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Fatigue behaviour of precipitation-hardening medium C steels containing Cu

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The presence of Cu precipitates counteracts the cyclic softening present in ordinary quenched and tempered steels. This is expected to result in an increase in fatigue limit. The fatigue crack propagation rate (dc/dN) at constant ΔK in the Cu–C steels was shown to depend on heat-treatment and carbon content. To maximize yield strength and minimize $|da/dN|_{\Delta K}$ for tempering at 500° C, one must choose a low C content and temper for a short time; $|da/dN|_{\Delta K}$ in 0.28 wt% C–1.45 wt% Cu tempered for 13 min was one-third that for 0.45 wt% C–1.45 wt% Cu tempered for 200 min. There is also an advantage in adding Cu while simultaneously lowering the C content. The dc/dN data are discussed in terms of the yield strength and the energy to form a unit area of fatigue crack, U, which was measured using foil strain gauges. The quantity ($|dc/dN|_{\Delta K}\, \sigma_y'^2 U$), where σ_y' is the cyclic yield stress, was found to be nearly constant. In the 0.28 wt% C–1.45 wt% Cu alloy, short ageing times at 500° C resulted in greater resistance to initiation of cracks at notches for low ΔKs than long ageing times.

1. Introduction

For hardness values below approximately 35 Rockwell C or tensile strengths below approximately 1300 MN m^{-2}, the fatigue limit of quenched and tempered steels is proportional to the ultimate tensile strength or hardness [1]. Beyond these strength levels there is considerable scatter in the data and the scatter increases with strength level. The scatter can be reduced somewhat by careful polishing of the specimens indicating that scratch or flaw sensitivity is playing a role. In view of the occurrence of substantial cyclic softening in quenched and tempered steels, it is more reasonable to compare the fatigue limit to the cyclic mechanical properties. The fatigue limit and the cyclic yield stress, σ_y', have been shown to be approximately equal [2] provided that the tempering temperature is not too low.

In a quenched and tempered medium carbon steel containing Cu and NiAl coherent precipitates (0.3 wt% C, 4 wt% Ni, 1 wt% Al, and 1 wt% Cu), the precipitates gave a hardening contribution to the cyclic stress curve compensating for the cyclic softening observed in quenched and tempered steels [3]. The cyclic and monotonic stress-strain curves were much closer together when the steel was tempered 10 h at 650° C followed by ageing 8 h at 550° C than when tempered 1.5 h at 350° C. The former structure consisted of very coarse carbides plus small coherent Cu and NiAl precipitates, while the latter structure, which gave essentially the same strength level, consisted of fine lenticular carbides but no Cu and NiAl precipitates. Samples with the coherent precipitates were much more resistant to fatigue crack initiation in side notched (1 mm deep V-shaped) specimens; however, the rates of fatigue crack propagation, dc/dN, where c is the crack length and N is the number of cycles, were nearly the same in both.

It has previously been shown that dc/dN is inversely proportional to the materials strength

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[4, 5] and to the energy required to make a unit area of fatigue crack, U [6]. In ordinary quenched and tempered steels, U appears to increase as the tempering temperature is increased but the strength decreases so that there is little change in dc/dN [7].

In the present research, the fatigue properties of steels containing 1.5 wt % Cu and 0.28 to 0.45 wt % C were investigated. The Al was omitted in order to determine the effects of Cu precipitates and carbides in the absence of NiAl precipitates and also because Al additions tend to embrittle steel. For comparison, some measurements were made with a 1040 steel.

2. Experimental procedures

For this research four experimental alloys were supplied by the Inland Steel Research Laboratory as 2.8 mm thick hot-rolled sheets. Their analyses are given in Table I.

The tempering–ageing characteristics of the steels were initially determined by hardness testing (Rockwell C) after austenitizing 1 h at 1000° C. In contrast to the Cu–Ni–Al–C steel previously studied, the hardness of the Cu–C steels decreased monotonically with ageing time at all ageing temperatures; however, the Cu–C steels softened more slowly when tempered at 450 to 550° C than the 1040 steel, indicating that Cu precipitated simultaneously with coarsening of the carbides. For tempering at 550° C, the increase in hardness from the Cu precipitates was 15 to 20 Rockwell C. Unlike the Cu–Ni–Al–C steel, after tempering the Cu–C steels at 650° C, no subsequent hardening was observed on ageing at any lower temperature showing that most of the Cu was out of solution at 650° C.

A tempering temperature of 500° C was selected for most of the research on the Cu–C steels. The Rockwell C hardness values versus ageing time at 500° C for the three steels are shown in Fig. 1. The hardness level, of course, is highest for the alloy with the highest C content. A lower tempering temperature, 450° C, was selected for the 1040 steel to place it in the same strength range as the Cu–C steels. The principal tempering times selected were 13 and 200 min, to give different stages in the ageing processes, i.e. small and large carbide size and under aged and fully aged Cu precipitates. The austenitizing at 1000° C and tempering were done in 95% argon–5% hydrogen. After austenitizing, the specimens were quenched in oil, except for the 0.28 wt % C–1.45 wt % Cu steel which was quenched into water to prevent pearlite formation. The specimens were also water quenched after tempering. All specimens were heat-treated after machining.

TABLE I Chemical analyses of alloys

Heat no.	Cu (wt %)	C (wt %)	Mn (wt %)	P (wt %)	S (wt %)	Si (wt %)
112V73	1.45	0.28	0.52	0.010	0.006	0.008
112V74	1.44	0.36	0.55	0.008	0.006	0.003
112V75	1.46	0.45	0.55	0.008	0.006	0.003
113V74	0.001	0.40	0.54	0.008	0.008	0.004

The monotonic tensile testing was done with specimens 6 x 2.8 mm x 25 mm in the gauge length. The specimens for determination of the cyclic stress–strain curves, where buckling must be prevented during the compression part of the cycle, were 7.8 mm long and 4 mm x 2.8 mm cross-section in the gauge section but much larger in the grip sections. The fatigue-crack initiation and propagation specimens were 20 mm wide, 100 mm long and 2.25 mm thick. The surfaces were mechanically polished after heat-treatment using 600 grit SiC. For fatigue-crack propagation studies the specimens were centre-notched by spark cutting using a copper tool 0.13 mm x 0.7 mm cross-section. The fatigue-crack initiation experiments were conducted on side-notched specimens. The side notches, 1 mm long with 0.10 mm radius of curvature were ground into the specimens using an abrasive wire saw and 600 SiC abrasive.

Fatigue-crack initiation and propagation rates and cyclic stress–strain curves were determined with an MTS closed-loop electrohydraulic servo-valve testing machine of 90 kN capacity. The fatigue-crack initiation and propagation specimens

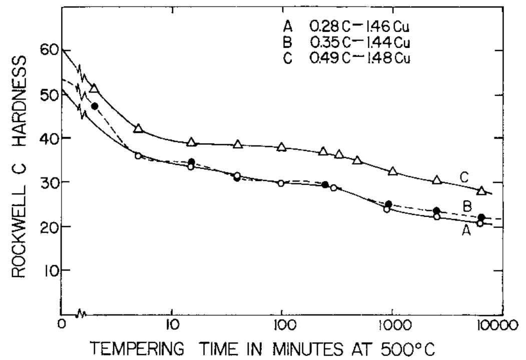

Figure 1 Hardness versus ageing time at 500° C for the three Cu–C steels. The austenitizing temperature was 1000° C.

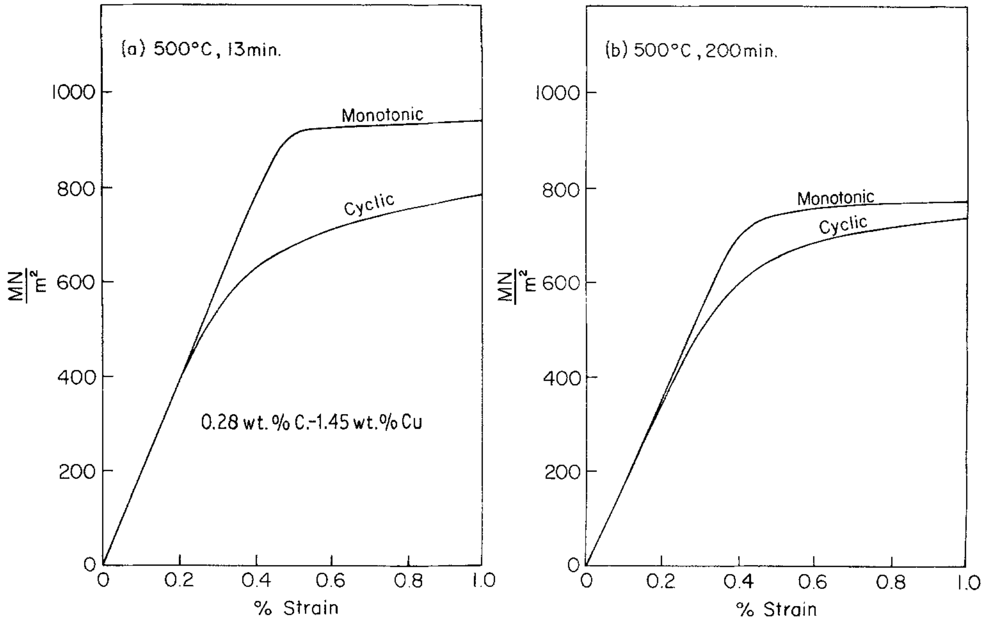

Figure 2 Monotonic and cyclic tension stress–strain curves for 0.28 wt % C–1.45 wt % Cu steel austenitized at 1000° C and aged for (a) 13 min and (b) 200 min at 500° C.

were solidly gripped and tested in a dry argon atmosphere in pull–pull with an R ratio ($\Delta\sigma_{\mathbf{min}}/\Delta\sigma_{\mathbf{max}}$) of 0.05 at a frequency of 30 Hz. Values of ΔK were calculated using the Isida finite width correction factor for the centre-notched specimens and the appropriate Pook correction factor for the side-notched specimens [7].

The monotonic stress–strain curves were determined partly using a screw-type Instron and partly with the MTS machine. The cyclic stress–strain curves were determined using the incremental step technique as described by Morrow and coworkers [8, 9]. Using the procedures described elsewhere [6], the plastic work in the plastic zone

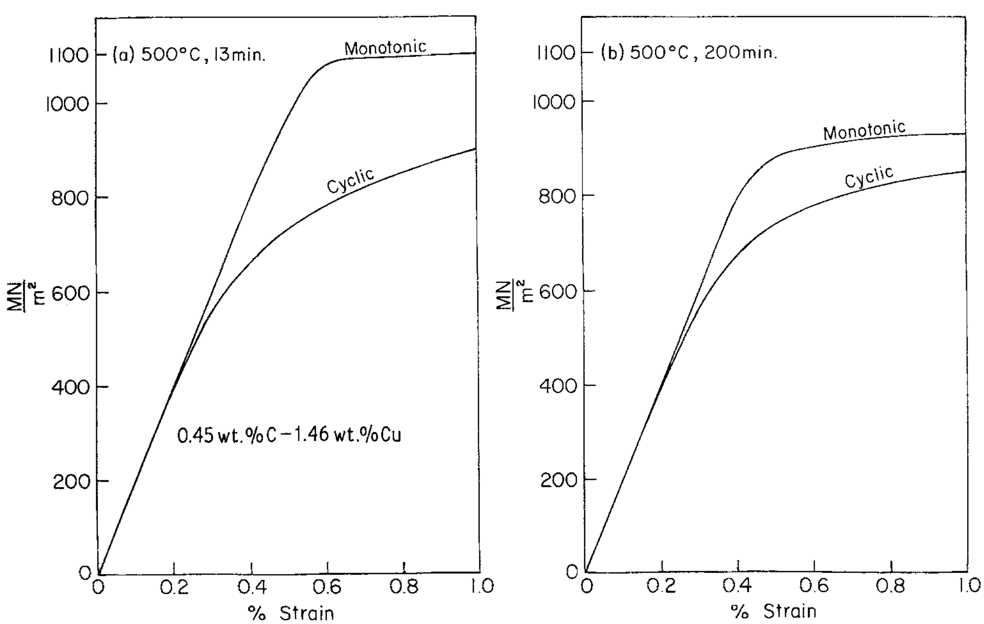

Figure 3 Monotonic and cyclic tension stress–strain curves for 0.45 wt % C–1.46 wt % Cu steel austenitized at 1000° C and aged for (a) 13 min and (b) 200 min at 500° C.

at the crack tip for a unit area of advancing crack was measured. In the procedure, small foil strain gauges (210 μm x 200 μm) are cemented ahead of the crack across and above the crack line. The nominal load and strain from the strain gauge are recorded on an x–y recorder during fatigue cycling as the crack approaches the strain gauge. The local stress at the strain gauge is estimated from a hysteresis loop on a bulk specimen equating the strains. The hysteresis loops are then integrated to give their areas and these are summed over space to give the total hysteretic energy per unit area of crack advance.

2.1. Cyclic stress–strain behaviour

Figs. 2 and 3 show the tension cyclic and monotonic stress–strain curves for the 0.28 wt % C–1.45 wt % Cu and 0.45 wt % Cu–1.46 wt % Cu steels tempered for 13 and 200 min at 500° C. The compression monotonic and cyclic stress–strain curves were somewhat above the tension curves due to the strength differential effect [10]. The amount of cyclic softening increases with C content and decreases with ageing time in keeping with the idea that the Cu precipitates counteract the cyclic softening in ordinary quenched and tempered steels. In particular, the 0.28 wt % C–1.45 wt % Cu steel aged for 200 min at 500° C shows the least cyclic softening particularly at large strains. The cyclic softening in 1040 steel tempered at 450° C is much larger than in the Cu–C steels. Curves for 1040 steel are not shown; however, the sizeable softening is indicated by the last column in Table II which gives the ratios of the cyclic and monotonic yield stresses. On tempering the 1040 steel at 500° C, σ_y is substantially less than for tempering at 450° C but the ratio of σ'_y/σ_y is somewhat higher. Following the previously reported correlation between the 0.2% cyclic offset yield stress and the fatigue limit, an improvement in fatigue limit is predicted from Cu precipitation in quenched and tempered medium carbon steels over that in steel without Cu precipitates heat-treated to the same monotonic yield stress. Kenneford and Oxlee [11] in fact observed that the fatigue endurance limit–ultimate tensile strength ratio is increased by copper additions.

2.2. Fatigue crack propagation

Plots of $\mathrm{d}c/\mathrm{d}N$ versus ΔK for the three Cu–C alloys tempered at 500° C for 13 and 200 min and for the plain carbon steel tempered at 450° C are given in Figs. 4 to 7. The data were fit to the Paris equation

$$\mathrm{d}c/\mathrm{d}N = C(\Delta K)^m. \quad (1)$$

The Paris exponents are given in Table II along with $\mathrm{d}c/\mathrm{d}N$ values at $\Delta K = 20\,\mathrm{MN\,m^{-3/2}}$. Data for the 0.4% C steel aged at 500° C is also given. The $\mathrm{d}c/\mathrm{d}N$ versus ΔK curves are mostly parallel with most of the m values being near 3. Except for the 0.40 wt % C steel tempered at 500° C, m decreases slightly with ageing time.

As discussed previously [4–6], $|\mathrm{d}c/\mathrm{d}N|_{\Delta K}$ is inversely related to σ_0 (an appropriate measure of the materials' strength), μ, the shear modulus and U, the energy to make a unit area of fatigue crack

TABLE II Paris equation exponents, $\mathrm{d}c/\mathrm{d}N$ at $\Delta K = 20\,\mathrm{MN\,m^{-3/2}}$ and strength properties

Steel*	T_t (° C)	t_t (min)	m	$\mathrm{d}c/\mathrm{d}N$ $\Delta K = 20\,\mathrm{MN\,m^{-3/2}}$ ($\mathrm{mm\,sec^{-1}} \times 10^5$)	σ_y ($\mathrm{MN\,m^{-2}}$)	σ_u ($\mathrm{MN\,m^{-2}}$)	σ'_y ($\mathrm{MN\,m^{-2}}$)	σ_y/σ'_y
0.28 C–1.45 Cu	500	13	3.4	1.7 to 1.8	880	940	710	0.81
		200	3.2	2.9 to 3.5	780	860	680	0.87
0.36 C–1.44 Cu	500	13	3.2	2.3 to 3.0	1020	1110		
		200	2.6	4.8 to 5.5	870	960		
0.45 C–1.45 Cu	500	13	3.0	4.0 to 4.4	1100	1180	790	0.72
		200	2.8	4.5 to 6.6	950	1050	770	0.81
0.40 C	450	12	3.0	1.5 to 1.6	990	1050	550	0.56
		240	2.8	3.0 to 3.6	850	920	410	0.48
0.40 C	500	13	3.4	3.0 to 4.0	740	760	480	0.65
		200	4.0	1.9 to 2.3	600	710	390	0.65

* Compositions in wt %.
σ_y and σ'_y = 0.2% offset monotonic and cyclic yield stress, σ_u = ultimate tensile strength.

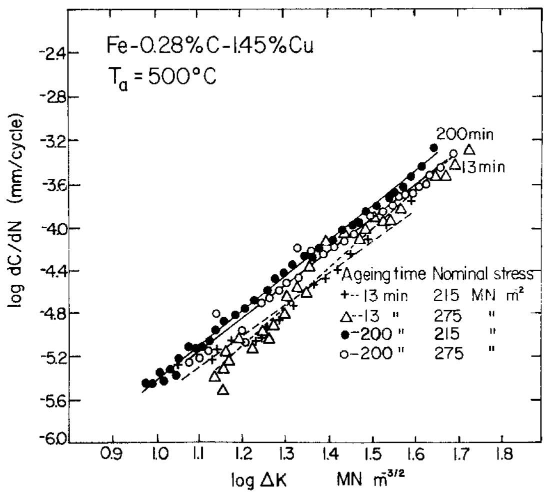

Figure 4 Fatigue-crack propagation versus ΔK for 28 wt% C–1.46 wt% Cu steel aged 13 and 200 min at 500° C. Specimens were austenitized at 1000° C for 1 h and then water quenched.

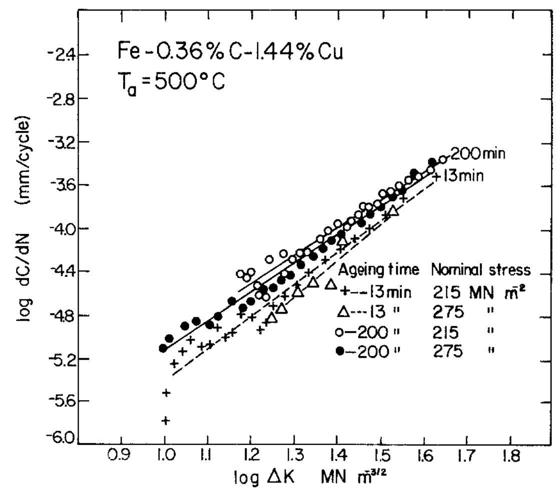

Figure 5 Fatigue-crack propagation versus ΔK for 0.36 wt % C–1.44 wt % Cu steel aged 13 and 200 min at 500° C. Specimens were austenitized at 1000° C for 1 h and then are water quenched.

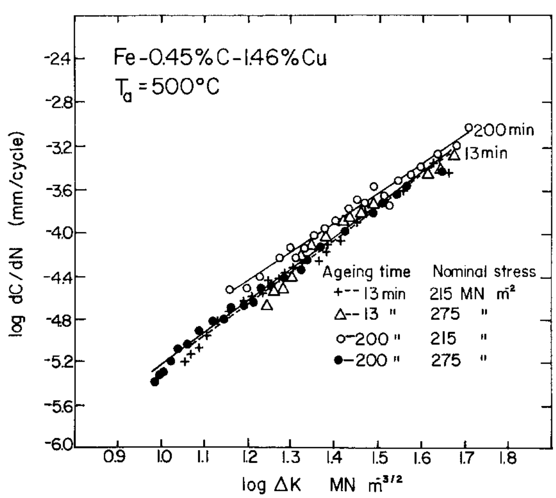

Figure 6 Fatigue-crack propagation versus ΔK for 0.45 wt %C–1.46 wt % Cu steel aged 13 and 200 min at 500° C. Specimens were austenitized for 1 h and then oil quenched.

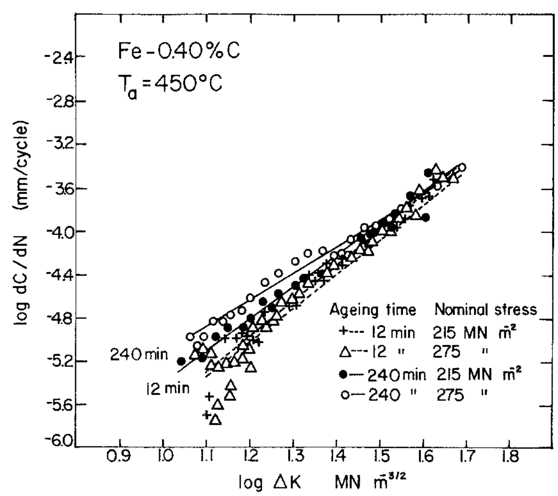

Figure 7 Fatigue-crack propagation versus ΔK for 0.40 wt %C steel tempered 12 and 240 min at 450° C. Specimens were austenitized at 1000° C for 1 h and then oil quenched.

$$|\,\mathrm{d}c/\mathrm{d}N\,|_{\Delta K} \sim 1/\mu\sigma_0^n U. \qquad (2)$$

The exponent n is thought to be approximately 2. Table II also shows average values of the 0.2% offset monotonic and cyclic yield stresses, σ_y and σ_y', and the ultimate tensile strength σ_u. The shear modulus is essentially the same for all of the alloys and all of the heat-treatments.

If U is constant then $|\,\mathrm{d}c/\mathrm{d}N\,|_{\Delta K}$ is expected to be inversely proportional to σ_0^n. The fatigue crack propagation rate increases with ageing time and decreases in σ_y, σ_y' and σ_u except for the plain C steel tempered at 500° C where the reverse behaviour was observed. Further insight into the relation between $|\,\mathrm{d}c/\mathrm{d}N\,|_{\Delta K}$ and σ_0 and how U might be playing a role was obtained by taking σ_0 to be σ_y or σ_y' and plotting $|\,\mathrm{d}c/\mathrm{d}N\,|$ at $\Delta K = 20$ MN m$^{-3/2}$ versus σ_y (Fig. 8) and versus σ_y' (Fig. 9). The data for the 0.40 wt%C steel for 500° C tempering is not shown, but will be discussed subsequently. As shown in Figs. 8 and 9, at constant Cu content, increasing the C from 0.28 wt% to 0.45 wt% increases the level of the $\mathrm{d}c/\mathrm{d}N$ versus σ_y or σ_y' curves. From Equation 2 a de-

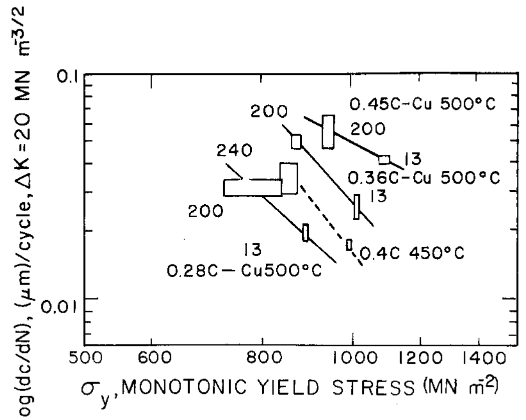

Figure 8 Log dc/dN versus monotonic yield stress for Cu–C steels aged at 500° C and 0.40% C steel tempered at 450° C. Temperatures on figure refer to tempering temperatures. Numbers refer to tempering times in minutes.

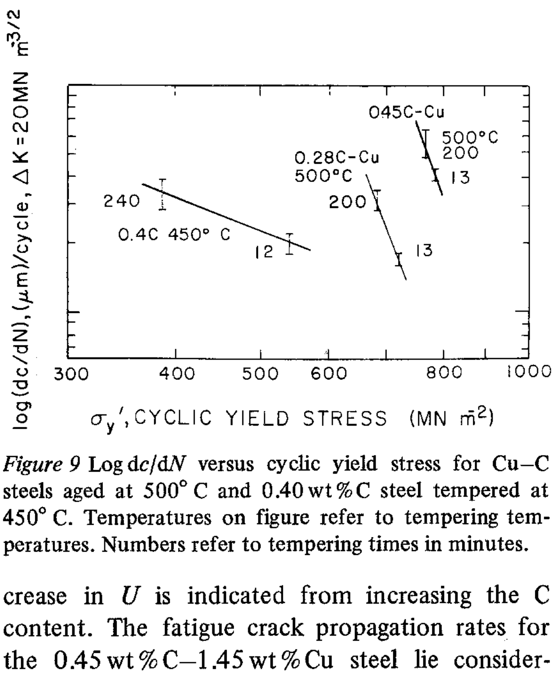

Figure 9 Log dc/dN versus cyclic yield stress for Cu–C steels aged at 500° C and 0.40 wt % C steel tempered at 450° C. Temperatures on figure refer to tempering temperatures. Numbers refer to tempering times in minutes.

crease in U is indicated from increasing the C content. The fatigue crack propagation rates for the 0.45 wt % C–1.45 wt % Cu steel lie considerably above those for the 0.40 wt % C steel, indicating that Cu additions also decrease U.

At constant monotonic or cyclic yield strength for lowest $|dc/dN|_{\Delta K}$, the present results for C–Cu steels tempered at 500° C show that one must use a low carbon content and temper for a short time. If high cyclic yield strength is desired, then the 0.28 wt % C–1.45 wt % Cu alloy represents a substantial improvement over a 1040 steel at the same monotonic yield strength level since a quite low tempering temperature would be required to give the 1040 steel a comparable range of σ_y' values. Such heat-treated 1040 steel would be quite brittle and highly sensitive to initiation of cracks at notches.

In order to examine further the relation between $|dc/dN|_{\Delta K}$ versus σ_y in the 0.40 wt % C steel, the range of tempering was extended to 700° C. The results are given in Fig. 10. With increase in ageing time at 600° C $|dc/dN|_{\Delta K}$ increases but is essentially independent of ageing time at 700° C where there is little change in σ_y or σ_y' between ageing 13 and 700 min, as shown in Fig. 11. Thus the reverse behaviour shown in Fig. 10 is characteristic of the tempering region around 500° C only. The fact that $|dc/dN|_{\Delta K}$ decreases with tempering time at 500° C may be associated with some structural change characteristic of short time tempering at 500° C giving a small U, i.e. temper embrittlement.

2.3. Experimental determination of U

In previously published research in this laboratory [6] U was experimentally determined from the total hysteretic plastic work associated with a unit area of fatigue crack propagation. The non-hysteretic plastic work in the plastic zone around the propagating crack is small and may be neglec-

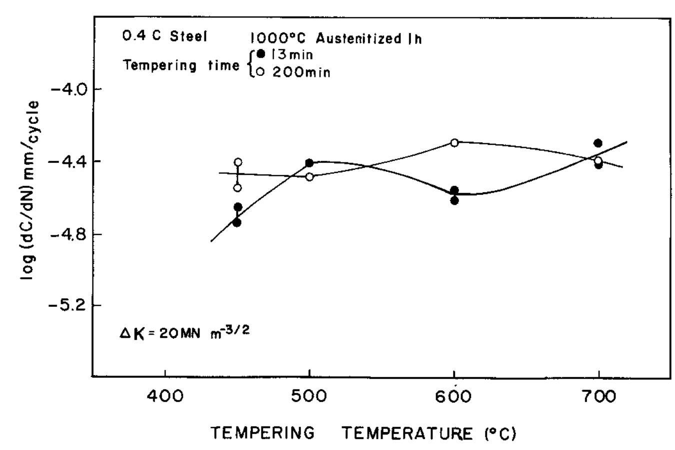

Figure 10 Fatigue-crack propagation rate (dc/dN) at $\Delta K = 20\,\mathrm{MN\,m^{-3/2}}$ for 0.40 wt % C steel tempered 13 and 200 min from 450 to 700° C.

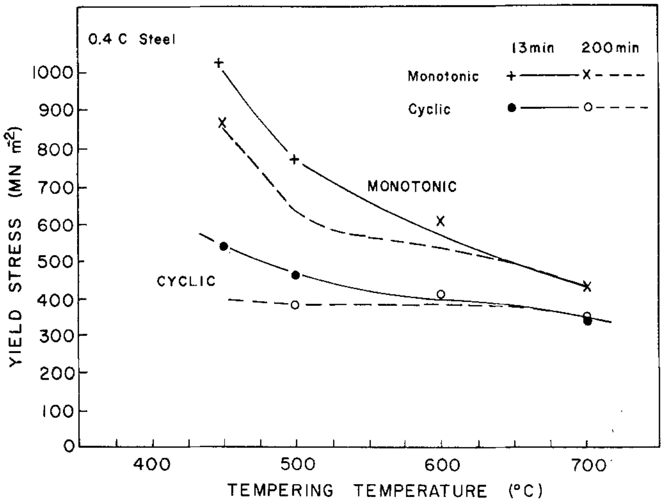

Figure 11 Cyclic and monotonic 0.2% offset yield stress of 0.40 wt%C steel tempered 13 and 200 min from 450 to 700° C.

ted for the ΔK values investigated. The measured U correlated well with the much smaller value of $|dc/dN|_{\Delta K}$ for a Nb-doped HSLA steel compared to 7050 Al base alloy even though σ_y' is actually larger for the 7050 Al alloy; $(|dc/dN|_{\Delta K}\sigma_y'^2 U)$ was nearly constant.

Measurements of U were made on 0.28 wt%C–1.45 wt%Cu aged 13 min at 500° C and 0.45 wt% C–1.46 wt%Cu aged 200 min at 500° C to determine whether the difference in $|dc/dN|_{\Delta K}$ could be explained by the difference in U. The method for measuring U is outlined under Section 2 and in detail in [6] where the approximations involved are discussed.

Defining x as the distance between the crack tip and the centre of the strain gauge along the line of the crack trace and y as the distance the gauge is above or below the crack line, U_{xy} is plotted versus x for several values of y in Fig. 12. Integration of these curves gives the hysteretic plastic work for a belt parallel to the crack line. These are plotted versus y in Fig. 13 for the steels and heat-treatments studied. Twice the area under these curves gives a measure of the hysteretic plastic work for a unit area advance of a fatigue crack. The U values so determined were 3×10^5 J m^{-2} for the 0.28 wt%C–1.45 wt%Cu steel tempered

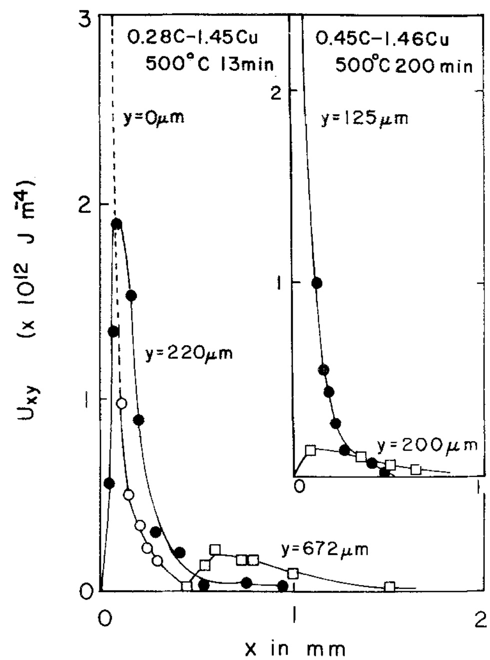

Figure 12 Hysteretic plastic work per cycle U_{xy} versus the distance between crack tip and strain gauge along x direction, i.e. crack direction which is normal to applied stress. Profiles are drawn for several values of y, the direction normal to the crack direction. $\Delta K = 20$ MN m$^{-3/2}$.

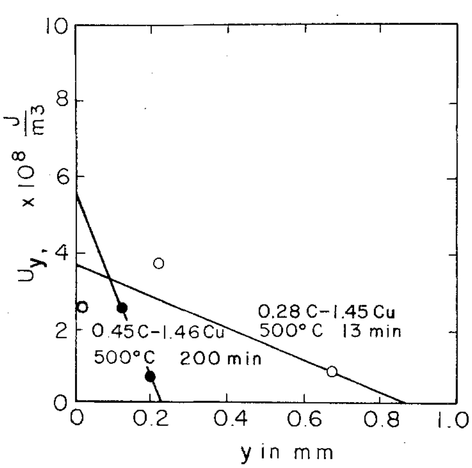

Figure 13 Areas under curves in Fig. 11 (U_y) versus y. $\Delta K = 20$ MN m$^{-3/2}$.

TABLE III Calculation of $[\,|\,dc/dN\,|_{\Delta K}(\sigma_y')^2 U]$

	0.28 C–1.45 Cu* 13 min at 500° C	0.45 C–1.46 Cu* 200 min at 500° C
σ_y ($MN\,m^{-2}$)	880	950
σ_y' ($MN\,m^{-2}$)	710	770
True fracture stress ($MN\,m^{-2}$)	1720	1850
dc/dN at $\Delta K = 20\,MN\,m^{-3/2}$ ($m\,sec^{-1}$)	1.8×10^{-8}	5.6×10^{-8}
U ($J\,m^{-2}$)	3×10^5	1×10^5
$[\,|\,dc/dN\,|_{\Delta K}(\sigma_y')^2 U]$ ($MN^3\,m^{-4}$)	3×10^{-3}	3×10^{-3}

* Compositions in wt %.

13 min at 500° C and $1 \times 10^5\,J\,m^{-2}$ for the 0.45 wt % C–1.46 wt % Cu steel tempered 200 min at 500° C. Table III shows a comparison of the results for the two steels and treatments. The quantity, $(|\,dc/dN\,|_{\Delta K}(\sigma_y')^2 U)$, is seen to be essentially constant for the two steels. The differences in U and σ_y' compensate for the difference in $|\,dc/dN\,|_{\Delta K}$. The agreement is not quite as good if σ_y or the true fracture stress are used instead of σ_y'.

By analogy with the monotonic case where the ductility is defined as the area under the stress–plastic strain curve, U may be defined as the cyclic ductility.

Fig. 13 also gives an indication of the size of the cyclic plastic zone in the y direction for $\Delta K = 20\,MN\,m^{-3/2}$. It is considerably larger for the lower carbon steel. The larger value of U is due to the larger plastic zone size.

2.4. Fatigue crack initiation

Some initial studies on fatigue crack initiation were completed for the 0.28 wt % C–1.45 wt % Cu steel aged for 13 min and 20 h at 450° C. This steel was selected for the initial study because it had the lowest fatigue-crack propagation rates. The austenitizing temperature was reduced to 830° C because it was thought that the smaller austenite grain size would be beneficial. The samples were side-notched (1 mm x 0.20 mm) using an abrasive wire saw and 600 grit SiC as the abrasive. The side of the sample at the notch was observed with a 40x telemicroscope. The minimum crack size observable was about 50 μm. Fatigue-crack initiation values so defined are plotted versus $\Delta K/\sqrt{\rho}$ [12] for the two treatments in Fig. 14, where ρ is the radius of curvature at the tip of the notch. For comparison, some of the strength properties are shown in Table IV. While there is scatter in the data for the 20 h ageing, the shorter ageing time which gives the higher strength is clearly more resistant to fatigue-crack initiation at low values of $\Delta K/\sqrt{\rho}$. The reverse is true at high values of $\Delta K/\sqrt{\rho}$. Thus for this alloy aged at 450° C, the short ageing time gives the highest resistance to fatigue-crack initiation at the "stan-

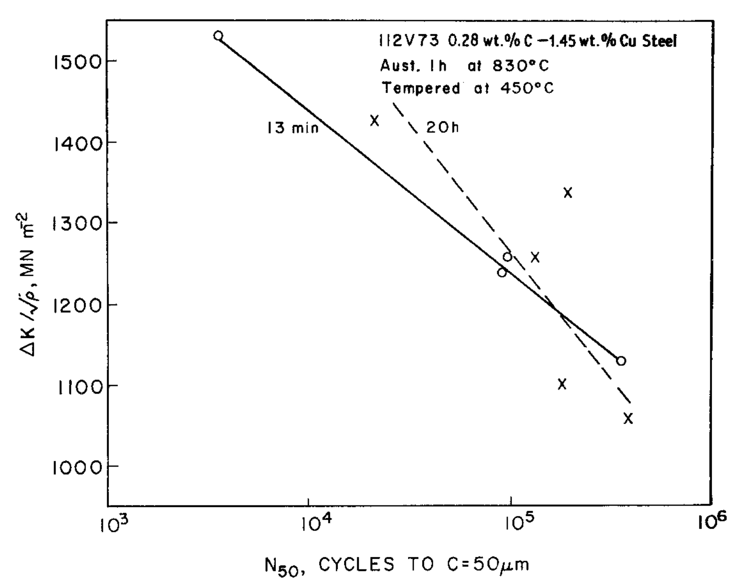

Figure 14 Number of cycles to 50 μm long fatigue crack versus $\Delta K/\sqrt{\rho}$ for 0.28 wt % C–1.45 wt % Cu steel aged 13 min at 20 h at 450° C.

TABLE IV Strength properties of 0.28 wt % C–1.45 wt % Cu steel aged at 450° C

	t_a	
	13 min	20 h
σ_y ($MN m^{-2}$)	980	770
σ'_y ($MN m^{-2}$)	690	520
σ_u ($MN m^{-2}$)	1090	960
% elongation	10	13

dard notch" used in this research for low $\Delta K/\sqrt{\rho}$, the lowest values of $|dc/dN|_{\Delta K}$ ($\Delta K = 20$ MN $m^{-3/2}$), and the highest strength values.

3. Conclusions

(1) In a previous study [2] the fatigue limit in steels was found to be approximately equal to the cyclic yield stress below about 1400 MN m^{-2}. Ordinary quenched and tempered steels show considerable cyclic softening. Cu precipitates counteract the cyclic softening so that the cyclic stress–strain curve is more nearly like the monotonic stress–strain curve, especially at large strains. This is expected to result in an increase in endurance limit.

(2) A definite variation of the fatigue-crack propagation rate at constant ΔK, $|dc/dN|_{\Delta K}$, was observed with changes in composition and heat-treatment. In the C–Cu steels tempered at 500° C for constant monotonic or cyclic yield stress, low carbon content and short ageing times give the lowest value of $|dc/dN|_{\Delta K}$. The difference in $|dc/dN|_{\Delta K}$ over the limited range of composition and heat-treatment studied was about a factor of four. This suggests that a modest decrease in $|dc/dN|_{\Delta K}$ may be obtained by optimizing composition and heat-treatment. Short ageing times also impart more resistance to initiation of cracks at notches at low ΔK than long ageing times in keeping with the higher strength.

(3) The plastic work expended in making a unit area of fatigue crack, U, was determined by measuring the hysteretic plastic work attendant to a unit area of fatigue crack advance following procedures reported elsewhere [6]. The quantity ($|dc/dN|_{\Delta K}\sigma'^2_y U$) was the same for the 0.28 wt % C–1.45 wt % Cu steel aged 13 min at 500° C and the 0.45 wt % C–1.46 wt % Cu steel aged 200 min at 500° C even though their values of $|dc/dN|_{\Delta K}$ differed by a factor of three. This difference was essentially compensated by a factor of 3 difference in U.

(4) The fatigue-crack propagation rate at constant ΔK is inversely related to the strength of the alloy and U which may be defined as the cyclic ductility. The cyclic ductility decreases as the C content of the Cu–C steels is increased. Comparison with results for the 0.40 wt % C steel showed the cyclic ductility also decreases on adding Cu.

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