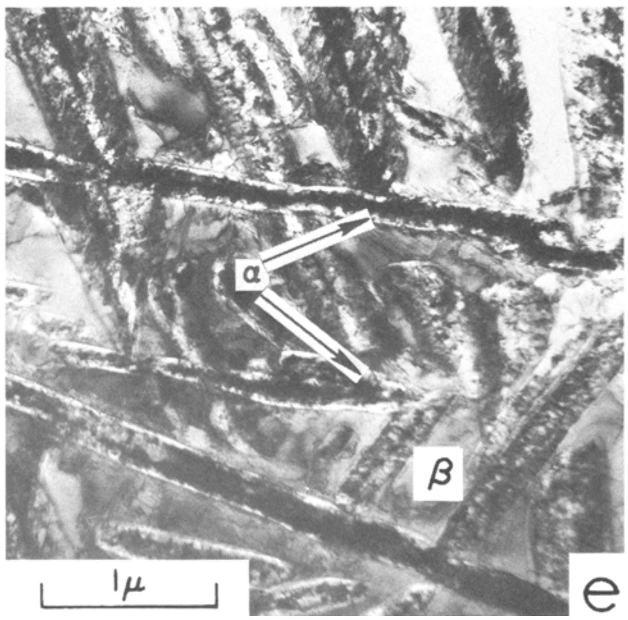

Fig. 2—Continued.

composition of the eutectoid determined during this investigation are the lowest reported to date. The explanation for such results, and for the large differences among the findings of the various investigations, appears to lie in three of the difficulties encountered during the present study. One is the previously mentioned near uselessness of optical microscopy for ascertaining the eutectoid temperature; previous investigators relied on this form of microscopy in varying, but often extensive degrees. It was also found that comparatively long annealing times were required in the vicinity of the eutectoid temperature in order to coarsen the microstructure sufficiently to obtain good diffraction patterns. Both of these difficulties can result in increased apparent values of the eutectoid temperature. The higher eutectoid compositions reported by previous investigators may also have been due indirectly to severe staining experienced during etching (98 pct HNO_3, 2 pct HF, lightly swabbed, or 5 pct HF in water were used in this study) of the 23.6 and 33.0 wt pct Pd alloys after annealing in the vicinity of the $\beta/(\beta + Ti_2Pd)$ transus. Although these microstructures were coarse, repeated polishing and etching at both the North American Rockwell and the Ford laboratories was required before the authors could be certain that the presence or absence of Ti_2Pd had been established. This difficulty, of course, also hampered determination of the $\beta/(\beta + Ti_2Pd)$ transus *per se*.

As a result of this study, the authors are convinced that in alloy systems in which thin foil preparation can be accomplished with reasonable facility, transmission electron microscopy should be looked upon as a principal technique for phase diagram determination. The combination of high resolution and electron diffraction capability make the results so much more certain that the added time required for specimen preparation and examination is more than offset.

During a portion of this investigation, one of us (H.I.A.) was a Visiting Scientist at the North American Rockwell Science Center, and wishes to thank the staff of the Center for their hospitality and for a stimulating visit.

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Recent Studies into the Mechanism of Ridging in Ferritic Stainless Steels

HUNG-CHI CHAO

THE ferritic stainless steels, such as AISI Types 430 (17 pct Cr) and 434 (17 pct Cr, 1 pct Mo), are used extensively in many applications requiring forming and drawing operations, such as kitchen sinks and automobile trim and other decorative items. However, an undesirable surface condition, known as "ridging" or "roping," often develops during forming. This undesirable defect, which always occurs parallel to the sheet rolling direction, appears on the surface of a formed part as narrow, raised areas similar to corrugations. Ridging is detrimental to the appearance of decorative items, and expensive grinding and polishing operations are required to eliminate it.

In recent years, a number of mechanisms[1-6] have been proposed to explain ridging, all based on considerations of crystal plasticity and texture observation. Most recently, a plastic-buckling mechanism was proposed[6] which seems to indicate that ridging is not caused solely by anisotropic plastic flow. The present paper presents the results of recent studies to clarify certain conflicting points among various proposed mechanisms.

Experimental Results and Discussion. As observed by Chao,[1,2] by Ohashi,[4] and by Pouillard and Osdoit,[7] bands of cube-on-face texture groups (CF) such as $\{001\}\langle 011\rangle$, $\{117\}\langle 011\rangle$, $\{115\}\langle 011\rangle$, and $\{113\}\langle 011\rangle$ and bands of cube-on-corner texture groups (CC) such as $\{111\}\langle 011\rangle$, $\{111\}\langle 112\rangle$, and $\{554\}\langle 225\rangle$[2] exist in the rolling direction of ferritic stainless sheet. During straining along any direction in the plane of the sheet, the anisotropic plastic flow of these highly banded structures can be described by using the strain ratio (r value). The r values for textures common to ferritic material are well documented in the literature.[8] Fig. 1 shows r values calculated for both CF and CC textures as a function of direction of testing.[9] For all directions of stressing, the CC-textured grain offers two to three times more thinning resistance (r = 2 to 3) than the highest value for the CF-textured grain

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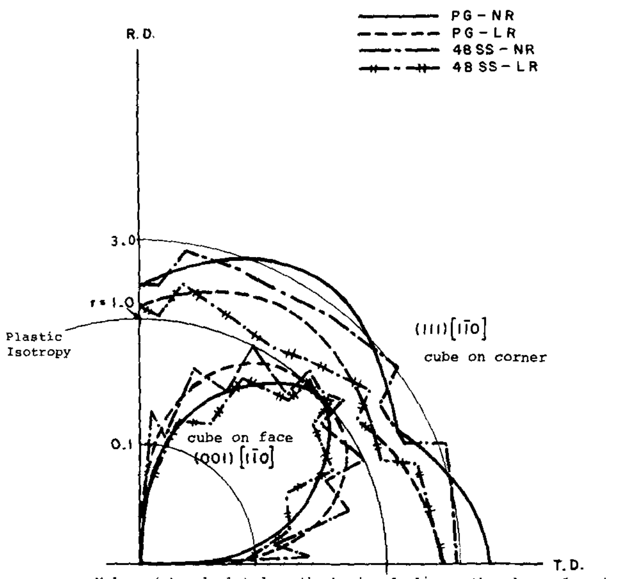
Table I. Chemical Composition of AISI Type 434 Stainless-Steel and Carbon-Steel Specimens Used in This Study

	Wt Pct		
	Type 434 Stainless	SK Steel	Rimmed Steel
C	0.088	0.040	0.030
Mn	0.36	0.30	0.37
P	0.019	0.007	0.004
S	0.007	0.015	0.007
Si	0.58	0.015	0.006
Cu	0.045	*	*
Ni	0.16	*	*
Cr	17.10	*	*
Mo	1.04	*	*
N	0.030	0.009	0.002
Al soluble	*	0.031	0.004
Al total	*	0.037	0.008

*Not determined.



Values (r) calculated on the basis of slip on the plane of maximum resolved shear stress, assuming both pencil-glide slip (P.G.) and slip on 48 slip systems encompassing <111> directions and the {110} {112} and {123} planes. Results calculated both on basis of initial orientation of crystal (N.R.) and after rotation of crystal axis (L.R.).

Fig. 1—Polar plot of calculated r-values for (001)[1$\bar{1}$0] and (111)[1$\bar{1}$0] textures. From Ref. 9.

(r = 1). Ridging should be expected for all directions of stressing.

Commercial Type 434 stainless-steel sheet (with the composition shown in Table I and texture components shown in Fig. 2) was strained in tension. The tensile axis of the specimens was along directions 0, 30, 45, 60, and 90 deg from the rolling direction. The surface undulations of the strained samples are shown in Fig. 3. Notice that all ridges are along the rolling direction. The appearance of the rolling-direction ridges on the transversely strained sample could not be produced by the plastic buckling mechanism for ridging as proposed by Wright.[6] A tensile stress applied in the transverse direction can hardly produce compressive stress to cause buckling of the (001)[110] texture to give ridges along the rolling direction.

In a previous study,[2] relatively little ridging was found in a sample (No. 436H) with a texture almost free of (100)[110] texture. Additional experimental

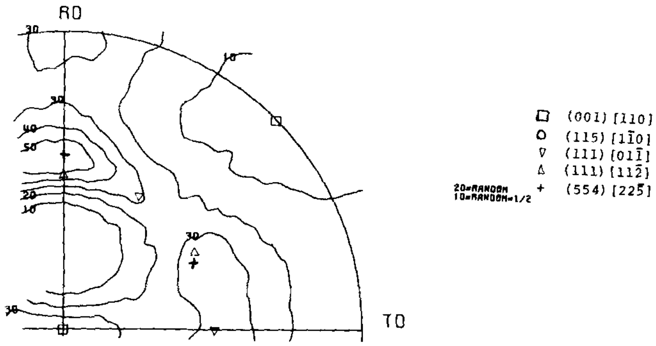


Fig. 2—Pole figure of specimen shown in Fig. 3.

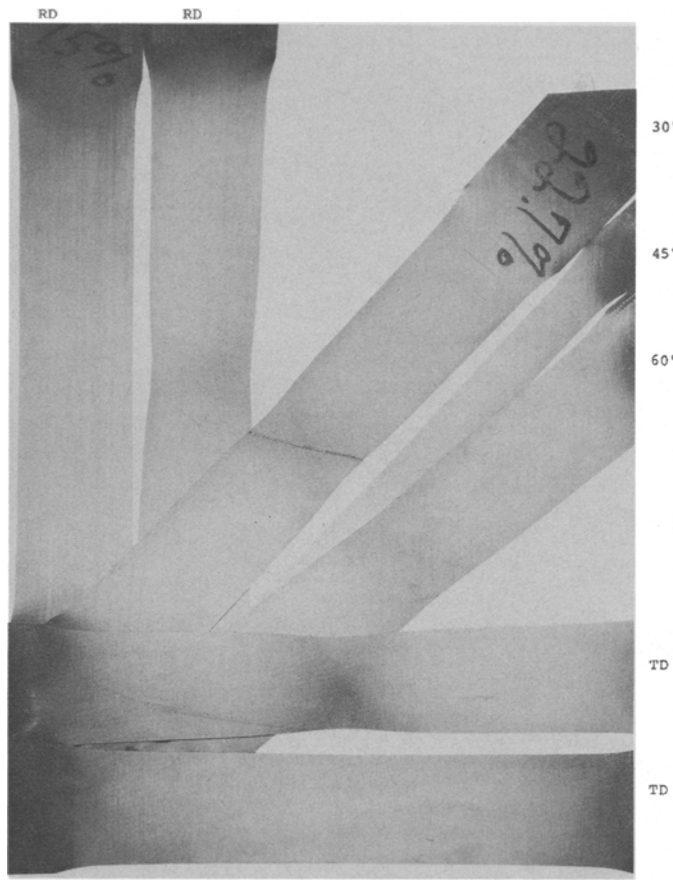


Fig. 3—Ridging of Type 434 specimens pulled along the rolling direction, along the transverse direction, and along directions 30, 45, and 60 deg away from the rolling direction. Specimens were obtained from the same sample sheet.

data from various commercial AISI 430-type stainless steel, Table II, offers further evidence to support this observation. Of the six ferritic stainless steels tested, those with CF and nearby texture components ridged whereas those with a well-developed CC fiber texture and no CF components showed little or no ridging. This same effect occurs in carbon steel. Fig. 4 compares the ridging behavior of low-carbon rimmed steel with that of low-carbon special killed (SK) steel, the compositions of which are shown in Table I. The rimmed steel, which has an extremely low r value, and a mixture of CC and CF texture, Fig. 5(a), ridges whereas the SK steel, which has a highly CC textured structure, Fig. 5(b), does not ridge.

Table II. Correlation of Textures With Degree of Ridging in Ferritic Stainless Steels

Sample No.[a]		Composition, Pct C	Cr	Ni	N	Mn	Si	Other Significant Alloying Elements	CC[e] Component	CC[e] Fiber-Background Texture	Other Texture Components[f]	Degree of Ridging Determined by the Room-Temperature Tensile Test
U-8X1938	S[b]	0.064	17.39	–	0.037	–	–	0.48 Cb	X2.8	X1.7	CE	1[d]
	M[c]								X2	X1.2	CE	
U-9X0766	S	0.064	17.5	0.25	0.037	0.43	0.26	0.50 Cb	X1.5	X1.2	CE	1
	M								X1.5	X1	CE	
U-9X0566	S	0.078	16.4	0.25	0.043	0.35	0.22		X1	X0.5	CE	2
	M								Not developed	X0.5	CF and its nearby textures	
U-9X0722	S	0.072	17.2	0.25	0.054	0.44	0.34		X1.5	X1	CE	3
	M								Not developed	X0.7	CF and its nearby textures	
C-430X	S	0.060	15.21	0.20	–	0.67	0.66	0.41Cb, 0.01Ti, 0.02Mo	X4	X3	CE	0
	M								X2.1	X1.6	CE	
C-CUCA	S	0.086	16.0	0.35	0.063	0.47	–	0.39Cb, 0.036Ti, 0.028Al	X1.7	X1.3	CE and some CF and its nearby textures	3
	M								X1.4	X0.8	CE and some CF and its nearby textures	

a. U group and C group are from two commercial producers.
b. S denotes texture of surface portion of sample.
c. M denotes texture of midthickness portion.
d. Degree of ridging is classified from 0 to 4 where 0 indicates no or very slight ridging and 4 indicates most severe ridging.
e. CC denotes cube-on-corner textures.
f. CE denotes cube-on-edge textures and CF denotes cube-on-face textures.

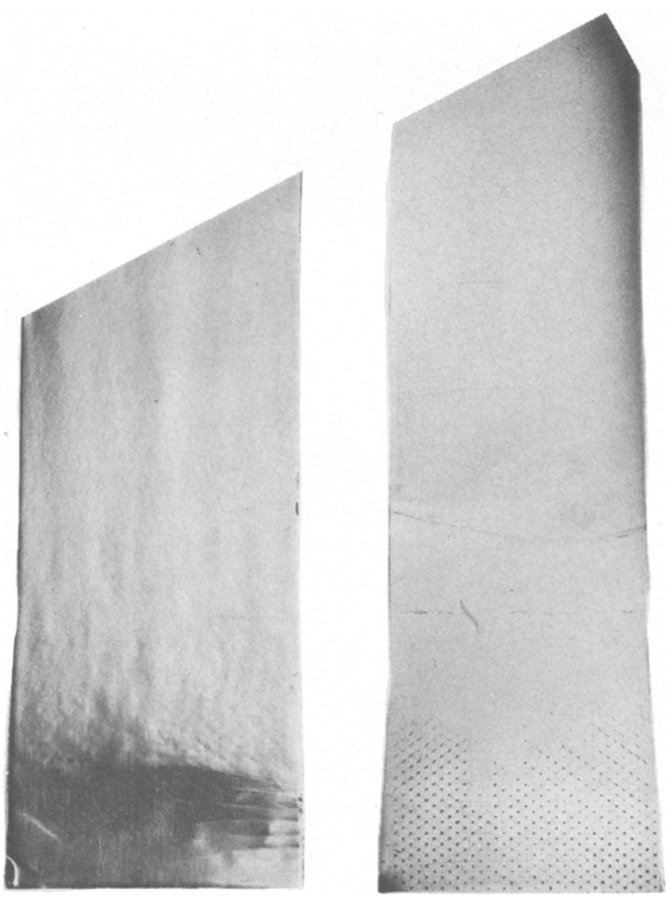


Fig. 4—Surface appearance of rimmed steel (left) and SK-steel (right) sheet after 15 pct straining.

The fact that ridging always appears with the banded CF texture in the CC texture matrix indicates that the mechanism proposed by Takechi and coworkers[5] can-

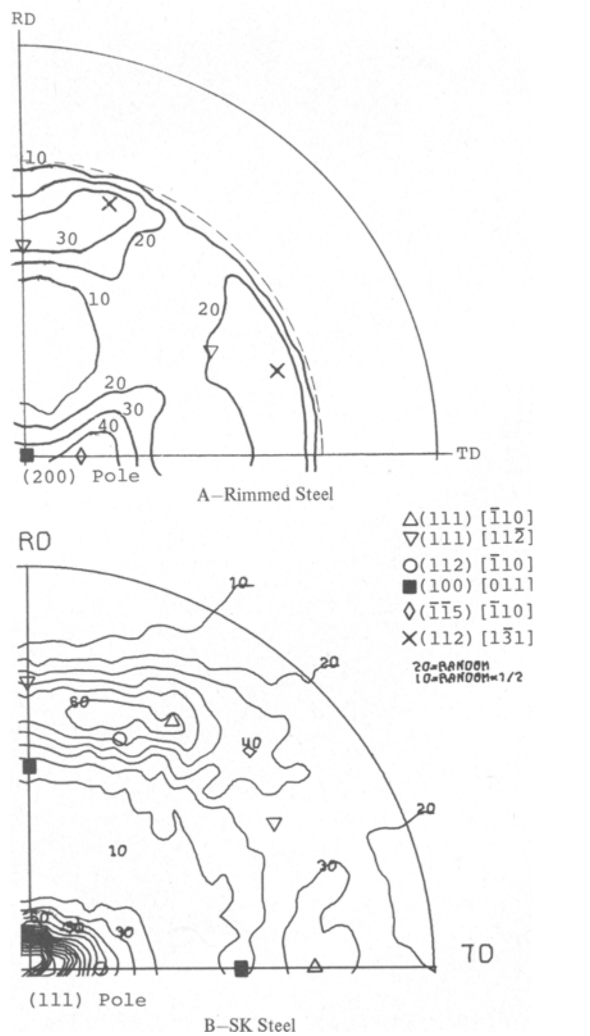


Fig. 5—Pole figures of rimmed and SK carbon-steel sheet.

not be generalized, although it might be true for a particular case. The samples they used had a strong (100)[011] texture component along with {111}⟨011⟩ and {211}⟨011⟩. Ridging must be shown to occur in a sample without a (100)[011] component before their proposed mechanism can be accepted as general.

Conclusion. New experimental evidence was obtained to verify current theories of the mechanism of ridging. The observation of ridging in transverse direction tension tests rules out the "buckling" theory.[6] The fact that ridging correlates with the relative amount of CF and CC texture in carbon steels as well as stainless steels supports the proposed "mixed-texture bands" of different component theory[2,3] but not the theory of "mixed texture rotations" of the same component.[5] Ridging should always appear in the rolling direction, irrespective of direction of strain because ridging is due to the anisotropic flow of the mixture of the banded structures of two or more highly anisotropic orientations coexisting along the rolling direction.

The author is grateful to Dr. Waldo Rall and Mr. P. A. Stoll for supplying the X-ray data on commercial stainless steels listed in Table II, to Dr. P. R. Mould for offering the carbon steel samples, and to Drs. K. G. Brickner and J. D. Defilippi for their helpful suggestions after reviewing the manuscript. The critical review of Professor W. F. Hosford is also deeply appreciated.

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Strength Differential Effect in α-Pu

M. D. MERZ

THE strength differential effect (SD effect) refers to a difference in the compressive and tensile strengths in a material. The effect has recently been studied in steels[1-4] and polymers.[5] This communication is a report of a very large strength differential effect in monoclinic α-Pu. The difference in compressive and tensile strengths was studied at strain rates from 8.3 $\times 10^{-3}$ to 6.7×10^{-5} s^{-1} at 22 to 100°C. A significant increase in σ_t/σ_c, the ratio of tensile to compressive flow stress, from 0.58 to 0.92, was observed as temperature increased; a similar increase was observed with decrease in strain rate at 100°C.

High-purity, electrorefined plutonium was extruded to minimize the effects of microcracks, which are always present in as-cast high purity metal.[6] This was necessary to attain reproducible tensile elongations, especially at the lower testing temperatures. Specimens for the tensile tests were cylindrical with 0.50 in. (1.25 cm) gage length and 0.125 in. (0.32 cm) diam. The compression specimens were also cylindrical with 0.40 in. (1.0 cm) gage length and 0.20 in. (0.50 cm) diam. Some contribution to the difference in tensile and compressive strength at the same crosshead speed was attributed to the gage length difference. The error in the ratio σ_t/σ_c due to this strain rate effect was no greater than 6 pct, and in most cases was smaller as estimated from the observed dependence of tensile and compressive flow stresses on strain rate in the present work.

The difference in compressive and tensile yield stress (1 pct offset) was 46,000 psi at 22°C and 0.002 in./min, Fig. 1, Table I. A large difference in flow stress persisted over the entire stress strain curve. With increased temperatures, at the same rate, the strength differential decreased, Table I. However, at the highest testing speed, 0.2 in./min, the effect was still appreciable at 100°C, Table I, Fig. 2. The apparent decrease in tensile stress beyond about 0.2 strain was attributed to gradual necking of the specimen.

Several possible causes for the strength differential in steels and cast iron have been reported, including transformation of retained phases, opening of microcracks, residual stresses, internal Bauschinger effect, and nonlinear solute-dislocation interactions.[1] Some of these are reasonably ruled out for α-Pu. Retained phases are negligible in the extruded α-Pu; moreover such phases would transform more readily under compressive stress than under tensile stress. Residual stresses due to deformation and phase transformation damage were very small as evidenced by very sharp X-ray diffraction peaks from extruded α-Pu, and such stresses could not be expected to have an effect beyond a few percent strain. The effect of impurities has not been investigated, though the total impurity content of the metal used was quite low, about 300 ppm. Microcracks are present in the microstructure of as-

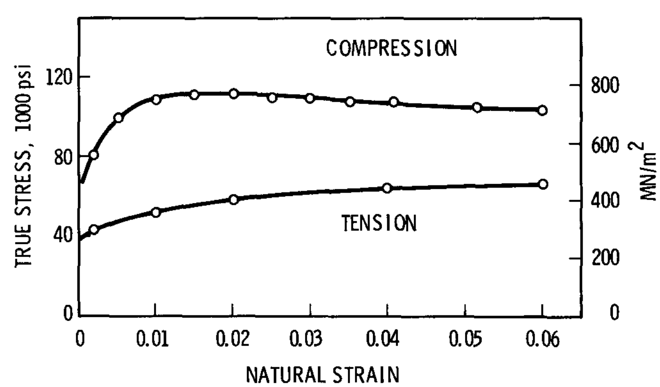

Fig. 1—Tension and compression true stress-strain curves for α-Pu at 22°C. Crosshead speed = 0.002 in./min (8.5 $\times 10^{-5}$ cm/s).

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