Mechanical behaviour of heavily deformed CuAgZr conductor materials

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Abstract. The mechanical properties of Cu−7 m. % Ag−0.05 m. % Zr-alloys have been improved by optimised thermo-mechanical treatments. After processing the conductor material shows an ultimate tensile strength of more than 1.1 GPa, a yield strength of about 1 GPa, a plastic strain of 0.7%, and a conductivity of 60 % IACS (at room temperature). The yield strength of a cold deformed Cu−7 m. % Ag−0.05 m. % Zr-alloy has been assessed on the basis of different hardening mechanisms: solid solution, grain boundary, precipitation and dislocation hardening. The critical shear stress which is determined from the experimentally observed critical shear strength by Schmid’s law is between a linear and a quadratic superposition of its individual contributions and hence is in good agreement with the theoretical prediction.

1. Introduction
CuAg-alloys have attracted interest because of their advantageous combination of excellent mechanical strength and relatively high electrical conductivity [1, 2]. The properties of CuAg-alloys originate from their microstructure, which arises from the chemical composition as well as from thermo-mechanical treatments [3, 4]. The microstructure of the alloys is mostly determined by the silver content, which determines the predominantly observed precipitation reaction. The microstructure of a homogenised Cu-7 m. % Ag alloy (mass percent) consists of a Cu-rich solid solution and discontinuous Ag-rich precipitates next to large angle grain boundaries. Adding 0.05 m. % Zr to a Cu-7m-% Ag alloy suppresses the discontinuous precipitation mode in these alloys and continuous precipitates which are favourable in the sense of maximum strength are observed [5].

2. Experimental Details
Polycrystalline samples of Cu−7 m. % Ag−0.05 m. % Zr (CuAgZr) were molten in an inductance furnace and cast into a graphite mould (D = 32 mm). The CuAgZr alloy has then been homogenised at 850° C for 5 h and subsequently water-quenched. Precipitates were formed during a further heat treatment at 400°C / 4 h. For details please refer to [5].

Vickers hardness HV0.5 (load time 10 s) values were obtained from a minimum of 15 indentations. The ultimate tensile strength σUTS and yield strength σYP were measured at room temperature on drawn wire pieces of a total length of 150 mm using an electro-mechanical Instron 8562 testing machine at a displacement rate of 0.2 mm/min.
3. Results and Discussion

The alloys show a reasonable strength after cold deformation up to a logarithmic drawing strain of \( \eta = \ln(A_0/A) = 5 \) (\( A_0 \) and \( A \) are the initial and final cross sectional areas). However, there is a demand to enhance the elongation from about 1.4% to 5%. This originates from the design of pulsed high field magnets [6, 7]. Finite Element Analysis simulations show that for achieving the highest possible stress in both, conductor and reinforcement, the conductor needs to be able to tolerate this mentioned strain [7]. A simple way of enhancing the strain in a metallic material is to heat treat the samples [8]. This, however, will at the same time cause a loss in strength. The effect of the temperature of annealing for 2 h on the strength and strain is shown in Fig. 1.

![Figure 1](image1.png)

**Figure 1.** Yield strength of CuAgZr alloys in dependence on the uniform elongation. The values are taken after a heat treatment for 2 h at the depicted temperatures (next to symbols).

Figure 1 shows that a temperature of 300°C causes a drop in strength to 800 MPa while the uniform elongation increases by 0.5%. Thus, this treatment must not be applied on the final drawn sample. We have shown earlier [5] that intermediate heat treatments are beneficial for a further enhancement of the strength. As shown in Fig. 2, a heat treatment (300°C / 2 h) of a cold drawn (\( \eta = 1.5 \)) sample causes a drop in hardness. Further cold deformation causes a larger increase of the hardness when compared to the sample without intermediate heat treatment [9]. Thus, at a later stage of deformation, the loss in hardness, that is caused by the thermal treatment, is balanced. For even higher deformation an enhanced hardness is observed when compared to the condition without thermal treatment. The final deformed wire is a compromise of either further enhancing the strength up to 1.4 GPa while keeping the strain to failure at a comparable level or to enhance the strain to failure by 0.5% while keeping \( \sigma_{UTS} \) at 1.2 GPa. Figure 3 summarises the optimisation of the strength in CuAg-alloys. For a Cu-7m.%Ag alloy \( \sigma_{UTS} \) increases with cold deformation as shown. Adding 0.05m.% Zr to this alloy enhances the overall strength level due to the suppression of the discontinuous precipitation reaction [5]. An intermediate heat treatment which has been applied at \( \eta = 4 \) allows further deformation yielding to \( \sigma_{UTS} = 1.4 \) GPa.

The final microstructure of CuAgZr following casting, solutionising and precipitation heat treatment bears four different hardening mechanisms: solid solution hardening, precipitation hardening, phase boundary hardening and work hardening. These individual contributions to the total strength are assessed in the following on the basis of general formulas [10]. The increase of the critical yield stress for solid solution strengthening \( \Delta \tau_{c,ss} \) is calculated by:
Table 1. Summary of the individual strength contributions (solid solution $\Delta \tau_{c,ss}$, grain boundaries $\Delta \tau_{c,HP}$, precipitates $\Delta \tau_{c,precip.}$, dislocations $\Delta \tau_{c,disloc.}$) to the total yield strength and their sum according to $\tau_{c,\text{total}} = \Delta \tau_c + \tau_{c,0}$ and Eq. 5.

| Component          | Value (MPa) |
|--------------------|-------------|
| $\tau_{c,0}$       | 40          |
| $\Delta \tau_{c,ss}$ | 116         |
| $\Delta \tau_{c,HP}$  | 31          |
| $\Delta \tau_{c,precip.}$ | 188        |
| $\Delta \tau_{c,disloc.}$ | 71         |
| $\tau_{c,\text{total}} (k=1)$ | 446        |
| $\tau_{c,\text{total}} (k=2)$ | 274        |

In Eq. 1, $\delta = \frac{\partial \ln a}{\partial x} = 0.1302$ is a factor of lattice change and $\eta = \frac{\partial \ln G}{\partial x} = 0.3485$ is a factor describing the change of the shear modulus due to alloying. $x_a = 0.59$ at.% is the Ag fraction within the solid solution. Lattice parameters of $a_{Cu} = 0.3620$ nm and $a_0 = 0.3618$ nm (solid solution) and shear moduli of $G_{Cu} = 46$ GPa and $G_0 = 45.91$ GPa have been used for modelling. The increase of the yield strength $\Delta \tau_{c,HP}$ due to the grain refinement is determined by:

$$\Delta \tau_{c,HP} = k_{\text{HP}}/\sqrt{d} = 31 \text{ MPa}$$

In Eq. 2, $d = 24.6 \, \mu\text{m}$ is the mean grain size and $k_{\text{HP}} = 0.15$ MPa$\sqrt{\text{m}}$ [10, 11], the Hall-Petch constant. The precipitates are aligned along the drawing axis. Shearing of precipitates becomes possible as the slip plane of the dislocations is oriented by an angle of $53^\circ$ to the long axis of the precipitates. In this case the increase of the strength $\Delta \tau_{c,\text{precip.}}$ is given by:

$$\Delta \tau_{c,\text{precip.}} = \frac{\gamma^{3/2}}{b^2} \sqrt{ rf \tau_{c,\text{total}} (k=1)} = 188 \text{ MPa}$$

In Eq. 3, $\gamma = 2.2449 \, \text{J/m}^2$ [12] is the phase boundary energy, $b = 0.2556$ nm the Burgers vector of the solid solution, $r = 15$ nm the radius and $f = 0.07$ the volume fraction of the precipitates. The total number of dislocations can be drastically increased by cold working. The CuAgZr alloys have been cold worked up to a logarithmic deformation strain of $\eta = \ln (A_0/A) = 4.07$. The increase of strength due to work hardening as given by:

$$\Delta \tau_{c,\text{disloc.}} = \alpha Gb\sqrt{\rho} = 71 \text{ MPa}$$

In Eq. 4, $\alpha = 0.6$ is a factor according to the geometry and $\rho = 1 \cdot 10^{14}$ m$^{-2}$ is the dislocation density, which has been obtained from TEM analysis and the linear intercept length method [13]. The total critical stress of the CuAgZr alloy is calculated by a superposition of the previously discussed individual contributions. However, there is no generally accepted formula available to assess this superposition. In consequence, an empirical equation has often been used [14, 15]:

$$\Delta \tau_{c,\text{total}} (k=2) = 274 \text{ MPa}$$
The exponent $k$ in Eq. 5 is typically between 1 and 2, depending if the individual strengthening mechanisms interact. The critical shear stresses are summarised in Tab. 1 for both $k$ values.

Now we make a Gedankenexperiment: Assuming a linear superposition ($k = 1$) of the individual strengthening mechanisms leads to a critical shear stress of $\tau_{c,\text{total}} = 446 \text{ MPa}$. For polycrystalline materials the Taylor factor $M$ is used to capture the orientation dependence of the strength from numerous sources of variability. It can be between 2 and 4. The critical shear stress and the experimentally measured shear strength $\sigma_{\text{yp}} = 1064 \text{ MPa}$ match for a Taylor factor of $M = 2.38$. Here we assume the strengthening mechanisms with different strength and interaction distance [15] and thus no interaction. In a second step, we assume a quadratic superposition ($k = 2$) and calculate a critical shear stress of $\tau_{c,\text{total}} = 274 \text{ MPa}$, which matches to the critical shear strength if a Taylor factor of $M = 3.88$ is assumed. In this case the individual strengthening mechanisms should have a similar strength or have similar obstacle distances, which is not the case. In addition this Taylor factor deviates somewhat from the usual value of 3.06 for polycrystalline materials. However, this can be rationalised by the strong $<111>$ fibre texture in the material. This will increase the Taylor factor to about $M = 3.7$ (assuming $M$ being the inverse Schmid factor calculated for a $<111>$ oriented fcc single crystal).

The microstructure of the alloys provide reasonable arguments for $k = 1$ as the obstacle distances range from about 2 nm (average spacing between the solute atoms in the case of solid solution hardening) to about 100 nm (distance between dislocations). In contrast to this, a $k$-value of 2 allows the explanation including texture and an interaction between the strengthening mechanisms. Equation 5 matches for a $k$-value of 1.76, providing a further argument for the later case. Although there is no physical explanation for a nearly quadratic superposition of the individual contributions to the total shear stress we find an accurate prediction of the actual measured yield strength of the CuAgZr material.

4. Summary
The mechanical properties of CuAgZr have been improved by optimised thermo-mechanical treatments. The optimised material bears a compromise of enhanced strength (plus 200 MPa) or strain (plus 0.5%) while keeping the other properties at a similar level. The yield stress of a cold drawn CuAgZr fits best to a quadratic superposition of the individual shear stresses that arise from the microstructure of the alloy.

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