QUASISTATIC DEFORMATION BEHAVIOUR OF DISPERSION HARDENED ALUMINIUM MATERIALS

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ABSTRACT   Dispersion hardened Aluminium materials with very fine oxides and carbides were plastically deformed under tensile as well as compressive loadings between 293 K and 873 K. The occuring strength-differential-effect (SD-effect) of the as received state at $T \leq 373$ K is negative due to extrusion induced back stresses. These back stresses are responsible for an anisotropic behaviour of the investigated Bauschinger-effect. However, at $T > 373$ K the SD-effect is positive due to the greater generation of volume expanding microstructurally defects under tensile loading in comparison with compressive loading, which reduce the effective specimen cross section and combined with the occuring stress concentrations lead to decreases of the tensile flow stress. The great temperature dependency of the 0.2% proof stress of the investigated materials is caused by a thermal activated recovery of the deformation induced back stresses.

Keywords: *Dispersions, SD-effect, back stresses, Bauschinger-effect, recovery*

1. INTRODUCTION

Aluminium materials with very fine dispersions of oxides (Al_2O_3) and carbides (Al_4C_3), which were produced by reaction-milling of powders of aluminium and graphite and a following consolidation by extrusion are characterized by a much better high temperature strength and creep resistance than conventional aluminium alloys [1,2]. Very fine grain sizes are found, which are stabilized by the dispersoids. Although many investigations about the tensile and creep deformation behaviour of dispersion strengthened aluminium alloys exist, the reasons for their good mechanical properties are still discussed controversially. The present investigation deals with some new aspects of the deformation behaviour of aluminium with 1.5 vol.-% Al_2O_3 (Al-1.5 Al_2O_3) and aluminium with 12 vol.-% Al_4C_3 and 1.5 vol.-% Al_2O_3 (Al-12 Al_4C_3 /1.5 Al_2O_3) under tensile as well as compressive loading in the temperature range 293 K $\leq$ T $\leq$ 873 K.

2. MATERIALS AND EXPERIMENTAL DETAILS

The materials were produced by PEAK, Germany, by reaction-milling of powders of aluminium and graphite and by extrusion at 500°C. Their microstructures are dominated by very fine grain sizes of about 4 μm (Al-1.5 Al_2O_3) and 0,5 μm (Al-12 Al_4C_3/1.5 Al_2O_3), respectively, which are stabilized by the dispersoids located at grain boundaries, so that no significant grain growth occurs in heat treatments up to 600°C. The diameter of rounded Al_2O_3-dispersoids is about 100 nm. The Al_4C_3-dispersoids have cylindric and platelike shapes with lengths between 30 and 200 nm. Inside the grains elevated dislocation densities are observed.

The materials were investigated in the as received conditions as well as after heat treatments at T $\geq$ 500 °C. Tensile and compressive tests were carried out in the temperature range 293 K $\leq$ T $\leq$ 873 K at the strain rate $\dot{\varepsilon} = 3 \cdot 10^{-4}$ s^{-1} with computer-controlled screw driven testing machines using radiation and induction heating systems [3]. Cylindrical specimens were used with a gauge length of 25 mm and a diameter of 5 mm for tensile tests and with a gauge length of 10 mm and a diameter of 5 mm for compression tests. Specimen strains were measured with a capacitive extensometer.
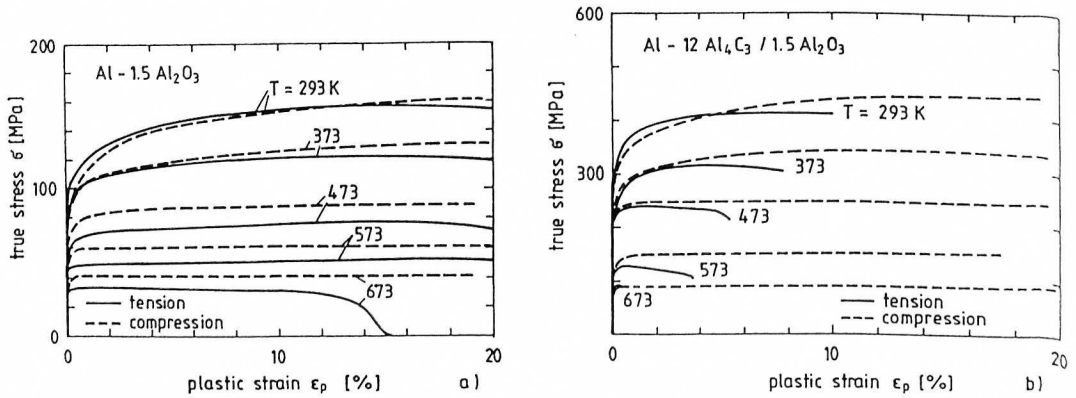
Fig. 1: True stress-plastic strain curves under tensile and compression loading, resp.
        a) Al-1.5 Al$_2$O$_3$, b) Al-12 Al$_4$C$_3$/1.5 Al$_2$O$_3$

3. RESULTS

An insight on the deformation behaviour of the materials Al-1.5 Al$_2$O$_3$ and Al-12 Al$_4$C$_3$/1.5 Al$_2$O$_3$ in the temperature range from 293 K to 673 K is given in Fig. 1, where the true stress is plotted versus the plastic strain. The flow stresses of both materials decrease steadily with increasing deformation temperature. The stress-strain curves determined under tensile or compressive loading at 293 K and 373 K show at smaller strains a distinct work hardening for Al-1.5 Al$_2$O$_3$ (Fig. 1a) as well as for Al-12 Al$_4$C$_3$/1.5 Al$_2$O$_3$ (Fig. 1b). For T > 373 K the flow stresses only change slightly with the plastic deformation.

On Fig. 2 the material properties 0.2% proof stress R$_{p0.2}$, ultimate tensile stress R$_m$ and percentage elongation at failure A$_5$ are plotted versus the temperature. Both materials show a steady decrease of R$_{p0.2}$ and R$_m$ with increasing temperature. Also the percentage elongation at failure decreases with increasing temperature up to minimum values at 673 K or 773 K. When the temperature is further increased the A$_5$ values again increase slightly. In the investigated temperature range the material Al-12 Al$_4$C$_3$/1.5 Al$_2$O$_3$ has clearly higher material resistances and considerably smaller percentage elongation at failure than Al-1.5 Al$_2$O$_3$.
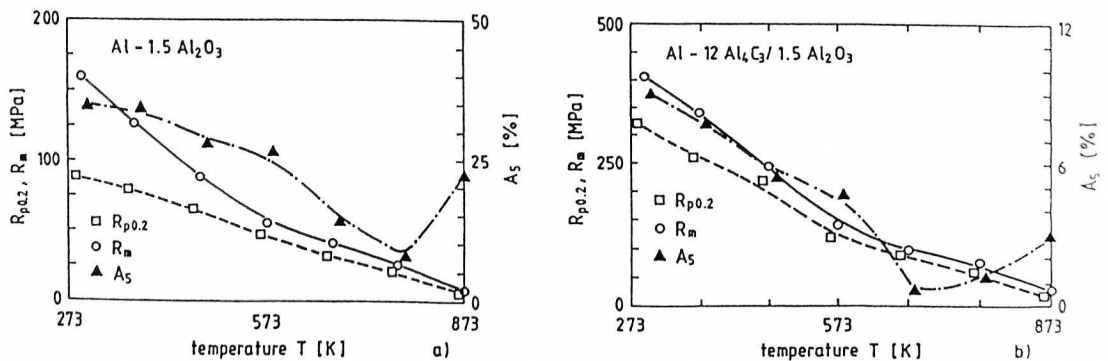

Fig. 2: 0.2% proof stress R$_{p0.2}$, ultimate strength R$_m$ and percentage elongation at failure A$_5$ as a function of temperature. a) Al-1.5 Al$_2$O$_3$, b) Al-12 Al$_4$C$_3$/1.5 Al$_2$O$_3$

The plotting of the absolute values of the true stress versus the plastic deformation during tensile and compressive loading, as Fig. 1 shows at different deformation temperatures, enables the evaluation of the occurring strength-differential-effect (SD-effect). At room temperature and 373 K the flow stresses of the compressive deformation curves are only above plastic deformations of 14 % and 2 %, respectively, at slightly higher values than the ones of the tensile stress-strain curves. However, at $T \geq$ 473 K with smallest deformations a considerable positive SD-effect occurs. The evaluation of the SD-effect as percentage flow stress difference according to

$$SD = \left. \frac{\sigma_c - \sigma_t}{\sigma_c} \right|_{\varepsilon_{p = const.}} \cdot 100 \%$$

(σ_c = flow stress in compression, σ_t = flow stress in tension) results for Al-1,5 Al$_2$O$_3$ at $\varepsilon_p = 0.5\%$ and 5% as a function of the deformation temperature in the correlations shown in Fig. 3a. The SD-effect
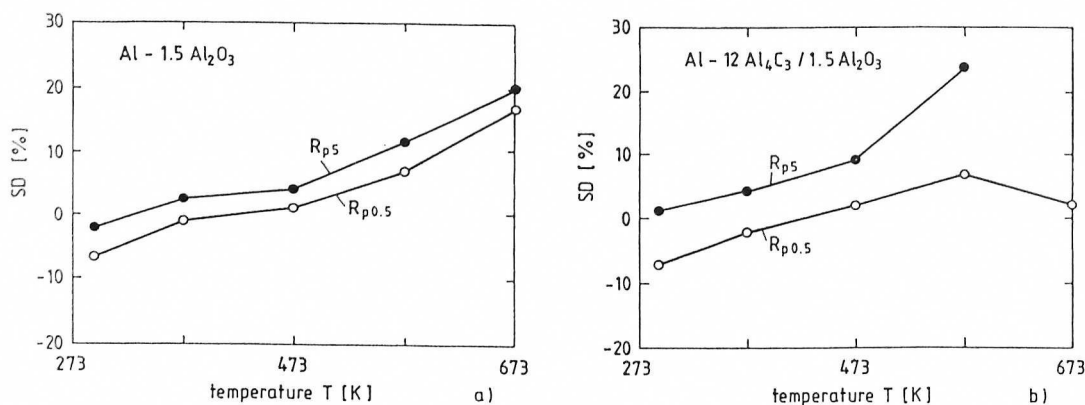

Fig. 3: Strength-differential-effect SD at $\varepsilon_p = 0.5\%$ and 5% as a function of temperature.
a) Al-1.5 Al$_2$O$_3$, b) Al-12 Al$_4$C$_3$/1.5 Al$_2$O$_3$
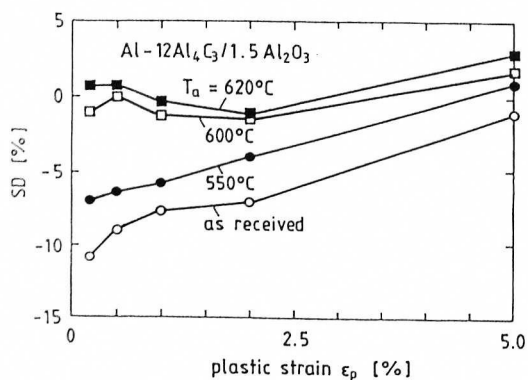

Fig. 4: SD-effect of Al-12 Al$_4$C$_3$/1.5 Al$_2$O$_3$ as a function of plastic deformation after various heat treatments

becomes absolutely greater with increasing temperature as well as with increasing plastic deformation. The determination of the SD-effect with Al-12 Al$_4$C$_3$/1.5 Al$_2$O$_3$ leads to the courses dependent on the temperature shown in Fig. 3b. This parameter increases with increasing temperature at $\varepsilon_p = 0.5$ % comparable with Al-1.5 Al$_2$O$_3$ and at $\varepsilon_p = 5\%$ considerably more distinctly than with Al-1.5 Al$_2$O$_3$ (see Fig. 3a).

As Al-12 Al$_4$C$_3$/1.5 Al$_2$O$_3$ was investigated at room temperature in different heat treatment conditions, an evaluation of the SD-effect could also be made for the corresponding specimens. Fig. 4 shows these SD-values as a function of the plastic deformation.

At first the SD-effect increases with the annealing temperature T_a and plastic strain. If $T_a \geq 600$ °C then there exists practically no negative SD-value.

4. DISCUSSION

The work-hardening observed at the stress-strain curves at room temperature and 373 K results from the increase of the dislocation density due to deformation. The constant flow stresses observed at $T \geq$ 473 K despite increasing plastic deformation are a consequence of a dynamic equilibrium between generation and annihilation of dislocations. The slight work softening occurring with greater deformations is due to the damage of the particle matrix interfaces as well as the reduction of the specimen cross section caused by the formation of voids. These mechanisms are also responsible for the decrease of the percentage elongation at failure A_5 with increasing temperature.

The 0.2% proof stress of both materials is relatively large and is characterized by a great temperature dependency (see Fig. 2). The proof stress is not dominated by particle hardening. TEM-investigations show that plastic deformation of Al-1.5 Al_2O_3 leads to some cell and subgrain formation whereas the deformation of Al-12 Al_4C_3 /1.5 Al_2O_3 is connected with nearly dislocation free grain interiors and some kind of pile-ups at the grain boundaries after deformation at 293 K [3,4]. Therefore, it is concluded that the very small grain size of the material Al-12 Al_4C_3 /1,5 Al_2O_3 prevent the development of three-dimensional dislocation structures and produce quasi-planar dislocation arrangements. These dislocation structures lead to a growing contribution of the grain boundaries to the proof stress with decreasing grain diameter of both materials and a reduction of the proof stress with growing temperature because of increasing possibilities of the thermal activated recovery of back stresses.

The negative SD-values at $T \leq 373$ K could be with both materials due to effects induced by the experiments (e. g. at the compression experiment: superposition of compression and bending stresses which cannot be completely avoided, as well as friction induced multiaxialities), machining induced surface residual stresses, and/or extrusion induced micro residual stresses of the as received state.

The positive SD-effects, which are extremely high at 673 K support the thesis of [5], that the formation of voids together with the separation of interfaces between matrix and dispersoids cause the SD-effect. With growing plastic deformation the increase of the concentration of excess vacancies is considerable, and the mobility of the vacancies becomes greater with increasing temperature, which results in better conditions for the development of voids under tensile loading. Existing voids reduce the effective specimen cross section and combined with the occurring stress concentrations lead to decreases of the flow stress. The voids are also responsible for the decrease of the percentage elongation at failure with increased temperature [6], as previouly mentioned, as the voids can coagulate quicker, which lead finally to microcrack initiation. In contrast to this, with compressive loading the hydrostatic stress component ($p = -1/3 \ \sigma_1$) works against the growth of voids and the bearing cross section of the specimens remains nearly unchanged. As a result there are greater absolute values of the flow stress than with tensile loading and thus, positive SD-effects occur.

X-ray residual stress measurements with Cu-Kα-radiation at the {511/333}-interference lines did not show any measurable machining induced residual stresses at the specimen surfaces in the as received state. As the different thermal expansion coefficients of the aluminium matrix ($\alpha_{th}^{Al} = 22 \cdot 10^{-6}$ m/K) and the dispersoids (Al_2O_3: $\alpha_{th}^{Al_2O_3} = 7 \cdot 10^{-6}$ m/K, Al_4C_3: $\alpha_{th}^{Al_4C_3} < \alpha_{th}^{Al}$) during cooling from extrusion to room temperature should result in a higher contraction of the aluminium matrix compared to the dispersoids, and thus, result in tensile residual stresses in the softer deformable matrix phase, they should reduce the material resistance with tensile deformation at room temperature at least at the beginning of the plastic deformation. This possible process, however, does not provide an explanation

for the observed material behaviour.

The unusual SD-behaviour can be better understood, if the history of the investigated specimens is considered. The material production was finished with a consolidation by extrusion. Although the extrusion was performed at relatively high temperatures (approx. 500°C) and during the deformation there dominated work hardening as well as dynamic recovery processes and dislocation structures with back stresses remained, which did not completely relaxate in spite of the high temperatures [4]. One has to expect that these back stresses originating from the history, which have the character of micro residual stresses, influence the quasistatic deformation behaviour. Investigations concerning the Bauschinger-effect should allow a further statement on the back stresses regarding the effect and sign.

Therefore, in the as received state of Al-12 Al_4C_3/1.5 Al_2O_3 the characteristic Bauschinger parameter $\Delta\epsilon_p$ was determined. $\Delta\epsilon_p$ is the difference between the Hooke's straight line lengthened during unloading to $\sigma = 0$ and the curve of back deformation observed during the experiment (compare for example [7]). $\Delta\epsilon_p$ also was investigated at specimens annealed for 25 hours at 640 °C, at which the back stresses should be reduced to a large extent by heat treatment. Fig. 5 shows the behaviour of $\Delta\epsilon_p$ dependent on the plastic predeformation at tensile or compressive loading. The $\Delta\epsilon_p$-values increase with the plastic predeformation. In the as received state the tensile predeformation leads to considerably greater back plasticities than a compressive predeformation. However, in the heat-treated condition the $\Delta\epsilon_p$-values are independent of the direction of loading. As $\Delta\epsilon_p$ is always caused by deformation induced back stresses, the differe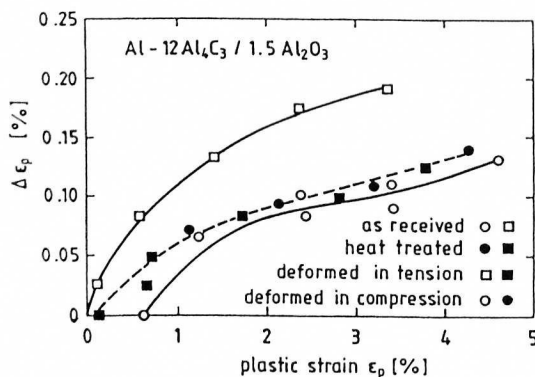nce of the $\Delta\epsilon_p$-values between the tensile and compressive predeformed specimens in the as received state implies that already before the plastic predeformation extrusion induced back stresses existed, which overlap the back stresses generated during the deformation and this leads to the different results. As $\Delta\epsilon_p$ reaches larger amounts after the tensile deformation this implies that the as received state already contains back stresses generated during extrusion, which work in the direction of the compression and support the back deformation. As a consequence, the as received state reacts qualitatively like a tensile predeformed condition. There are no longer any extrusion induced back stresses, as after a 25 hour heat treatment at 640°C $\Delta\epsilon_p$ does no longer depend on the direction of predeformation. The back stresses existing in the as received state relaxate obviously to a large extent during the heat treatment.

Fig. 5: Bauschinger parameter $\Delta\epsilon_p$ of Al-12 Al_4C_3/ 1.5 Al_2O_3 as a function of plastic predeformation of the as received state and a 25 h at 640°C heat treated state

Thus, the extrusion induced back stresses are responsible for the above treated negative SD-effect. It becomes smaller during the heat treatment due to the thermally activated relaxation. At $T_a = 550°C$ the 25 hour heat treatment obviously is not sufficient for a complete relaxation, as there are still negative SD-values. After heat treatments of 25 hours at $T \geq 600°C$ one can assume that the back stresses are completely relaxed.

The negative SD-effect, as a result of the extrusion induced back stresses, only occurs at deformation

temperatures $T \geq 373$ K. As the SD-effect becomes positive at higher temperatures one can assume that these back stresses relaxate quickly under mechanical loading and therefore do not affect the SD-effect. However, occurring void formations and void growth processes which lead to a positive SD-effect could also overcompensate the effect of the back stresses.

5. ACKNOWLEDGEMENT

6. REFERENCES

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