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⑦ Applicant: **USS ENGINEERS AND CONSULTANTS, INC.**
600 Grant Street
Pittsburgh Pennsylvania 15230(US)

⑦ Inventor: **Hu, Hsun**
3560 Mayer Drive
Murrysville Pennsylvania 15668(US)

⑦ Representative: **Ranson, Arthur Terence et al,**
W.P. Thompson & Co. Coopers Building Church Street
Liverpool L1 3AB(GB)

⑤ Method for producing high-strength deep-drawable dual-phase steel sheets.

⑦ High-strength, deep-drawable, steel sheet exhibiting dual-phase properties is produced by (i) initially annealing the sheet to achieve crystallographic textures yielding high deep drawability, (ii) heating the sheet to a temperature above the A₁ for a time sufficient to produce from 2 to 10% austenite, and thereafter (iii) rapidly cooling to transform at least a portion of the austenite to martensite or bainite. To achieve such transformation of austenite, the composition of the steel must be selected to provide a requisite degree of hardenability.

EP 0 053 913 A1

-1-

DESCRIPTION

METHOD FOR PRODUCING HIGH-STRENGTH DEEP-DRAWABLE

DUAL-PHASE STEEL SHEETS

This invention is directed to a method for the production of dual-phase steel sheet or strip products with improved deep drawability.

The past decade has seen the development of dual-phase steels, characterized by a microstructure which is basically a matrix of ferrite containing from about 10 to 30% martensite or lower bainite. The resulting microstructure provides a steel with an initially low yield strength, typically in the range of 50 to 60 ksi, thereby affording ease of formability; but which as a result of the strains developed by forming, more closely approaches the ultimate tensile strength of the steel, typically in the range of 80 to 90 ksi. A variety of thermo-mechanical processing methods have been proposed for the production of such dual-phase steels. In general, these methods involve heating the

- 2 -

steel to a temperature in the two-phase, alpha + gamma region (e.g. U. S. Patents 3,502,514 or 3,914,135); or to above the A_3 region for a brief period, avoiding undesirable grain growth (e.g. U. S. Patents 3,914,135 or 4,033,789), to produce a minimum amount of austenite which when rapidly cooled will decompose to martensite or bainite. The formability of dual-phase steels, coupled with their relatively high-strength to weight ratios, have led to the adoption of such steels for the production of automotive parts such as bumpers, wheel racks, brackets--primarily as hot-rolled products. To date, however, cold-rolled, dual-phase steels have not been employed for the production of automobile body panels, one reason being that steel sheets for body panel applications must have good deep-drawing capabilities. It is recognized that the performance of a steel sheet or strip during deep-drawing operations, is closely associated with r_m value, which is the ratio of true width strain to true thickness strain when the sheet is strained in tension. Apparently, the r_m values of dual-phase steel sheets are invariably poor, i.e. around unity (see, for example, Hayami et al., Formable HSLA and Dual-Phase Steels, Conference Proceedings, TMS-AIME, 1979, pages 167 to 180). Thus, in comparison with the

- 3 -

solution strengthened, deep-drawing steels having r_m values of the order of 1.6 to 2.0 and yield strengths of the order of 40 ksi, the presently available, dual-phase steels are much inferior. For deep drawing applications, crystallographic texture must be strongly (111) with very little (100) and other undesirable orientations. Although strong (111) textures can be produced successfully in various low-carbon sheet steels, development of an equally strong (111) texture in the ferrite + martensite aggregate would be much more difficult to accomplish. This is because it is difficult to achieve a sharp texture in martensite due to the nature of the transformation variance. It has now been found that if proper orientations [e.g. the predominant (111) texture] are first developed or provided to the steel, and if the steel is thereafter briefly intercritically annealed so that local regions with a high carbon content could become austenite pools at the intercritical temperature, these austenized regions will, upon rapid cooling, transform to martensite or bainite without unduly affecting the ferrite matrix - if the amount of martensite formed is small.

The production of a requisite minimum amount of martensite necessitates that the steel contain at

- 4 -

least about 0.02% carbon in uncombined form; whereas
if the carbon is present in uncombined form, in
amounts significantly in excess of 0.15%, the desired
deep drawability will not be achieved. Similarly,
5 good deep drawability will not be obtained if
manganese is in excess of 0.4% (preferably less than
0.3%), or if oxygen is in excess of 500 ppm
(preferably less than 300 ppm). Steel compositions
within the above parameters known to provide good deep
10 drawability, are phosphorus containing or
silicon-phosphorous containing steels described in
U.S. Patent 3,954,516, as well as certain of the
compositions described in U.S. Patent 3,827,924, the
disclosures of which patents are hereby incorporated
15 by reference. Examples of three such steels were
treated in the laboratory to provide an illustration
of the properties which could be achieved. The
chemical composition of these illustrative steels is
provided in Table I below.

TABLE I

Chemical Composition of Air-Melted Silicon-Phosphorus Steels, Wt.%

<u>Steel</u>	<u>C</u>	<u>Mn</u>	<u>P</u>	<u>S</u>	<u>Si</u>	<u>Cu</u>	<u>Ni</u>	<u>Cr</u>	<u>Al Sol</u>	<u>Al Total</u>	<u>N</u>	<u>Oppm</u>
A	0.038	0.135	0.066	0.018	0.408	0.007	0.005	0.018	0.001	0.002	0.006	251
B	0.040	0.134	0.068	0.017	0.724	0.010	0.002	0.018	0.001	0.004	0.006	223
C	0.044	0.154	0.070	0.017	1.071	0.010	0.002	0.020	0.001	0.004	0.006	178

- 6 -

Initially, samples were given a simulated box-annealing treatment at 780°C for a period of four hours. Although the use of such a shortened soak period will result in some sacrifice of deep-drawing texture, it was deemed to be preferential, in order to reduce grain size and thereby attain a more uniform distribution of martensite during the subsequent transformation. After this four-hour hold at temperature, the specimens were cooled by removing them from the hot zone to the colder zone of the furnace. This simulated an accelerated cooling rate attainable in box-annealing and was intended further to reduce the size of the carbides or pearlite colonies at the grain boundaries. Thus, it was hoped that, upon subsequent intercritical annealing, small austenite pools would form at these regions and thereby be more uniformly distributed throughout the ferrite matrix. Subsequently, to simulate continuous annealing, selected sets of specimens were heat treated in a lead bath at 740°C, 760°C, and 780°C for one minute and thereafter quenched. The resulting mechanical properties of these specimens are reported in Table II.

As may be noted from the tabulation under treatment (2) in Table II, the intercritical heat

- 7 -

treatment at 740°C followed by quenching in water resulted in substantial increases in the yield and tensile strengths and a corresponding decrease in ductility. The yield point elongation or Luders strain in the quenched specimens was appreciably smaller, even though the grain size was slightly smaller and the amount of martensite produced was only about 1%. Although r_m values decreased slightly, i.e. to about 1.7, it should nevertheless be noted that at such high strength levels (71 to 76 ksi yield strength and 95 to 100 ksi tensile strength) r_m values at a level of about 1.7 are quite impressive.

For the specimens heat treated at higher intercritical temperatures, i.e. 760° and 780°C, mechanical properties of microstructural features are summarized under treatments (3), (4) and (5) in Table II. It can be noted that the yield point elongation was completely eliminated in specimens having higher silicon content, i.e. steels B and C. The amount of martensite in these specimens was substantially higher being about 3 to 4%, than those specimens quenched from 740°C. For Steel A which contained 0.41% silicon, no change of the yield point elongation was observed when the heat treating temperature was increased to 760 or 780°C and the amount of

- 8 -

martensite showed only insignificant variations. The r_m values, in particular, for the steels having relatively high silicon contents, were in the range 1.7 to 1.85. The yield and tensile strengths of these specimens were in the range 55 to 65 ksi and 75 to 100 ksi, respectively. The change in initial strain hardening rate, resulting from the intercritical heat treatment can be noted by the increase in flow stress after 4% elongation (column labeled " $\sigma_{0.04}$ "). With the presence of yield point elongation of about 1 to 2%, the increase in flow strength at 4% elongation ranged from 10 to 14 ksi. When the yield point elongation was completely eliminated, the corresponding increase in flow stress ranged from 17 to 20 ksi. Nevertheless, in comparison with properties characteristic of dual-phase steels having substantially enriched alloy compositions and greater volume fractions of martensite, the initial strain hardening rate and the total elongation were much lower in the present specimens. However, both ductility and initial strain-hardening rate could be improved by changing the steel composition to increase the hardenability, for example, by utilizing elements such as nickel, which would have little or no effect on the annealing texture or the r_m value. Such

increased hardenability would enable the employment of less severe cooling rates from the intercritical anneal, thereby minimizing supersaturation and quenched aging of the ferrite matrix.

TABLE II
PROPERTIES OF SILICON-PHOSPHORUS STEELS, HEAT-TREATED
AT VARIOUS INTERCRITICAL TEMPERATURES THEN QUENCHED

<u>Steel</u>	<u>YS*</u> <u>ksi</u>	<u>00.04</u> <u>ksi</u>	<u>TS</u> <u>ksi</u>	<u>Yp El</u> <u>%</u>	<u>Unif El</u> <u>in 1 in., %</u>	<u>Tot El</u> <u>in 1 in., %</u>	<u>r_m</u>	<u>α, %</u>	<u>G.S.*</u> <u>μm</u>
Simulated box-annealed to 780°C (1436°F), held at temp for 4 hr, cooled in cool zone, then									
1. No additional Heat Treatment									
A	41.5	45.4	59.8	3.0	25.8	33.7	1.94	0	17
B	46.5	50.7	65.1	3.0	25.5	34.2	1.78	0	17
C	49.5	53.4	67.8	3.2	25.9	33.6	1.89	0	17
2. Heat-treated in Lead Bath at 740°C (1364°F) for 1 min, Water-quenched									
A	71.4	84.6	94.7	1.1	13.3	15.6	1.66	1.0	16
B	76.0	89.2	100.4	1.0	13.0	16.5	1.73	1.0	16
C	73.4	84.5	96.0	1.7	14.3	16.4	1.77	1.0	16
3. Heat-treated in Lead Bath at 760°C (1400°F) for 1 min, Water-quenched									
A	55.5	67.4	77.5	1.0	12.9	17.1	1.52	2.0	18
B	59.8	78.1	89.9	0	13.9	17.3	1.75	4.0	17
C	64.2	83.3	95.1	0	14.1	16.2	1.71	4.0	17

(Continued)

TABLE II (Continued)

Steel	YS* ksi	00.04 ksi	TS. ksi	Yp El %	Unif El in 1 in., %	Tot El in 1 in., %	r _m	α' %	G.S. μm
4. Heat-treated in Lead Bath at 760°C (1400°F) for 1 min, Oil-Quenched									
A	57.4	67.5	76.5	1.1	13.6	16.5	1.46	1.0	16
B	59.0	76.2	88.1	0	14.7	18.9	1.85	2.8	16
C	67.7	87.2	99.3	0	12.8	15.2	1.75	2.6	16
5. Heat-treated in Lead Bath at 780°C (1436°F) for 1 min, Oil-quenched									
A	62.0	76.2	86.0	1.1	12.3	14.0	1.73	1.7	17
B	64.6	83.7	95.5	0	14.2	17.9	1.85	3.7	17
C	67.4	86.3	97.9	0	15.1	18.7	1.69	3.7	17

* Lower yield strength or 0.2% offset

Conversion Factor:

1 ksi = 6.895 MPa

- 12 -

Although box annealing has most widely been employed for the development of good deep-drawing properties in low carbon sheet steels; annealing with relatively rapid heating rates, such as in continuous annealing, can produce equally satisfactory results if (i) the prior hot processing conditions (such as the finishing and coiling temperature which influence the carbide size and the distribution in the hot rolled band), and (ii) the amount of subsequent cold rolling reduction are appropriately adjusted. Thus, as shown by Matsudo et al., Texture of Crystalline Solids, Volume 3, 1978, pages 53 to 72, when the carbide size is comparatively large, the r_m value increases with increasing heating rate. Thus, annealing to produce a high r_m value and the subsequent intercritical heat treatment for the production of from about 2 to 10% austenite can be accomplished in one continuous anneal -- eliminating the separate intercritical heat treatment utilized in the aforementioned examples. In either case, the steel sheet will be annealed to provide a crystallographic texture capable of yielding an r_m value greater than 1.5 and preferably greater than 1.7. The next heating phase, whether performed as part of the initial anneal for texture formation or as a discrete step, will be conducted for a time and

- 13 -

temperature (preferably within the range A_1 to A_3)
sufficient to produce from about 2 to 10% austenite,
preferably less than 7% austenite. Thereafter, the
sheet will be cooled at a rate sufficient to transform
5 all or a major portion of the austenite to martensite
or bainite--the most preferred range of such
decomposition products being about 3 to 5%.

Claims:-

1. A method for the production of a steel sheet which exhibits a yield strength/tensile strength ratio less than 0.75, a yield strength increase of at least 10 ksi after straining 4% and an r_m value greater than 1.4, which comprises

providing a sheet steel with an r_m value greater than 1.5,

heating said sheet to a temperature above the A_1 thereof for a time and temperature sufficient to produce from 2 to 10% austenite therein,

cooling the sheet at a rate sufficient to transform at least a major portion of said austenite to decomposition products selected from the group of martensite, lower bainite, or combinations thereof, so as to produce such decomposition products in an amount of 2 to 10%, and

in which said steel has a composition consisting essentially of 0.02 to 0.15% uncombined carbon, less than 0.4% manganese and less than 500 ppm oxygen, the total of the alloy elements therein being sufficient to provide a degree of hardenability effective to transform said austenite to said amount of decomposition products.

2. The method of claim 1 wherein said sheet

- 15 -

is heated to a temperature below the A_3 temperature thereof, for a time sufficient to produce less than 7% austenite.

3. The method of claims 1 or 2 in which said
5 sheet is cooled at a rate sufficient to produce said decomposition products in an amount of 3 to 5%.

4. The method of claim 3 wherein said decomposition product is martensite.

5. The method of claim 1 wherein the sheet
10 so provided has an r_m value greater than 1.7, is heated to a temperature within the range A_1 to A_3 for a time sufficient to produce from 2 to 10% austenite and is thereafter cooled to transform said austenite to 3 to 7% martensite.

15 6. A method for the production of a sheet steel which exhibits a yield strength to tensile strength ratio less than 0.75, a yield strength increase of at least 15 ksi after straining 4% and an r_m value greater than 1.5, which comprises

20 continuously annealing a steel sheet which has been cold reduced at least 50%, by heating said sheet to a temperature above the A_1 thereof but below the A_3 thereof to recrystallize the grains therein to an extent sufficient to provide a texture
25 which upon cooling would yield an r_m value of at

- 16 -

least 1.5, said time and temperature being sufficient to produce 2 to 10% austenite therein, and

cooling the sheet at a rate sufficient to transform at least a major portion of said austenite to decomposition products selected from the group of martensite, bainite or combinations thereof and produce such decomposition products in an amount of 2 to 10%, and

in which said steel has a composition consisting essentially of 0.02 to 0.15% uncombined carbon, less than 0.4% manganese and less than 500 ppm oxygen, the total of the alloy elements therein being sufficient to provide a degree of hardenability effective to transform said austenite to said amount of decomposition products.



DOCUMENTS CONSIDERED TO BE RELEVANT			CLASSIFICATION OF THE APPLICATION (Int. Cl. 3)
Category	Citation of document with indication, where appropriate, of relevant passages	Relevant to claim	
X	<u>GB - A - 2 028 690</u> (KAWASAKI) * the entire document * --	1-6	C 21 D 9/48 8/04
X	<u>WO - A - 79/00644</u> (INLAND STEEL) * claim 1; page 16 * --	1-6	
X	<u>FR - A - 2 236 945</u> (NIPPON KOKAN) * the entire document * & <u>US - A - 3 904 446</u> --	1,6	TECHNICAL FIELDS SEARCHED (Int.Cl. 3)
A,D	<u>US - A - 4 033 789</u> (E.G. HAMBURG et al.)		C 21 D
A,D	<u>US - A - 3 827 924</u> (H. TAKECHI et al.)		
A	<u>FR - A - 2 286 205</u> (USS ENG. & CONS.)		CATEGORY OF CITED DOCUMENTS
A,D	& <u>US - A - 3 954 516</u>		
A	<u>FR - A - 2 176 130</u> (NIPPON KOKAN)		X: particularly relevant if taken alone Y: particularly relevant if combined with another document of the same category A: technological background O: non-written disclosure P: intermediate document T: theory or principle underlying the invention E: earlier patent document, but published on, or after the filing date D: document cited in the application L: document cited for other reasons
A,D	& <u>US - A - 3 914 135</u>		
A	<u>DE - A - 1 904 162</u> (US STEEL)		
A,D	& <u>US - A - 3 502 514</u> -----		&: member of the same patent family, corresponding document
The present search report has been drawn up for all claims			
Place of search		Date of completion of the search	Examiner
The Hague		17-03-1982	MOLLET