INFLUENCE OF FREQUENCY ON FATIGUE LIMIT AND FATIGUE CRACK GROWTH BEHAVIOR OF POLYCRYSTALLINE Cu

B. Weiss*, H. Müllner*, R. Stickler*, P. Lukas and L. Kunz****

**University of Vienna, Vienna, Austria*

***Institute of Physical Metallurgy, Czechoslovak Academy of Sciences, Brno, CSSR*

ABSTRACT

The effects of test frequency on the cyclic stress strain response, the formation of PSBs, the fatigue limit, and on the nucleation and propagation of fatigue cracks in polycrystalline Cu were studied. Tests were carried out at 100 Hz and 20 kHz, at R = -1, at room temperature. The cyclic plastic deformation was found to be the controlling parameter. Practically independent of test frequency a fatigue limit occurred at a finite plastic strain amplitude on the order of 10^{-5} while a PSB-threshold was observed at $\varepsilon_{a,\,pl} = 10^{-6}$; the conditions for the propagation of nucleated micro-cracks appear to govern the fatigue limit. The minor effects of test frequency may be explained by the thermally activated stress component in addition to a frequency dependence of the rate of damage accumulation.

KEYWORDS

Fatigue limit, cyclic stress strain, persistent slip band, fatigue microcrack, plastic strain fatigue limit, frequency effect, statistical evaluation, high cycle fatigue.

INTRODUCTION

Recent advances in experimental techniques permit now to study the cyclic stress-strain (CSS) behavior of single and polycrystalline metals down to very low plastic strain amplitudes, e.g. of less than 10^{-5} . This possibility evoked the - in fact - old question of the nature of the fatigue limit (FL) in fcc metals and alloys. As the typical nucleation sites for fatigue cracks in pure fcc polycrystalline metals are related to the formation of PSBs, it appears logical to relate the FL of such metals to the threshold conditions of PSB formation (1,2). On the other hand, sufficient experimental evidence shows the existence of microcracks in specimens cyclically loaded well below the FL (e.g. Ref.3). These observations offer the possibility to understand the FL as the threshold condition for the propagation of nucleated microcracks.

The FL in fcc polycrystalline metals is often defined in a more or less conventional manner, e.g. as the stress or strain amplitude corresponding to the number of cycles to fracture exceeding 10^7. As recently shown on the basis of a statistical data evaluation a FL occurs in polycrystalline Cu for strain amplitudes leading to an excess of 10^8 loading cycles (3). Economic considerations prevent conventional fatigue tests under the required closely controlled conditions to be routinely extended to such high number of loading cycles. For this reason several investigators have attempted to increase the test frequency (3,4) up to more than 20 kHz. Improved test methods made it possible to conduct such high-frequency tests under total-strain controlled conditions whereby the associated plastic strain could be measured in the range of approx. 1×10^{-6} to 5×10^{-4} (6).

It may be speculated that the increase in test frequency might affect the test results (5). Most of the conventional CSS-data have been obtained at strain rates between 10^{-4} and 10^{-2}/sec. Testing at 20 kHz is associated with average strain rates between 1 and 10/sec for which - with the exception of one very recent study (6) - no stress-strain data have been published. However, in reviewing published information, Laird (7) came to the conclusion that only a weak strain rate effect (SRE) if any at all should be noticeable in the conventional range of strain rates. Laird also postulated that only a small SRE may occur up to the high test frequencies (7). Deducing this information over such a large range of strain rates from literature data may appear somewhat questionable because of the possible overriding effects of variations in composition and pretreatments of the specimen material, and in experimental conditions.

The objectives of the present investigation included a detailed study of the effects of the test frequency on CSS-response, on the formation of PSBs, on the FL, and on the nucleation and propagation of fatigue cracks in commercial pure Cu. Cyclic tests were carried out at 100 Hz ("low frequency", LF) and 20 kHz ("high frequency", HF) with special emphasis on sufficient data to reveal minor differences in test results by means of a statistical data evaluation.

SPECIMEN MATERIAL AND EXPERIMENTAL CONDITIONS

A single lot of commercial pure polycrystalline Cu (99,98 % Cu) was selected to provide identical specimen material for all the required experiments. After machining and mechanical polishing of the gauge section all specimens were heat treated 600 C/1h in vacuum to obtain a uniform microstructure with a grain size of approx. 70 µm. The mechanical properties of the heat treated material at RT were as follows:
$R_{p0,2}$ = 37 MPa, R_m = 220 MPa, dyn. Youngs modulus 142 000 MPa for rod shaped specimens and 126 500 MPa for plate shaped specimens (determined at 20 kHz).

The geometry of the specimens for both the LF tests (100 Hz) and the HF tests (20 kHz) were essentially identical. Two types of specimens were used. The fatigue life curves and the CSS- curves were determined on cylindrical rod-shaped specimens with a diameter of 4mm. For the measurement of the crack propagation rate and the threshold stress intensity range, centrally notched plate specimens were used (thickness 5mm, width 20mm).

All the LF-tests were carried out in a modified Amsler pulsator. The tests on the cylindrical specimens were run at constant total strain amplitudes ($\varepsilon_{a,tot}$). The hysteresis loops, i.e. the data on the $\varepsilon_{a,pl}$ and σ_a were continuously recorded. The values of $\varepsilon_{a,pl}$ used in Fig. 2 correspond to the values measured at 1/2 of the number of cycles to failure. In tests for the crack propagation rate (da/dN) and threshold (ΔK_{th}) measurements the stress amplitude was controlled. The crack length was monitored by means of travelling microscopes.

The HF-tests were carried out with a commercial test unit in an essentially total-strain controlled manner (3). The total strain amplitude ($\varepsilon_{a,tot}$) was measured by means of miniature strain gauges. The plastic strain amplitude ($\varepsilon_{a,pl}$) and the cyclic saturation stress (for N approx. 10^6) were determined as described in Ref. 6.The statistical evaluation of the fatigue data from both the LF and HF tests was carried out following established procedures (3).Details of the experimental set-up for the HF crack growth measurements are given in Ref. 8. The crack length was determined by means of a light-microscope/closed-circuit-TV system. The computation of the ΔK-values was carried out by a relationship published recently (9).
The dislocation arrangement was observed by means of conventional TEM. Thin foils were prepared from sections both oblique and parallel to the specimen axis. To identify non-propagating microcracks the fatigue exposed specimens were strained by a small amount (ε = approx. 2%) in a tensile test machine, then heated in air for 1h at 400 C to oxidize the surface of the microcracks opened up by the prestrain, and finally fractured by fatigue loading at a high strain amplitude for fractographic evaluation.

EXPERIMENTAL RESULTS AND DISCUSSION

The fatigue life (N_f) of specimens cyclically loaded under $\varepsilon_{a,tot}$-controlled condition at both test frequencies is plotted in the diagram of Fig. 1. For most of the strain levels 10 specimens were tested. Computed fracture probability data are shown in Fig. 1 as p = 10% and p = 50% curves. The $\varepsilon_{a,tot}$-N curve for the HF-tests is clearly shifted towards

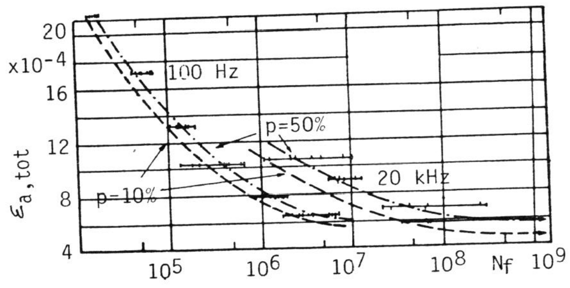

Fig.1: $\varepsilon_{a,tot}$-N_f data for test frequencies of 100 Hz and 20 kHz (curves computed for 10% and 50% fracture probability), Cu recrystallized, R = -1, RT.

higher $\varepsilon_{a,tot}$ values with respect to the curve for the LF tests. From the 10% fracture probability curves one may deduce for the HF test conditions a FL at $\varepsilon_{a,tot}$ = 4,6x10^{-4} for N exceeding 10^8. For the NF tests a FL may be indicated near $\varepsilon_{a,\ tot}$ = 5x10^{-4} for N exceeding 1x10^7.
The CSS-curves shown in Fig. 2 were obtained as follows. For the LF test, although carried out under $\varepsilon_{a,tot}$ control, σ_a and $\varepsilon_{a,\ pl}$ were recorded for each specimen so that a large number of data points was available for constructing the CSS-curve. For the HF tests only the $\varepsilon_{a,\ pl}$ ranges corresponding to each $\varepsilon_{a,tot}$ level could be plotted. As shown in Fig. 2 the CSS-curve for the HF tests is shifted towards higher stress amplitudes compared to the LF results. In addition, data published by Mughrabi (2) for Cu specimens (grain size 25 µm) obtained at test frequencies of less than 1 Hz are included in Fig. 2. The latter data indicate the frequency effect even in the conventional testing range. An explanation of this frequency effect on the cyclic saturation stress may be obtained from information in the published literature in which the role of point defects on cyclic hardening (10) at conventional test frequencies and the effects of the increased production rate of point defects with increasing test frequencies are reported (7, 11).

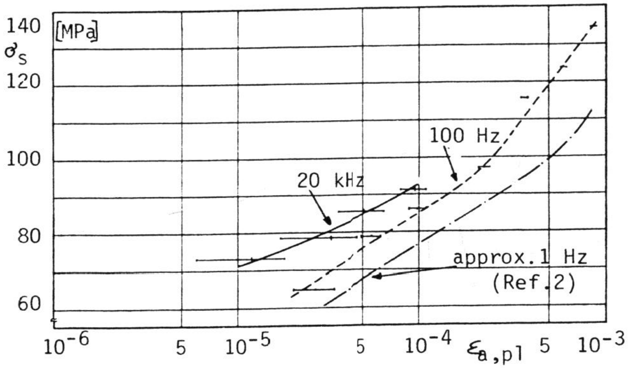

Fig.2: CSS-curves determined at various test frequencies (Cu-recrystallized, R = -1, RT)

Plotting the fatigue life data in a Coffin-Manson type diagram yields the curves shown in Fig. 3. The p = 50 % lines for each test frequency appear parallel to each other and may be expressed by the equations included in Fig. 3. The pre-exponential term indicates the frequency effect. From these results a FL under the LF test conditions may be deduced at $\varepsilon_{a,\ pl}$ = approx. 3x10^{-5}. The statistical analysis of the HF data reveals that the FL (assumed to correspond with the p = 10% value) occurs at $\varepsilon_{a,\ pl}$ = approx. 1x10^{-5} for N larger than 10^8, in agreement with the results of an earlier investigation (3). It should be pointed out, however, that the HF tests have been extended well beyond the computed Coffin-Manson line while the LF tests have been terminated just in the range of this line. If we surmise that in prolongating the LF tests some specimens would have failed, the LF-FL would be reduced by a small amount. Thus, the influence

of the test frequency on the FL expressed in terms of $\varepsilon_{a,\ pl}$ may be negligible. At this time one may only speculate about the reasons why in the Coffin-Manson plot the HF - FL is attained at higher numbers of loading cycles compared to the LF tests. One possibility could lie in the fact that the rate of damage accumulation is affected during HF testing so that a larger number of loading cycles is required to accumulate the microstructural damage as compared to the LF testing.

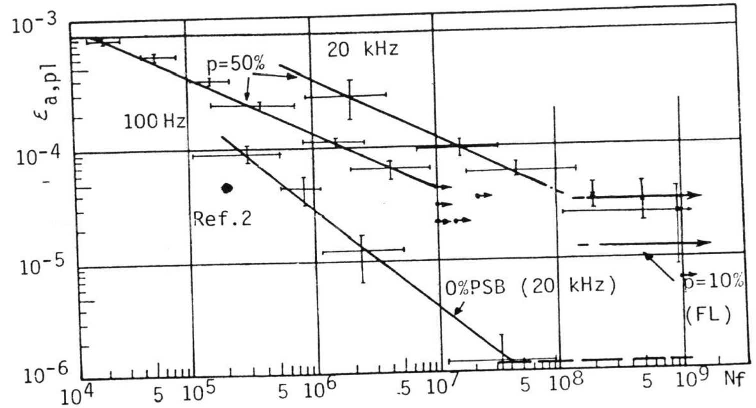

Fig.3: $\varepsilon_{a,pl}$-N_f relationship determined at 100 Hz and 20 kHz, curves computed for 50% fracture probability. Lower curve indicates appearance of PSBs, single data point shows PSB threshold reported in Ref. 2.
100 Hz ... $\varepsilon_{a,pl}$ (p = 50%) = 0,125 x $N_f^{-0,5}$
20 kHz ... $\varepsilon_{a,pl}$ (p = 50%) = 0,5 x $N_f^{-0,5}$

As reported earlier (3) PSBs are formed well below the strain amplitude of the FL. A limited number of LF-results is in agreement with these findings. Mughrabi (12) found for polycrystalline Cu a threshold strain for PSB formation at approx. $\varepsilon_{a,\ pl}$ = 2,5x10^{-5} for N = 4x10^5, leading to a cumulative plastic strain ($\varepsilon_{pl,\ cum}$ = 4N $\varepsilon_{a,\ pl}$) of $\varepsilon_{pl,\ cum}$ = 40. The PSB threshold referred to by Mughrabi is shown by the single data point in Fig. 3. It falls only slightly short of our experimental 0 %-PSB line which corresponds to approx. $\varepsilon_{pl,\ cum}$ = 60. The cumulative plastic strain corresponding to a 10 %-PSB coverage was to average $\varepsilon_{pl,\ cum}$ approximately 1600. Our experiments revealed a PSB-threshold to occur at approx. $\varepsilon_{a,\ pl}$ = 10^{-6}, at which tests were extended for $\varepsilon_{pl,\ cum}$ to exceed 3000. We speculate from these observations that for the LF tests reported in Ref. 12 lower PSB-thresholds would have been observed if the cyclic exposure would have been extended to larger N-values, equivalent to larger values of $\varepsilon_{pl,cum}$.

Dislocation structures and microscopic observations:

In the overlapping region of cyclic exposure (see Fig. 3), the dislocation arrangement in HF and LF test specimens (run well into saturation) were found to be the same, namely to consist of the vein structure. For the high plastic strain amplitudes in the case of the LF tests, the structure consisted of cells or transition veins-cells. The lowest plastic strain amplitudes in the HF tests were characterized again by the loose vein (or bundle) structure. What we consider to be of great importance is the infrequent observation of the ladder-like PSB structure (in sections parallel with the specimen axis) in the interior of both, the run-out LF specimens ($N \geq 10^7$) and HF specimens ($N \geq 10^9$). Examination of run-out specimens (with N larger than 10^9, at $\varepsilon_{a,\ tot}$ below the FL) revealed the presence of microcracks in isolated surface areas, with length of approx. 100 µm. Fractography indicated a semi-elliptical shape. These findings can be considered as an unambiguous proof for the presence of non-propagating microcracks. We may surmise, therefore, that the FL is related to that critical strain amplitude below which microcracks are formed but do not propagate (1).

In view of the critical role of the propagation behavior of existing cracks in determining the FL, an investigation of the frequency effect on the crack growth behavior appeared essential.The crack growth rate, da/dN, near threshold is shown for both test frequencies as function of the stress intensity range, ΔK, in Fig. 4. The growth curve for the HF tests falls somewhat below that for the LF tests. The respective threshold values, ΔK_{th} (assumed to correspond to a growth rate of less than 10^{-12} m/cycle), were found to be 2,15 ± 0,1 and 2,55 ± 0,1 $MPam^{1/2}$ for the LF and HF tests, respectively. The ratio of these two values is comparable to the ratio of the saturation stress amplitudes for HF and LF specimens for the same plastic strain amplitude (see the CSS-curves in Fig. 2). In other words, the frequency effect on ΔK_{th} may be explained on the basis of the thermally activated stress component.

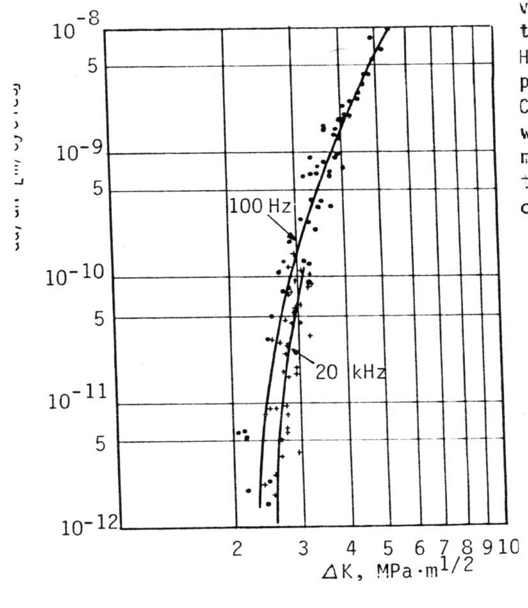

Fig.4: Fatigue crack growth rates and threshold values determined at 100 Hz and 20 kHz. (Cu-recrystallized, R = -1, RT)

CONCLUSIONS

(i) The controlling parameter of the fatigue process both in the HF and the LF exposure is the cyclic plastic deformation. The fatigue life curves can be described in terms of the plastic strain amplitude up to the number of cycles to fracture on the order of 10^9 cycles.

(ii) The fatigue life curves exhibit towards higher N a horizontal portion the onset of which depends on the test frequency, (for LF tests near $5x10^7$, for HF tests beyond $1x10^8$). This makes it possible to define a fatigue limit (corresponding in our evaluation to a 10 % fracture probability bound) to occur at a finite plastic strain limit on the order of $1x10^{-5}$, practically independent of test frequency.

(iii) The observation of PSBs both at the surface (by light and scanning electron microscopy) and in the interior (by transmission electron microscopy) of specimens cycled for very large N-values at $\varepsilon_{a,\ pl}$ lower than the FL indicates that the formation of PSBs is probably not the critical event in determining the FL. We believe that the conditions for the propagation of nucleated microcracks are decisive . The reasons for the higher critical $\varepsilon_{a,\ pl}$-values cited in the literature may be explained by the N-dependence of the formation of PSBs, which appears to require a certain critical $\varepsilon_{pl,\ cum}$. The PSB-threshold strain may be as low as $\varepsilon_{a,\ pl} = 10^{-6}$.

(iv) The difference in the HF and LF test results (CSS-curves, thresholds for crack propagation) can be attributed mainly to the thermally activated stress component with point defect reactions and point defect clusters as significant factors. The shift of the Coffin-Manson curve towards higher N-values cannot be explained in this way alone, probably frequency-dependent processes (which reduce the rate of damage accumulation) should be taken into consideration.

LITERATURE

1. M.Klesnil and P.Lukas, Fatigue of metallic materials, Elsevier Sc.Publ.Co., Oxford 1980
2. H.Mughrabi and R.Wang, Cyclic deformation of fcc polycrystals: a comparison with observations on single crystals, Proc. 2nd Risø Int.Symp.,Sept.1981
3. W.Hessler, H.Müllner and B.Weiss, High cycle fatigue and threshold behavior of polycrystalline Cu, Proc. ICF-5, Cannes (1981) 1349
4. J.Awatani and others, A study of the fatigue limit of Cu, Met.Trans. 6A (1975) 1029
5. V.A.Kuzmenko, Fatigue strength of structural materials at sonic and ultrasonic loading frequencies, Ultrasonics (Jan.1975) 21
6. W.Hessler, H.Müllner, R.Stickler and B.Weiss, Aufnahme von Spannungs-Dehnungs-Hysteresekurven während einer Wechselbeanspruchung bei 20 kHz, Z.Metallkunde 73 (1982) 570

7. C.Laird and P.Charsley, Strain rate sensitivity effects in cyclic deformation and fatigue fracture, in Ultrasonic Fatigue 1981, J.M.Wells et al, Eds., AIME, USA, 1982, p. 188
8. R.Stickler and B.Weiss, Review of the application of US-fatigue test methods for the determination of crack growth and threshold behavior of metallic materials, ibid., p.135
9. A.F.Blom, A.Hadrboletz and B.Weiss, Effect of crack closure on near-threshold crack growth behavior in a high-strength Al-alloy up to ultrasonic-frequencies, Proc. ICM-4, Stockholm (1983), p.755
10. J.Piqueras, J.C.Grosskreutz and W.Frank, The influence of point defects clusters on fatigue hardening of Cu single crystals, Phys.Stat.Sol.11 (1972) 567
11. W.Hoffelner and B.Weiss, Cyclic hardening of Al at various test frequencies, Scripta Met. 12 (1978) 1947
12. H.Mughrabi and R.Wang, Cyclic strain localization and fatigue crack initiation in PSBs in fcc metals and single phase alloys, Proc.Internat.Symp. Tucno, PL (1980)