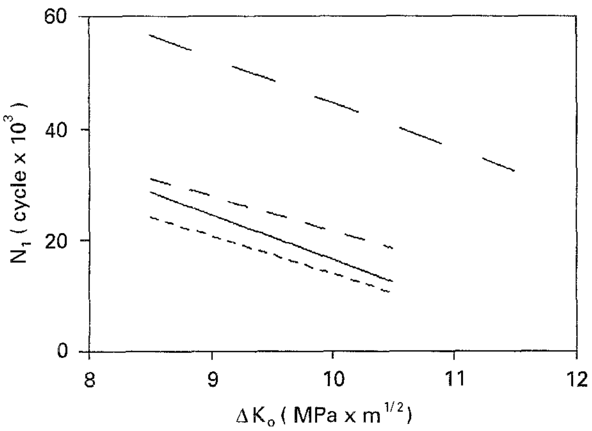


*Figure 12* Cycles required for the formation of a fatigue crack 1 mm long, $N_1$, versus the range of initial stress intensity factor, $\Delta K_o$, for CT-2 (— LT and ---- TL specimens) and SPT (— — LT and – – – TL specimens) states of 1960 alloy.

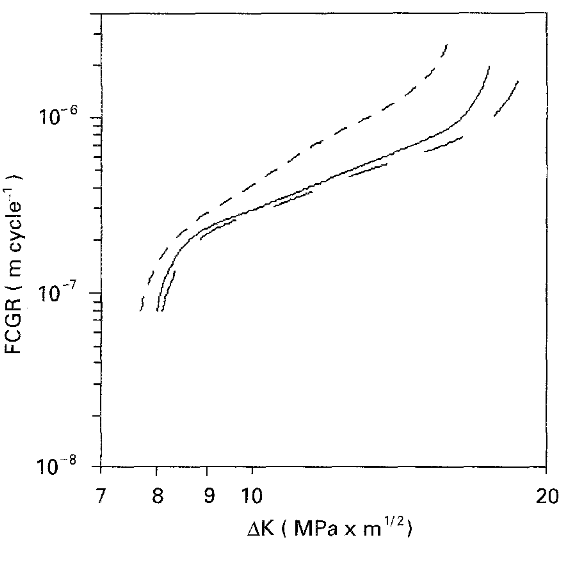


*Figure 13* Transverse fatigue crack growth rate (FCGR) versus the range of stress intensity factor, $\Delta K$, for CT-1 (– – –), CT-2 (——) and SPT (– – – –) states of 1960 alloy.

TABLE V Influence of treatment and radius in the top of the notch, $r$, on the durability of longitudinal specimens of the 1960 alloy

| $\sigma_{max}$ (MPa) | Cycles to failure | | | |
|---|---|---|---|---|
| | CT-1 | | SPT | |
| | $r = 0.05$ mm | $r = 0.30$ mm | $r = 0.05$ mm | $r = 0.30$ mm |
| 555 | 77 | 330 | 74 | 312 |
| 437 | 222 | 547 | 154 | 689 |
| 281 | 514 | 1468 | 583 | 1641 |

## 4. Conclusions

1. Grain refinement, transition from the coarse grained structure to the fine grained and substitution of conventional treatment by superplastic formation influence the mechanical properties of aluminium alloys.
2. The advantages of an FG structure in comparison with coarse recrystallized structures are: the reduction in mass due to the increase of yield and tensile stresses, the increase of ductility, the decrease of toughness and crack resistance anisotropy, the increase of fatigue limit, the increase of crack formation resistance under various loading conditions.
3. The negative features of FG structure are: probable increased mass due to lower strength properties compared with the unrecrystallized structure, the decrease of crack growth resistance under different loading conditions.
4. Less crack growth resistance in fine grained alloys is the reason for probable decrease of impact toughness, critical stress intensity factor and alloy endurance under low cycle fatigue, and the increase of sensitivity to sharp stress concentrators.

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